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Summary of Modified 9Cr-1Mo Steel Development Program: 1975–1985

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MARTIN MARIETTA ENERGY SYSTEMS, INC.
FOR THE UNITED STATES
DEPARTMENT OF ENERGY

Printed in the United States of America. Available from
the U.S. Department of Energy
Technical Information Center
P.O. Box 62, Oak Ridge, Tennessee 37830

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ORNL-6303
Distribution
Categories UC-79Th,
-Tk, -Tr

Metals and Ceramics Division

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Date Published - October 1986

[redacted]

Prepared for
Office of Technology Support Programs
U.S. Department of Energy

Prepared by the
OAK RIDGE NATIONAL LABORATORY
Oak Ridge, Tennessee 37831
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MARTIN MARIETTA ENERGY SYSTEMS, INC.
for the
U.S. DEPARTMENT OF ENERGY
under contract DE-AC05-84OR21400

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FOREWORD AND ACKNOWLEDGMENTS

This report documents the planning, testing, and technology transfer activities that have been carried out in developing modified 9Cr-1Mo steel (T-91) for liquid-metal reactor (LMR) applications. Successful development of a new material for critical high-temperature energy applications has required considerable effort, coordination, and staying power. In such an endeavor important contributions have been made by many organizations and individuals, and it would be impossible to acknowledge them all. However, several deserve special mention. First of all, the alloy composition was the brainchild of C. T. Ward at Combustion Engineering, Inc. R. T. King led the Oak Ridge National Laboratory (ORNL) Task Group on development of alternate materials for advanced LMR applications and was the first modified 9Cr-1Mo steel program manager. P. Patriarca provided needed direction and leadership of the program at ORNL through much of the 1975 to 1985 period. Staying power was provided chiefly through the support of the U.S. Department of Energy (DOE) Office of Nuclear Energy and in particular through the efforts of E. E. Hoffman, DOE/Oak Ridge, and C. C. Beals, DOE Headquarters, who made specific recommendations for financial support and provided enthusiastic encouragement. Continued technical justification of the program was provided by various industrial companies interested in advanced LMR development; of particular note were the efforts of C. L. Storrs at Combustion Engineering. Undoubtedly, however, the most significant contribution was that of V. K. Sikka, who, as task leader throughout most of the program, tested, coordinated, and promoted the alloy. Without his effort and expertise the program could not have succeeded.

Although the report contains much detailed information (including over 160 figures and 200 references), we have also included an Executive Summary and a Summary and Recommendations (Chapter 4) to more succinctly describe the program. The sheer volume of material to be included presented a special problem for Karen Perry, who was responsible for providing a typed draft for review. Special thanks are due for her patience and perseverence.

J. R. DiStefano
LMR Program Manager

SUMMARY OF MODIFIED 9Cr-1Mo STEEL ALLOY DEVELOPMENT PROGRAM: 1975-1985*

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EXECUTIVE SUMMARY

In 1974 a task force was authorized by the U.S. Department of Energy (DOE) to conduct a three-phase study directed at (1) evaluating the deficiencies in types 304 and 316 stainless steel for liquid metal reactor (LMR) applications, (2) selecting candidate alloys with potential for improved performance, and (3) developing a plan for acquisition of the data base that would be required for use of the alloy in a future LMR. Responsibility for the study was assigned to a task force at Oak Ridge National Laboratory (ORNL), and the findings of the task force ultimately led to a recommendation for developing an improved ferritic steel. The projected costs for developing a ferritic steel alloy by 1985 were estimated to be $17.74 million, and the recommended alloy development program was initially justified by forecasts of a significant LMR economy by the 1980s. After consideration of several ferritic alloy compositions, DOE selected modified 9Cr-1Mo steel for further development.

Based on creep strength and microstructural considerations, many 9Cr-1Mo types of steels were considered promising; however, alloy development studies carried out by Combustion Engineering, Inc., led to an alloy modified by the addition of vanadium and niobium. By 1980 the alloy composition was fixed, and criteria (melting, fabrication, mechanical properties, thermophysical properties, heat treatment, weldability, environmental effects, and high-temperature design behavior) that the alloy would have to meet were established. We recognized that appropriate

*Research sponsored by the Office of Technology Support Programs, DOE/RSDT AF 20 10 20 2, BRIM05, U.S. Department of Energy, under contract DE-AC05-84OR21400 with Martin Marietta Energy Systems, Inc.

codes and standards, as well as significant industrial experience, would also be required before a new material would be used for important structural applications in an LMR or, for that matter, in any high-temperature pressure boundary application. In general, the strategy to achieve these codes, standards, and industrial experience consisted of developing a strong data base in the DOE program and concomitantly involving both U.S. and foreign industry in a wide variety of developmental activities. Technology transfer among governmental and industrial organizations has been an important part of the modified 9Cr-1Mo steel development program.

A national program plan was first published in 1980 and then revised in 1982. In addition to identifying all of the program task elements, the plan provided detailed milestone schedules for each element and test matrices for the required design data.

Since 1980 considerable information on modified 9Cr-1Mo steel has been generated both through government-sponsored work and through industrial programs. These can be summarized as follows.

MELTING AND FABRICATION

Over 800 tons of material have been melted and processed, and commercial procedures are well established. The alloy is now advertised as available from B&W Tubular Products Group, Nippon Kokan K.K., and Sumitomo Metals Industries. In applications where toughness properties are critical, electroslag-remelted material is recommended.

MECHANICAL PROPERTIES

An extensive mechanical properties data base on modified 9Cr-1Mo steel in air has been developed. Data packages containing information on tensile, creep, fatigue, and toughness properties have been approved by the American Society of Mechanical Engineers (ASME) Code. However, additional data on creep-fatigue damage interactions are still required.

Modified 9Cr-1Mo steel is resistant to swelling from irradiation, and effects of irradiation on tensile and toughness properties have been determined. Based on existing data and data from other ferritic alloys, modified 9Cr-1Mo should have good long-term performance for out-of-core

structural applications in an LMR. However, further studies are recommended to provide data at the low irradiation damage levels relevant to its intended structural application.

Limited testing during and after exposure to sodium does not indicate significant deleterious effects on the mechanical properties of modified 9Cr-1Mo steel. Testing for longer times is recommended, however, to better define the effects of sodium on creep-rupture properties.

JOINING BEHAVIOR

Studies have indicated no major problems with the weldability of modified 9Cr-1Mo steel. Differences in microhardness between base metal and a tempered zone in modified 9Cr-1Mo steel are about the same as for other ferritic steels. However, results indicate that tempering at 621°C before welding followed by a postweld heat treatment at 760°C will essentially eliminate hardness variations across the weldment.

The gas tungsten-arc, submerged-arc (SA), and shielded metal-arc processes have been used to satisfactorily join modified 9Cr-1Mo steel. Only two areas are recommended for further work. A flux that can be used in SA welding with alternating current is needed. Slag removal from the weld bead has been difficult, and low Charpy-impact properties have resulted with the commercial fluxes used thus far. In addition, more information is needed on the mechanical properties of weldments, especially the effects of long-term aging.

STRUCTURAL DESIGN

LMR components present unique structural design requirements because of the significant thermal transient loadings and temperatures where time-dependent deformation processes can occur. Because specific failure criteria and design rules are not available for modified 9Cr-1Mo steel, the task has been to adopt, to the extent possible, the existing methodology for austenitic stainless steel and 2.25Cr-1Mo steel, while at the same time recognizing and accounting for the unique behavioral features of modified 9Cr-1Mo steel. A preliminary analytical assessment of the design characteristics of modified 9Cr-1Mo steel was carried out relative to other alloys (type 304 stainless steel, 2.25Cr-1Mo steel, alloy 800, and

standard 9Cr-1Mo steel); the modified steel proved to have lower thermal loadings and better resistance to creep damage. However, much work remains to be done. The present NE F 9-5T constitutive framework is inadequate, and its behavior appears to require a nonlinear viscoplastic model that makes no distinction between creep and plastic strain. Only a few scoping studies have been conducted as a basis for failure criteria assessments. Behavioral features of modified 9Cr-1Mo that will likely have to be given special consideration in developing ratchetting and creep-fatigue rules include strain softening, low strain at onset of tertiary creep, and rate dependence of yielding and plastic flow. Material behavior modeling tests are planned to permit validation of design procedures.

CORROSION BEHAVIOR

Corrosion studies of modified 9Cr-1Mo steel have included stress corrosion cracking in caustic and chloride environments, tests in sodium, and steam and air oxidation tests. Although testing in all of these environments is limited, the results can be compared with the much greater data base available for 2.25Cr-1Mo steel and standard 9Cr-1Mo. Results generally indicate that stress corrosion cracking under LMR service conditions is unlikely, no significant changes are expected from carburization/decarburization reactions, and steam corrosion rates should be slightly better than those of 2.25Cr-1Mo steel.

INSPECTION

Only limited efforts have been made to apply nondestructive examination methods to modified 9Cr-1Mo; however, prior work on a similar material (2.25Cr-1Mo) provides confidence that present technology can be adapted to the alloy. Development of ultrasonic and eddy-current techniques is presently under way to allow in-place inspection of the intermediate heat exchanger and steam generator tubing of the Sodium Advanced Fast Reactor. A major task is to develop ultrasonic and eddy-current probes that will withstand the hot standby temperature of ~200°C. It is anticipated that the inspection techniques and instrumentation systems required will be

similar to those developed for the Westinghouse double-wall steam generator after modifications for geometry, temperature, and material characteristics.

TECHNOLOGY TRANSFER

One objective of the modified 9Cr-1Mo steel development program has been to bring the alloy to a state of commercialization that will allow it to be used in advanced LMR applications. Thus, technology transfer is an important element of the program. Technology transfer activities have consisted of establishing codes and standards [American Society for Testing and Materials (ASTM), ASME, and *Nuclear Systems Materials Handbook* (NSMH)]; providing material and components for testing in prototypical end-use environments; and coordinating cooperative test programs with universities, other national laboratories, and industrial companies.

ASTM specifications have been established for all product forms of modified 9Cr-1Mo steel except bar and castings. ASME Code Cases 1943 and 1992 (Sect. I) and 1973 (Sect. VIII) have been approved, and an inquiry has been requested by Rockwell International Corporation for revisions to Cases N-47 and N-253 of Sect. III to permit use of modified 9Cr-1Mo steel. Baseline tensile and creep properties of modified 9Cr-1Mo steel have been published in DOE's NSMH.

Tube sections of modified 9Cr-1Mo steel have been installed in operating steam power plants in the United States, England, Canada, and Japan. The longest time has been achieved in the Tennessee Valley Authority's Kingston Steam Plant (6 years), and some samples are scheduled for removal and examination in 1987. Several test components have been operated in sodium, including 9Cr-1Mo steel spoolpieces in the Experimental Breeder Reactor (EBR)-II secondary sodium Yardline circuit and in the Sodium Component Test Loop at the Energy Technology Engineering Center. ORNL has also supplied the Oil Well Division of United States Steel Corporation with material for switch valves that are being evaluated for service at 510°C.

One of the most successful examples of government-to-industry technology transfer has been the cooperative programs for the exchange of materials, results, etc. among ORNL and universities and other U.S. and foreign private companies.

After 10 years of development, modified 9Cr-1Mo steel is a commercially available material that has been specified for use in an advanced LMR and is being used worldwide in fossil plants and other applications. Although there is some work needed before this steel is actually used in a nuclear system, the main objectives of the program have been essentially achieved at a cost less than originally estimated.

1. INTRODUCTION

Nuclear plants for electrical power production have traditionally required large capital investments in component equipment. Costs of such installed components in light-water reactors have been estimated to be as high as 20 to 25% of the total cost of power generation. Reductions in these costs are possible through design simplifications and material improvements.

In 1974 the principal reference materials for structural applications in liquid metal reactors (LMRs) were types 304 and 316 stainless steel (cladding, cold leg and hot leg piping, and ducts and pressure vessels), 2.25Cr-1Mo steel (steam generator), and alloy 718 (special applications, such as upper internals). There were sound technical justifications for the selection of these materials, and their widespread use for a range of structural applications had generated a large data base and significant industrial experience. However, there were some recognized problems that provided an incentive to seek materials with even better properties. Some of these problems are

1. susceptibility of types 304 and 316 stainless steel to stress corrosion cracking,
2. wide variability in the properties of types 304 and 316 stainless steels because of allowable compositional specifications,
3. high thermal stresses in austenitic stainless steels because of high coefficient of thermal expansion and low thermal conductivity,
4. difficulties with in-service inspection of austenitic stainless steel weld joints,
5. atmospheric rusting of 2.25Cr-1Mo steel,
6. low design stress allowables for 2.25Cr-1Mo steel, and
7. difficulties with austenitic to ferritic weld transition joints.

1.1 DOE TASK FORCE STUDY AND RECOMMENDATIONS

To take advantage of past experience and new developments in materials technology, the U.S. Department of Energy (DOE) established a task force at the Oak Ridge National Laboratory (ORNL) to conduct a study that would lead to the recommendation of alternate structural materials for liquid metal fast breeder reactor (LMFBR) applications. The objectives of the study, which was divided into three phases, were to (1) evaluate the deficiencies of types 304 and 316 stainless steel, (2) select candidate alloys with the promise of improved performance, and (3) develop a plan for acquisition of the data base that would support the use of an improved alloy in future LMFBRs. The task force, with considerable help from U.S. industry, selected three classes of alloys as prime candidates for future LMFBR application: ferritic steels, austenitic steels, and nickel-bearing solid solution alloys.[1] Within the three classes of alloys, hundreds of specific alloys were evaluated according to the following criteria: corrosion resistance, fabricability, weldability, mechanical properties, engineering code requirements, design costs, and availability. As a result, 12 alloys were recommended for further investigation and development. The most promising alloys from this critical evaluation were as follows:

Alloy class		Recommendation
Ferritic steel	High priority:	Jethete M154 and Timken modified 9Cr-1Mo;
	Low priority:	Sandvik HT9, CRBN R-8, and EM-12
Austenitic steel	High priority:	B&W 15-15N, Sandvik 12R72, and stabilized types 304 and 316 stainless steel
Nickel-bearing solid solution alloys	High priority:	Alloy 800H and alloy 800;
	Low priority:	Alloy 600

It was concluded that the ferritic steels could solve or alleviate all of the problems cited above for types 304 and 316 stainless steel. The development schedule recommended by the task force to make a ferritic steel alloy available for design of an LMFBR by 1985 is shown in Table 1, and the costs required to complete the eight tasks were estimated to be $17.74 million. After further consideration of the several alloys listed above, DOE selected modified 9Cr-1Mo steel for further development.

Table 1. Ferrite Steel Alloy Development Program

TASK	1976	1977	1978	1979	1980	1981	1982	1983	1984	1985
	FISCAL YEAR									
A. Program Guidelines and Management Objectives										
1. Complete evaluation and program planning studies by Task Force on Alternate Structural Materials.	▽									
2. Complete preliminary scoping survey and eliminate unsuccessful candidate materials and thermomechanical treatments; continue any long-time evaluations on materials of good potential.			▽							
3. Complete "small heat" evaluation study and eliminate unsuccessful candidate materials and thermomechanical treatments; continue any long-time evaluations on materials of good potential.				▽						
4. Complete "large heat" evaluation study; continue any long-time evaluations on materials of good potential.								▽		
5. Complete correlation of evaluation results and prepare data packages for *Nuclear Systems Materials Handbook* and submission to ASME Section III.									▽	

Table 1. (continued)

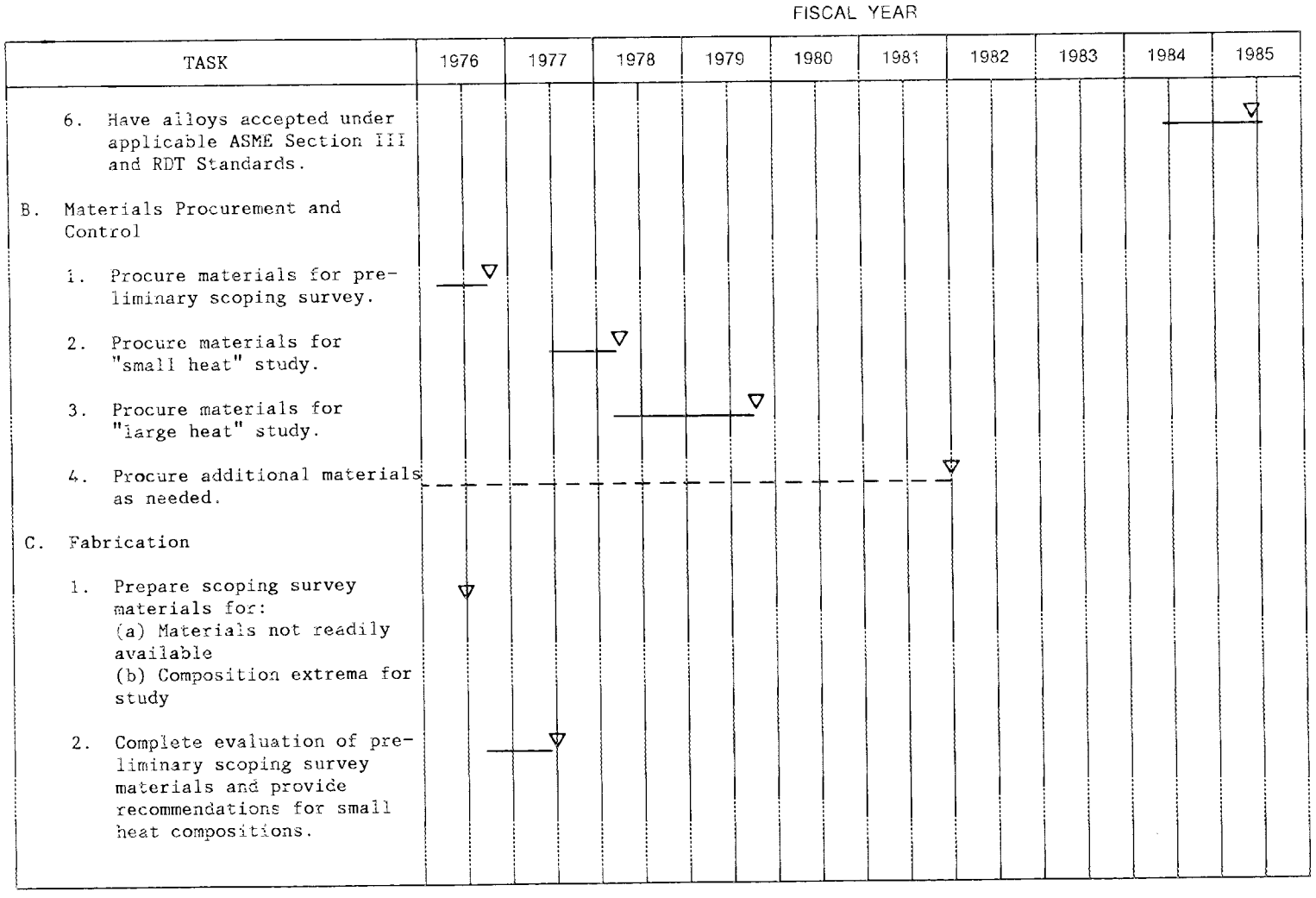


Table 1. (continued)

FISCAL YEAR

TASK	1976	1977	1978	1979	1980	1981	1982	1983	1984	1985
3. Complete survey of prior fabrication history and industrial capability for making large heats and product forms.										
4. Provide recommendations for large heat procurement.										
5. Complete evaluation of ferritic steels by following fabrication of large heats from melting through product form manufacturing and fabrication of components for testing.										
D. Welding										
1. Complete review and evaluation of existing weldability information.										
2. Complete evaluation of weldability of scoping survey materials and provide recommendations for filler metal procurement for small heat study. Continue weldability studies as needed.										

Table 1. (continued)

FISCAL YEAR

TASK	1976	1977	1978	1979	1980	1981	1982	1983	1984	1985
3. Provide scoping survey weldment specimens for evaluation under subtasks C, D, E, F.										
4. Provide recommendations for filler metal procurement for "large heat" study.										
5. Complete "small heat" weldability study.										
6. Complete "large heat" weldability study, including provision of assistance for fabrication of components for testing.										
E. Mechanical and Physical Properties										
1. Complete review of literature and industrial data.										
2. Complete short-time evaluation of scoping survey materials; continue any long-time testing, aging, and evaluations on materials of good potential.										
3. Complete short-time evaluation of "small heat" materials; continue any long-time testing, aging, and evaluations on materials of good potential.										

Table 1. (continued)

FISCAL YEAR

TASK	1976	1977	1978	1979	1980	1981	1982	1983	1984	1985
4. Complete short-time evaluation of "large heat" materials; continue any long-time testing, aging, and evaluations on materials of good potential.										
5. Complete evaluation and correlation of all results and submit data packages for *Nuclear Systems Materials Handbook* and ASME Section III.										
F. Corrosion and Environmental Effects										
1. Review existing data on ferritic steels in sodium and related environments; interface with Steam Generator and Advanced Cladding Corrosion Programs.										
2. Perform preliminary testing of scoping survey materials in existing sodium corrosion loop systems and industrial environments.										
3. Construct special facilities for testing ferritic materials.										

Table 1. (continued)

FISCAL YEAR

TASK	1976	1977	1978	1979	1980	1981	1982	1983	1984	1985
4. Complete testing of "small heat" materials in ferritic corrosion loop systems and industrial environments as required.										
5. Complete testing and evaluation of "large heat" materials and develop design data for *Nuclear Systems Materials Handbook*. Complete testing in industrial environments as required.										
G. Design Methods and Criteria										
1. Specify and perform minimum set of uniaxial tests to obtain data for scoping analyses based on scoping survey materials.										
2. Complete scoping analysis of ratchetting and creep-fatigue behavior.										
3. Specify and perform uniaxial exploratory tests for interim constitutive equations based on "small heats."										
4. Develop proposed constitutive relations and interim design guidelines.										

Table 1. (continued)

TASK	FISCAL YEAR 1976	1977	1978	1979	1980	1981	1982	1983	1984	1985
5. Perform minimum set of multiaxial tests and simple structure tests to evaluate interim methods and identify shortcomings.										
6. Define and perform additional uniaxial exploratory tests, multiaxial tests, and simple structure tests.										
7. Provide verified or improved constitutive relations.										
8. Test more realistic component-like structure and analyze to verify analysis methods.										
9. Specify and perform failure tests to develop and verify design criteria.										
10. Submit design methods to applicable ASME and RDT Standards-making bodies and gain approval.										
H. Component Fabrication and Testing										
1. Fabricate small, simple pressure vessel in controlled shop for sectioning and evaluation, using scoping survey materials.										

Table 1. (continued)

FISCAL YEAR

TASK	1976	1977	1978	1979	1980	1981	1982	1983	1984	1985
2. Fabricate small, simple pressure vessel in commercial shop for sectioning and evaluation, using scoping survey materials.										
3. Design system components for testing at liquid metal test facility (pump, valve, piping, and heat exchanger) according to interim design guidelines and constitutive relations, using "large heat" materials.										
4. Fabricate and ship system components for testing at liquid metal test facility.										
5. Install and test system components at liquid metal test facility.										

Source: R. T. King et al., *Report of Task Force on Alternate Structural Materials for Liquid Metal Fast Breeder Reactors*, ORNL-5076, May 1976, Table 4, pp. 138-42.

1.2 PRIOR WORK ON MODIFICATIONS OF 9Cr-1Mo ALLOYS

Over the years, considerable effort was devoted to developing steels with chromium concentrations lower than austenitic stainless steels for elevated temperature service. A compilation by G. V. Smith[2] listed data on 73 heats of 9Cr-1Mo types of steels. In the United States, Timken Roller Bearing Company developed for superheater tubing in fossil power plants a modified 9Cr-1Mo that has good properties for LMFBR applications.[3] This alloy contains additions of niobium and vanadium and in the normalized-and-tempered condition (N&T) exceeds the average yield strength and ultimate tensile strength of type 304 stainless steel to 593°C (1100°F). However, the Timken modified 9Cr-1Mo steel was never used commercially, and the alloy was not patented. Several foreign steels in this composition range have also been developed. The chemical compositions of Tempaloy F-9 (ref. 4) (Nippon Kokan), CRBN R-8 (ref. 5), and EM-12 (Vallorec) are compared in Table 2. Sandvik[6] produced a series of 7 to 9% chromium-1 to 1.5% molybdenum steels with creep-rupture properties similar to the Timken alloys. The highest strength Sandvik steel was a normalized and tempered alloy containing 9.4% Cr, 1.08% Mo, 0.53% Mn, 0.44% Nb, 0.30% Si, and 0.15% C with a martensitic microstructure.

Based on creep strength and microstructural considerations, many of these alloys were considered promising for LMR applications. However, a review of the properties of alloys developed to this point indicated that those with high creep strength exhibited either a decrease in ductility with increasing time to rupture, a high ductile-to-brittle transition (DBTT) temperature, or both. In addition, the use of any new alloy in systems for electric power generation requires that the material be fully commercialized with a significant data base and industrial experience. After optimizing the composition for LMR applications, it would be necessary to (1) demonstrate melting and fabrication of commercial heats, (2) set optimum heat treatments, (3) develop appropriate design methods and a related properties data base, (4) demonstrate weldability, (5) determine the effect of environmental interactions on properties, and (6) develop appropriate codes and standards for its intended use. In addition, significant industrial experience was considered a prerequisite before a new material would likely be specified for important structural applications in an energy

system. Thus, although the potential benefits of an improved 9Cr-1Mo steel to an LMFBR (and other industries as well) had been identified, much work remained to be done.

Table 2. A comparison of the Timken 9Cr-1Mo modified steel with similar foreign steels

Element	Characteristic chemical composition (%)			
	Timken 9Cr-1Mo modified (United States)	Nippon Kokan Tempaloy F-9 (Japan)	CNRM R-8 (Belgium)	EM-12 (France)
C	0.15 max.[a]	0.04-0.08	0.08-0.12	0.15 max.
Mn	0.30-0.60	0.40-0.80	0.9-1.2	0.8-1.3
Si	0.50-1.00	0.25-1.00	0.2-0.4	0.75 max.
Cr	8.0-10.0	8.0-9.5	9-10	8.5-10
Mo	0.90-1.10	0.90-1.10	1.8-2.2	1.7-2.3
V	0.25-0.35	0.15-0.45	0.25-0.45	0.2-0.4
Nb	0.04-0.15	0.20-0.60	0.4-0.6[b]	0.3-0.6[b]
N	0.01			
B	0.007	0.010 max.		
Zr	0.01			
P		0.03 max.	0.03 max.	0.03 max.
S		0.03 max.	0.03 max.	0.03 max.

[a]max. = maximum.

[b]Nb + Ta.

2. PROGRAM DESCRIPTION

2.1 OBJECTIVE

The objective of the modified 9Cr-1Mo steel program is to develop and bring to a state of commercialization a structural alloy with better properties for LMR and other advanced energy applications than standard austenitic stainless steels and other low-alloy ferritic steels. In general, the strategy to achieve commercialization consisted of the early involvement of industry in a wide variety of developmental activities. As discussed later, much of the success of this program can be attributed to interactive government/industry technology transfer.

2.2 CRITERIA

Although a modified 9Cr-1Mo steel was expected to have advantages over austenitic stainless steels in LMR applications, its significant strength advantage over 2.25Cr-1Mo steel was expected to provide a substantial increase in the margin of safety against steam generator failures caused by fatigue and thermal stresses. In addition, the superior strength of modified 9Cr-1Mo steel over 2.25Cr-1Mo steel was expected to result in substantial cost savings (reduced tubesheet and shell thicknesses) and improved reliability. Thus, criteria that an improved 9Cr-1Mo alloy must satisfy were first established for LMFBR steam generator application. To be acceptable, the improved alloy was required to meet criteria in the areas of melting and fabricability, mechanical properties, thermophysical properties, heat treatment, welding, environmental effects, and high-temperature design behavior. Specific criteria selected are shown in the Appendix.

2.3 TASK ELEMENTS

To meet the objective and criteria cited above, the program is organized into six task elements (Fig. 1). Within the DOE-funded program,

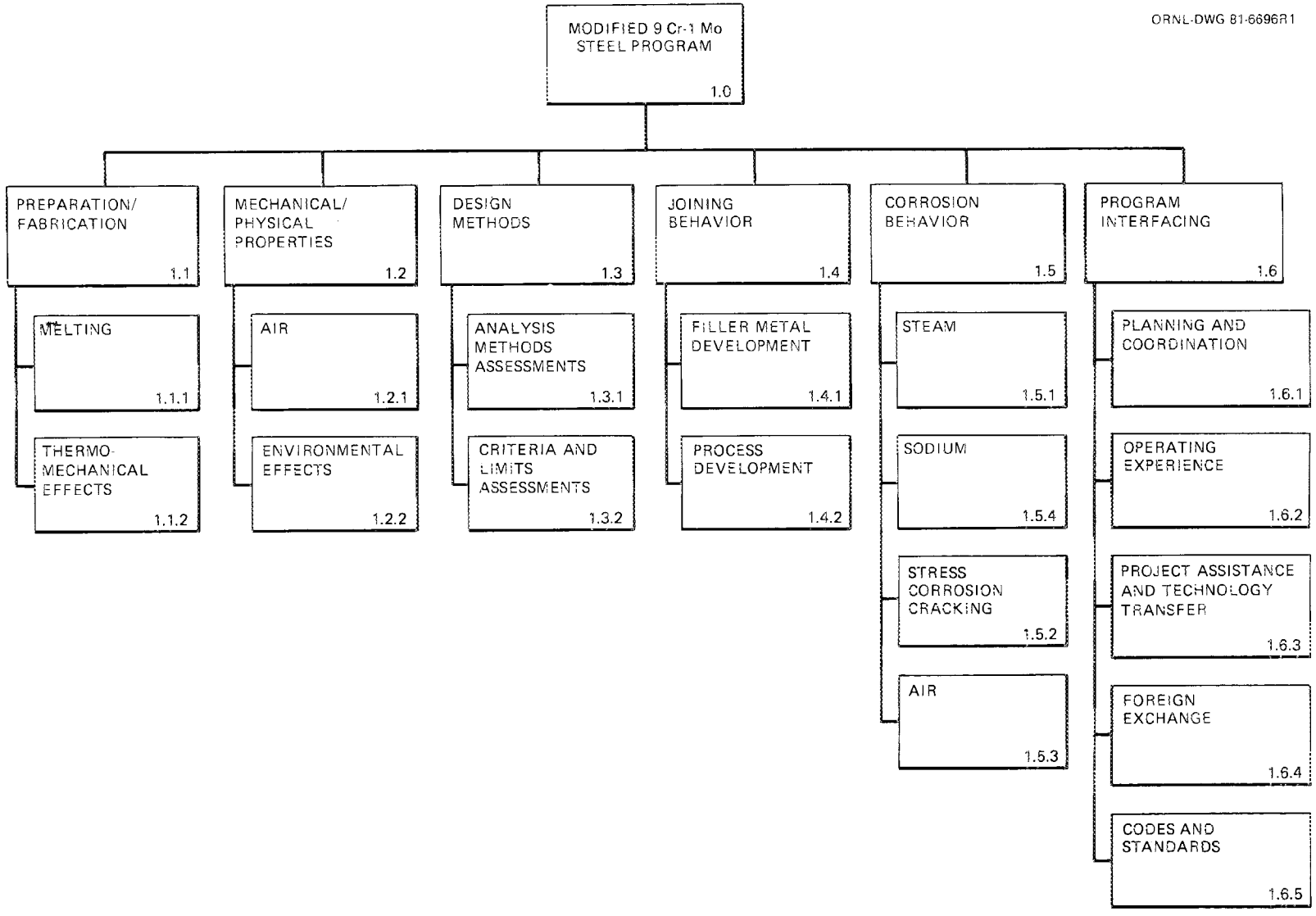


Fig. 1. Task elements for modified 9Cr-1Mo steel program.

overall management of the modified 9Cr-1Mo steel is the responsibility of ORNL, and ORNL personnel have been the principal contributors to task elements 1.1 (Preparation/Fabrication), 1.2 (Mechanical/Physical Properties), 1.3 (Design Methods), and 1.6 (Program Interfacing). Mechanical properties tests in sodium have been conducted by the Argonne National Laboratory (ANL) and the Westinghouse Advanced Energy Systems Division (WAESD). Joining behavior was studied by ORNL principally through a subcontract with Combustion Engineering, Inc. (CE), Chattanooga, Tennessee. Stress-corrosion cracking studies in caustic and chloride environments have been conducted by the General Electric Nuclear Systems Technology Organization (NSTO), and effects of sodium were determined by ANL.

2.4 PROGRAM PLAN

A national program plan for the development and commercialization of a modified 9Cr-1Mo steel was published in September 1980 (ref. 7) and revised[8] in December 1982. In addition to identifying the program objectives and task elements, the plan summarizes the advantages and limitations of modified 9Cr-1Mo steel for specific LMR component applications and serves as an interim report on the status of developmental activities and remaining needs. Detailed milestone schedules are included for each of the task elements, and Appendix D in the revised plan[8] contains a detailed set of test matrices for the required American Society of Mechanical Engineers (ASME) Code and design data on modified 9Cr-1Mo steel.

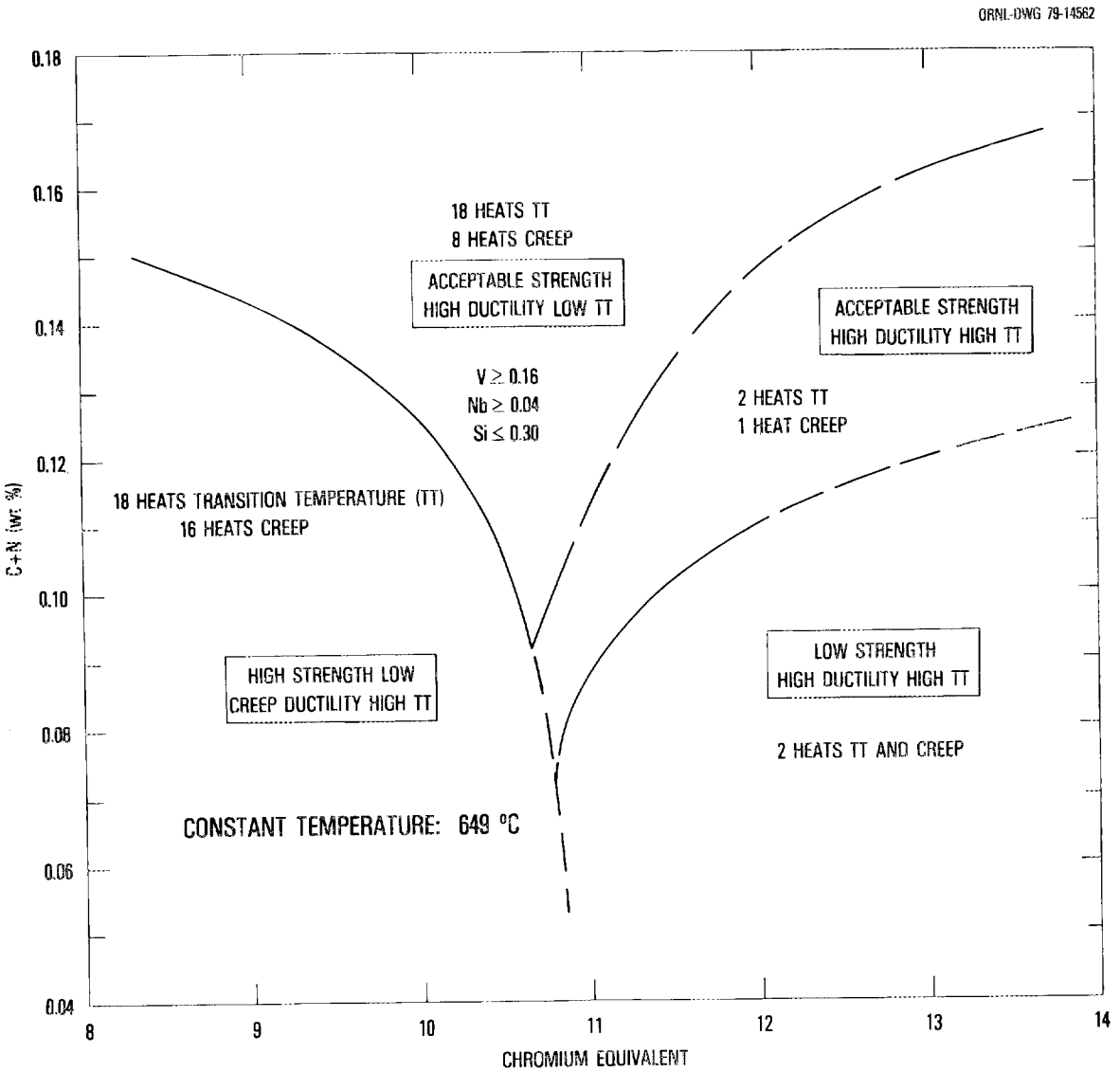
3. RESULTS AND ACCOMPLISHMENTS

3.1 SELECTION OF ALLOY COMPOSITION

Initial alloy development studies on modified 9Cr-1Mo steel for LMR applications were carried out by CE.[9] To maintain the microstructure free of δ-ferrite (reduces strength and toughness), the influence of compositional variations on microstructure and properties was evaluated. An equation for chromium equivalent developed by Newhouse et al.[10] and expanded by CE was used to design systematic alloy modifications:

$$\begin{aligned} \text{Cr eq.} = {} & \text{Cr} + 6\text{Si} + 4\text{Mo} + 1.5\text{W} + 11\text{V} + 5\text{Nb} + 8\text{Ti} + 12\text{Al} \\ & - 40\text{C} - 30\text{N} - 4\text{Ni} - 2\text{Mn} - 1\text{Cu}. \end{aligned} \tag{1}$$

Because of a previous report that chromium contributes little to strengthening iron base alloys,[11,12] a nominal chromium concentration of 9% was maintained in the various experimental alloys. Argent et al.[12] had also shown that molybdenum, though a very effective strengthening element, appears to saturate the dislocation networks in ferritic steels at ~1% and further additions of molybdenum would not produce a significant increase in strength. Thus, systematic additions of C, Mn, P, S, Si, V, Nb, Ti, Al, B, Zr, and N were made to a nominal 9Cr-1Mo alloy; Ni, Co, Cu, W, As, Sn, and O concentrations were controlled at low levels. Tensile and impact tests were conducted, hardness measurements were made, and some metallographic examinations were conducted on these laboratory heats. Results of this initial study and follow-on testing of four intermediate-sized (4-8 ton) commercial heats at both CE and ORNL[13-16] are summarized in Fig. 2. With chromium equivalent as the abscissa and percent (C + N) as the ordinate, the graph shown in Fig. 2 was constructed. Correlation of strength, ductility, and toughness properties with microstructural observations indicated that heats with the best properties contained little δ-ferrite, (C + N) ≥ 0.14%, Nb ≥ 0.04%, V ≥ 0.16%, and Si ≤ 0.30%. If the chromium equivalent was <10, δ-ferrite was usually not present, and if the



[a]Cr EQ. = % Cr + 6%, Si + 4% Mo + 1.5% W + 11% V + 5% Nb + 8% Ti + 12% Al − 40% C − 30% N − 4% Ni − 2% Mn − % Cu − 2% Co.

[b]Cr EQ. OF ≤ 10, δ IS USUALLY NOT PRESENT, 10–12 SOME δ MAY BE PRESENT, > 12 δ WILL BE PRESENT.

Fig. 2. Chromium equivalent,[a,b] C + N, combination of V and Nb, and Si contents are critical in producing heats of modified 9Cr–1Mo steel with acceptable creep strength and ductility and toughness properties.

Table 3. Commercial chemical composition specifications for modified 9Cr-1Mo steel

Element	Concentration (wt %)		Element	Concentration (wt %)	
	Target	Allowed range		Target	Allowed range
C	0.10	0.08–0.12	Nb	0.08	0.06–0.10
Mn	0.40	0.30–0.50	N	0.05	0.03–0.07
Si	0.20	0.20–0.50	Al	<0.02	0.04 max.[a]
P	<0.01	0.02 max.[a]	Ti	<0.01	0.01 max.[a]
S	<0.01	0.01 max.[a]	B	<0.001	<0.001
Cr	8.5	8–9	W	<0.01	<0.01
Ni	<0.10	0.2 max.[a]	Zr	<0.01	<0.01
Mo	0.95	0.85–1.05	O	<0.02	<0.02
Cu	<0.10	0.2 max.[a]	Sb	<0.001	<0.001
V	0.21	0.18–0.25			

[a]Max. = maximum.

chromium equivalent was >12, a significant amount of δ-ferrite was usually present. Based on these results, final commercial specifications for a modified 9Cr-1Mo steel were selected (Table 3).

3.2 MELTING AND FABRICATION OF COMMERCIAL HEATS

3.2.1 Melting

After specifying the chemical composition of the alloy, the next step was to demonstrate that it could be melted and processed commercially, as well as to provide material for generating the properties data required for American Society for Testing and Materials (ASTM) specifications and ASME Code approval. A total of 13 heats have been commercially melted, and details are summarized in Table 4. Additional information on processing is available in a paper by Bodine and McDonald.[17] Chemical analyses of all 13 heats are presented in Table 5. Although no special problems were encountered, some of the heats missed the allowed range specified for some elements. Heats that did not meet specifications are described below.

Table 4. Details on melting of commercial heats of modified 9Cr-1Mo steel

Heat identification	Melter	Heat size (tons)	Melting practice[a]
XA3602	Quaker	4	AOD
F5349	CarTech	2.5	Electric/ESR
91887	CarTech	15	AOD
30383	CarTech	15	AOD/ESR
30394[b]	CarTech	15	AOD/ESR
30176[c]	CarTech	15	AOD/ESR
30182[c]	Electralloy	15	AOD and AOD/ESR
10148[d]	Electralloy	23	AOD and AOD/ESR
14361[d]	Timken	65	Electric furnace
82631	Combustion Engineering	0.5	Air induction
YYC982C	Sumitomo[e]	2.0	VIM
59020	NKK[f]	5.0	VIM
A231001	Sumitomo[e]	50	VOD

[a]AOD = argon-oxygen decarburization; ESR = electroslag remelting; VIM = vacuum induction melting; VOD = vacuum oxygen decarburization.

[b]Heat 30394 is half of heat 30383 and was electroslag remelted.

[c]Heats 30176 and 30182 are same heat; their numbers are different because CarTech identifies each electroslag remelted ingot by a different number.

[d]Heats 10148 and 14361 were originally AOD processed.

[e]Located in Japan.

[f]Nippon Kokan K. K., located in Japan.

Table 5. Chemical analyses of various commercial heats of modified 9Cr-1Mo steel

Element	Content (wt %) for each heat[a]															
	F5349	30383		30394		30176	30182	10148	XA3602	91887	YYC982C	59020	A231001	82631	14361	Specifications
		Plate	Tube[b]	Plate	Tube[c]	Plate	Tube[c]									
C	0.10	0.083	0.075	0.084	0.094	0.081	0.081	0.089	0.089	0.097	0.105	0.096	0.097	0.09	0.12	0.08–0.12
Mn	0.39	0.46	0.46	0.46	0.45	0.37	0.36	0.47	0.40	0.38	0.43	0.38	0.38	0.37	0.48	0.30–0.60
P	0.007	0.010	0.009	0.010	0.014	0.010	0.013	0.021	0.005	0.007	0.006	0.003	0.015	0.010	0.014	0.02 max.[d]
S	0.015	0.004	0.004	0.003	0.004	0.003	0.003	0.006	0.009	0.005	0.006	0.003	0.005	0.006	0.009	0.010 max.
Si	0.35	0.41	0.40	0.40	0.37	0.11	0.11	0.28	0.19	0.08	0.42	0.39	0.40	0.41	0.33	0.02–0.50
Ni	0.10	0.09	0.08	0.09	0.09	0.09	0.09	0.16	0.05	0.08	0.02	0.02	0.11	0.10	0.12	0.40 max.
Cr	8.80	8.46	8.17	8.57	8.11	8.61	8.32	9.24	9.21	9.22	8.52	8.30	8.46	8.59	8.62	8.00–9.50
Mo	0.94	1.02	0.99	1.02	1.03	0.89	0.90	0.96	0.93	1.01	0.98	0.93	0.96	0.94	0.96	0.85–1.05
V	0.205	0.198	0.195	0.198	0.200	0.209	0.208	0.21	0.197	0.22	0.20	0.21	0.23	0.33	0.22	0.18–0.25
Nb	0.060	0.072	0.064	0.073	0.072	0.072	0.076	0.054	0.059	0.15	0.075	0.082	0.076	0.10	0.09	0.06–0.10
Ti	0.006	0.005	0.005	0.005	0.002	0.004	0.002	0.002	0.005	<0.01		Trace				
Co	0.018	0.055	0.055	0.055	0.056	0.010	0.011	0.019	0.019	0.018					0.02	
Cu	0.09	0.03	0.03	0.04	0.04	0.04	0.04	0.08	0.02	0.01	0.003	Trace	0.02	0.08		
Al	<0.001	0.002	0.003	0.014	0.024	0.007	0.004	0.002	0.003	0.003	0.003	0.016	0.005	0.025	0.02	0.04 max.
B	0.001	<0.001		<0.001		<0.09		<0.001	<0.001							
W	<0.01	0.05		0.05		<0.01		0.01	<0.01	<0.01		0.002				
As	0.002	<0.001		<0.001		0.001		0.002	<0.001							
Sn	0.004	<0.001		<0.001		<0.001		0.004	0.001			0.001			0.004	
Zr	<0.001	<0.001		<0.001		<0.001		<0.001	<0.001	<0.001		0.002				
N	0.011	0.051	0.050	0.053	0.053	0.055	0.053	0.035	0.065	0.038	0.0465	0.0329	0.039	0.033	0.038	0.03–0.07
O	0.012							0.008		0.005						

[a]All chemical analysis was carried out at Combustion Engineering, Inc., Chattanooga, Tenn.

[b]Hot extruded.

[c]Hot rotary pierced.

[d]Max. = maximum.

Heat F5349 was low in nitrogen, which generally, is added while the melt is in the ladle. However, when this heat was melted, it was decided to pour the melt into ingots rather than take the time near the end of the day to make the nitrogen addition. Because this was the first commercial scale-up heat, it was accepted on a best-efforts basis. In retrospect, this heat has served a very useful purpose in providing material for determining the effect of low nitrogen content on the mechanical properties of modified 9Cr-1Mo steel.

Heats 30176 and 30182 (same as 30176; see Table 4, footnote *c*) are low in silicon content. During the scale up at CarTech, it was decided to target silicon contents for 0.20 and 0.50 wt % (low and high end of the allowed specification range). Having limited experience in the recovery of silicon during electroslag remelting (ESR) of material such as modified 9Cr-1Mo steel, CarTech missed the target silicon contents (see Table 5). However, in subsequent heats 30383 and 30394, the allowed range of concentration for all elements in the specifications were met. Once again, however, out-of-specification heats served a useful purpose by providing material to evaluate the effects of low silicon level on the mechanical and oxidation properties of the alloy.

Heat 91887, the very first heat made at CarTech, missed the allowed range for both silicon and niobium. This heat was melted before setting a final compositional specification and, thus, was not required to meet the chemistry range. The reason that it is included in this compilation is that extensive test data were generated on it before the availability of other commercial heats. Because of its low silicon content, this heat had poorer oxidation resistance, but its high niobium content yielded very high creep strength.

Heat 82631, the first electric furnace heat of modified 9Cr-1Mo steel, was melted at Timken. Unfortunately, it analyzed 0.33% vanadium, which missed the allowed range specified because a ferro-vanadium alloy containing a higher-than-expected vanadium content was used for alloying.

Although all heats were used in generating mechanical properties data, most of the data were generated on heats F5349, 30383, 30394, 30176, 30182, 10148, XA3602, 91887, and 14361.

3.2.2 Fabrication

Commercial heats of modified 9Cr-1Mo steel were subjected to extensive fabrication processes (Table 6), and no unusual problems were encountered. Detailed reports on tubing,[17,18] extruded pipe,[19] and large forgings[20] are available. Photographs of an extruded pipe and a saddle forging are presented in Figs. 3 and 4 as examples of primary fabrication product forms.

Fabrication of double-wall tubing for steam generator applications was initiated during 1985. Initial work was completed at Thermo Electron Corporation. Double-wall tubing was sent to ANL for interfacial stress measurements, but the data are not yet available.

Several sand castings of the modified 9Cr-1Mo steel were made in the United States[21] and Japan.[22] Results show that modified 9Cr-1Mo steel can be sand cast without porosity only when nitrogen is at the low end of the specification range (N = 0.03–0.07 wt %). In addition, several electroslag castings of modified 9Cr-1Mo steel were also prepared and tested.[23–25] Electroslag castings were not sensitive to nitrogen content, and impact properties were much superior to the sand castings. However, both sand and electroslag castings showed poorer mechanical properties than the wrought material.

3.3 MECHANICAL PROPERTIES (BASE METAL)

3.3.1 Heat Treatment – Microstructure and Properties

Modified 9Cr-1Mo steel is a hardenable steel, and, thus, a knowledge of its continuous cooling transformation (CCT) diagram is required. Data were developed for several commercial heats of modified 9Cr-1Mo steel at both Climax Molybdenum Company[26] and the University of Tennessee (UT).[27] The CCT diagrams of heats 30182, 30383, and 30394 developed by Climax Molybdenum Company are presented in Figs. 5 through 7. The specimens were first austenitized at 1000°C for 15 min before programmed cooling. The CCT diagram developed for heat 30176 at UT is presented in Fig. 8. The diagrams for heats 30176 and 30182 are nearly the same, but there are some differences in the exact location of various transformations (Fig. 5 vs Fig. 8) from the two sources.[26,27] The higher austenitizing temperature

Table 6. Summary of information on fabrication of commercial heats of modified 9Cr-1Mo alloy

Heat	Melter	Melting practice[a]	Product	Product size (mm)		Fabrication method	Fabricator
				OD	Thickness		
F5349	Quaker	AOD	Plate		16	Hot forged, hot rolled	ORNL,[b] Y-12
30383	CarTech	AOD	Plate		51	Hot forged, hot rolled	Jessop
30383	CarTech	AOD	Bar	95		Hot forged, hot rolled	CarTech
30383	CarTech	AOD	Tube	102	15	Hot extruded	Amax
30394	CarTech	AOD-ESR	Plate		25	Hot forged, hot rolled	Jessop
30394	CarTech	AOD-ESR	Bar	232		Hot forged	Cartech
30394	CarTech	AOD-ESR	Tube	102	15	Hot extruded	Amax
30394	CarTech	AOD-ESR	Tube	76	13	Hot rotary pierced	Timken
30182	CarTech	AOD-ESR	Plate		16	Hot forged, hot rolled	Combustion Engineering
30182	CarTech	AOD-ESR	Bar	232		Hot forged	CarTech
30182	CarTech	AOD-ESR	Tube	102	15	Hot extruded	Amax
30182	CarTech	AOD-ESR	Tube	76	13	Hot rotary pierced	Timken
30176	CarTech	AOD-ESR	Plate		25	Hot forged, hot rolled	Jessop
10148	Electralloy	AOD	Plate		16	Hot forged, tempered, cold rolled	ORNL
10148	Electralloy	AOD	Bar	44		Hot extruded	ORNL
10148	Electralloy	AOD	Bar	107		Hot rolled	Bethlehem
10148	Electralloy	AOD	Pipe	245	25	Hot pilgered	Phoenix
10148	Electralloy/ Universal Cyclops	AOD-ESR	Plate	203		Hot rolled	Universal Cyclops
10148	Electralloy Universal Cyclops	AOD-ESR	Plate	51		Hot rolled	Universal Cyclops
10148	Electralloy Universal Cyclops	AOD-ESR	Octagon box	200		Hot forged	Universal Cyclops
10148	Electralloy Universal Cyclops	AOD-ESR	Round	232		Hot forged	Universal Cyclops

Table 6. (continued)

Heat	Melter	Melting practice[a]	Product	Product size (mm)		Fabrication method	Fabricator
				OD	Thickness		
10148	Electralloy Universal Cyclops	AOD-ESR	Tube	54	9.45	Hot extruded, cold reduced	TI Stainless[c]
10148	Electralloy Universal Cyclops	AOD-ESR	Pipe	406	25	Hot extruded	Cameron Iron Works
14361	Electralloy/ Special Metals	AOD-ESR	Saddle forging	781	166	Hot forged	National Forge
14361	Electralloy/ Special Metals	AOD-ESR	Pipe	584	102	Hot extruded	Cameron Iron Works
XA3602	Combustion Engineering	Air induction	Tube	51	8.03	Centrifugally cast and cold pilgered	Combustion Engineering
YYC982C	Sumitomo[d]	VIM	Tube	76	13	Hot extruded, cold drawn	Sumitomo
YYC982C	Sumitomo	VIM	Tube	51	6.4	Hot extruded, cold drawn	Sumitomo
59020	NKK[d]	VIM	Plate		25	Hot forged, hot rolled	NKK
59020	NKK	VIM	Tube	76	13	Hot extruded, cold drawn	NKK
59020	NKK	VIM	Tube	51	6.4	Hot extruded, cold drawn	NKK

[a]AOE = argon-oxygen decarburization; ESR = electroslag remelting; VIM = vacuum induction melting.

[b]Oak Ridge National Laboratory.

[c]Located in the United Kingdom.

[d]Located in Japan.

CYN-5815

(*a*)

CYN-5816

(*b*)

Fig. 3. Hot-extruded pipe of 387-mm ID × 22-in. wall (15.25-in. ID × 0.875-in. wall) of modified 9Cr-1Mo steel (heat 10148) from Cameron Iron Works: (*a*) straight section; (*b*) section after induction bending.

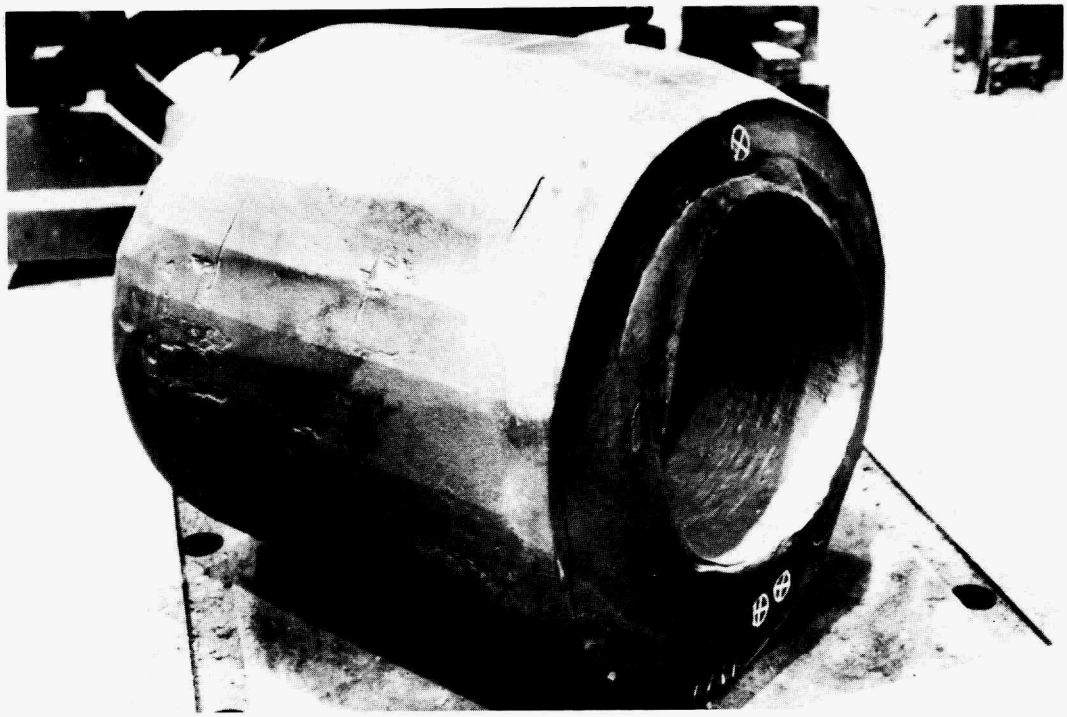


Fig. 4. Saddle forging of 77.5-cm OD × 32.5-cm wall (31-in. OD × 13-in. wall) of modified 9Cr-1Mo steel (heat 14361) from National Forge Company.

of 1030°C used by UT as opposed to 1000°C used by Climax Molybdenum Company is probably responsible for the difference in CCT diagrams from the two sources. We believe that for modified 9Cr-1Mo steel, dissolution of NbC(N) requires an austenitization temperature as high as 1040°C. Thus, the CCT diagram developed at UT may be a better representation of transformation in modified 9Cr-1Mo steel.

The modified alloy is recommended for use in the N&T condition. Normalizing consists of heating the alloy to 1040°C, holding for 1 h (up to 25-mm thickness), and then air cooling to room temperature. This treatment produces a fully martensitic structure. The typical hardness in this condition is Rockwell C40 [Brinell hardness number (BHN) of 400]. A typical tempering treatment consists of heating the normalized steel to 760°C, holding for 1 h (up to 25 mm thick), followed by air cooling to room temperature. The typical hardness in this condition is Rockwell B95 (220 BHN).

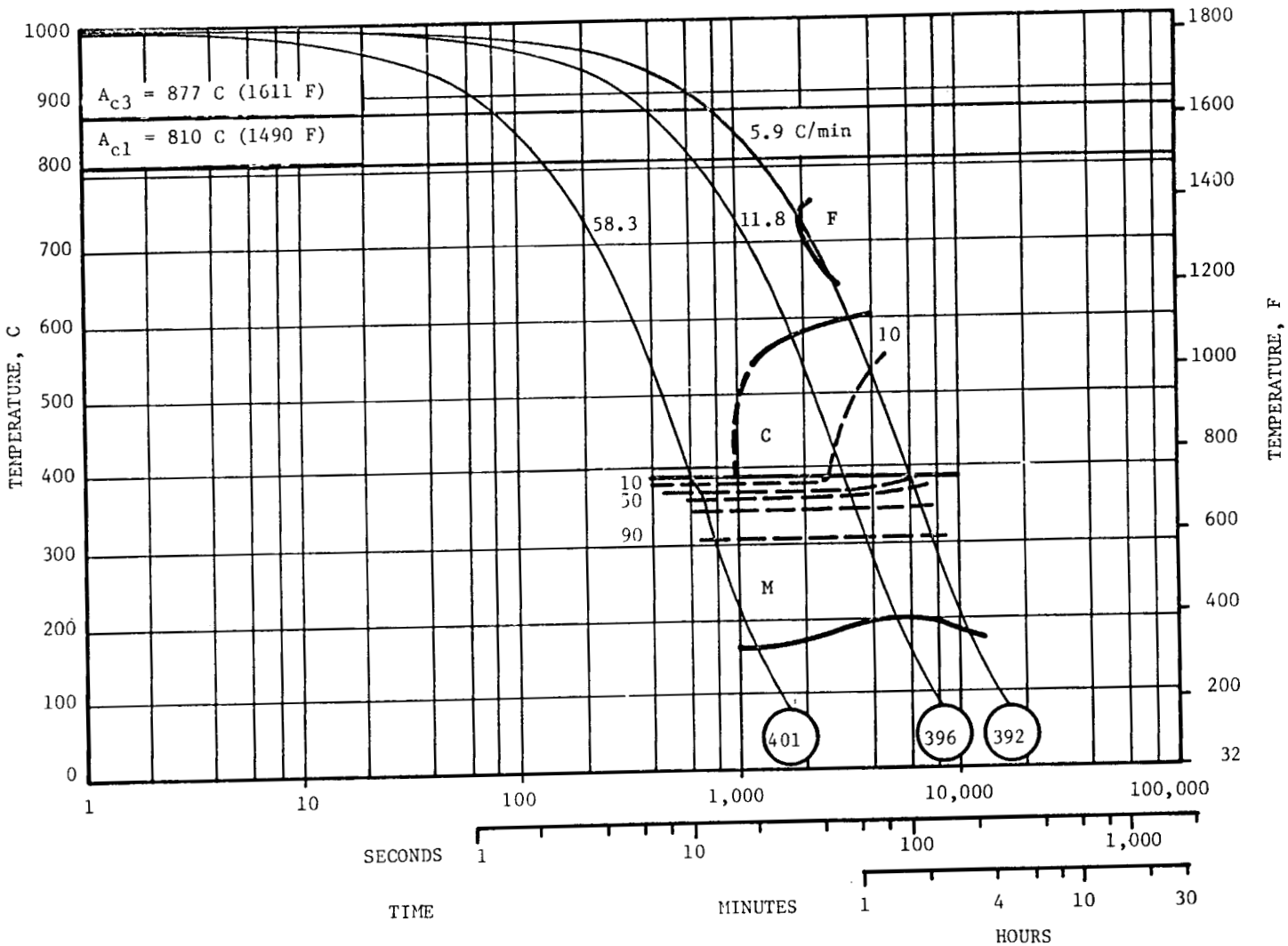


Fig. 5. Continuous cooling transformation (CCT) diagram of heat 30182 of modified 9Cr-1Mo steel. (Austenite grain size: ASTM No. 9.5.) *Source*: Climax Molybdenum Company.

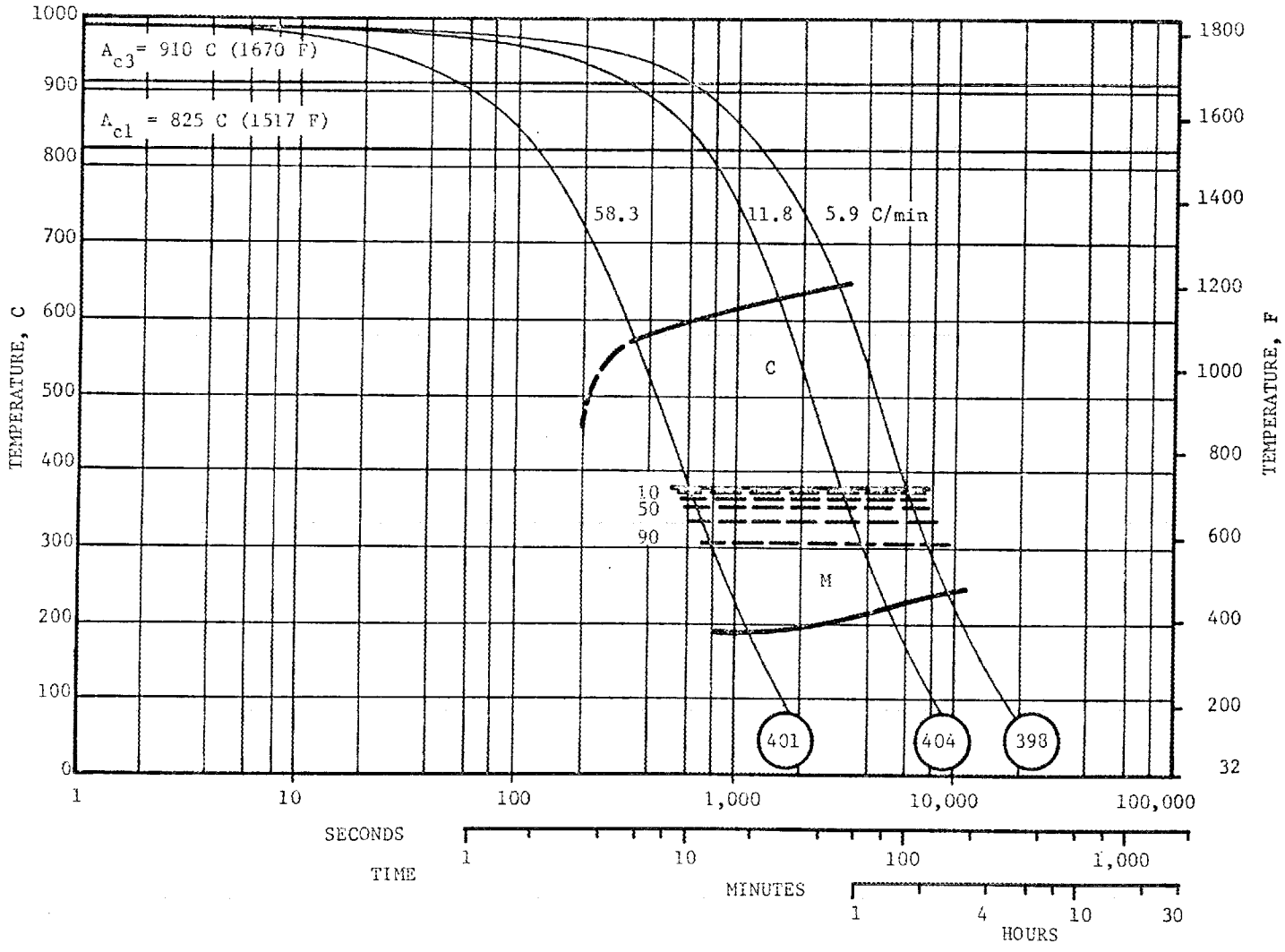


Fig. 6. CCT diagram of heat 30383 of modified 9Cr-1Mo steel. (Austenite grain size: ASTM No. 9.5.) *Source*: Climax Molybdenum Company.

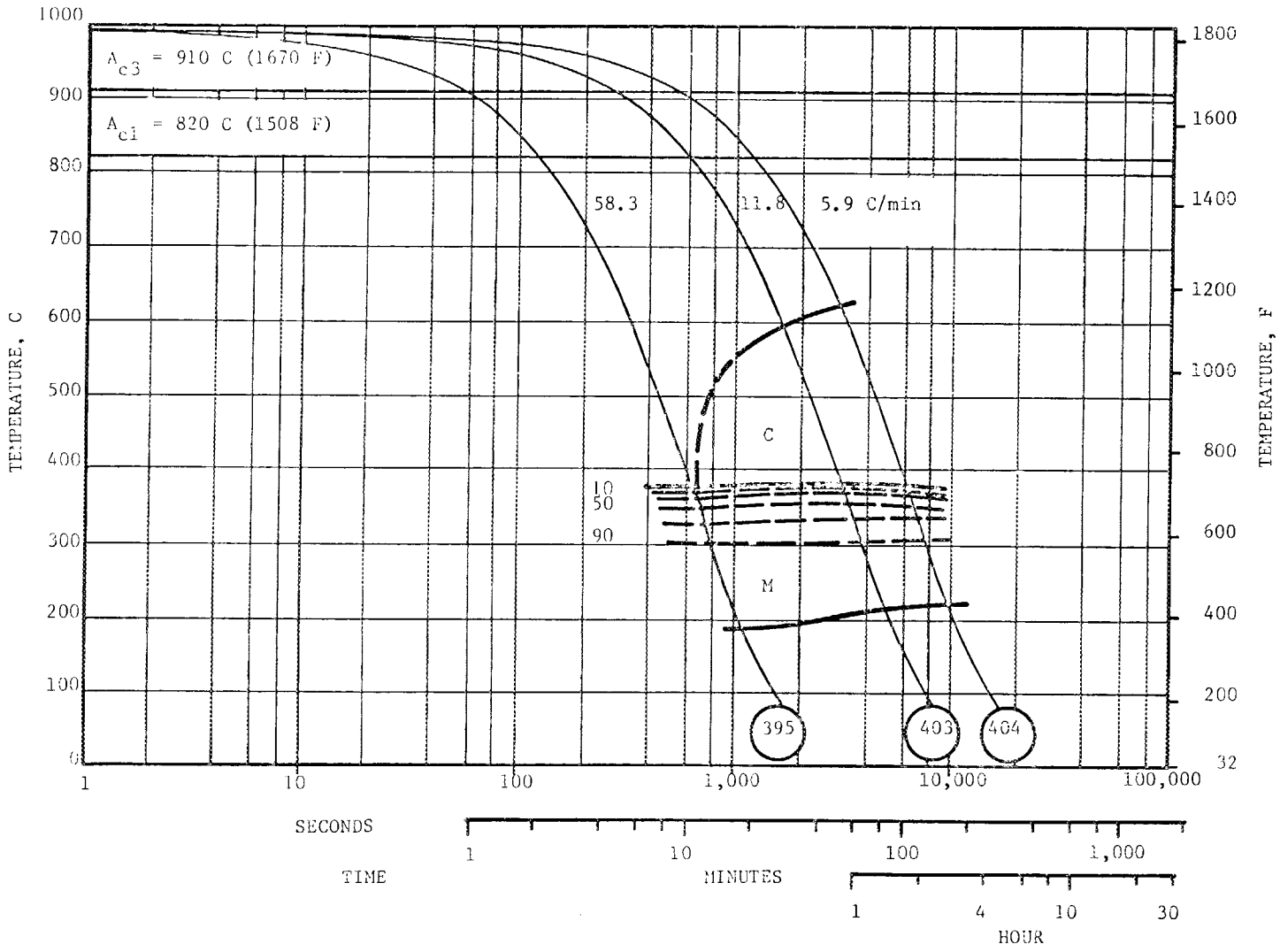


Fig. 7. CCT diagram of heat 30394 of modified 9Cr-1Mo steel. (Austenite grain size: ASTM No. 9.5.) *Source*: Climax Molybdenum Company.

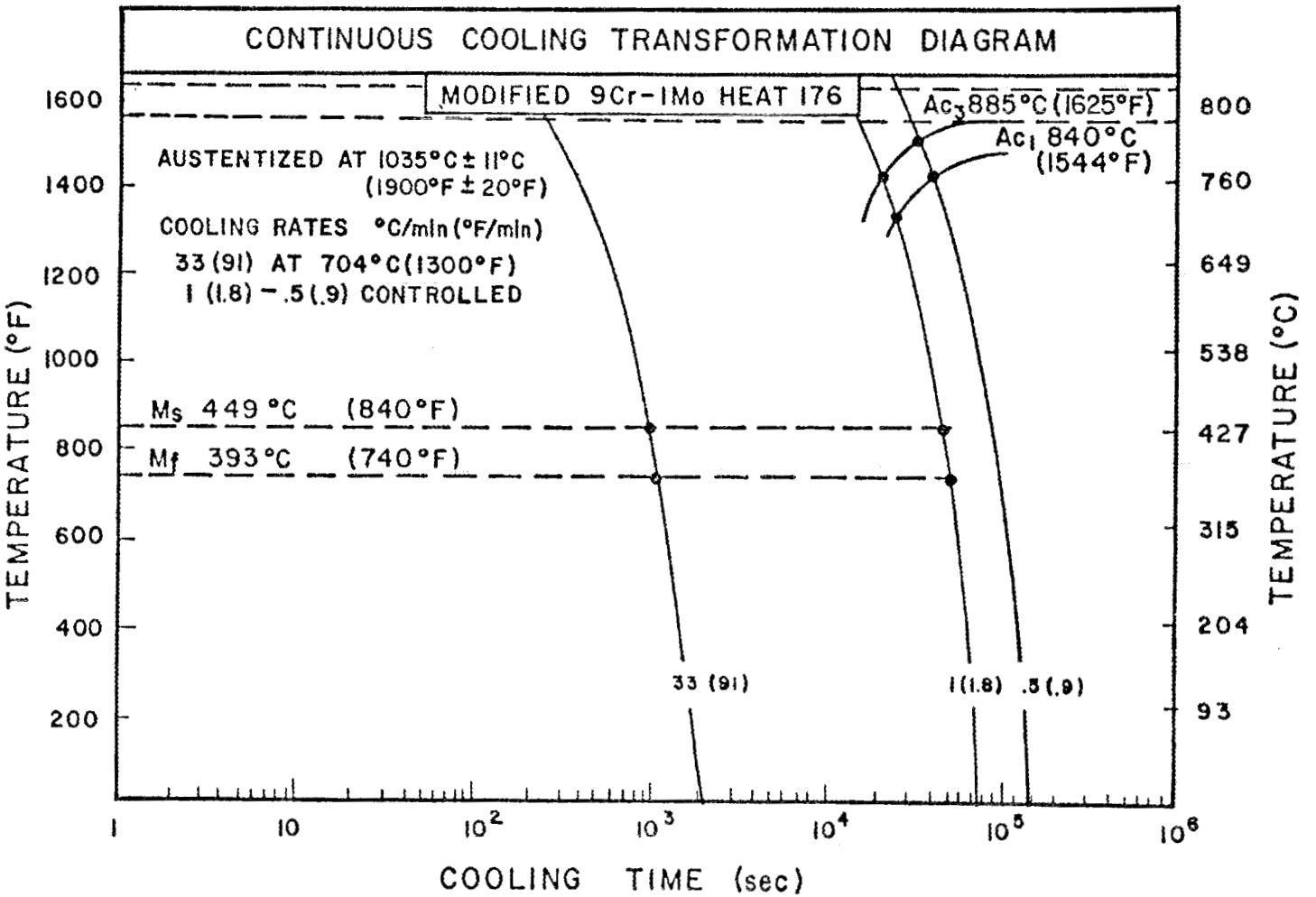


Fig. 8. CCT diagram of heat 30176 of modified 9Cr-1Mo steel.
Source: University of Tennessee.

The detailed tempering behavior of modified 9Cr-1Mo steel was investigated by Hammond.[28] The tempering response of commercial heats of modified 9Cr-1Mo steel can best be represented by a plot of hardness vs tempering temperature (Fig. 9). The tempering response of the alloy can also be described by the Hollomon-Jaffe (HJ) tempering parameter.[29]

$$HJ = T\ (C + \log t)\ , \qquad (2)$$

where

T = temperature in Kelvin,

t = time in hours,

C = constant.

The optimized value of C for two commercial heats (30176 and 30394) was determined[29] to be 22.3.

In addition, the normalizing treatment was also optimized.[30,31] Figure 10 shows the effect of different normalizing temperatures on the grain growth behavior of modified 9Cr-1Mo steel.

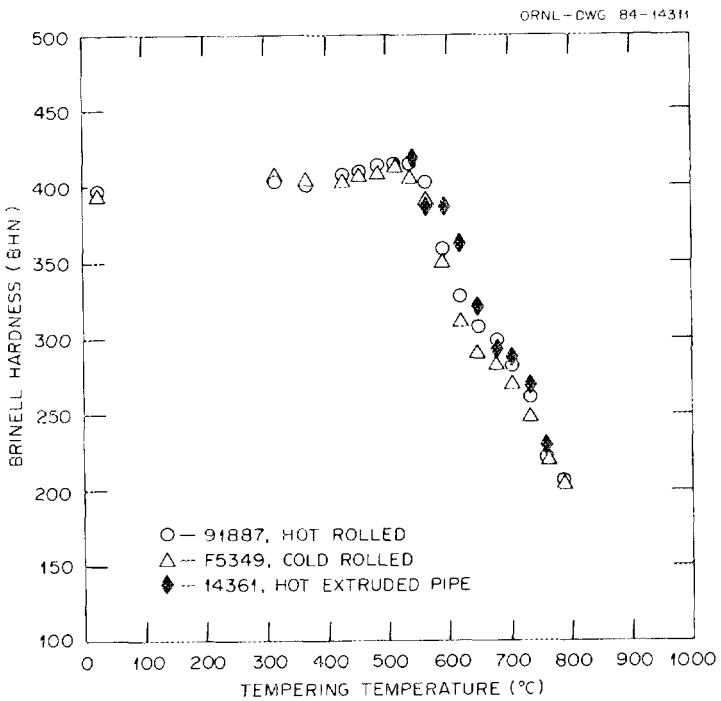


Fig. 9. Effect of tempering temperature on hardness of commercial heats of modified 9Cr-1Mo steel.

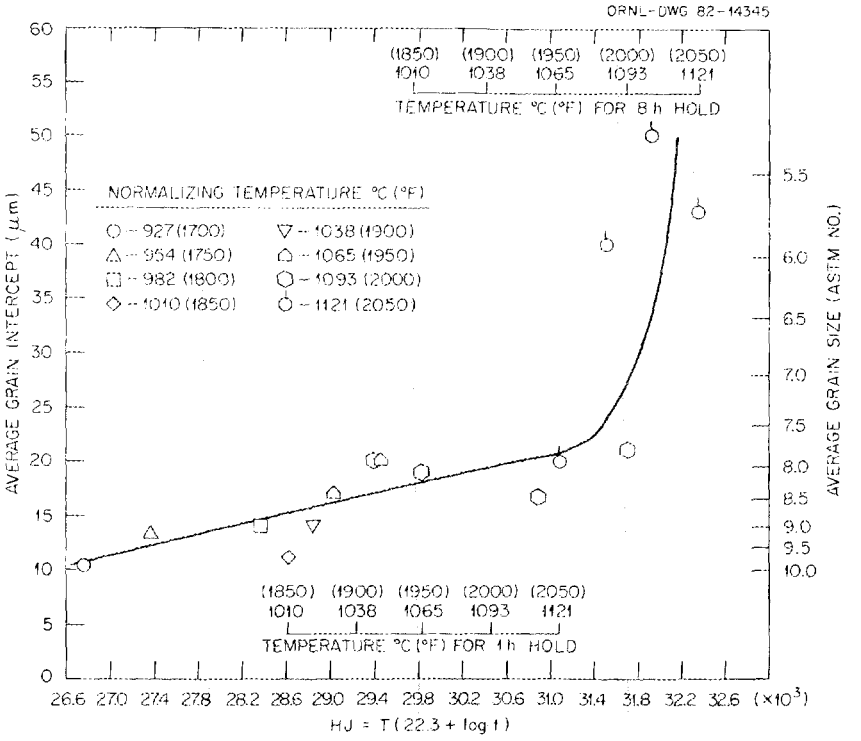


Fig. 10. Grain-coarsening behavior of modified 9Cr-1Mo steel as function of normalizing treatment.

Optical and transmission electron micrographs of tempered martensite in modified 9Cr-1Mo steel are shown in Figs. 11 and 12. Figure 11 shows that the alloy is single phase (no δ-ferrite) and has a fine grain size (ASTM 8-9). The transmission electron micrograph shows the presence of high-dislocation-density subboundaries in the matrix. The subboundaries are stabilized by the precipitation of MC, and $M_{23}C_6$ precipitates form at the prior austenite grain boundaries.

The effect of heat treatment (both normalizing and tempering) on mechanical properties has been investigated.[29,32] The effect of various tempering temperatures on hardness, Charpy V-notch, and tensile properties is summarized in Fig. 13. The effect of different tempering temperatures on creep properties after normalizing at 1040°C is presented in Fig. 14. Both Figs. 13 and 14 show that tempering below 760°C has a significant effect on all properties. However, temperatures higher than 760°C have a minimal effect. Of course, the tempering temperature must not exceed the lower critical A_{C1} temperature (see Figs. 5-8).

The effect[30] of normalizing temperatures on impact properties at room temperature is presented in Table 7. Room temperature impact energy

Y-185741

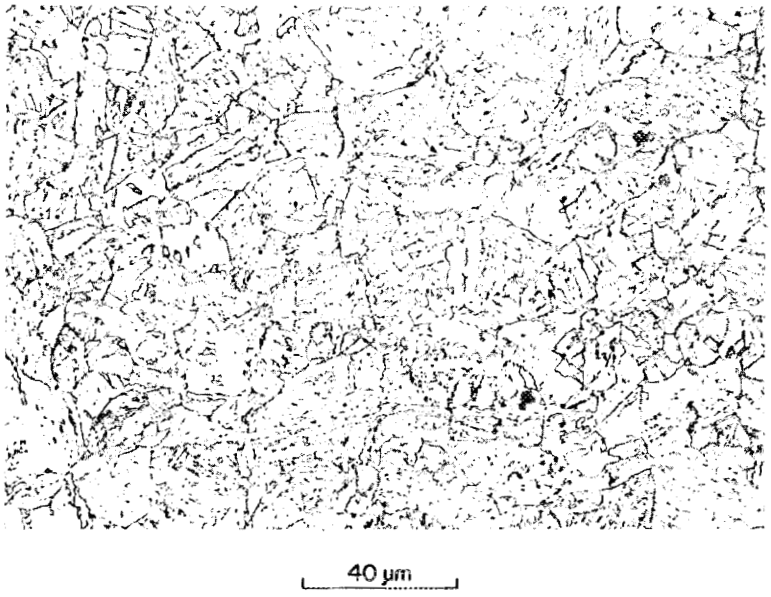


Fig. 11. Typical microstructure of modified 9Cr-1Mo steel after nominal normalizing and tempering treatment (1040°C for 1 h; 760°C for 1 h). Note that grain size is very fine.

Y-181757

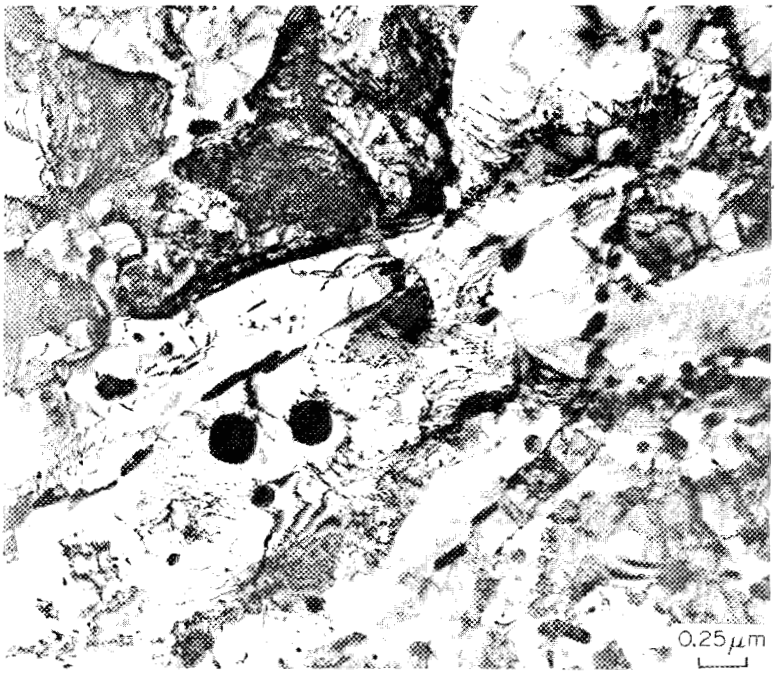


Fig. 12. Transmission electron micrograph of specimen normalized at 1040°C for 1 h and tempered at 760°C for 1 h. The micrograph shows that tempered martensite has dislocation substructure and there are carbides ($M_{23}C_6$ and MC) on both grain and subgrain boundaries.

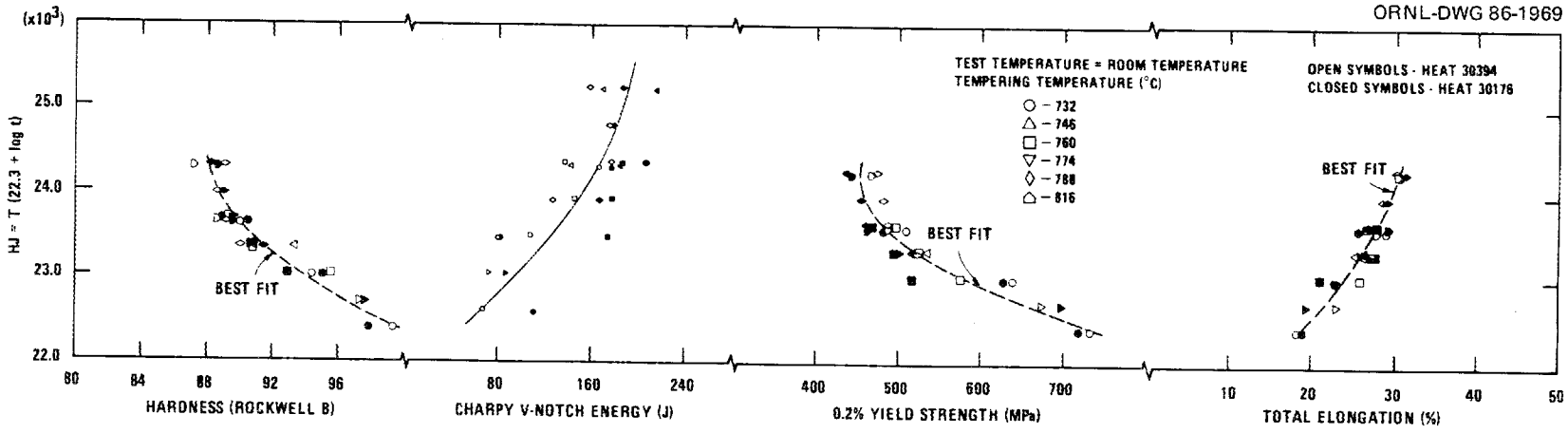


Fig. 13. Tempering parameter as function of hardness, Charpy energy, 0.2% yield strength, and total elongation of modified 9Cr-1Mo steel at room temperature.

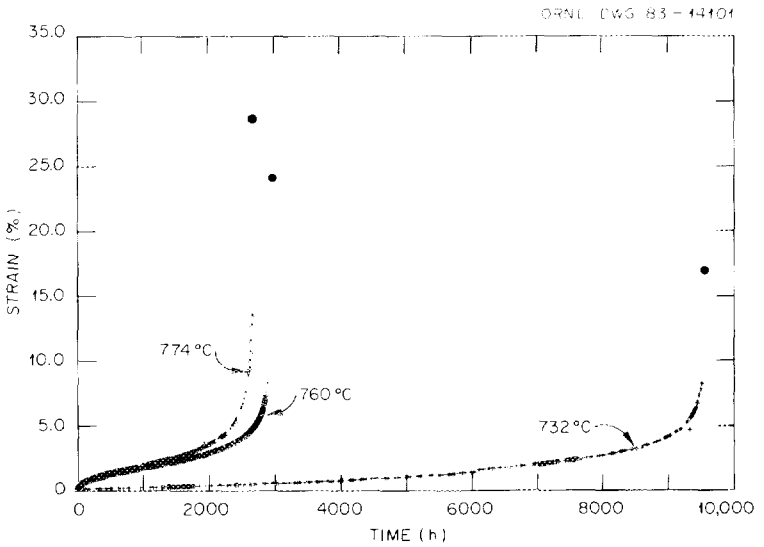


Fig. 14. Effect of tempering temperature on creep curves of modified 9Cr-1Mo steel. All specimens were subjected to same normalizing temperature (1040°C) before tempering. Tests were conducted at 593°C and 159 MPa.

decreases when the normalizing temperature is increased from 1038 to 1121°C. This drop is probably related to grain growth associated with the higher normalizing temperatures (Fig. 10). The effects of normalizing temperatures on tensile properties at room temperature and at 593°C are presented in Figs. 15 through 18. For a typical tempering temperature of 760°C, Figs. 15 through 18 show that at both room temperature and 593°C, strength is more sensitive than ductility to normalizing treatment.

The effects of normalizing temperature on hardness and time-to-rupture are shown in Figs. 19 and 20. A normalizing temperature of 1040°C was selected on the basis of optimum creep strength.

Table 7. Effect of normalizing temperature and time on room-temperature Charpy impact energy of heat 10148[a]

Normalizing conditions		Room-temperature Charpy impact energy (J)
Temperature (°C)	Time (h)	
1038	1	249
1121	1	177
1121	8	175

[a]All specimens were tempered at 760°C for 1 h before testing.

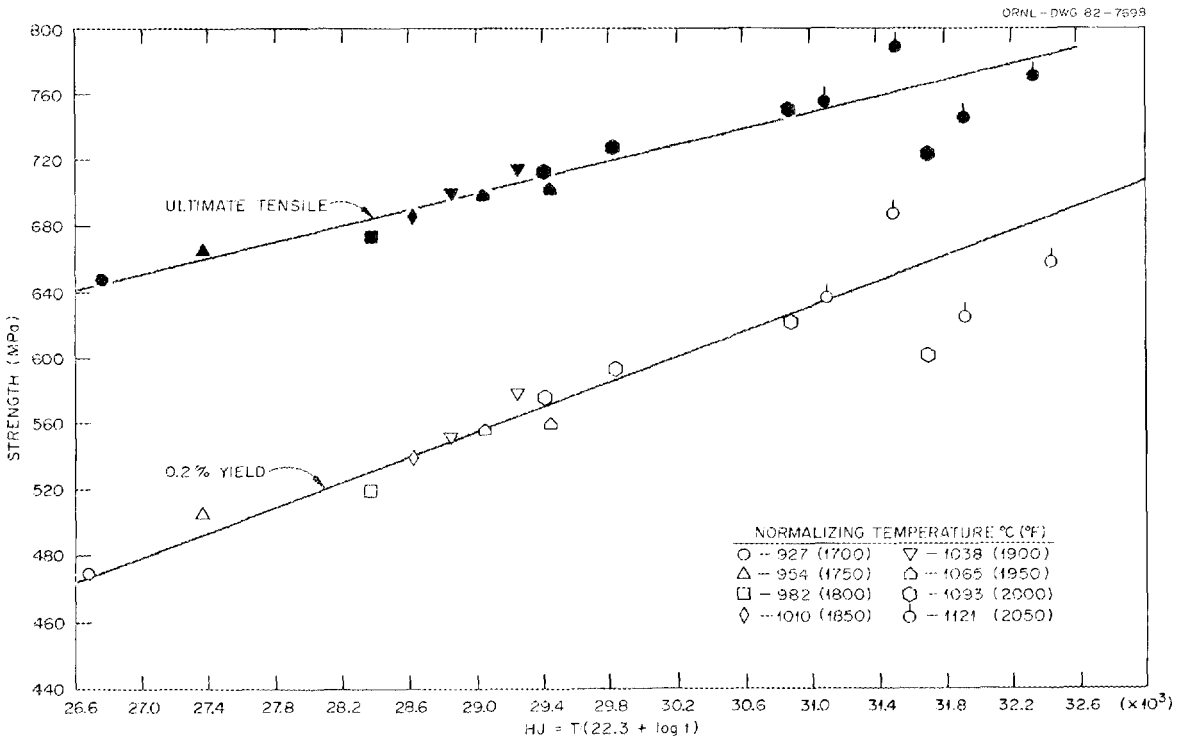


Fig. 15. Room-temperature yield and ultimate tensile strengths as functions of Holloman-Jaffee-parameter for various normalizing temperature and time combinations for commercial heats of modified 9Cr-1Mo steel.

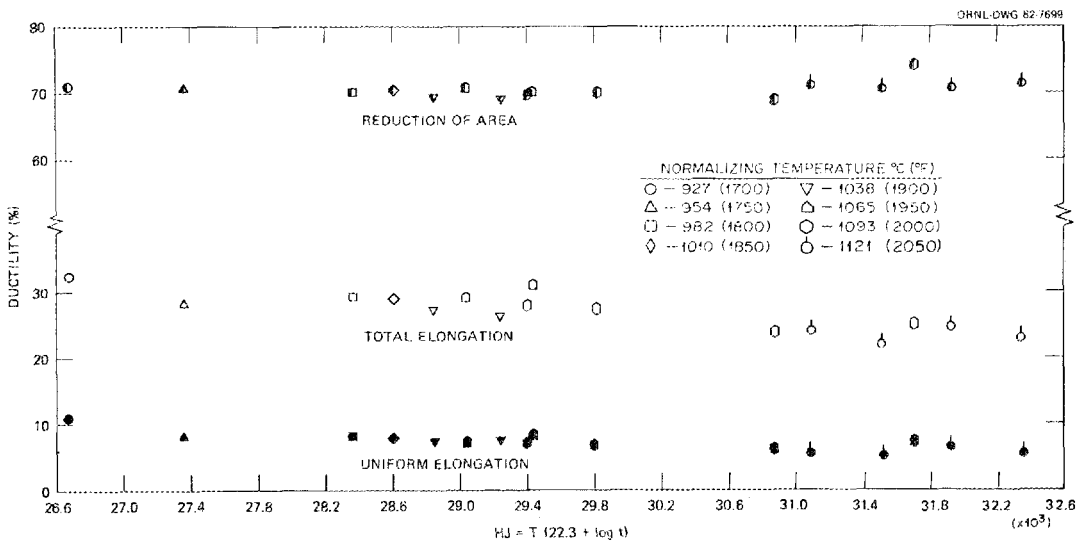


Fig. 16. Room-temperature ductility as function of Holloman-Jaffee-parameter for various normalizing temperature and time combinations for commercial heats of modified 9Cr-1Mo steel.

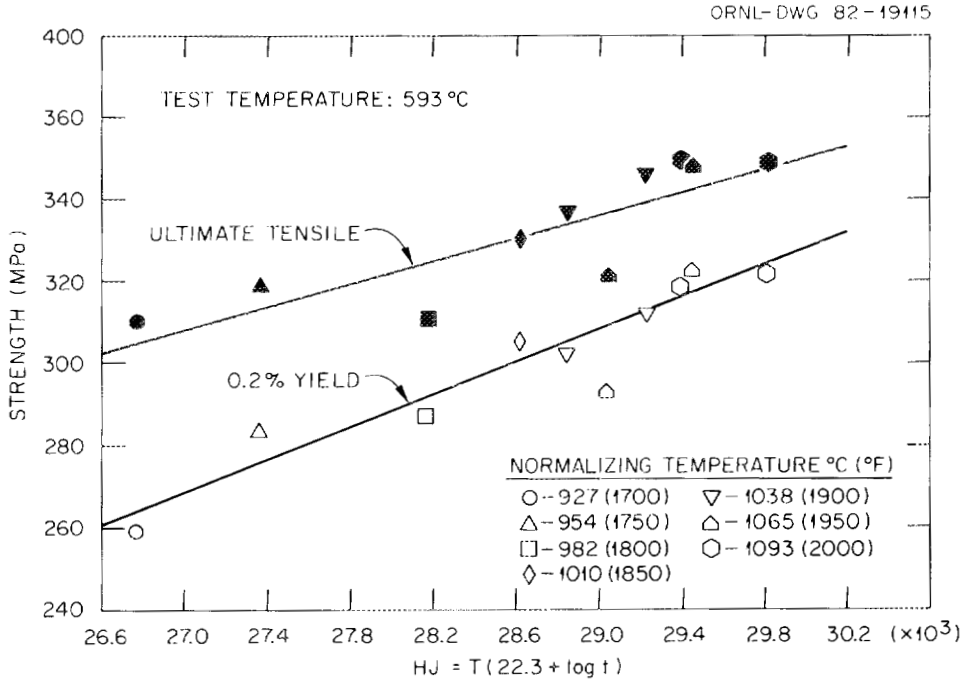


Fig. 17. Yield and ultimate tensile strengths at 593°C as functions of Holloman-Jaffe-parameter for various normalizing temperature and time combinations for commercial heats of modified 9Cr-1Mo steel.

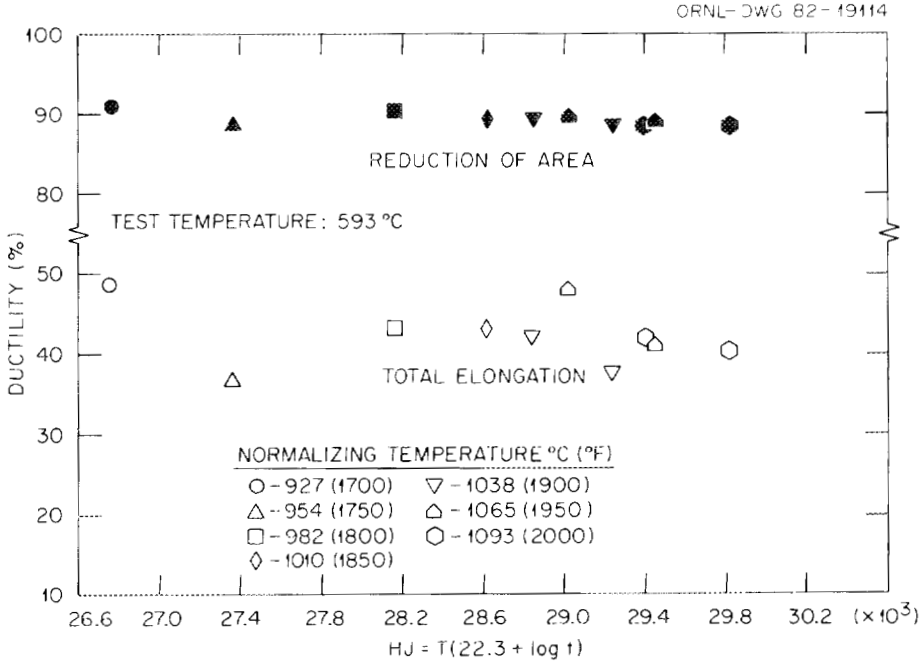


Fig. 18. Ductility at 593°C as function of Holloman-Jaffe-parameter for various normalizing temperature and time combinations for commercial heats of modified 9Cr-1Mo steel.

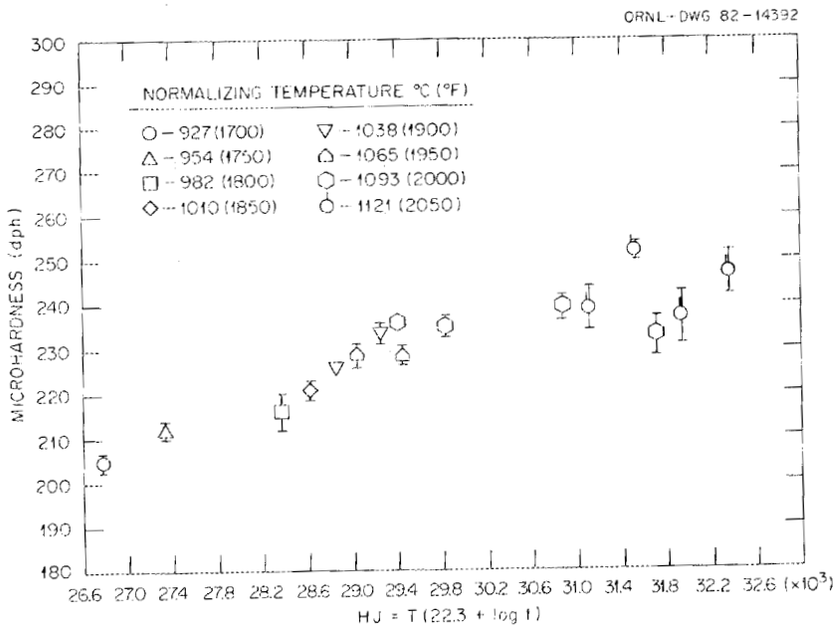


Fig. 19. Microhardness as function of Holloman-Jaffe-parameter for various normalizing temperature and time combinations for commercial heats of modified 9Cr-1Mo steel.

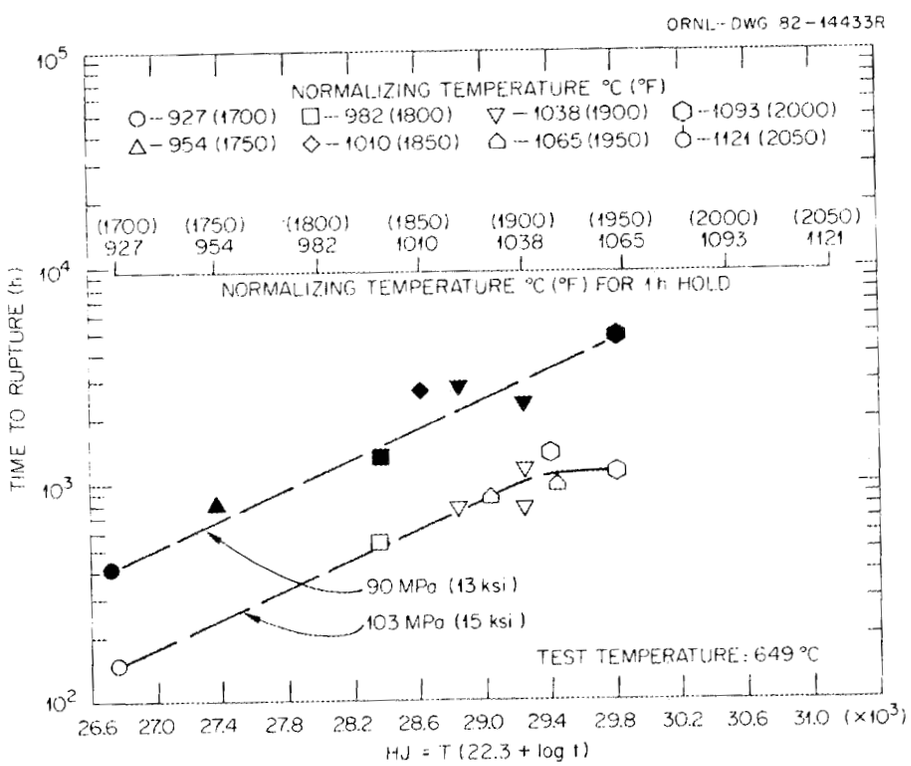
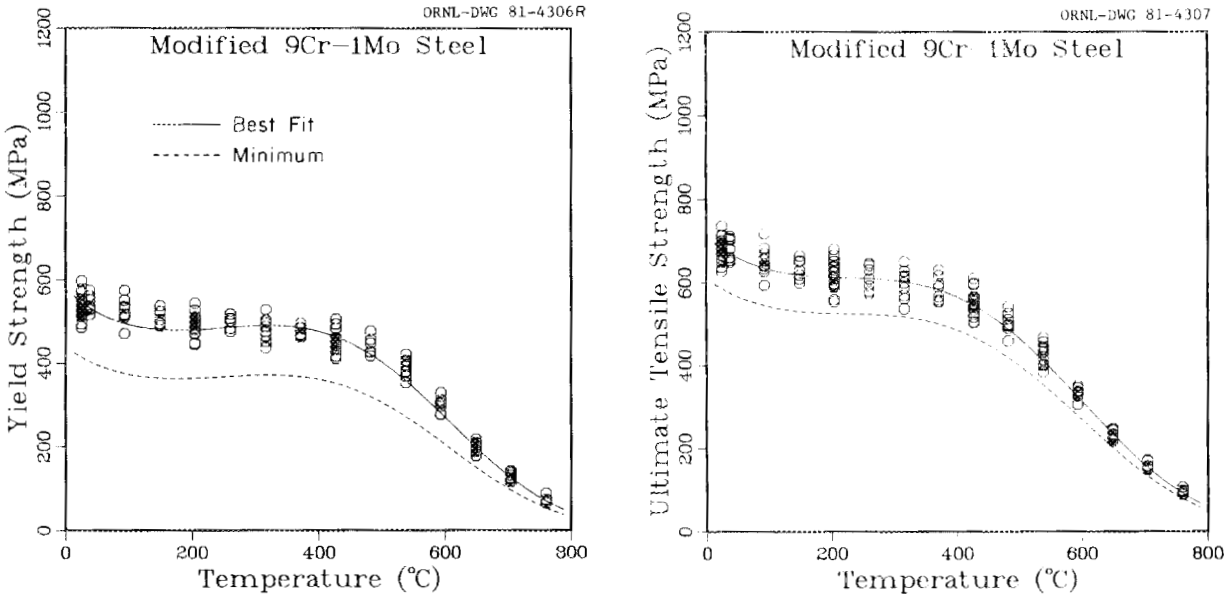


Fig. 20. Time-to-rupture data at 649°C as function of Holloman-Jaffe-parameter for various normalizing temperature and time combinations for commercial heats of modified 9Cr-1Mo steel.

3.3.2 Design Data (Base Metal)

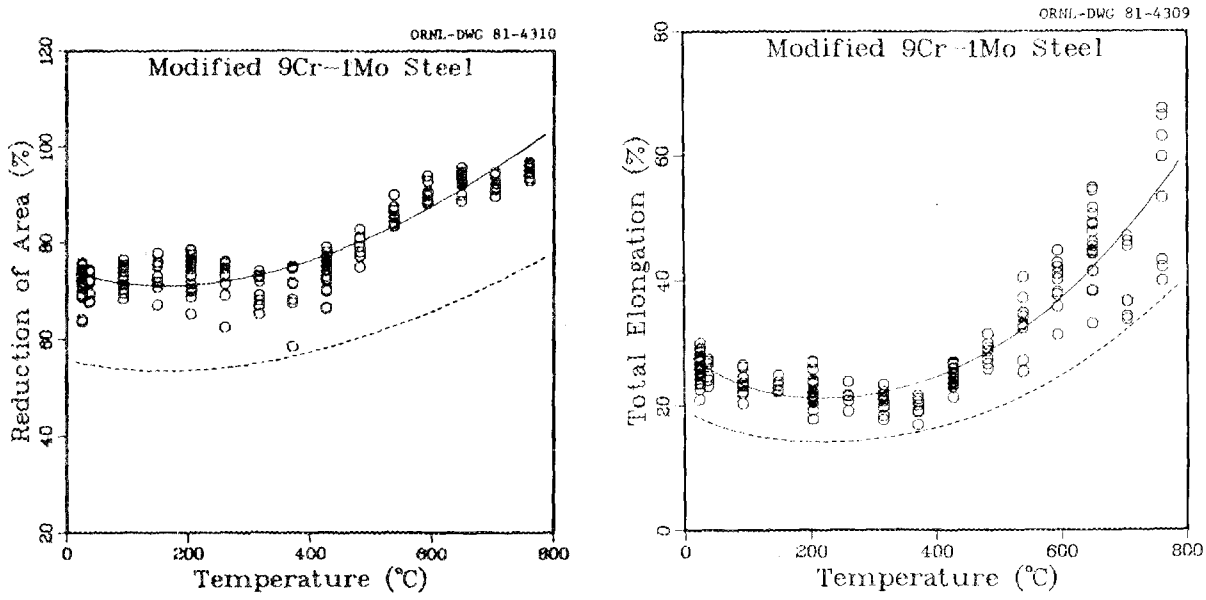
3.3.2.1 Tensile Properties

To satisfy ASME Code requirements, all commercial heats were subjected to tensile tests from room temperature to 760°C. Tables of the raw data are available in a package submitted to the ASME Code.[33] Tensile data have been analyzed and reported in several publications.[19,21,29,31,34–38] Typical tensile behavior of modified 9Cr–1Mo steel is represented in Figs. 21 and 22. Note that these results are on plate, bar, and tube products from commercial heats of modified 9Cr–1Mo steel. Furthermore, all tests were conducted on material that was normalized at 1040°C for 1 h and tempered at 760°C for 1 h. The tensile test strain rate for all tests was 0.05 in./(in.·min). Selected specimens from one of the commercial heats (30176) were also tested to determine strain rate effects on tensile



(*a*) (*b*)

Fig. 21. Plots of (*a*) 0.2% offset yield strength and (*b*) ultimate strength as functions of test temperature for plate, bar, and tube of commercial heats of modified 9Cr–1Mo steel. The best-fit average and minimum curves based on room-temperature specified minimum yield and ultimate tensile strength of 414 and 586 MPa are also included.



(*a*) (*b*)

Fig. 22. Plots of (*a*) total elongation and (*b*) reduction of area as functions of test temperature for plate, bar, and tube of commercial heats of modified 9Cr-1Mo steel. The best-fit average and minimum curves based on the room-temperature specified minimum total elongation and reduction of area of 18 and 55% are included also in these figures.

properties.[38] Strain rate was varied from 8 to 0.00008 in./(in.·min) at room temperature to 649°C. Tensile property curves for various strain rates are presented in Figs. 23 through 26, which show that the tensile properties of modified 9Cr-1Mo steel are sensitive to strain rate at room temperature, 427, 538, and 649°C. The yield and ultimate tensile strengths decrease with decreasing strain rate at room temperature and without any change in ductility. At 427 and 538°C, strength goes down, and ductility goes up with decreasing strain rate. At 649°C, strength goes down, but there is some decrease in ductility. The effects of melting practice, fabrication methods (product forms), compositional differences, heat treatment [isothermal anneal vs N&T, tempering temperature and time, and post-weld heat treatment (PWHT)], and thermal aging were also investigated.[38] Table 8 shows the room temperature tensile properties specified for various product forms. Table 9 shows the comparison of average yield and ultimate tensile strengths for standard and modified 9Cr-1Mo steel.

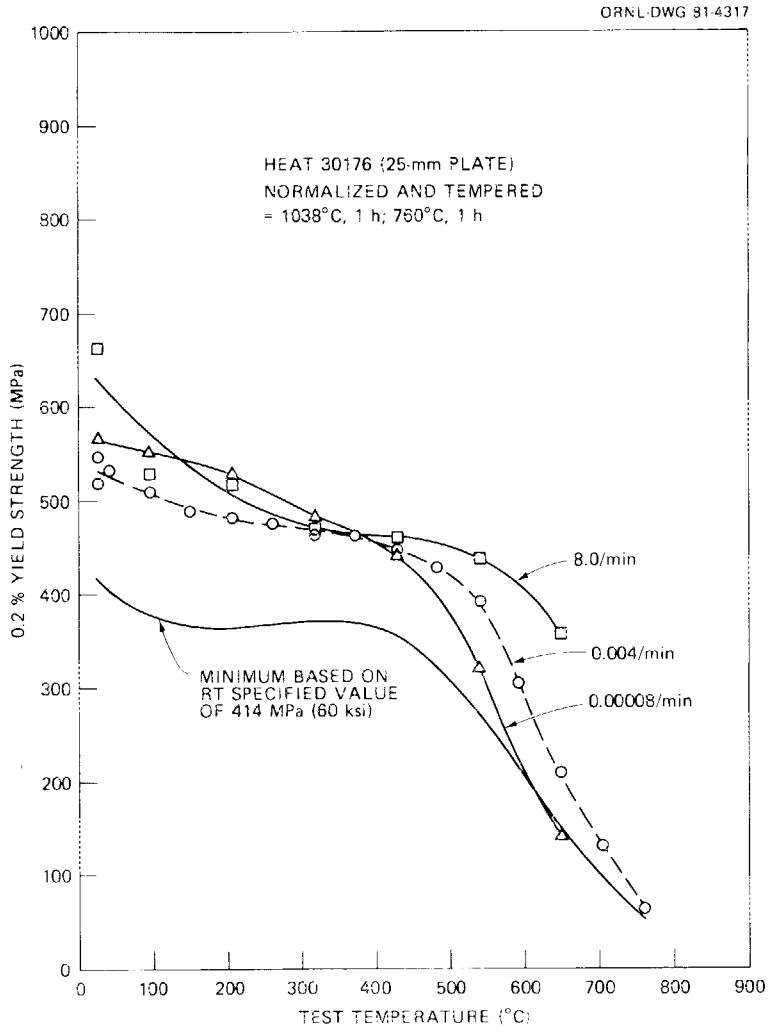


Fig. 23. The 0.2% offset yield strength as function of test temperature at highest, nominal, and lowest strain rates used in this investigation. Minimum curve based on the room-temperature specified value of 414 MPa (60 ksi) for yield strength is also included.

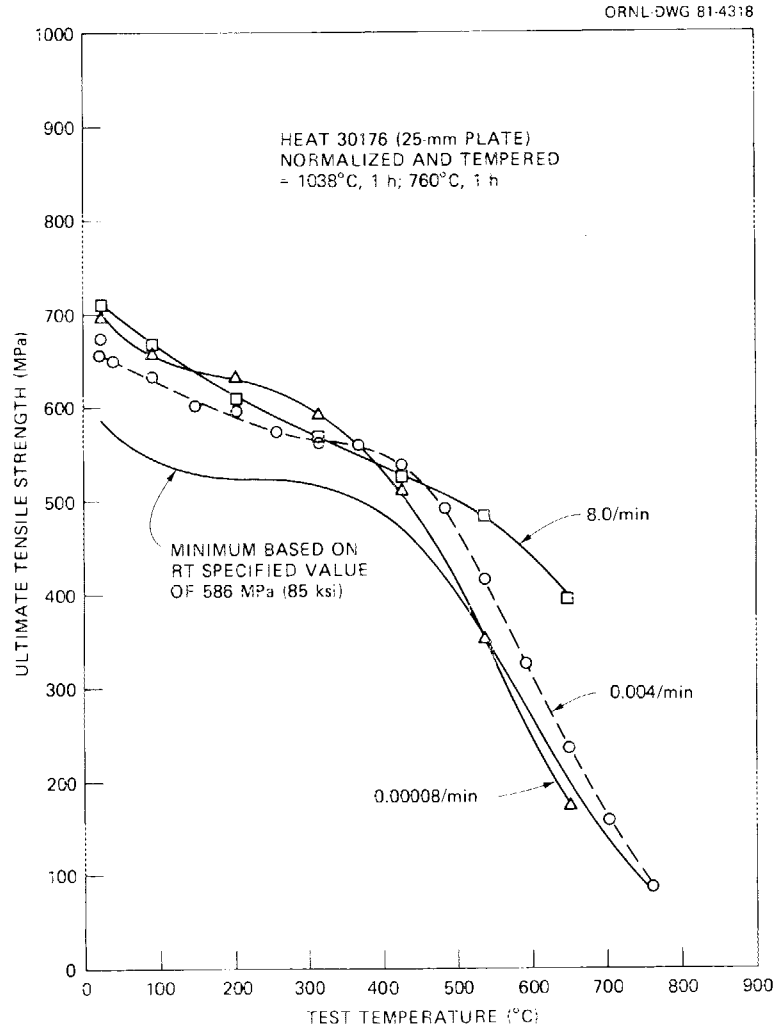


Fig. 24. Ultimate tensile strength as function of test temperature at highest, nominal, and lowest strain rates used in this investigation. Minimum curve based on room-temperature specified value of 586 MPa (85 ksi) for ultimate tensile strength is also included.

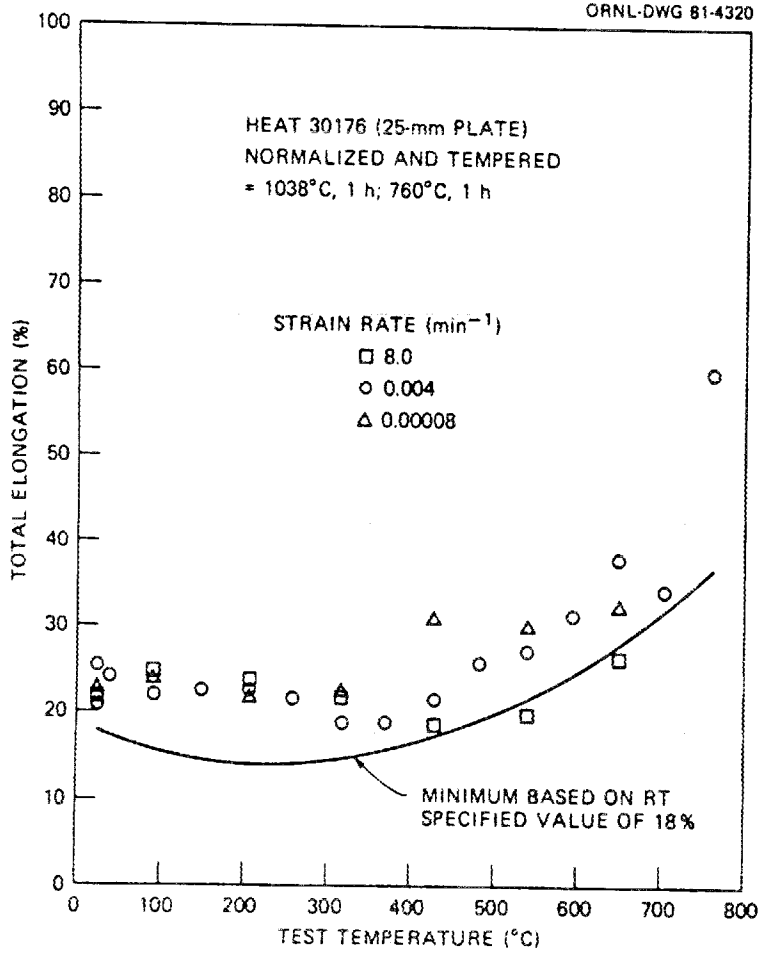


Fig. 25. Total elongation as function of test temperature at highest, nominal, and lowest strain rates used in this investigation. Minimum curve based on room-temperature specified value of 18% for total elongation is also included.

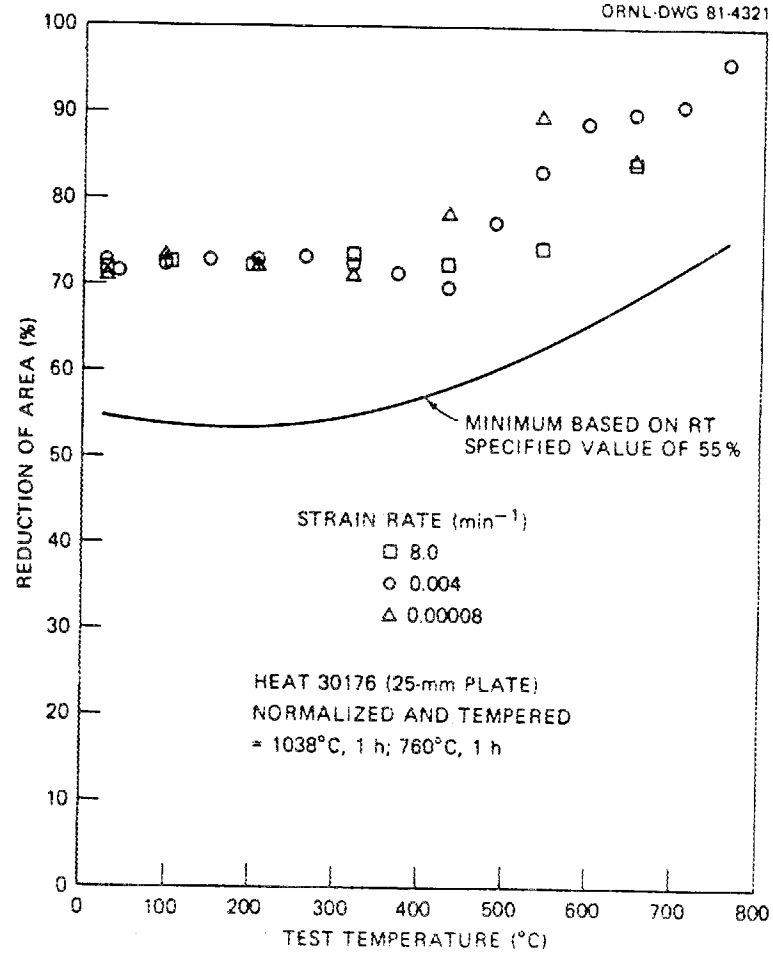


Fig. 26. Reduction of area as function of test temperature at highest, nominal, and lowest strain rates used in this investigation. Minimum curve based on room-temperature specified value of 55% for reduction of area is also included.

Table 8. Minimum tensile properties at room temperature specified for various products of modified 9Cr-1Mo steel

Calculation	Strength (MPa)		Total elongation (%)	Reduction of area (%)
	0.2% Yield	Ultimate		
		Plate		
Average	545	683	24.7	26.8
Average − 2SD[a]	496	634	20.6	61.6
Average − 3SD[a]	476	607	18.5	57.5
Recommended min	414	586	18	55
		Tube		
Average	524	669	27	74
Average − 2SD[a]	483	627	23	71
Average − 3SD[a]	462	607	21	70
Recommended min	414	586	18	55
		Bar		
Average	531	676	28	73
Average − 2SD[a]	469	614	26	70
Average − 3SD[a]	434	586	25.5	68
Recommended min	414	586	18	55

[a]SD = standard deviation.

The bulk of the tensile properties (Figs. 21 and 22) was generated on 25-mm-thick plate, 75-mm-diam bar, and 50-mm-OD × 6-mm-wall-thickness tubing. Data on a large forging (Fig. 4), a large diameter extruded pipe (Fig. 3), and castings have been generated more recently.[19-21] Comparisons of data on the large forging with average and minimum curves for tube, plate, and bar are presented in Figs. 27 through 30, which show that the tensile properties of the large forging (Fig. 4) fall within the average and average − 1.65 standard error of estimate (SEE) curves for the other product forms of several commercial heats. Similar results were found for the large diameter hot extruded pipe, Figs. 31 through 34. Tensile properties of various sand castings are plotted in Figs. 35 through 38. Both

Table 9. Comparison of average yield and ultimate tensile strengths for standard and modified 9Cr-1Mo steels

Temperature		0.2% Yield strength (MPa)		Ultimate tensile strength (MPa)	
(°C)	(°F)	Modified	Standard	Modified	Standard
Room temperature		546.5	481.3	682.0	666.4
38	100	531.9	470.7	668.8	650.2
93	200	494.6	442.4	633.5	604.8
149	300	480.6	429.4	617.7	581.8
204	400	480.3	425.5	612.4	571.9
260	500	486.1	425.4	609.7	567.9
316	600	490.3	424.0	602.0	562.8
371	700	485.1	416.3	582.3	550.1
427	800	463.4	397.9	544.9	524.1
482	900	420.7	366.0	487.2	481.2
538	1000	357.2	320.0	411.1	420.6
593	1100	279.3	262.8	323.3	345.9
649	1200	197.9		234.1	
704	1300	125.1		154.1	
760	1400	69.4		91.1	

strength and ductility of the castings tend to be somewhat lower than for wrought material. Note, however, that the exact composition of the castings compared with that allowed by wrought material specifications is still uncertain.[21]

Tensile properties of modified 9Cr-1Mo steel are well established, and the data base on the commercial heats have been used in establishing the maximum allowable stresses in ASME Boiler and Pressure Vessel Code Case numbers 1943, 1973, and 1992 (refs. 39–41).

3.3.2.2 Creep

Creep testing of modified 9Cr-1Mo steel was conducted at Battelle Memorial Institute (BMI), CE, WAESD, and ORNL. Tables of detailed creep data on commercial heats of modified 9Cr-1Mo steel are included in the

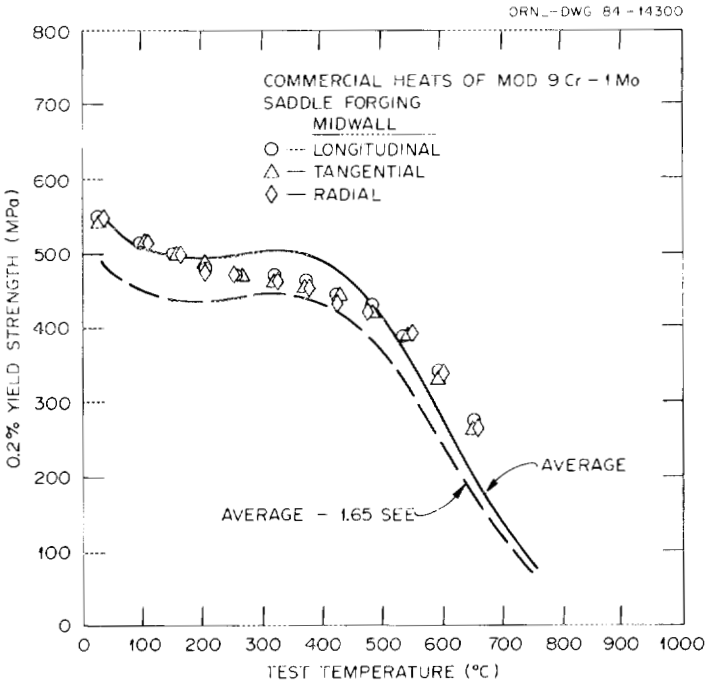


Fig. 27. Yield strength vs test temperature for longitudinal, tangential, and radial specimens from midwall section of saddle forging. Average and average − 1.65 SEE curves for various products form of several commercial heats are included for comparison.

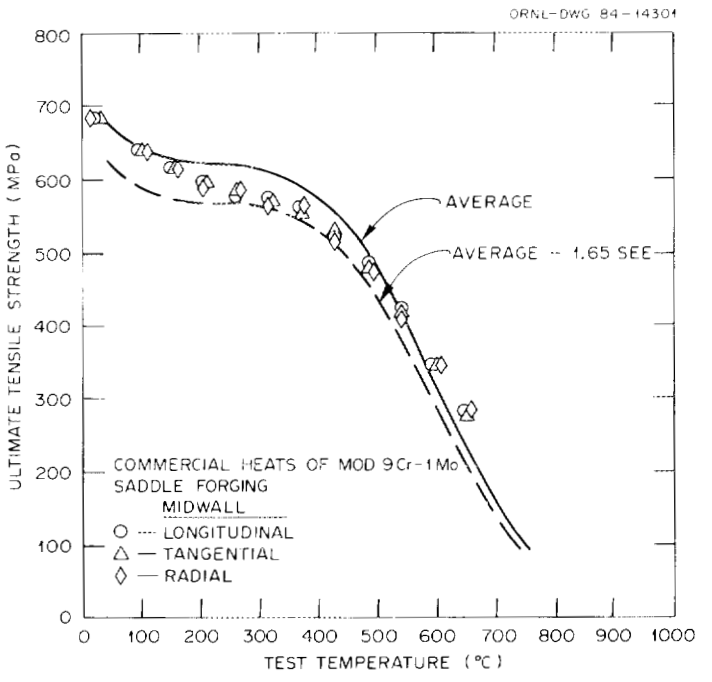


Fig. 28. Ultimate tensile strength vs test temperature for longitudinal, tangential, and radial specimens from midwall section of saddle forging. Average and average − 1.65 SEE curves for various product forms of several commercial heats are included for comparison.

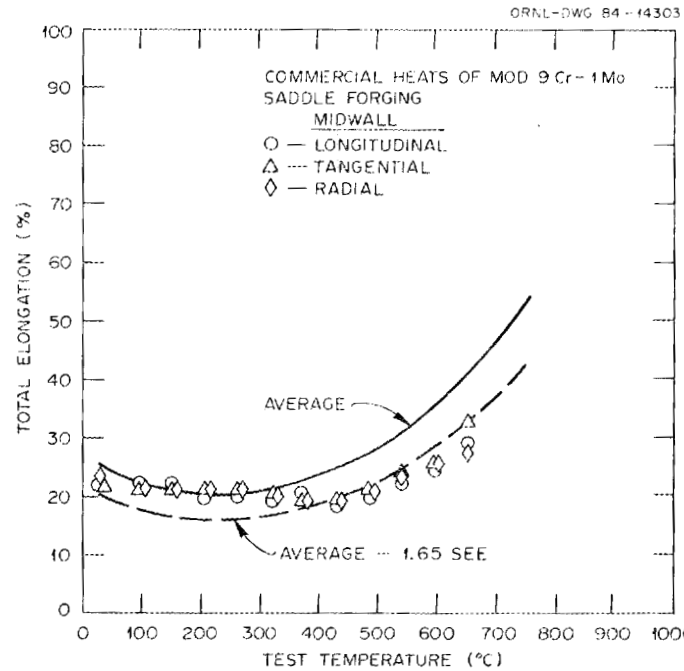


Fig. 29. Total elongation vs test temperature for longitudinal, tangential, and radial specimens from midwall section of saddle forging. Average and average − 1.65 SEE curves for various product forms of several commercial heats are included for comparison.

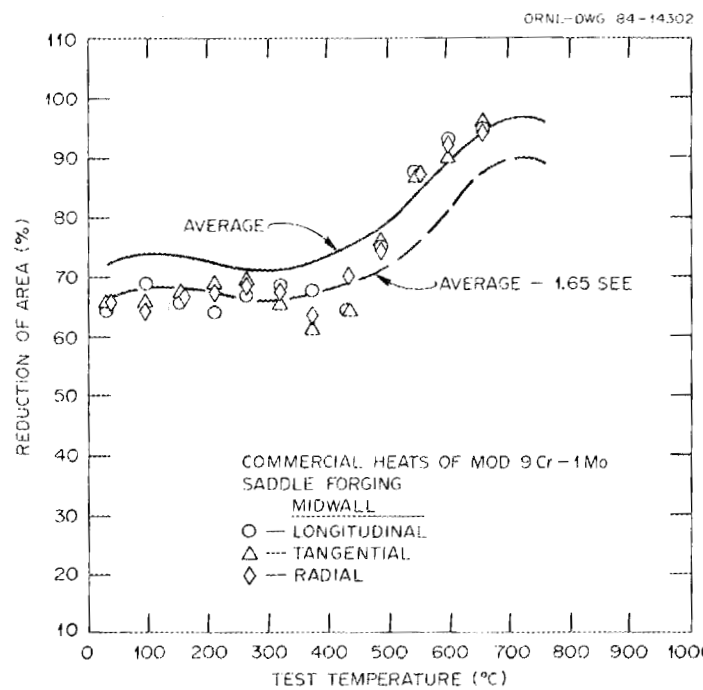


Fig. 30. Reduction of area vs test temperature for longitudinal, tangential, and radial specimens from midwall section of saddle forging. Average and average − 1.65 SEE curves for various product forms of several commercial heats are included for comparison.

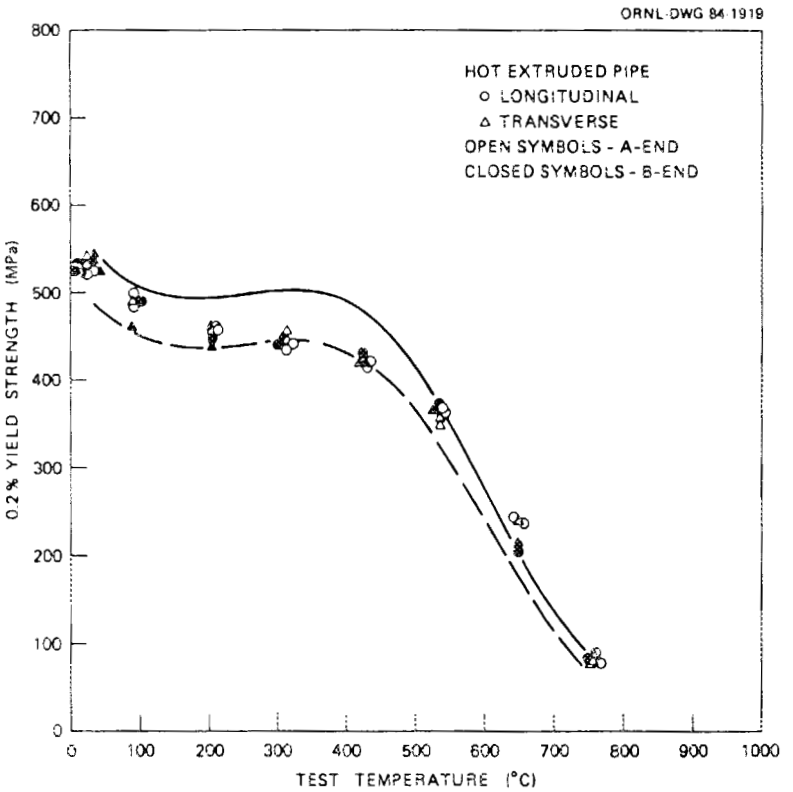


Fig. 31. Yield strength vs test temperature for hot-extruded pipe. Average (solid) and average − 1.65 SEE (dashed) curves for various product forms of several commercial heats are included for comparison.

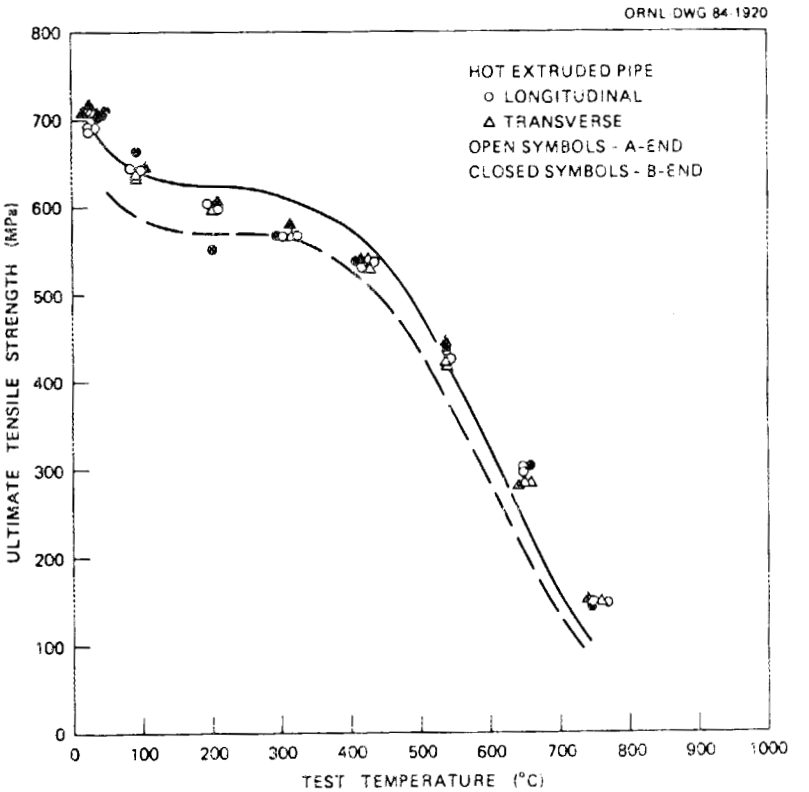


Fig. 32. Ultimate tensile strength vs test temperature for hot-extruded pipe. Average (solid) and average − 1.65 SEE (dashed) curves for various product forms of several commercial heats are included for comparison.

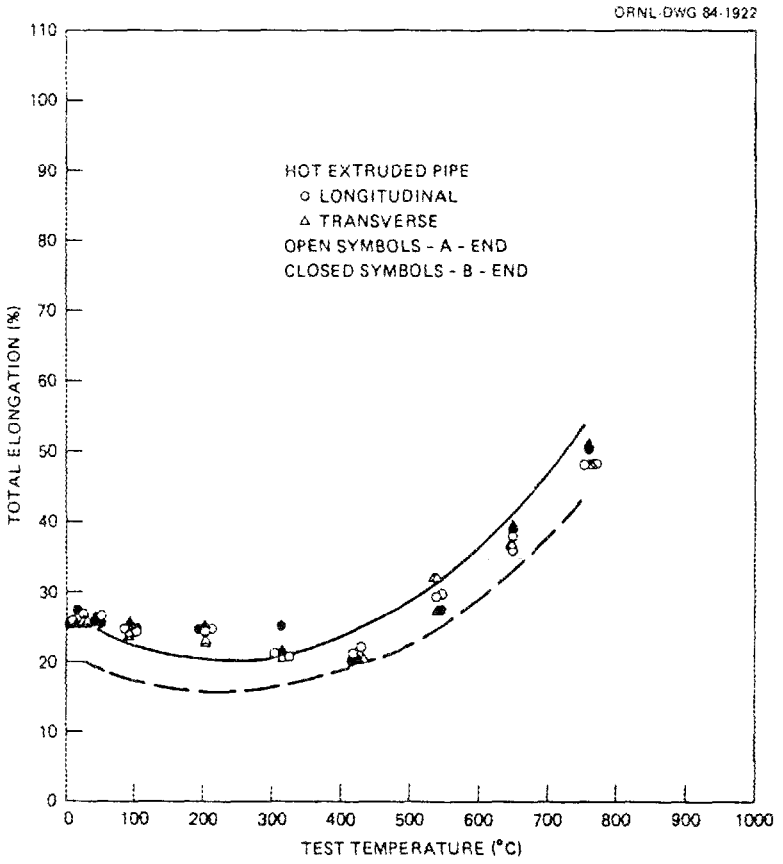


Fig. 33. Total elongation vs test temperature for hot-extruded pipe. Average (solid) and average − 1.65 SEE (dashed) curves for various product forms of several commercial heats are included for comparison.

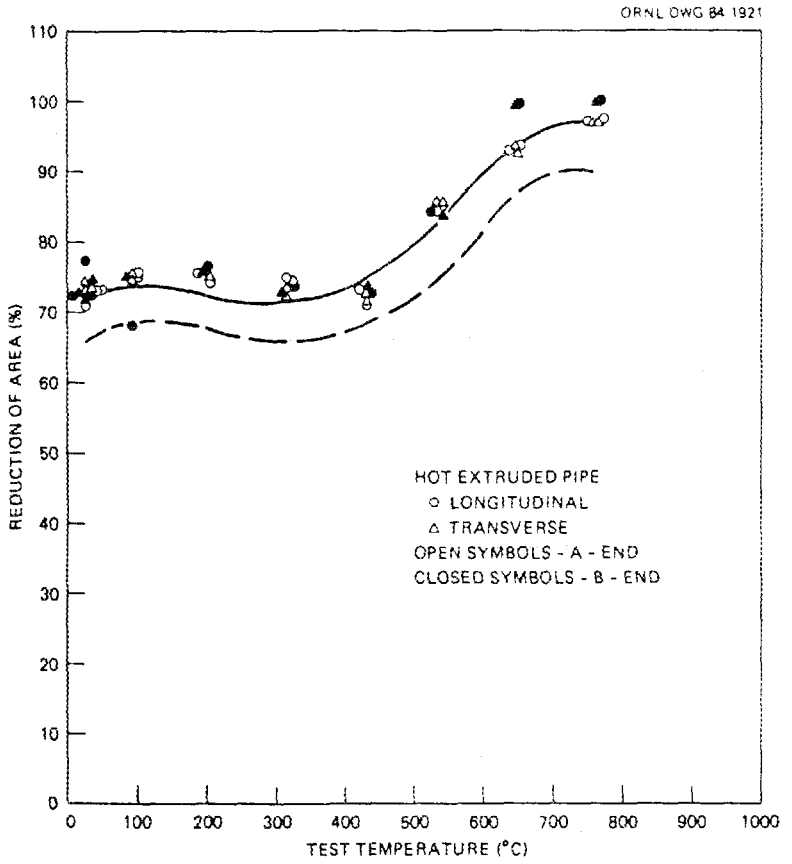


Fig. 34. Reduction of area vs test temperature for hot-extruded pipe. Average (solid) and average − 1.65 SEE (dashed) curves for various product forms of several commercial heats are included for comparison.

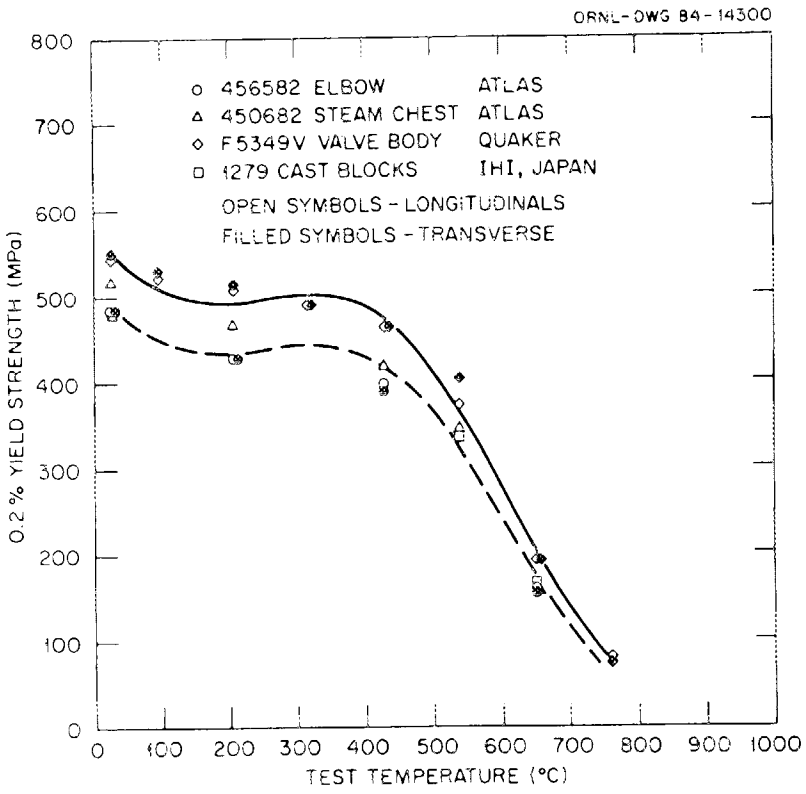


Fig. 35. Yield strength (0.2% offset) vs test temperature for various sand castings of modified 9Cr-1Mo steel. Average and average − 1.65 SEE curves for the wrought material are included for comparison.

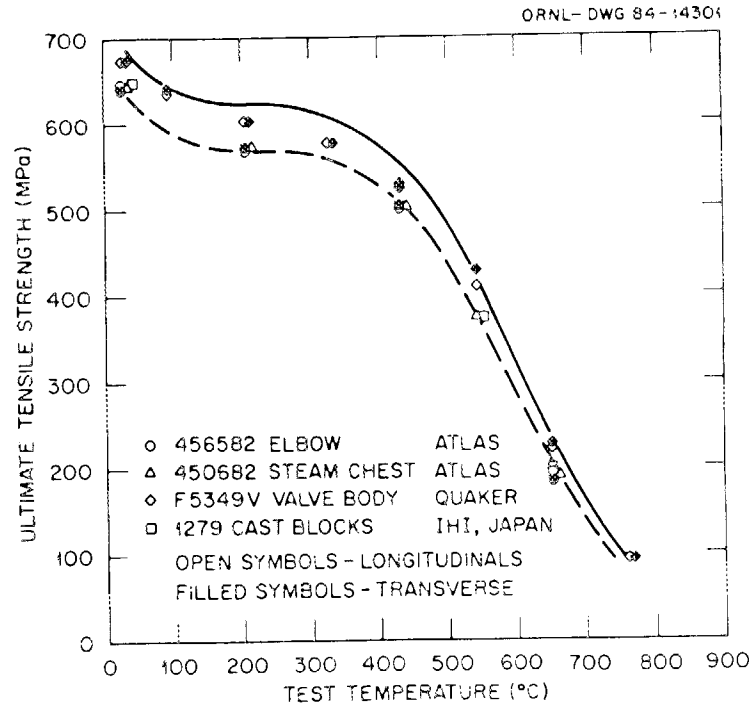


Fig. 36. Ultimate tensile strength vs test temperature for various sand castings of modified 9Cr-1Mo steel. Average and average − 1.65 SEE curves for wrought material are included for comparison.

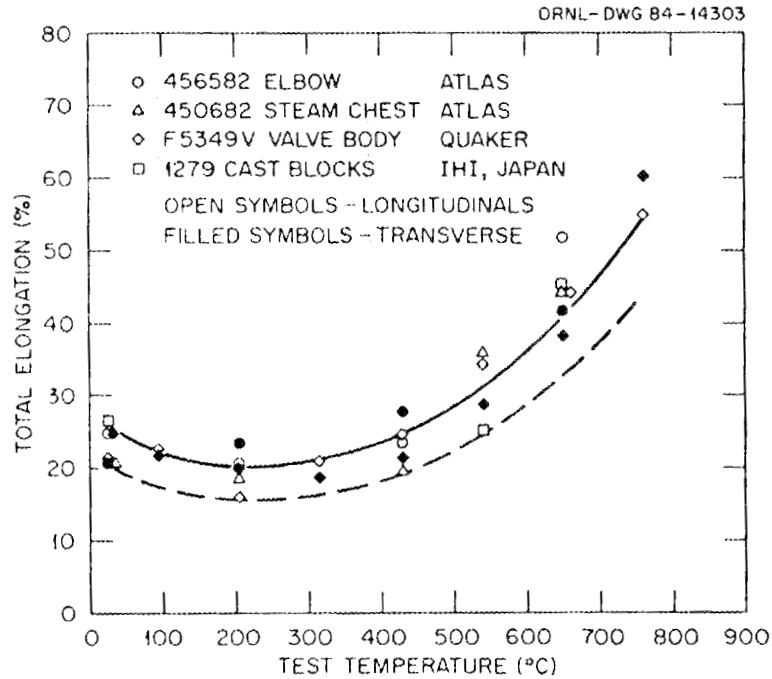


Fig. 37. Total elongation vs test temperature for various sand castings of modified 9Cr-1Mo steel. Average and average − 1.65 SEE curves for wrought material are included for comparison.

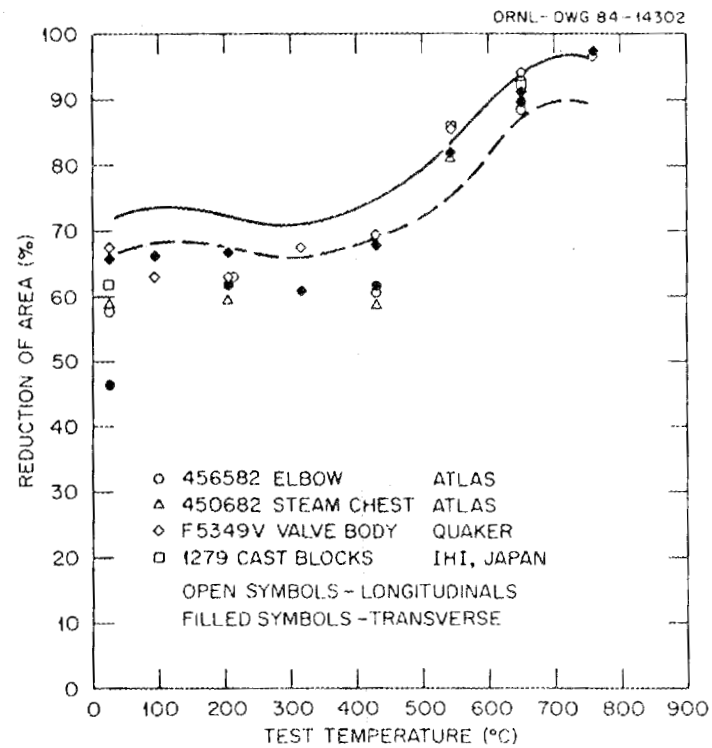


Fig. 38. Reduction of area vs test temperature for various sand castings of modified 9Cr-1Mo steel. Average and average − 1.65 SEE curves for wrought material are included for comparison.

Table 10. Status of creep-rupture tests on modified 9Cr-1Mo steel

Heat	Test temperature (°C)	Stress (ksi)	MPa	Test time[a] (h)	Primary creep strain (%)	Minimum creep rate (%/h) (10^{-4})	Total creep strain (%)	Total rupture strain[b] (%)	Rupture life (h)
394-E	704	4	28	21,809	1.00	2.00	14.24	14.33	21,809
143-61	593	18	124	11,893	0.09	0.03	0.14		
176[c]	538	30	207	7,511	0.45	0.58	0.93		
101	649	9	62	10,057	0.38	0.68	0.56		
143-61	649	11	76	14,321	0.50	1.25	2.85		
394-E	593	15	103	34,558	0.25	0.30	1.45		
394-E	649	11	76	6,620	0.06	9.67	21.35	21.48	6,620
5349-Y	538	26	179	56,606	1.85	0.10	2.17		
582	649	8	55	18,673	0.62	1.25	3.03		
394-E	593	16	110	13,154	0.40	0.53	1.07		
383-B	593	16	110	35,872	0.50	7.14	3.39		
582	593	16	110	11,738	1.20	1.88	3.26		
394-E	538	22.5	155	33,308	0.25	5.55	3.90		
394-E	538	34.0	234	33,295	0.15	0.17	0.71		
143-61	593	21.0	145	12,273	0.7	3.12	21.50	21.64	12,273
394-B	538	27.0	186	49,019	0.87	2.22	2.02		
394-E	454	55	379	36,765	0.23	0.08	0.97		
176	593	16	110	12,651	0.18	0.67	0.88		
001	593	18	124	19,275	0.70	1.15	3.19		
394-E[c]	538	30	207	7,728	0.80	0.63	1.22		
148-A	538	27	186	35,791	0.80	0.10	1.1		
394-B	427	60	413	48,144	0.10	0.04	0.33		
394-E	427	55	379	21,424	0.08	0.51	0.15		
176[c]	538	27	186	7,226	0.40	0.17	0.52		

[a]Test time through November 30, 1985.

[b]Total rupture strain includes creep strain plus loading strain.

[c]Sample aged for 2.5×10^3 h at 538°C before test.

ASME Code data package.[33] The status of some of the long-term creep tests still in progress is presented in Table 10. Note that the longest test time is 56,600 h, two tests have reached 49,000 h, and several have times of 35,000 h.

The creep-rupture and minimum creep rate data on commercial heats, tested in the N&T condition, were analyzed by lot-centered regression[42] with a generalized model selection procedure.[43] This procedure demonstrated that rupture data could be described quite well by a model of the form

$$\log t_r = C_h + a_1\sigma + a_2 \log \sigma + a_3/T \ , \qquad (3)$$

where

t_r = time to rupture (h),

σ = applied stress (MPa),

T = test temperature (K),

a_1 to a_3 = regression constants.

The term C_h is a lot constant that describes variations in creep-rupture strength among the different lots. The equation with actual constants, including C_h averaged for all heats studied, can be written as

$$\log t_r = -24.244 - 0.02374\sigma - 2.4871 \log \sigma + 31{,}876/T. \quad (4)$$

The minimum creep rate data were fitted to the following equation:

$$\log \dot{e}_m = 27.300 + 0.025265\sigma + 3.2172 \log \sigma - 35{,}594/T, \quad (5)$$

where minimum creep rate $\dot{e}_m$ is in percent per hour. The coefficient of determination R^2 for minimum creep rate data was 90.1%. Equations (4) and (5) are only slightly different from similar equations reported in ref. 37.

Isothermal stress-rupture data on commercial heats tested in the N&T condition are plotted in Figs. 39 and 40. Average curves predicted from Eq. (4) are also included in these figures. Note that the average curves describe the stress-rupture data of commercial heats extremely well. Also, note that data at 788°C was not used in developing the parametric model. Figure 41 shows the isothermal plots of stress vs minimum creep rate data on commercial heats tested in the N&T condition. Average curves from Eq. (5) are included in Fig. 41. Once again, the average curves describe the experimental data quite well.

The isothermal data for total elongation and reduction of area obtained on specimens tested in the N&T condition are plotted as a function of time-to-rupture in Figs. 42 and 43. In each case, the lines represent the minimum value for the parameter. Figure 42 shows that at all test temperatures total elongation exceeds 20%. Reduction-of-area exceeds 60% for all test temperatures. Neither total elongation nor reduction-of-area shows a systematic decrease with increasing rupture time up to 20,000 h.

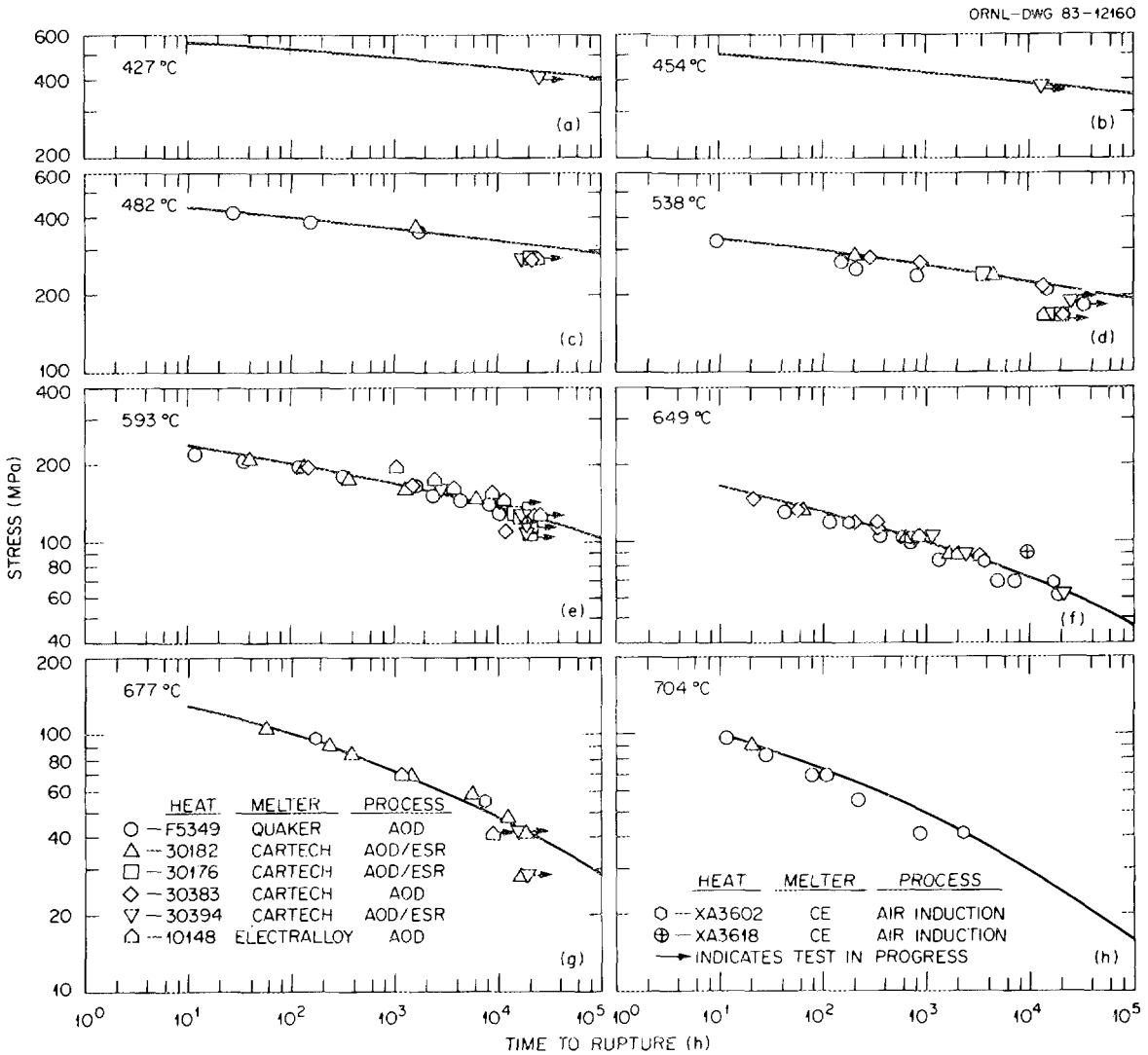


Fig. 39. Stress-rupture data for various heats of modified 9Cr-1Mo steel from 427 through 704°C. Average curves from parametric model based on several commercial heats are shown for comparison.

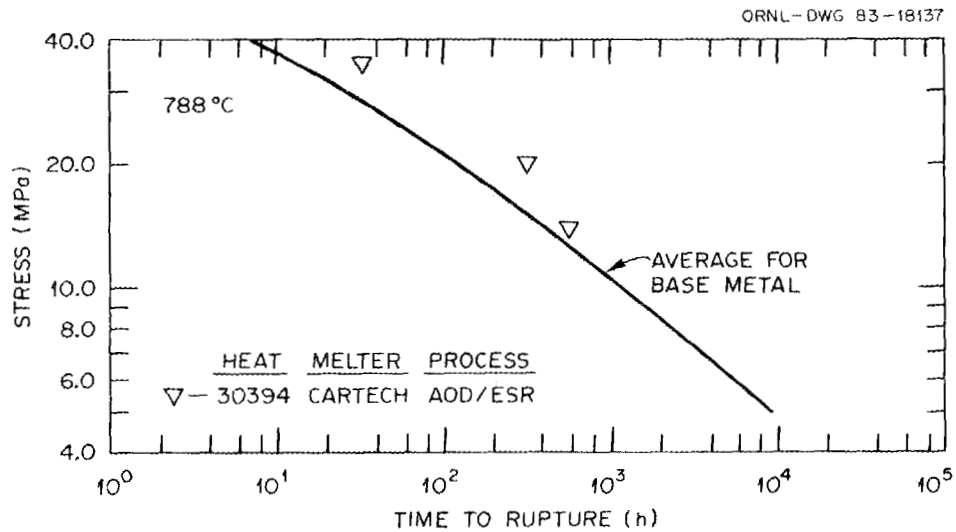


Fig. 40. Stress rupture data for CarTech heat 30394 of modified 9Cr-1Mo steel at 788°C, compared with average curve based on data from several heats at lower temperatures.

Creep rupture data at 593°C on commercial heats of modified 9Cr-1Mo steel are plotted in Fig. 44. In this figure data on modified 9Cr-1Mo are compared with the average curves for standard British 9Cr-1Mo (N&T), standard American 9Cr-1Mo (annealed), and 2.25Cr-1Mo steel (annealed). The following observations can be made:

1. Modified 9Cr-1Mo steel is significantly stronger than standard 9Cr-1Mo and 2.25Cr-1Mo. The strength difference becomes even more evident for rupture times exceeding 1000 h.
2. Comparison of data for the standard 9Cr-1Mo steels indicates that changing the heat treatment from annealed to N&T improves the creep-rupture strength only a small amount. However, for the same heat treatment (N&T) the compositional modifications in our alloy produced a very large improvement in creep-rupture strength.

The 10^5-h creep-rupture strength at temperatures from 427 to 732°C of various alloys is compared in Fig. 45. Modified 9Cr-1Mo steel is stronger than standard 9Cr-1Mo and 2.25Cr-1Mo steel (N&T) over the entire temperature range. The strength differences become greater for temperatures exceeding 500°C. The alloy also exceeds the strength of type 304 stainless steel up to 600°C.

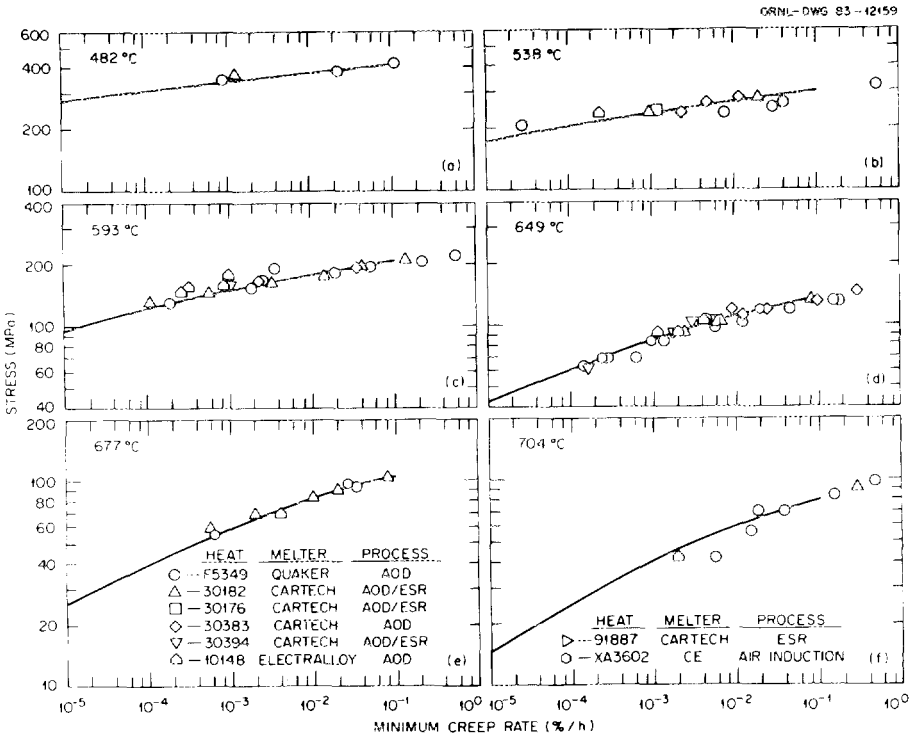


Fig. 41. Stress-minimum creep rate data for commercial heats of modified 9Cr-1Mo steel tested in normalized and tempered condition. Average curves from time-temperature parameter are also included in these plots.

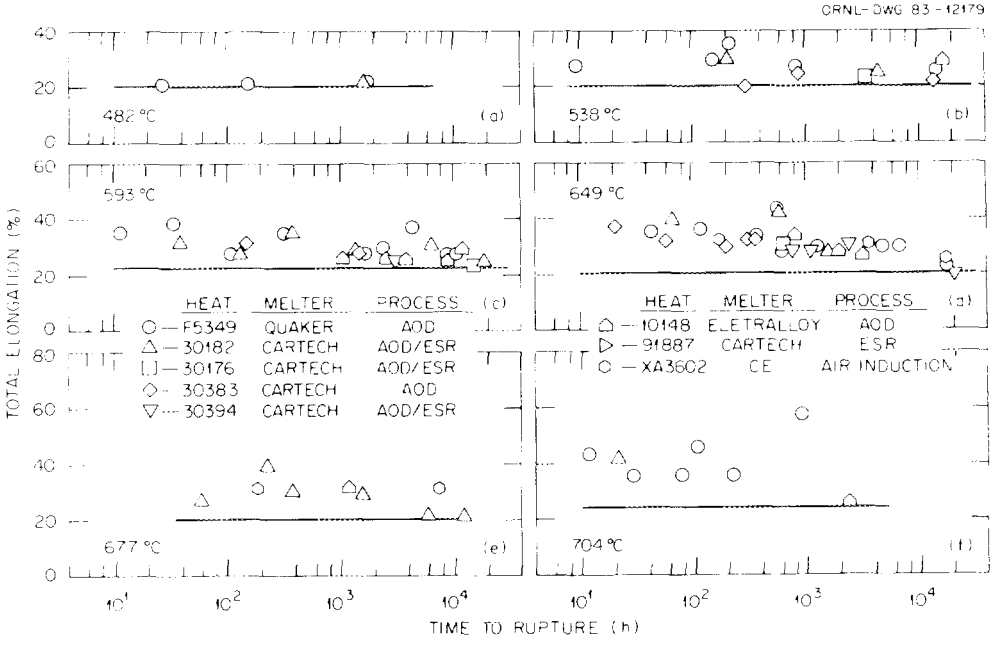


Fig. 42. Total elongation as function of time to rupture for commercial heats of modified 9Cr-1Mo steel tested in normalized and tempered condition. Lines in these plots represent estimated minimum value of total elongation for each test temperature.

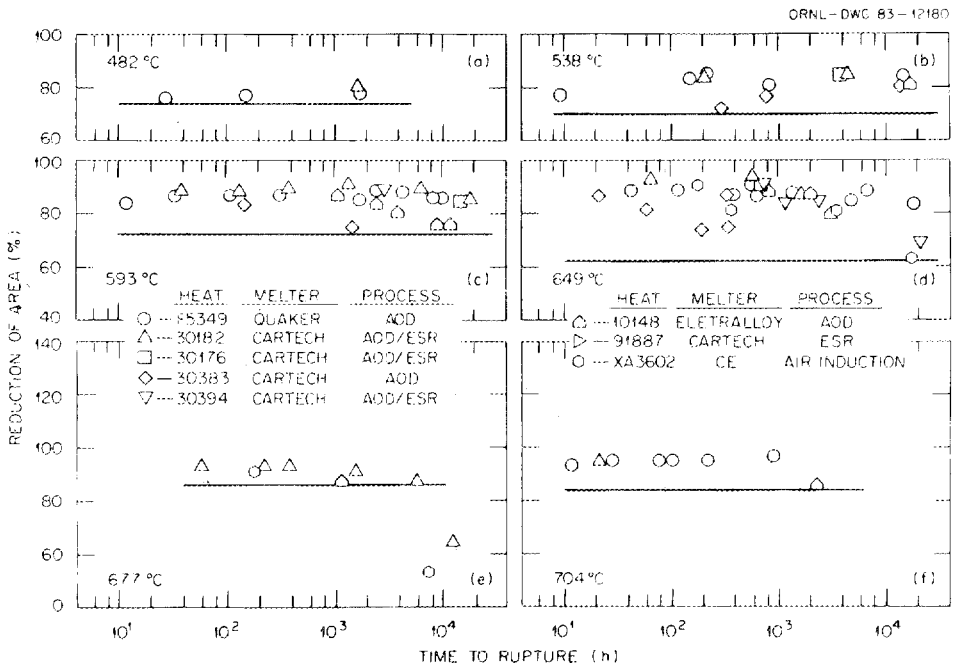


Fig. 43. Reduction of area as function of time to rupture for commercial heats of modified 9Cr-1Mo steel tested in normalized and tempered condition. Lines in these plots represent estimated minimum value of reduction of area for each test temperature.

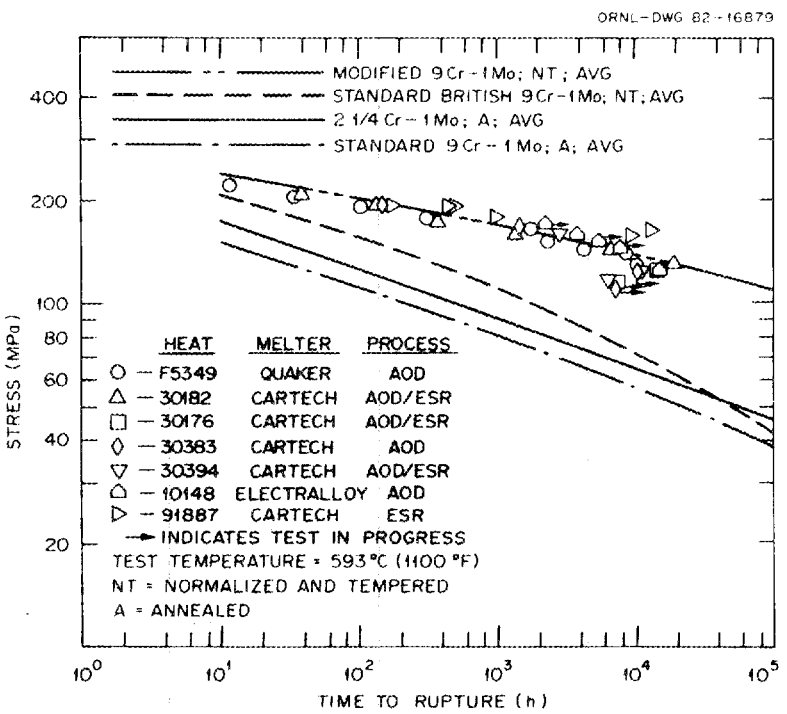


Fig. 44. Stress-rupture plot at 593°C comparing data on commercial heats of modified 9Cr-1Mo steel (tested in normalized and tempered condition) with average curves for standard British 9Cr-1Mo steel tested in normalized and tempered condition and standard American 9Cr-1Mo steel and 2.25Cr-1Mo steel tested in annealed condition.

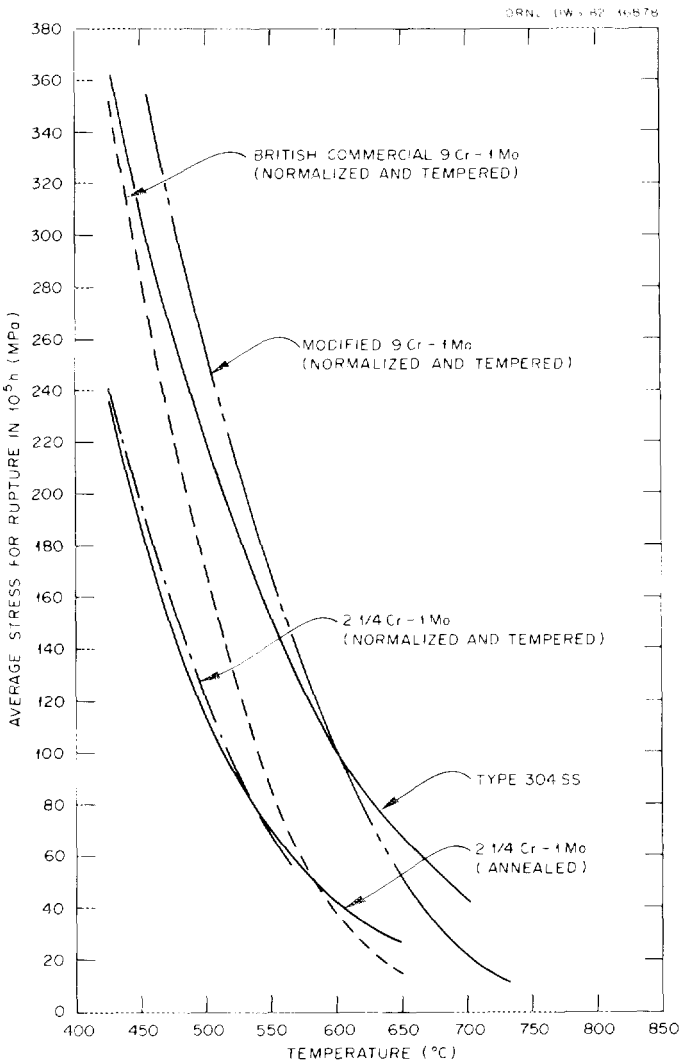


Fig. 45. Variation of 10^5-h creep-rupture strength with temperature for several materials.

There has been some concern expressed about the loss in ductility that sometimes occurs when the creep strength of an alloy is increased. The total elongation and reduction of area from creep-rupture tests conducted thus far on the commercial heats of modified 9Cr-1Mo steel are plotted in Fig. 46. Note that the total elongation for most heats exceed 15% and the reduction of area exceeds 70%. Furthermore, for all of the test temperatures these data show no indication of a drop in ductility with increasing rupture time.

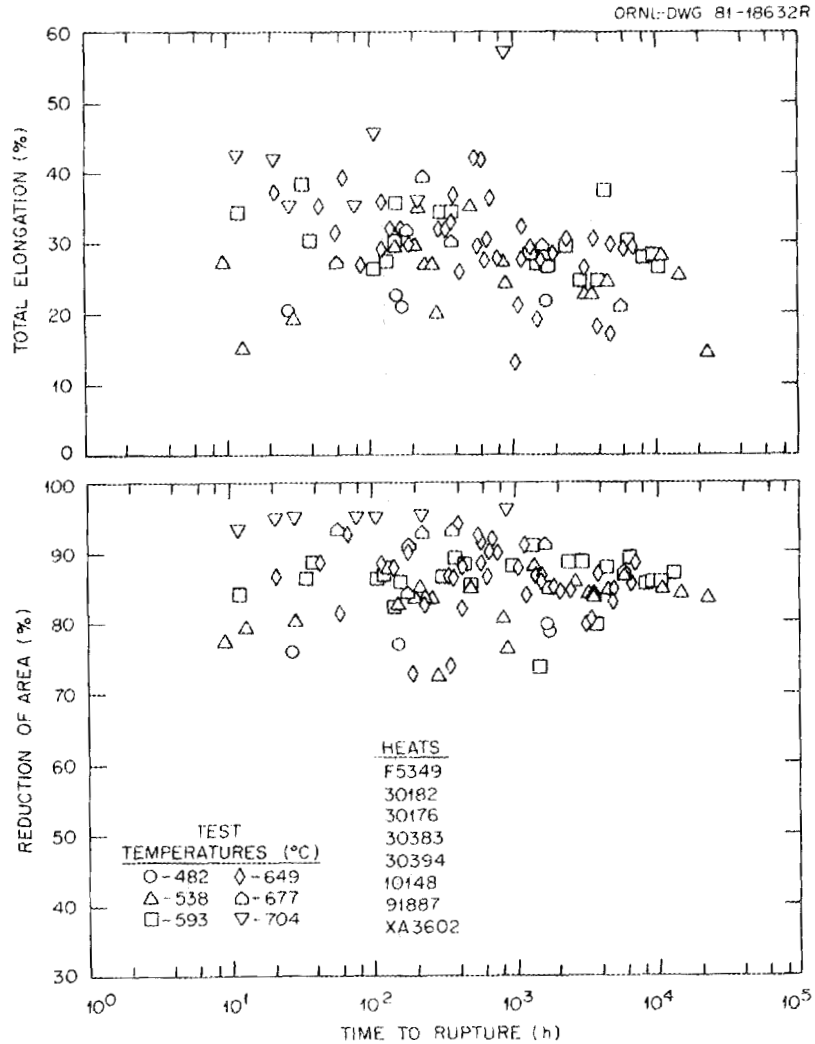


Fig. 46. Plots of (*a*) total elongation and (*b*) reduction of area functions of rupture time for commercial heats of modified 9Cr-1Mo steel tested in normalized and tempered condition. Data are for test temperatures in the range 482 to 704°C and for eight commercial heats.

The effect of other variables – such as isothermal annealing, normalizing temperature, tempering temperature, cold work, notch in the specimen, and biaxiality (tube burst) – on creep properties have also been studied and reported in ref. 35. Conclusions are summarized below.

1. Creep and rupture strengths of isothermally annealed material were slightly lower than the average for N&T material at 538°C; however, the strength values were the same for two heat-treated conditions at

593, 649, and 677°C. The total elongation and reduction of area data for isothermally annealed material exceeded the minimum value curves for the N&T material.

2. Cold work (compressive or cold rolled) to 20% has an unimportant effect on the creep-rupture strength at 593 and 649°C. The reduction of area for the long-term test at 593°C fell below the minimum for N&T material; however, the lowest value still exceeded 40%.
3. Increasing the normalizing temperature and/or time increased the rupture time. Normalizing at 1040°C and tempering at 760°C produced the optimum strength and ductility properties (Fig. 20).
4. Tempering above 760°C has an insignificant effect on creep strength. However, tempering below 760°C (i.e., 732°C) significantly improves creep strength.
5. A notch with a stress concentration factor of 3.9 slightly strengthens this alloy.
6. Tubular specimens showed creep strengths similar to those of uniaxial specimens.
7. Modified 9Cr-1Mo alloy is insensitive to several material, heat treatment, and specimen design variables. The creep process occurs by rearrangement of the dislocation structure, thus leading to a cleaner matrix and causing final fracture to occur by ductile void tearing.

Creep data in Figs. 39 and 40 were from tube, plate, and bar products of commercial heats. Figures 47 and 48 compare creep data from a large forging (Fig. 4) and a large diameter extruded pipe (Fig. 3). Note that data on these products are comparable to the average behavior of other product forms. Figure 49 compares the creep-rupture data of castings with the average of the wrought material. The sand castings appear to be somewhat weaker.

Figure 50 compares the creep data on heats melted in Japan with those melted and processed in the United States. Once again, there appears to be no difference in creep properties between the two sets of materials.

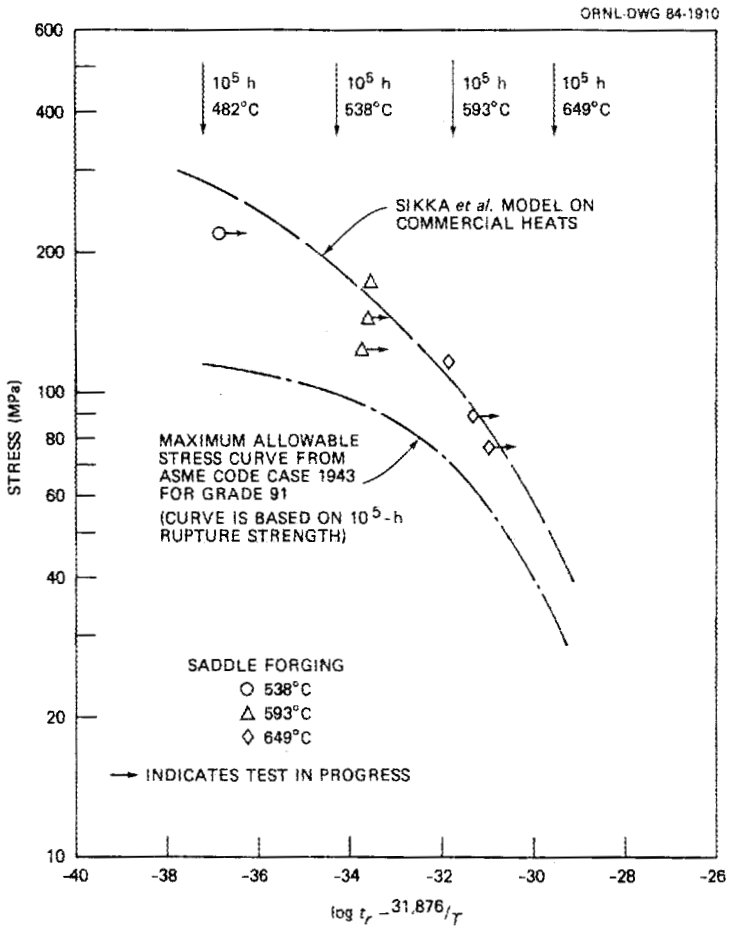


Fig. 47. Comparison of creep-rupture data on saddle forging with average curve developed from model for various product forms of several commercial heats. Maximum allowable design stress curve from ASME Code Case 1943 for Grade 91 is also included for comparison.

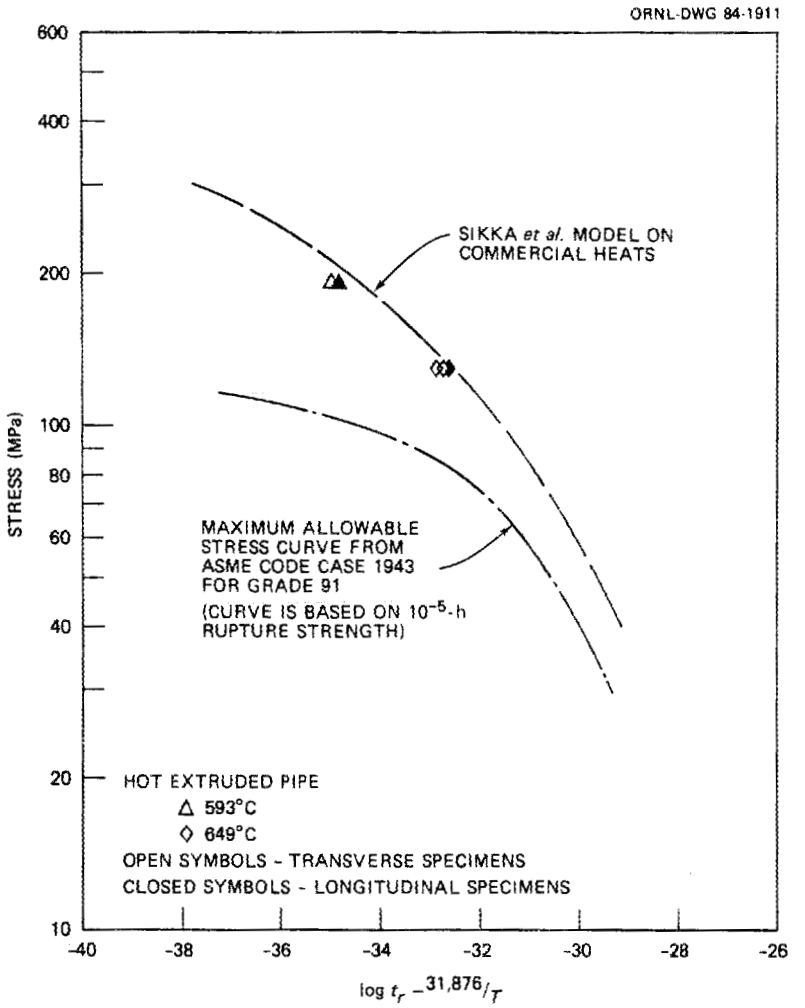


Fig. 48. Comparison of creep-rupture data on hot-extruded pipe with the average curves for various product forms of several commercial heats of modified 9Cr-1Mo steel.

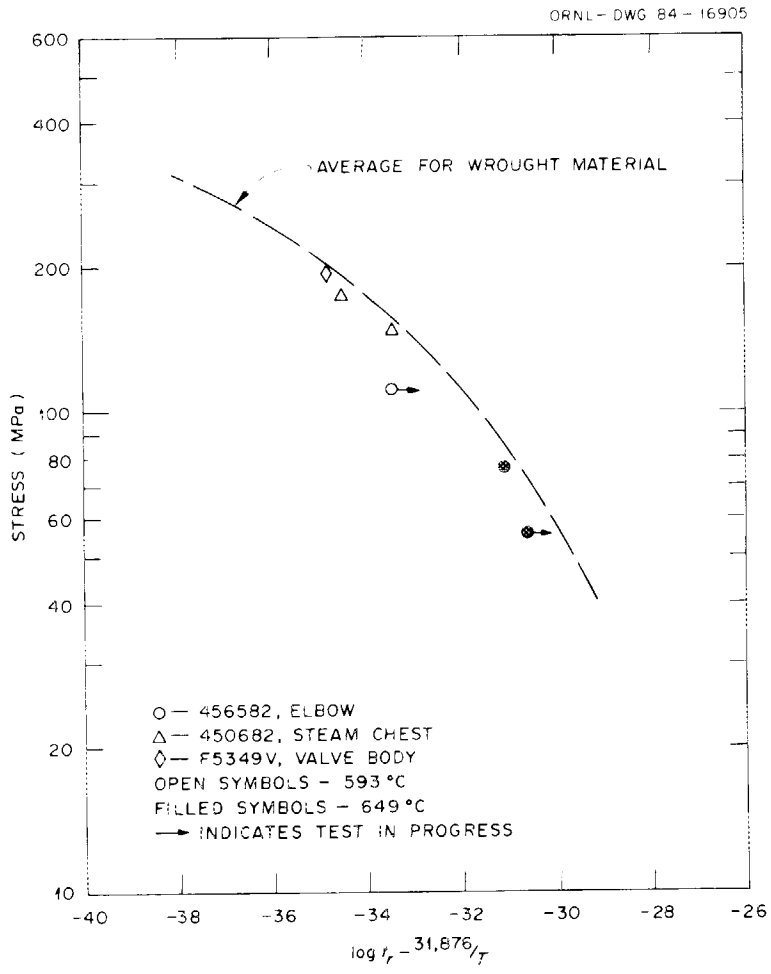


Fig. 49. Comparison of creep-rupture strength of modified 9Cr-1Mo sand castings with average curve for wrought material.

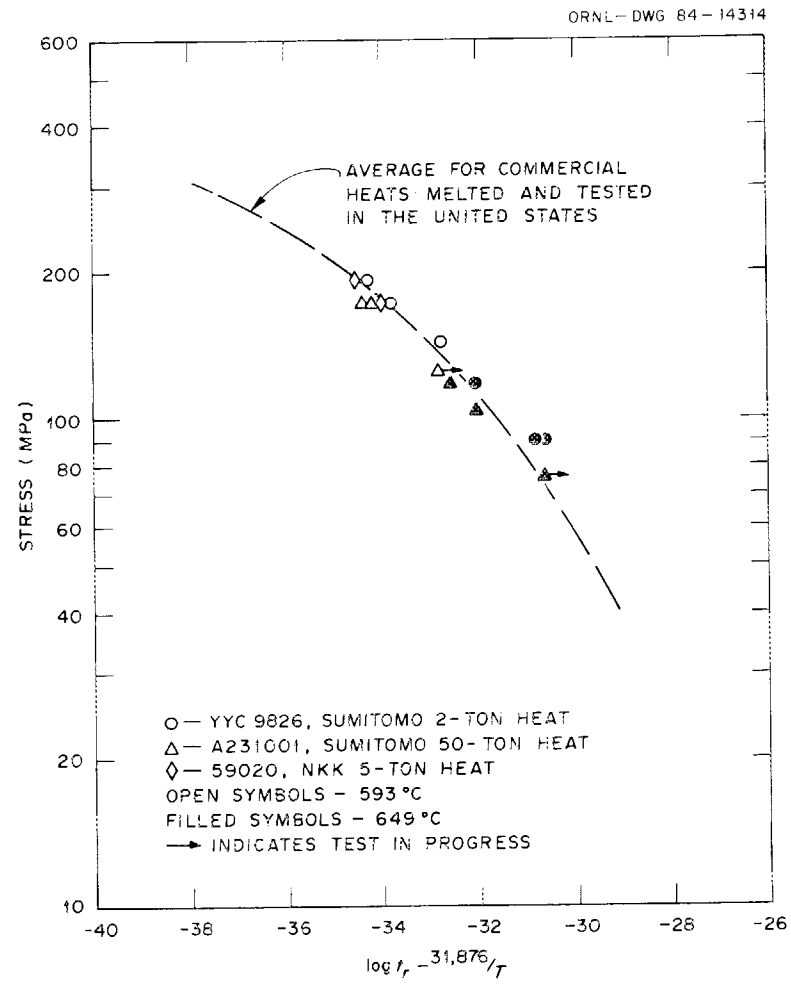


Fig. 50. Comparison of creep-rupture data on heats of modified 9Cr-1Mo steel melted and fabricated in Japan with average curve for commercial heats melted and tested in the United States.

3.3.2.3 Fatigue and Creep-Fatigue Behavior

The fatigue and time-dependent (creep-fatigue) behavior of this alloy has been studied in some detail by a number of investigators. Their conclusions have generally been as follows.

The continuous-cycle fatigue behavior of wrought material as determined from strain-controlled, isothermal, fully reversed tests is similar to that of other common structural alloys in the low-cycle range. In the high-cycle range, differences caused by strength variations in materials are apparent (Fig. 51). Similar data show that modified 9Cr-1Mo steel has superior fatigue life in comparison with the standard 9Cr-1Mo steel.[44] Continuous-cycle data have been generated at 25, 371, 482, 538, and 593°C primarily by ORNL,[44-48] but also by Idaho National Engineering Laboratory (INEL),[49] and ANL.[50,51] These data have shown that there is not a strong effect of temperature on fatigue life, but heat-to-heat variations have been observed. For example, heat 30176 appears characteristically weak in comparison with other heats shown in Table 11.

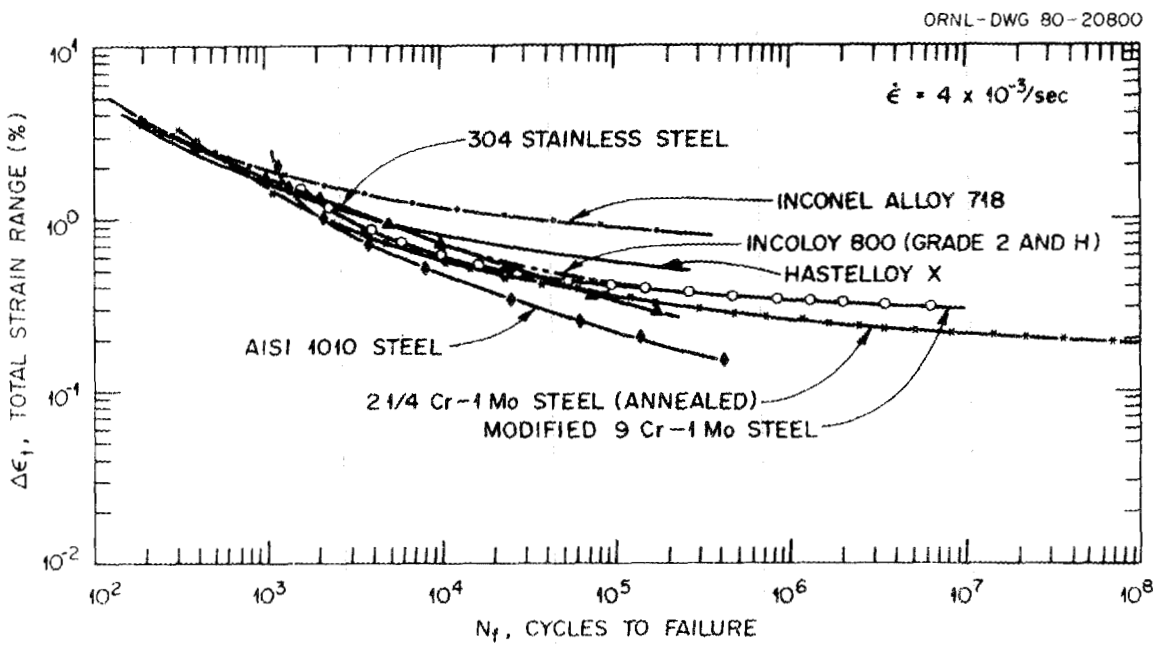


Fig. 51. Comparison of fatigue behavior of several materials at 538°C (1000°F). Lines represent best-fit values of actual data. Data for type 304 stainless steel include tests conducted at 538°C (1000°F) and 566°C (1050°F).

Table 11. Comparison of high-cycle fatigue lives and ultimate strength values of modified 9Cr-1Mo steel at 538°C

Heat	Total strain range (%)	Cycles to failure	Ultimate tensile strength (MPa)
10148	0.3	5,812,800	452
30176	0.3	190,733	434
30394	0.3	4,055,050	462

Tests in sodium indicate that fatigue life is superior to that in air. Short-term exposure to sodium (~6000 h at 538°C) had little or no effect on fatigue behavior of this steel.[50] Other investigators have similarly shown the beneficial effects of excluding oxygen from the environment surrounding specimens being subjected to elevated-temperature cyclic loading.[52,53]

The effects of product form and processing history on the continuous-cycle fatigue life of this material have been studied. Figure 52 compares data obtained by Ishikawajima-Harima Heavy Industries Company, Limited, on specimens taken from sand castings and forgings with a best fit curve from the ORNL data base. Note that specimens from cast material show somewhat lower fatigue resistance in comparison with those from wrought product forms.[45] Ebi and McEvily[54] compared the continuous cycle fatigue lives at 538°C of hot-forged (31-μm grain size) and hot-rolled (19-μm grain size) material. In both cases, specimens were normalized at 1038°C for 1 h in air and then tempered for 1 h at 760°C. The hot-rolled material, with a smaller prior austenite grain size, had the highest fatigue life, particularly at cycle lives in excess of about 4×10^3 cycles. Therefore, minimizing the prior austenite grain size of modified 9Cr-1Mo steel is beneficial to fatigue life.

The influence of tempering temperature on low-cycle (10^3–10^5) fatigue behavior from 760 to 820°C has been determined. Little or no effect of tempering temperature was observed.[47]

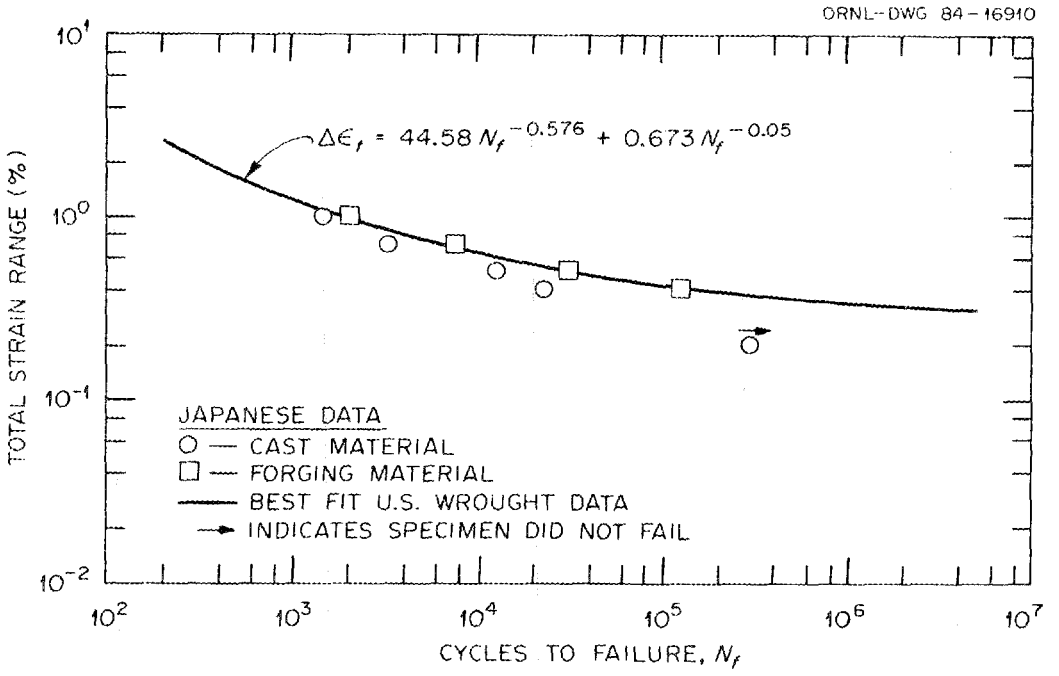


Fig. 52. Comparison of room-temperature continuous-cycle fatigue behavior for modified 9Cr-1Mo steel from two sources (ORNL and Japan) and several product forms. Note that cast form shows somewhat lower fatigue resistance.

Under strain-controlled cyclic-loading conditions, modified 9Cr-1Mo steel in the N&T condition (tempered martensite microstructure) undergoes considerable cyclic softening[44,48,52,54-56] (i.e., the stress required to maintain a constant strain range decreases with increasing cycles). Depending upon the strain range and strain rate, this stress may drop to nearly one-half of its starting value.[52] This softening is also indicated by a hardness change that occurs in the gage section. For example, after a specimen was fatigued to failure at a strain range of 0.5% at 538°C, the hardness in the gage section was 196 dph; that in the head section was 267 dph. Only limited microstructural characterization has been performed on cyclically softened material.[56] Additional information concerning cyclic stress-strain response can be found elsewhere.[55]

Comparisons have been made between the total strain range vs cycles to failure and results from thermal striping tests in which a tubular 9Cr-1Mo steel specimen was rotated about its major axis while being subjected to streams of flowing sodium at different temperatures from

diametrically opposed positions. Results from these latter tests[57,58] show good agreement with results from standard isothermal fatigue tests. Further, the results clearly demonstrate the superior thermal striping resistance of modified 9Cr-1Mo steel in comparison with type 316 stainless steel because the 9Cr-1Mo material was able to withstand ΔTs of ~83°C higher than the stainless steel.

Ferritic alloy steels are generally more resistant to creep/fatigue interaction (intergranular void formation and subsequent crack propagation) under cyclic loading conditions than types 304 and 316 stainless steel at temperatures less than about 565°C.[53] Limited tensile hold time tests conducted on modified 9Cr-1Mo steel have shown that grain boundary cavities can form under tensile-strain-hold time conditions at 593 and 700°C.[52] However, no indication of cavitation has been reported at 538°C where most of the testing has been done.[44-46,53] Compressive hold periods reduce fatigue life when the hold period occurs at the peak compressive strain in fully reversed tests (Table 12). Limited comparisons are also made with results from tests conducted on 2.25Cr-1Mo steel under identical test conditions. Although there is considerable data scatter, similar reductions in life occur in the two materials. The reason compressive hold periods are deleterious to fatigue life can be seen from a plot of

Table 12. Comparison of the fatigue life of 2.25Cr-1Mo steel (annealed) and modified 9Cr-1Mo steel (normalized and tempered) at 538°C and at a strain range of 0.5%

Compressive hold period (h)	Fatigue life (h)	
	2.25Cr-1Mo[a]	9Cr-1Mo[b]
0	17,329	52,400
0.01	6,775	19,291
0.1	4,496	
0.25		8,840
0.5		5,173
1.0	4,133	

[a]Annealed heat 3P5601.

[b]Modified heat 30394.

both tensile and compressive hold time data for modified 9Cr-1Mo steel in Fig. 53. Compressive hold periods introduce a shift in the hysteresis loop such that a tensile mean stress develops, the magnitude of which is dependent upon strain range and duration of hold time. Figure 53 shows that the lower the strain range and the longer the compressive hold period, the greater the magnitude of the tensile mean stress with resultant lower fatigue life. The reverse tends to occur with tensile hold periods at this temperature (538°C). At higher temperatures (i.e., 593°C), tensile hold periods become more detrimental as classical creep damage begins to occur (Fig. 54). Long-term tensile hold time tests are needed at 538°C to determine if low strain rates induce intergranular cavitation with subsequent pronounced reduction in fatigue life of this material.

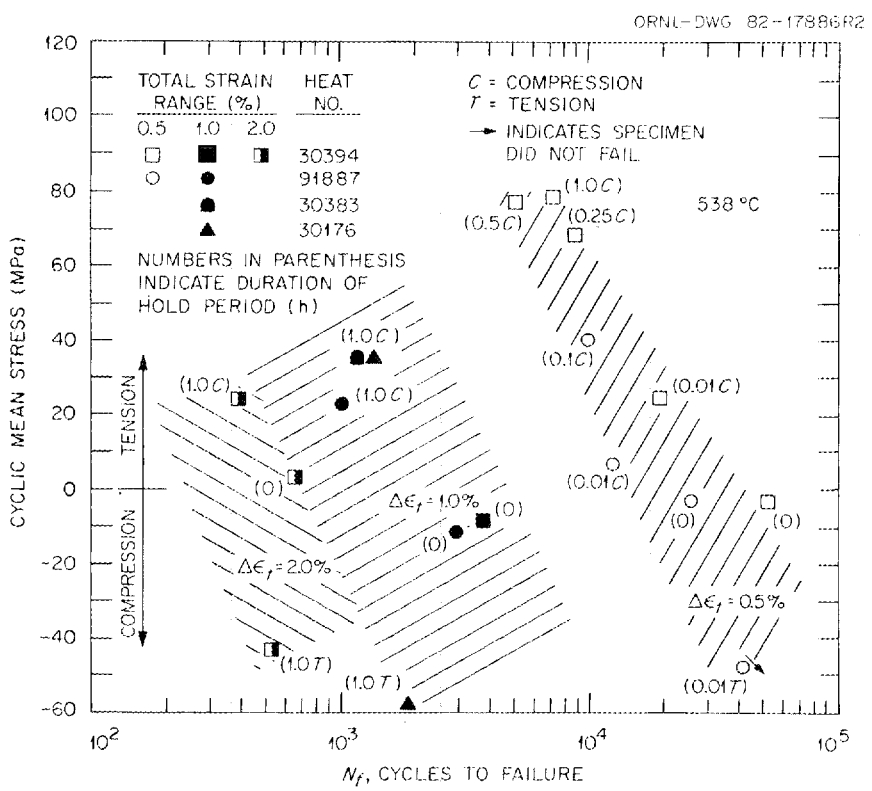


Fig. 53. Compressive hold periods result in cyclic mean tensile stress, magnitude of which depends on duration of hold period and on strain range.

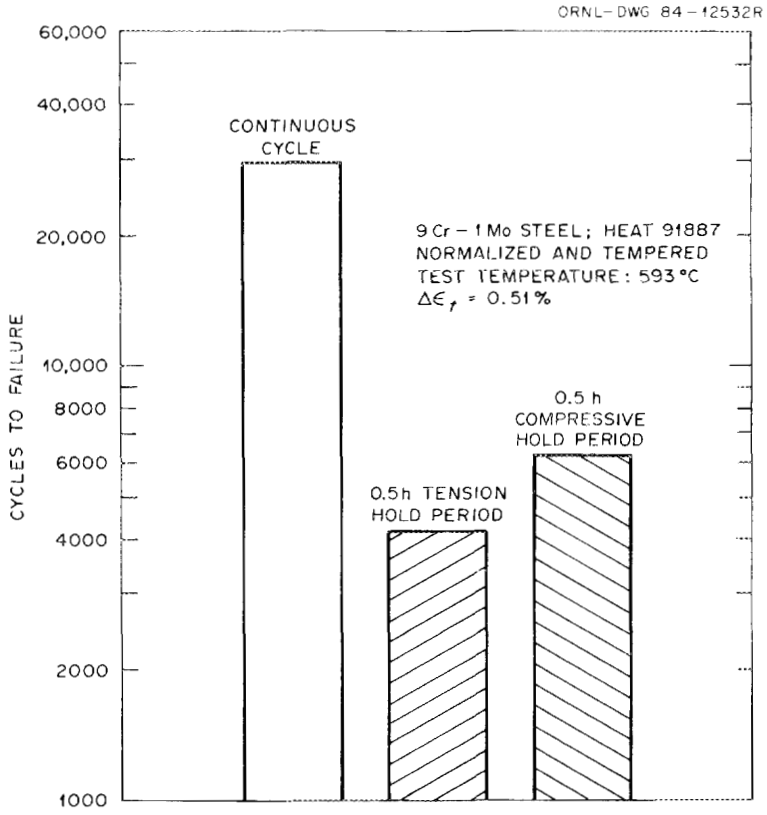


Fig. 54. Comparison of fatigue life of three specimens of modified 9Cr-1Mo steel tested at strain range of 0.51% at 593°C. Note that tensile hold time produced a somewhat lower fatigue life than corresponding compressive hold period.

A series of load-controlled fully reversed fatigue tests have been conducted on notched specimens at 25 and 538°C. The tests were fully reversed (R = −1) with notch geometries of k_t = 1.3, 2.6, or 4.0. Tests were conducted both with and without tensile hold periods to determine if rules given in ASME Code Sect. III, N-47, are adequate for this material. Although the tests are still ongoing, results thus far indicate that specimens with a notch accuity k_t of 1.3 have only a slight effect on fatigue life; specimens with a k_t of 4.0, however, have a markedly reduced cycle life in the intermediate- to high-cycle range.

3.3.2.4 Creep and Fatigue Crack Growth

Limited creep crack growth studies have been conducted with precracked specimens at 593°C. Results show similar crack growth rates to that of type 316 stainless steel at this temperature.[59] Fatigue crack

growth studies[60] have been conducted on three heats (plate) of this material with cyclic tests employing standard modified compact specimens at 24, 316, 427, 538, and 593°C. The material tested came from heats melted by argon-oxygen decarburization (AOD) or AOD/ESR processing. No effect of melt practice or heat-to-heat variations was found. Several specimens were also aged for 10,000 h at 474 to 482°C before test to determine if temper embrittlement that would accelerate crack growth rates occurred. No effect of this prior aging treatment was found on fatigue crack growth behavior. Stress ratio effects were also studied in the range of $0.05 < R < 0.667$, and the effects were not large. Crack propagation studies were also conducted on gas tungsten-arc weldments, and growth rates were found to be equivalent to those in wrought material. A similar conclusion was drawn by Ebi, who used low-cycle fatigue specimens to study fatigue behavior of electron beam welded material.[61]

3.3.2.5 Charpy-Impact Data

Improved Charpy-impact properties were one of the primary requirements for modified 9Cr-1Mo steel. As a result, each experimental heat and the commercial heats were tested to generate Charpy-impact energy vs test temperature curves. In addition to Charpy-impact energy, specimens were also used to determine lateral expansion and 100% shear fracture appearance data. Data on specimens tested at ORNL have been included in previous progress reports.[62-66] CE conducted Charpy-impact testing on the experimental heats, and their results are summarized in a report by Bodine et al.[67]

The effect of trace element impurities on the impact properties of modified 9Cr-1Mo steel was studied at Auburn University.[68] Conclusions from this work can be summarized as follows.

1. Phosphorus or sulfur additions increased DBTT and decreased the upper-shelf energy (USE). The effect of sulfur was more pronounced than that of phosphorus.
2. Phosphorus caused no change in the microstructure; sulfur, however, promoted the formation of nonmetallic inclusions.
3. Silicon and copper also embrittled modified 9Cr-1Mo steel. The DBTT increased, and USE decreased after the addition of either element.

4. Silicon favored formation of delta ferrite; copper seemed to increase the amount of carbides present.

All of the concentrations of trace elements used in the Auburn study were outside the range specified for modified 9Cr-1Mo steel. In spite of the above conclusions, the study suggested that the specified concentrations of P, S, Si, and Cu in modified 9Cr-1Mo steel could be raised without adverse effects on the overall toughness of the alloy.

The curves of Charpy V-notch impact energy vs test temperature for two commercial heats of modified 9Cr-1Mo steel are compared with a heat of standard 9Cr-1Mo steel in Fig. 55. All tests were conducted on electric-furnace-melted standard 9Cr-1Mo specimens with the V-notch parallel to the rolling direction. One modified 9Cr-1Mo steel heat was melted by both the AOD and AOD-ESR combination. The second heat was melted by the AOD-ESR process. Data show that the 68-J (50-ft-lb) DBTTs for both heats of modified 9Cr-1Mo are lower than that for the standard 9Cr-1Mo heat. The USEs for the AOD-ESR-melted modified 9Cr-1Mo heats are much higher than those for either AOD-melted modified 9Cr-1Mo steel or electric-furnace-melted standard 9Cr-1Mo steel. These data also show that Charpy properties are affected by both the silicon content and the melting practice.[69]

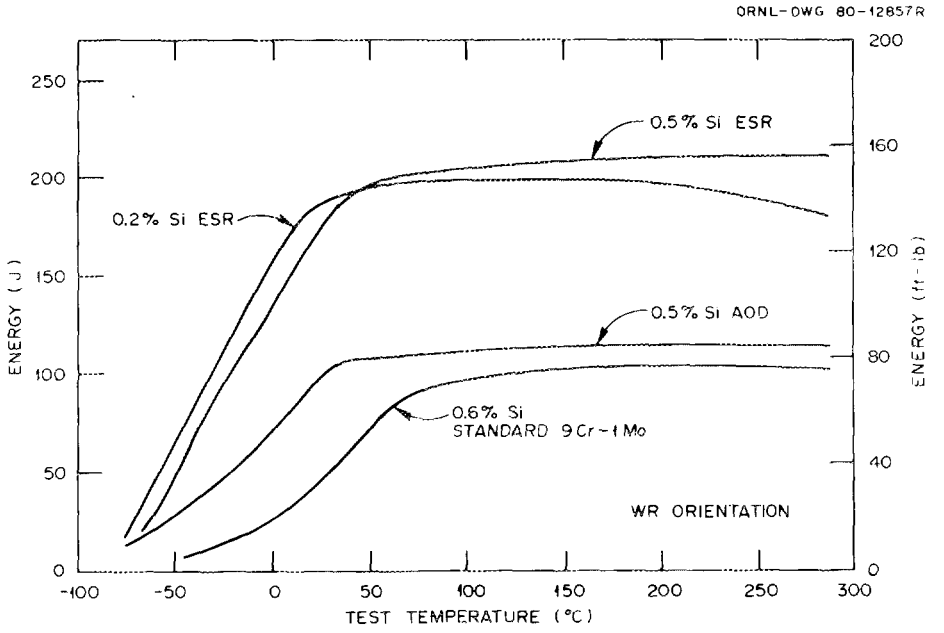


Fig. 55. Effect of melting process on Charpy V-notch toughness from two 14-Mg (15-ton) commercial ESR heats of modified 9Cr-1Mo steel and one standard 9Cr-1Mo steel plate.

Lower silicon content or ESR produces a lower transition temperature and higher USE. However, between the two factors, melting practice appears to have a much more potent effect than silicon.

Charpy-impact energy results for two modified 9Cr-1Mo heats processed by the AOD process are compared in Fig. 56. Heat 30383 was top poured into the ingot; heat 10148 was bottom poured. Data in Fig. 56 show that excellent impact properties of modified 9Cr-1Mo steel can be achieved even after AOD processing. However, to obtain superior properties, ingots should be bottom poured, a commercial practice apparently becoming more standard for large heats of steels. Figure 57 shows that ESR can further improve the impact properties of bottom-poured AOD ingots of modified 9Cr-1Mo steel.

Tubing made from 2- and 50-ton heats of modified 9Cr-1Mo steel melted at Sumitomo Metals Ltd., Japan, were also tested for impact properties at ORNL (Fig. 58). Note that not only are the impact properties excellent, but they also are the same for small and very large heats. This very important fact indicates that the results from small commercial heats are extrapolatable to large heats.

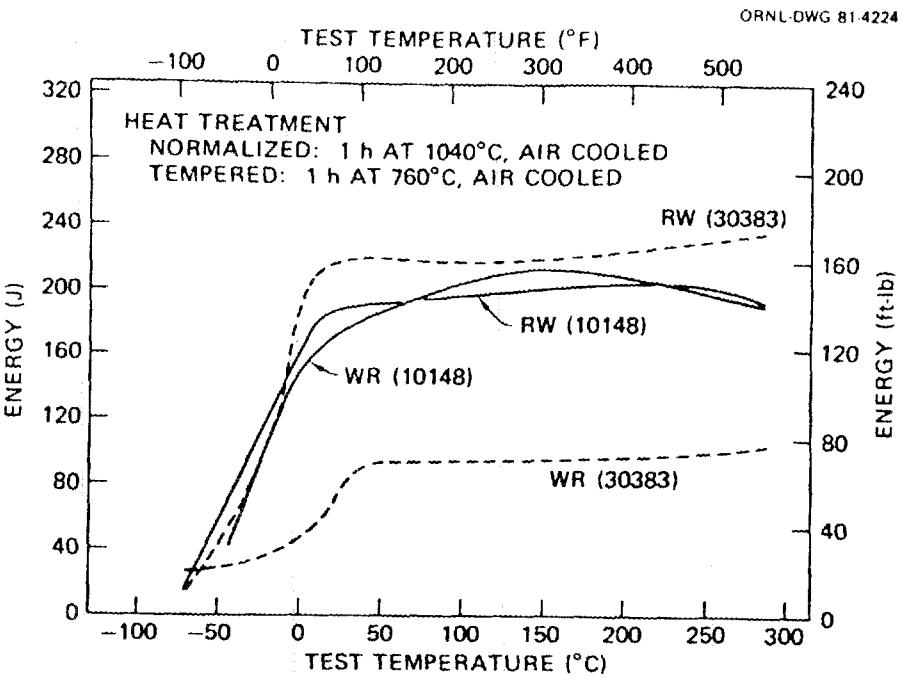


Fig. 56. Effect of specimen orientation on Charpy-V energy of 16-mm-thick Electralloy AOD 10148 and 15-mm-thick CarTech AOD 30383 modified 9Cr-1Mo steel plate.

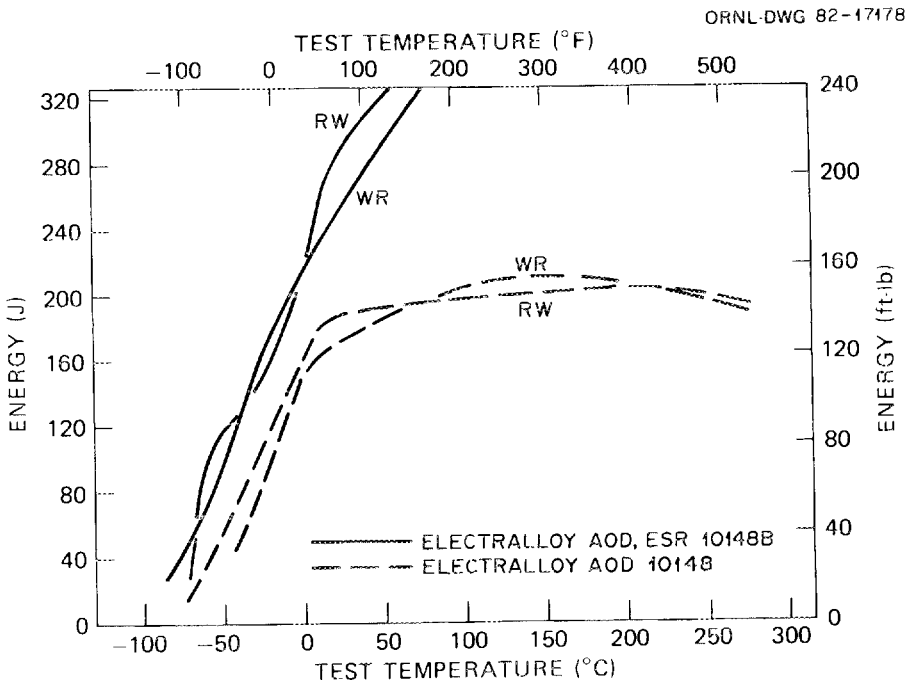


Fig. 57. Comparison of impact energy values of same heat (10148) of modified 9Cr-1Mo steel, processed from bottom-poured AOD ingots and after electroslag remelting.

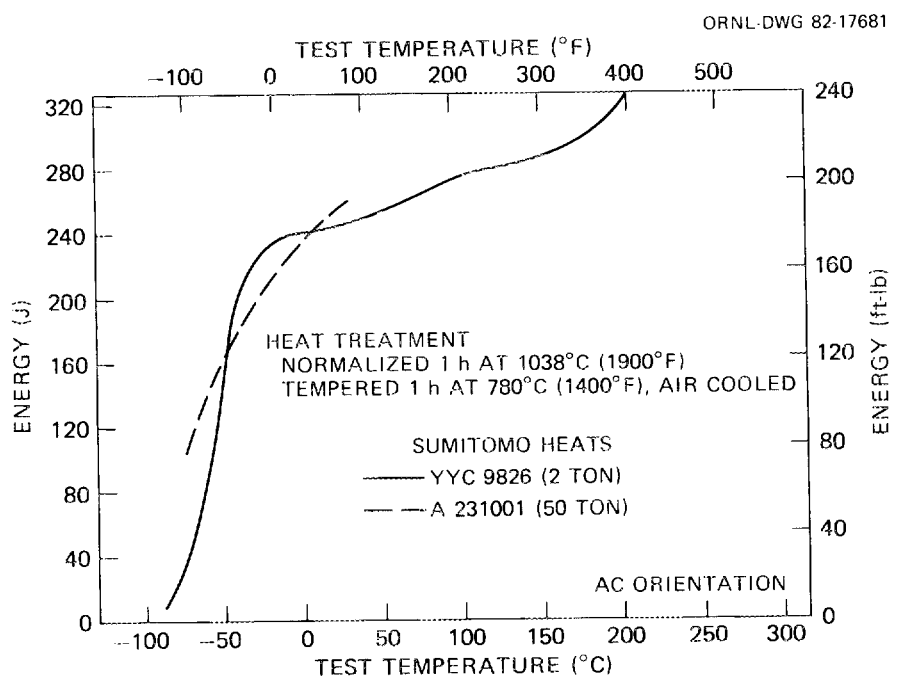


Fig. 58. Impact energy vs temperature plot for tubing made out of 2- and 50-ton heats of modified 9Cr-1Mo steel at Sumitomo Metals Ltd., Japan.

Impact properties of 222-mm-thick (8 3/4-in.) forged plates were also studied (Figs. 59 and 60). The impact properties of the forged plate were excellent and were independent of specimen location: surface, one-fourth thickness, and one-half thickness. Figures 61 through 64 compare the Charpy-impact results on the saddle forging (Fig. 4) with the forged plate. Impact energy values were identical for specimens with longitudinal, tangential, and radial orientations from the one-half thickness location of the saddle forging, and these properties were similar to those from the forged plate. The Charpy-impact properties of the extruded pipe (Fig. 3) from the same heat used to make the saddle forging are shown in Figs. 65 and 66. Note that the impact properties for axial, radial, and circumferential orientations were nearly identical. Furthermore, the impact properties for the extruded pipe were excellent.

The impact properties of sand castings[21] of modified 9Cr-1Mo steel are plotted in Figs. 67 and 68. Properties for the castings show a large

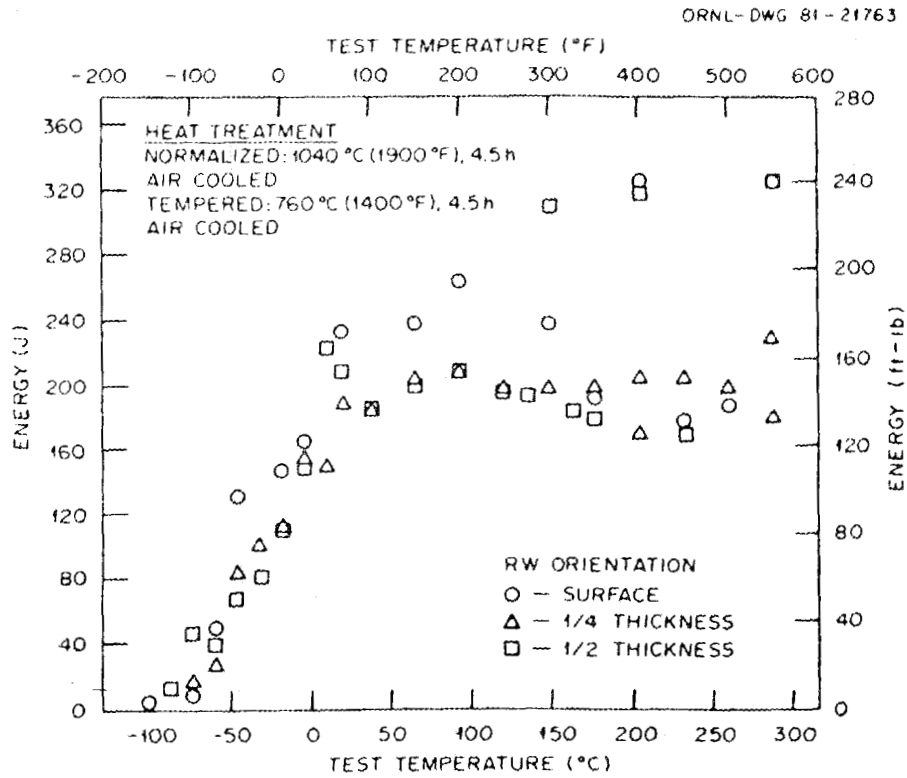


Fig. 59. Effect of location on Charpy-V impact energy of 222-mm-thick (8 3/4-in.) modified 9Cr-1Mo forged steel plate from CarTech ingot ESR 30182. RW orientation.

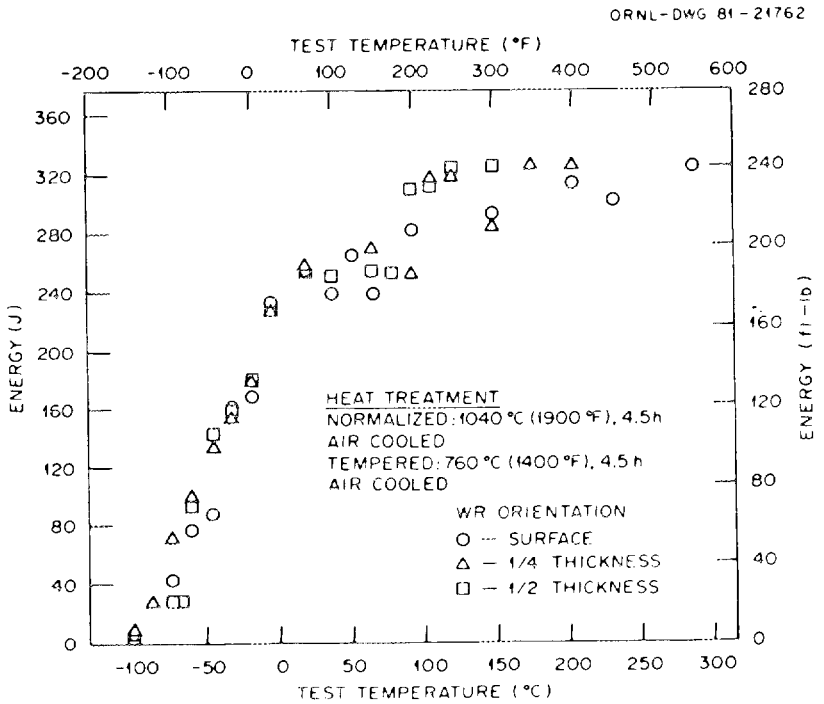


Fig. 60. Effect of location on Charpy-V impact energy of 222-mm-thick (8 3/4-in.) modified 9Cr-1Mo forged steel plate from CarTech ingot ESR 30182. WR orientation.

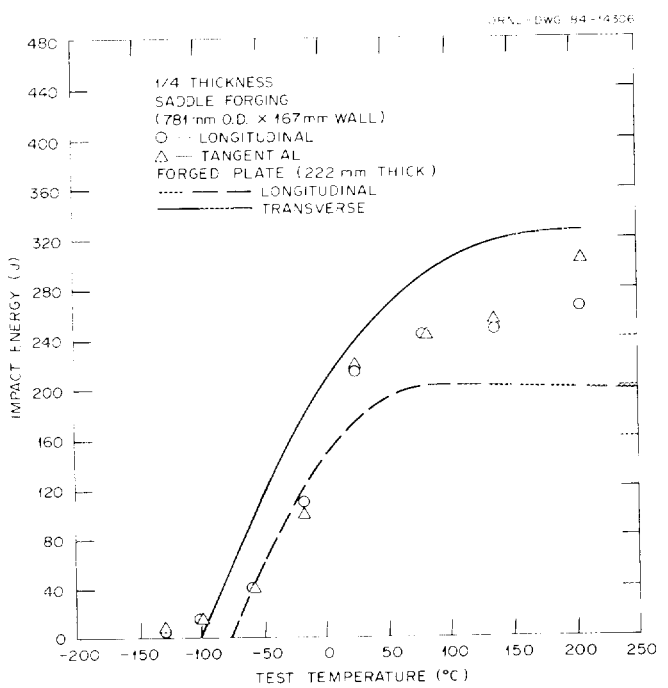


Fig. 61. Charpy impact energy vs test temperature of longitudinal and tangential specimens from 1/4-thickness section of saddle forging. Curves for longitudinal and transverse specimens from 1/4-thickness section of forged plate from another heat are included for comparison.

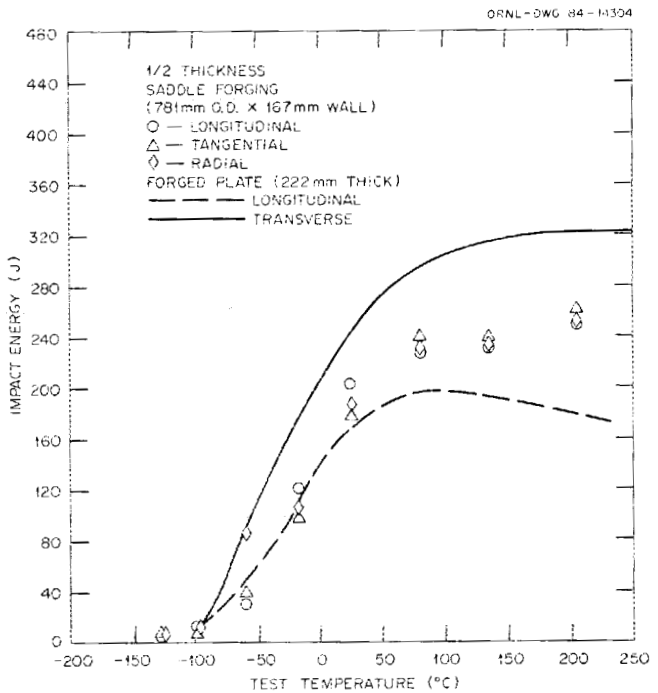


Fig. 62. Charpy impact energy vs test temperature of longitudinal, tangential, and radial specimens from 1/2-thickness section of saddle forging. Curves for longitudinal and transverse specimens from the 1/2-thickness section of forged plate from another heat are included for comparison.

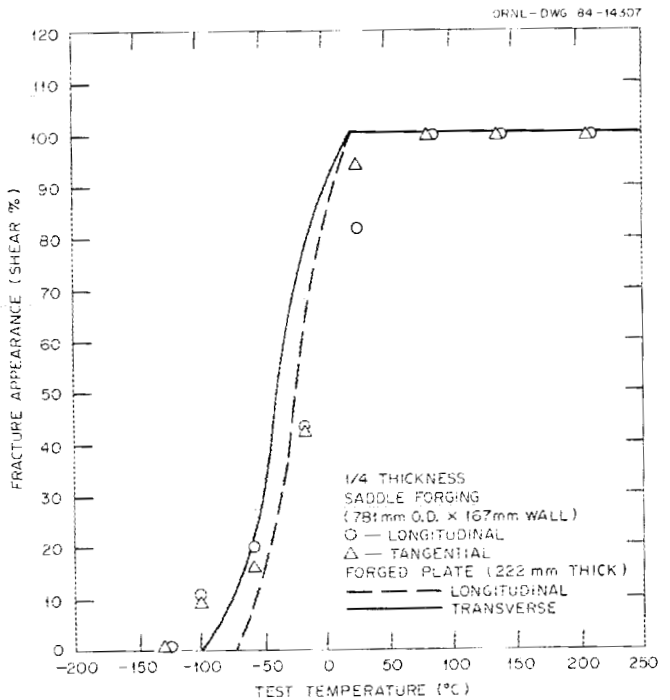


Fig. 63. Fracture appearance vs test temperature of longitudinal and tangential specimens from 1/4-thickness section of saddle forging. Curves for longitudinal and transverse specimens from 1/4-thickness section of forged plate from another heat are included for comparison.

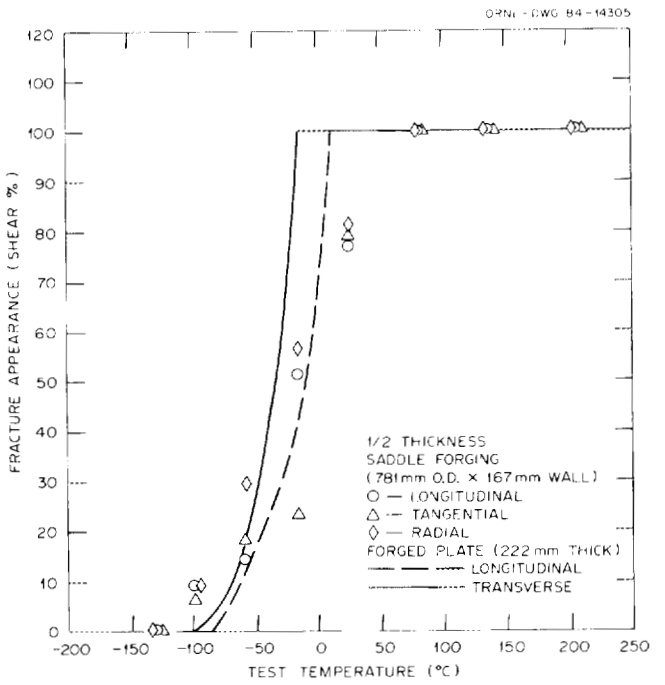


Fig. 64. Fracture appearance vs test temperature of longitudinal, tangential, and radial specimens from 1/2-thickness section of saddle forging. Curves for longitudinal and transverse specimens from the 1/2-thickness section of forged plate from another heat are included for comparison.

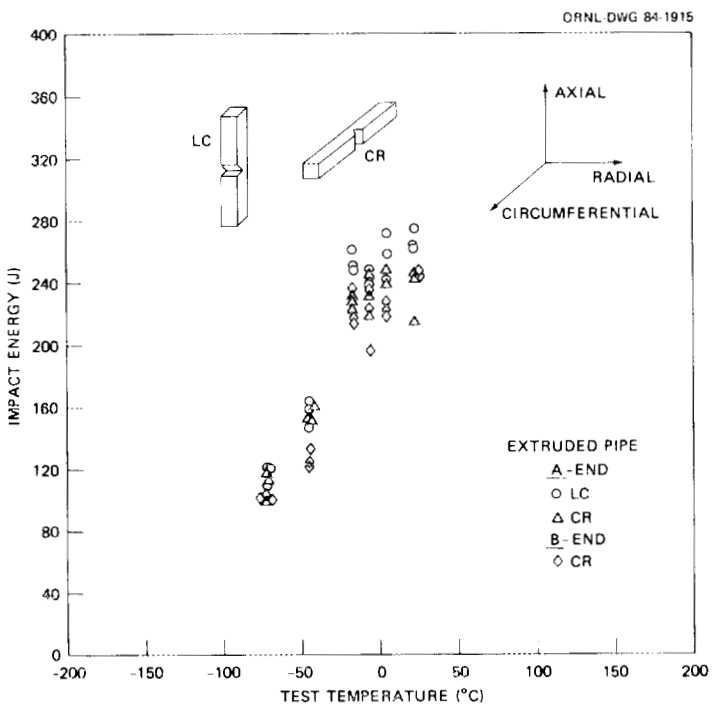


Fig. 65. Impact energy vs test temperature for longitudinal and circumferential specimens of hot-extruded pipe.

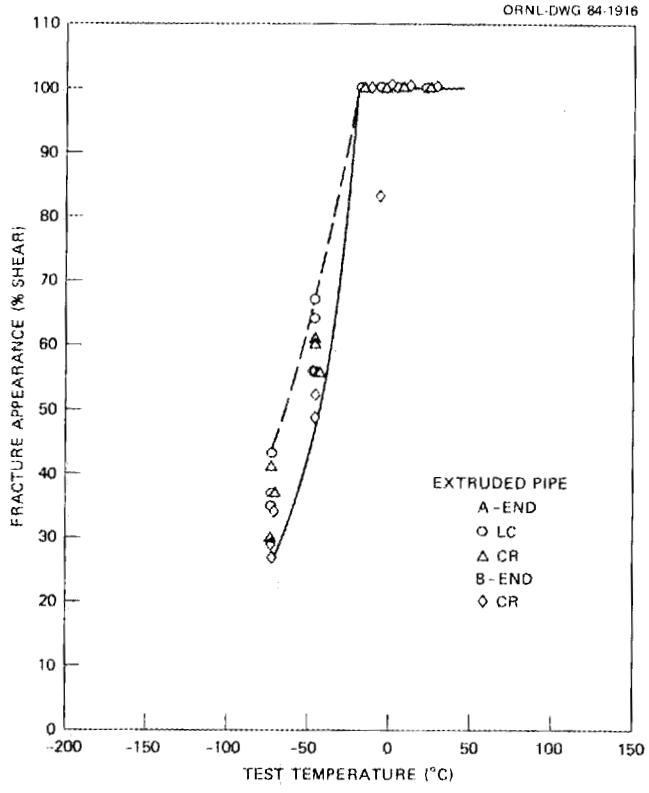


Fig. 66. Fracture appearance (% shear) vs test temperature for longitudinal and circumferential specimens of hot-extruded pipe.

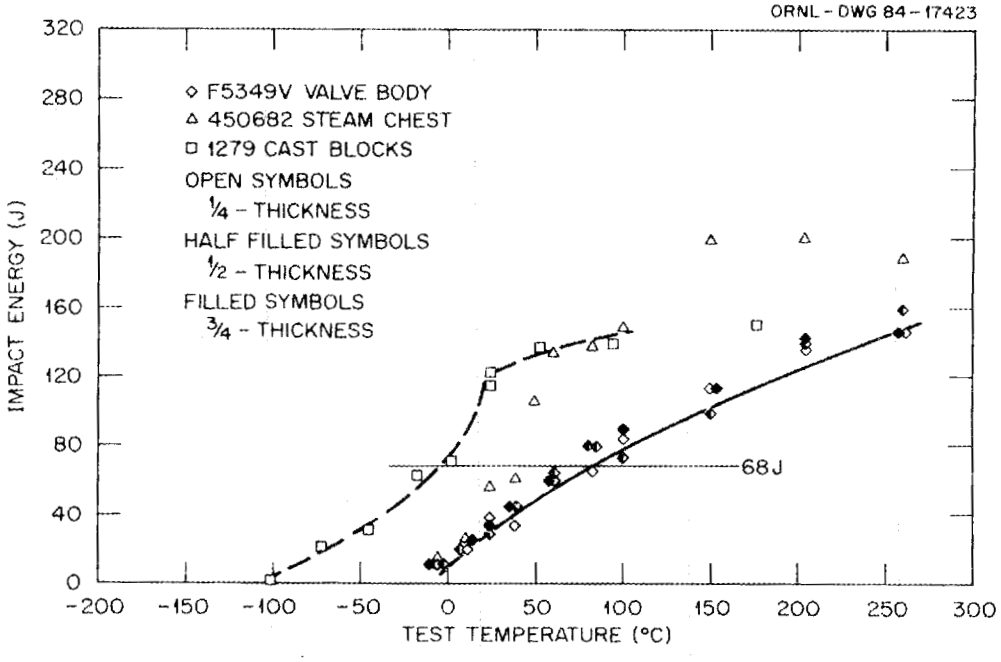


Fig. 67. Charpy impact energy as function of test temperature for various sand castings of modified 9Cr-1Mo steel.

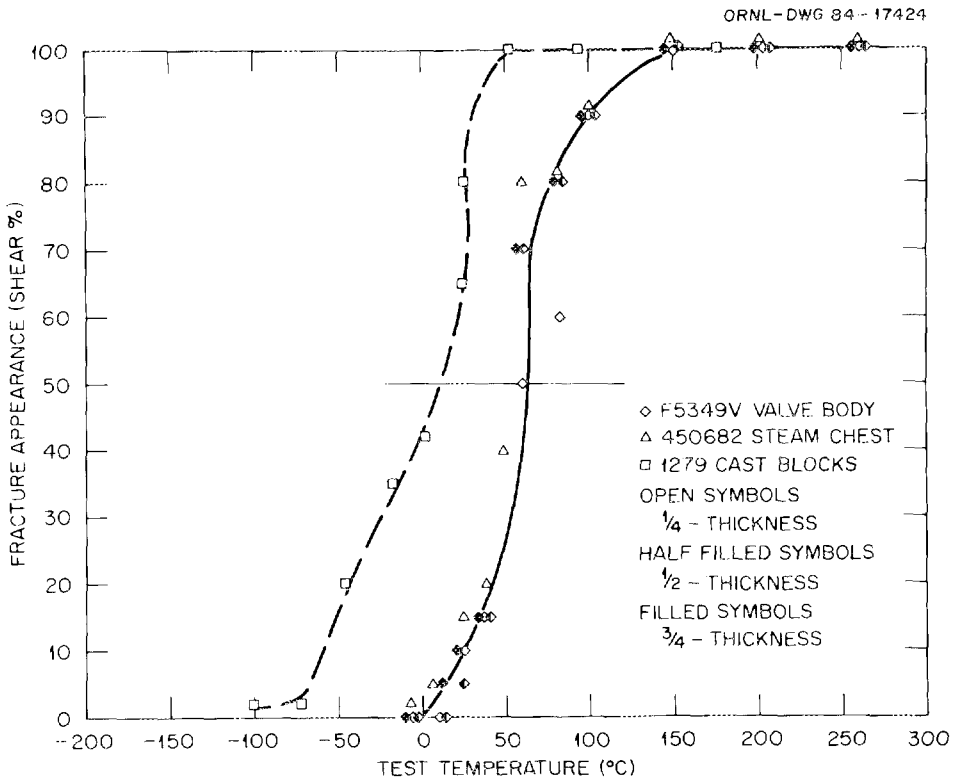


Fig. 68. Fracture appearance (% shear) from Charpy impact tests as function of test temperature for various sand castings of modified 9Cr-1Mo steel.

scatter and are poorer for castings that had low nitrogen content. Impact properties of sand castings will likely improve as foundries learn to control the composition and the processing parameters. Electroslag castings have been observed to possess excellent impact properties.

3.3.3 Thermal Aging Effects

To determine the effects of thermal aging on microstructure and mechanical properties, specimen blocks from several heats were aged at 482, 538, 593, 649, and 704°C. Charpy-impact,[62-67] tensile,[38,70] and creep tests[70] have been completed for material aged up to 25,000 h. Long-term aging is still continuing toward a planned 100,000 h. Microstructural evaluations have been conducted[71-73] on both unaged and aged material.

Charpy-impact results on aged and unaged material are compared in Figs. 69 through 78. These figures show that impact energy curves are

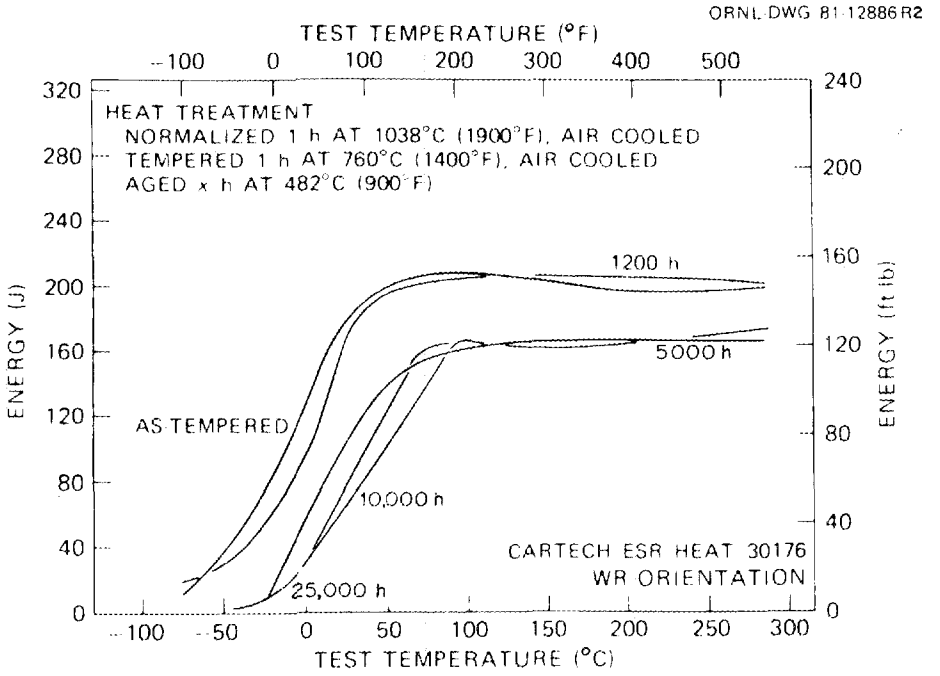


Fig. 69. Effect of aging at 482°C on Charpy-V energy of 27-mm-thick (1 1/16-in.) commercially processed modified 9Cr-1Mo steel plate, CarTech ESR 30176.

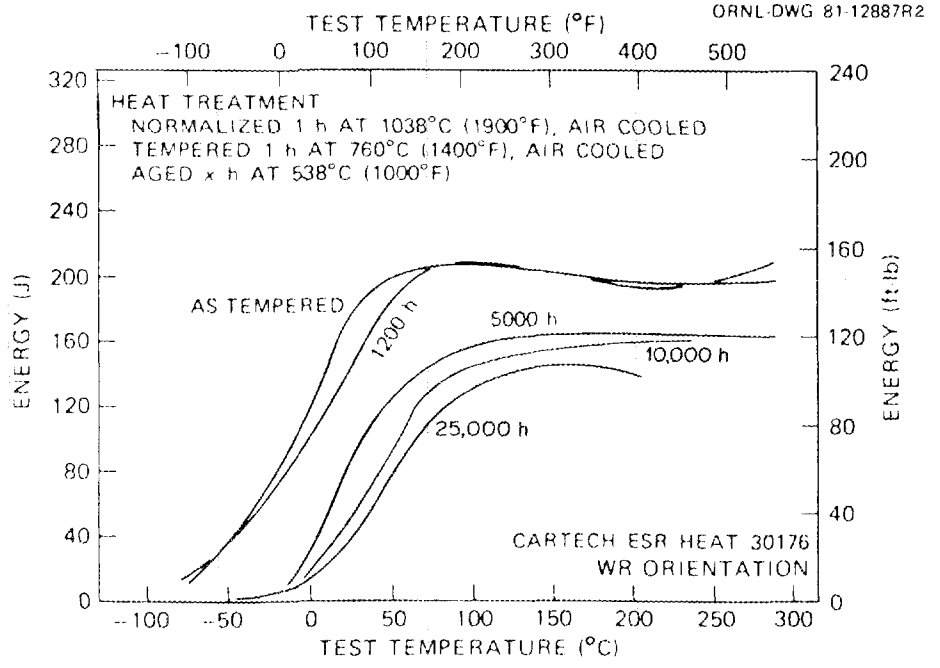


Fig. 70. Effect of aging at 538°C on Charpy-V energy of 27-mm-thick (1 1/16-in.) commercially processed modified 9Cr-1Mo steel plate, CarTech ESR 30176.

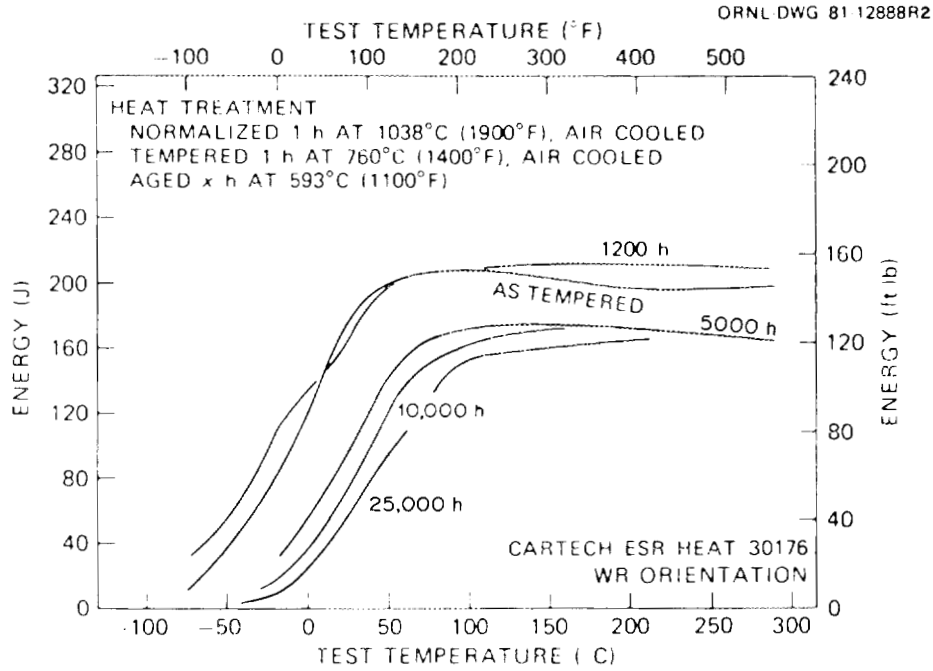


Fig. 71. Effect of aging at 593°C on Charpy-V energy of 27-mm-thick (1 1/16-in.) commercially processed modified 9Cr-1Mo steel plate, CarTech ESR 30176.

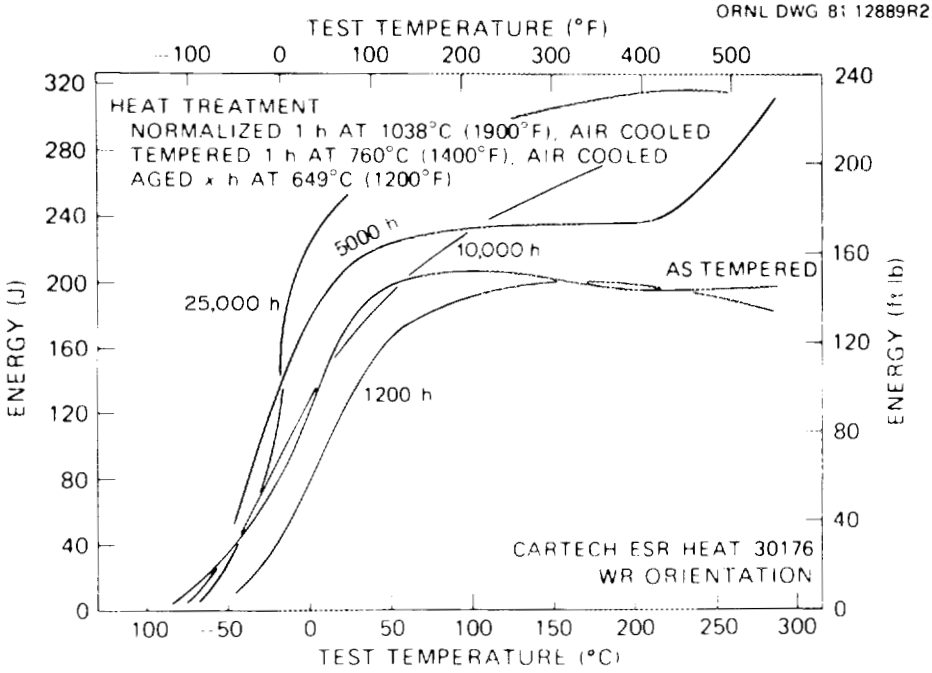


Fig. 72. Effect of aging at 649°C on Charpy-V energy of 27-mm-thick (1 1/16-in.) commercially processed modified 9Cr-1Mo steel plate, CarTech ESR 30176.

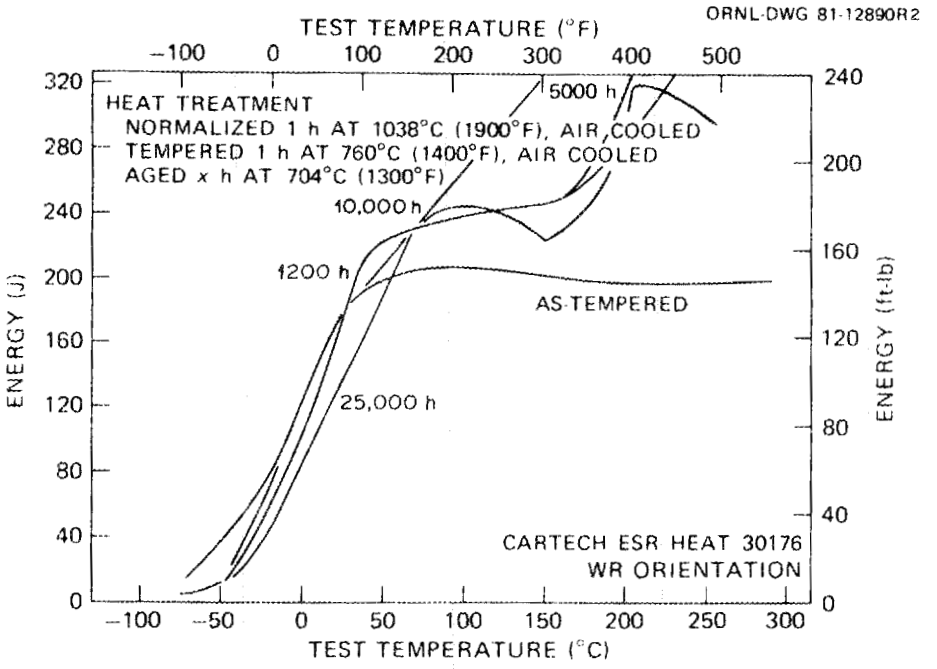


Fig. 73. Effect of aging at 704°C on Charpy-V energy of 27-mm-thick (1 1/16-in.) commercially processed modified 9Cr-1Mo steel plate, CarTech ESR 30176.

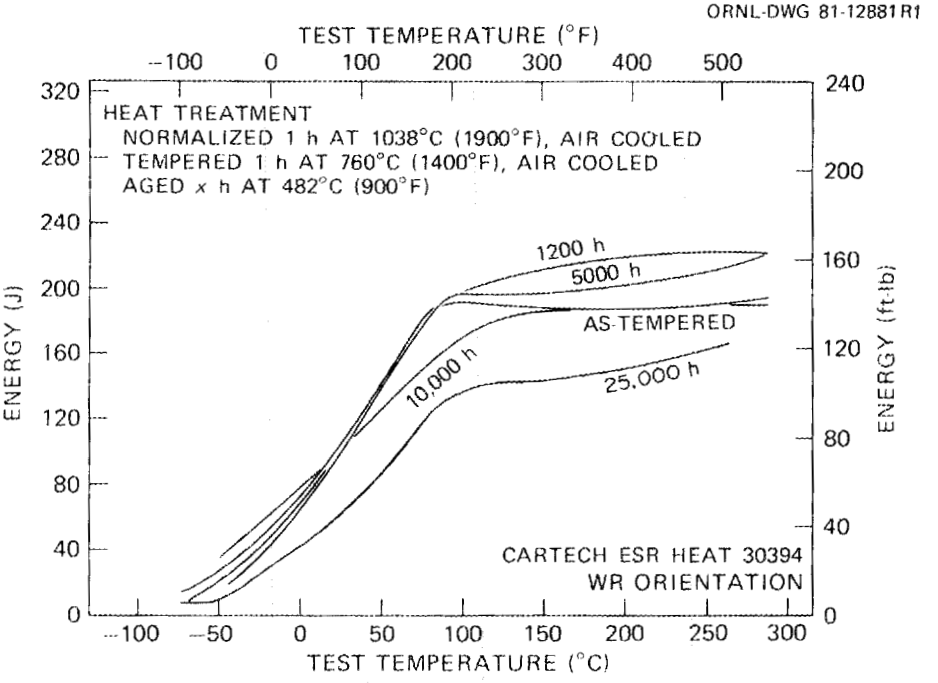


Fig. 74. Effect of aging at 482°C on Charpy-V energy of 27-mm-thick (1 1/16-in.) commercially processed modified 9Cr-1Mo steel plate, CarTech ESR 30394.

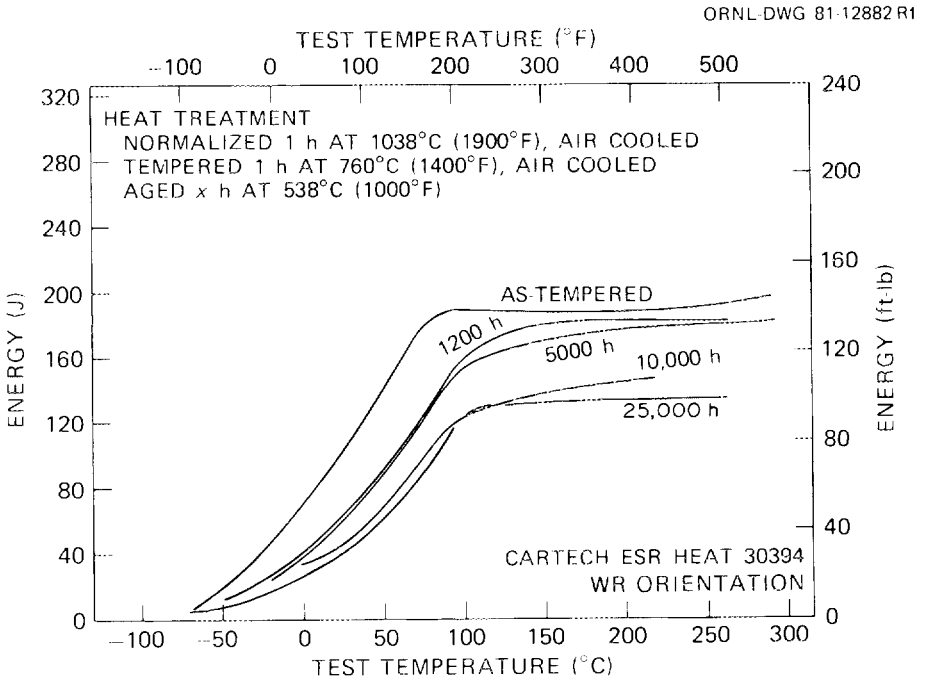


Fig. 75. Effect of aging at 538°C on Charpy-V energy of 27-mm-thick (1 1/16-in.) commercially processed modified 9Cr-1Mo steel plate, CarTech ESR 30394.

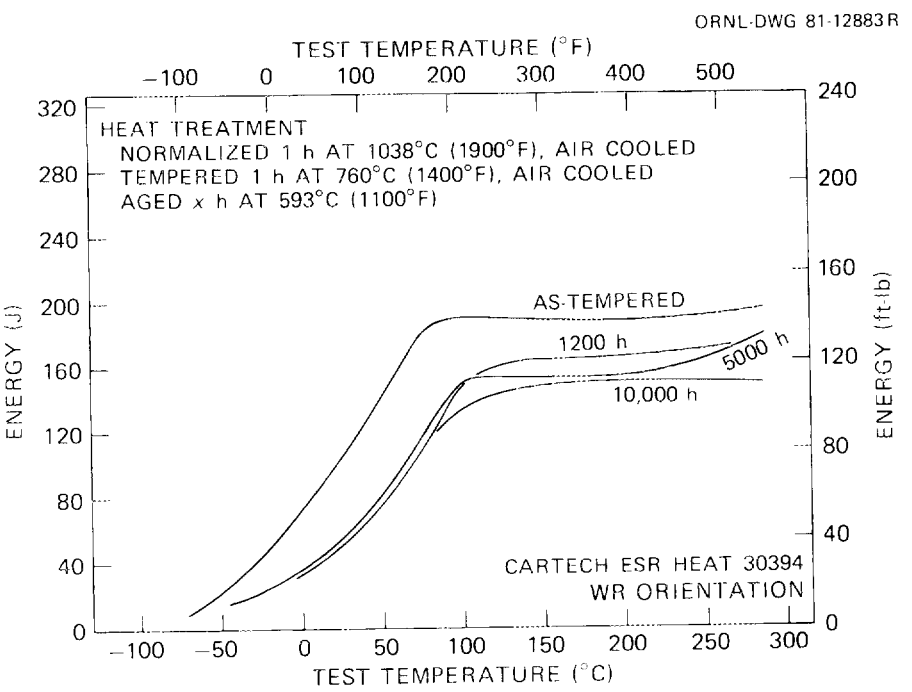


Fig. 76. Effect of aging at 593°C on Charpy-V energy of 27-mm-thick (1 1/16-in.) commercially processed modified 9Cr-1Mo steel plate, CarTech ESR 30394.

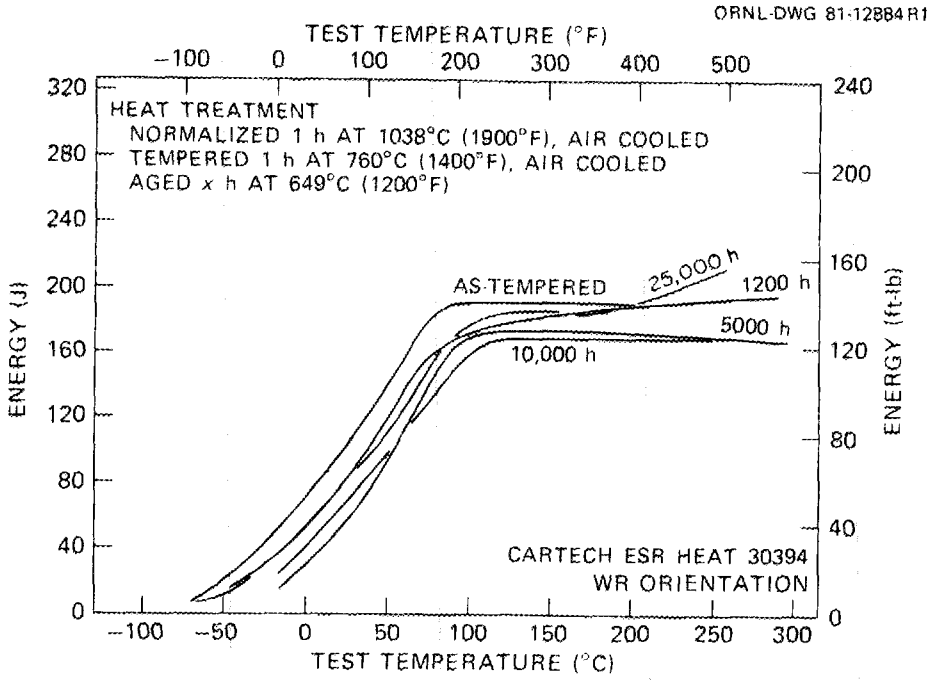


Fig. 77. Effect of aging at 649°C on Charpy-V energy of 27-mm-thick (1 1/16-in.) commercially processed modified 9Cr-1Mo steel plate, CarTech ESR 30394.

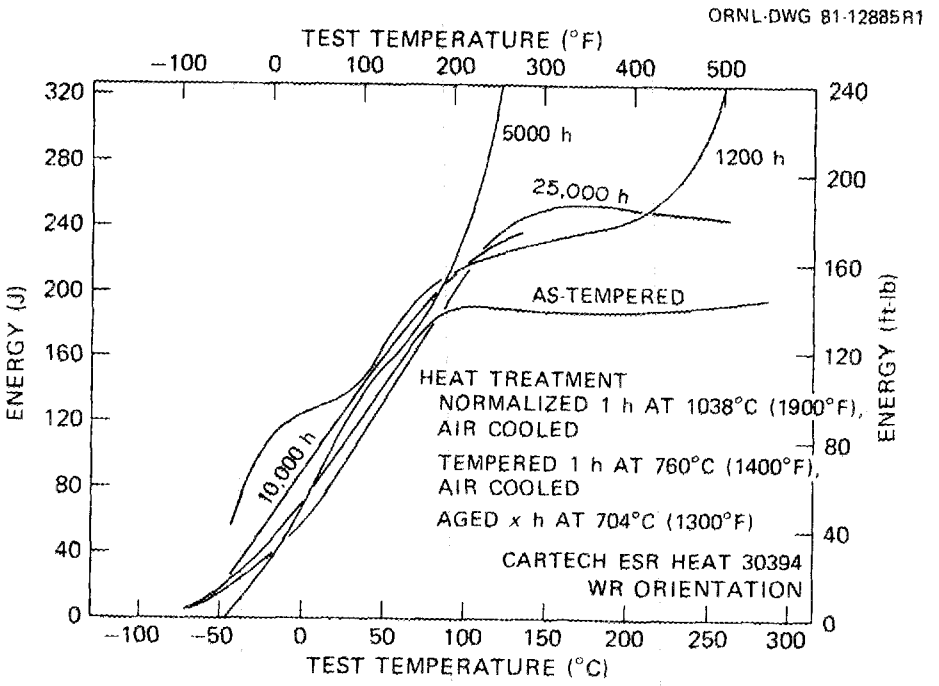


Fig. 78. Effect of aging at 704°C on Charpy-V energy of 27-mm-thick (1 1/16-in.) commercially processed modified 9Cr-1Mo steel plate, CarTech ESR 30394.

affected by thermal aging. In general, the DBTT increases and USE decreases after aging. These changes are more pronounced for aging temperatures of 538 and 593°C. At the higher temperatures, aging appears to have a beneficial effect on impact properties.

Tensile properties after various aging treatments are plotted in Figs. 79 through 84. Up to 600°C aging has minimal effect on room temperature strength and ductility. However, at higher aging temperatures, the room temperature strength can fall below the minimum values. Figs. 82 through 84 show that degradation in elevated-temperature properties also is not of concern for aging temperatures up to 600°C.

Creep data on the aged material are presented in Table 13. Time-to-rupture data on specimens aged for 25,000 h are compared with the average curve for the unaged material in Fig. 85. The data show that aging at 538, 593, and 649°C for times to 25,000 h has a minimum effect on the creep properties of modified 9Cr-1Mo steel. Some long-term creep tests on aged material are still continuing.

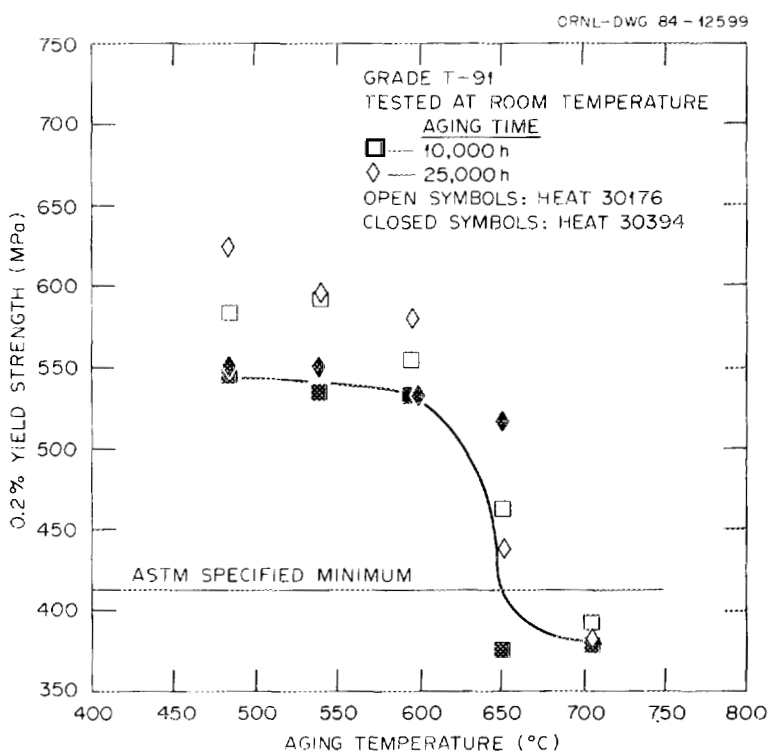


Fig. 79. Thermal aging for 25,000 h has minimal effect on room-temperature yield strength of modified 9Cr-1Mo steel for aging temperatures up to 600°C.

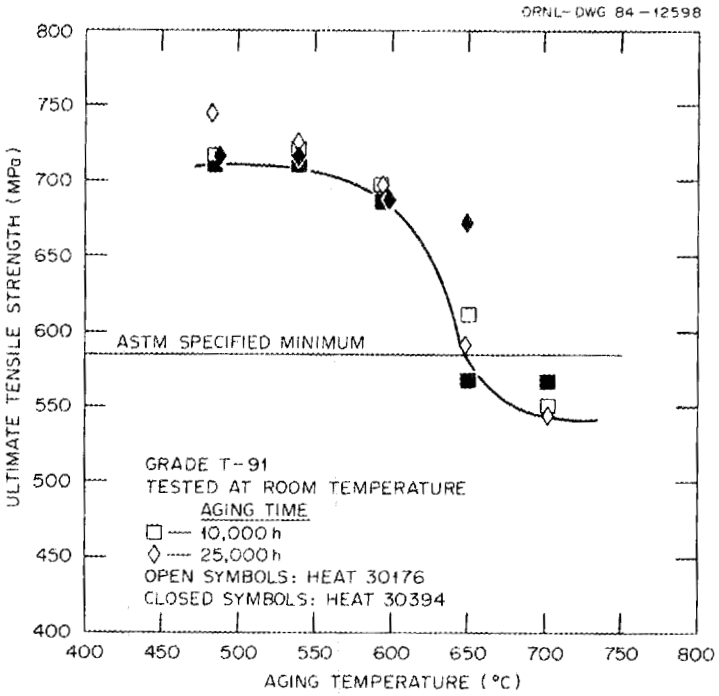


Fig. 80. Thermal aging for 25,000 h has minimal effect on room-temperature ultimate tensile strength of modified 9Cr-1Mo steel for aging temperatures up to 600°C.

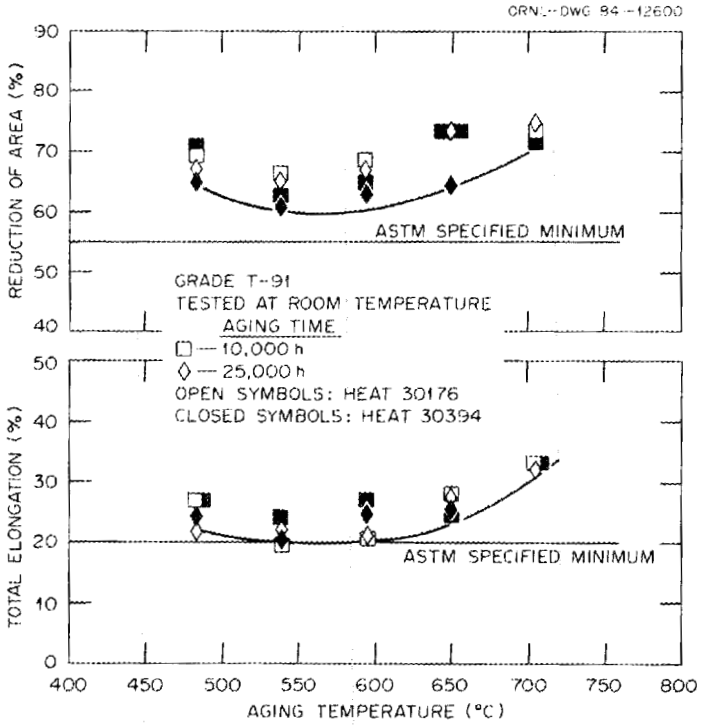


Fig. 81. Thermal aging for 25,000 h has minimal effect on room-temperature total elongation and reduction of area of modified 9Cr-1Mo steel for aging temperatures up to 600°C.

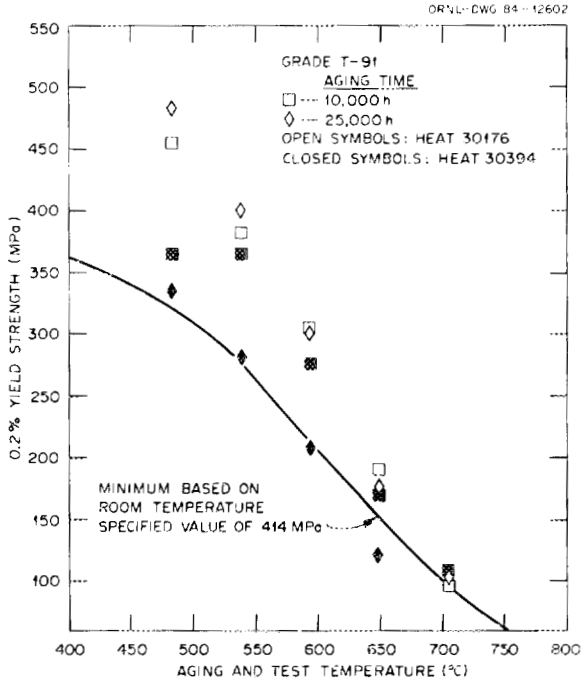


Fig. 82. Testing at aging temperatures after 25,000 h aging showed different response for two heats. One heat showed essentially no change, whereas aging brought the yield strength to the minimum curve for the second heat.

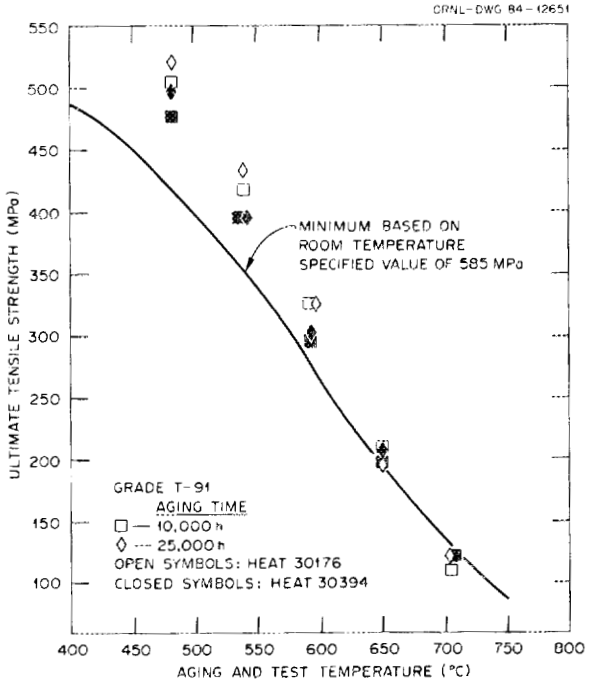


Fig. 83. Testing at aging temperatures after 25,000 h aging showed minimal change in ultimate tensile strength.

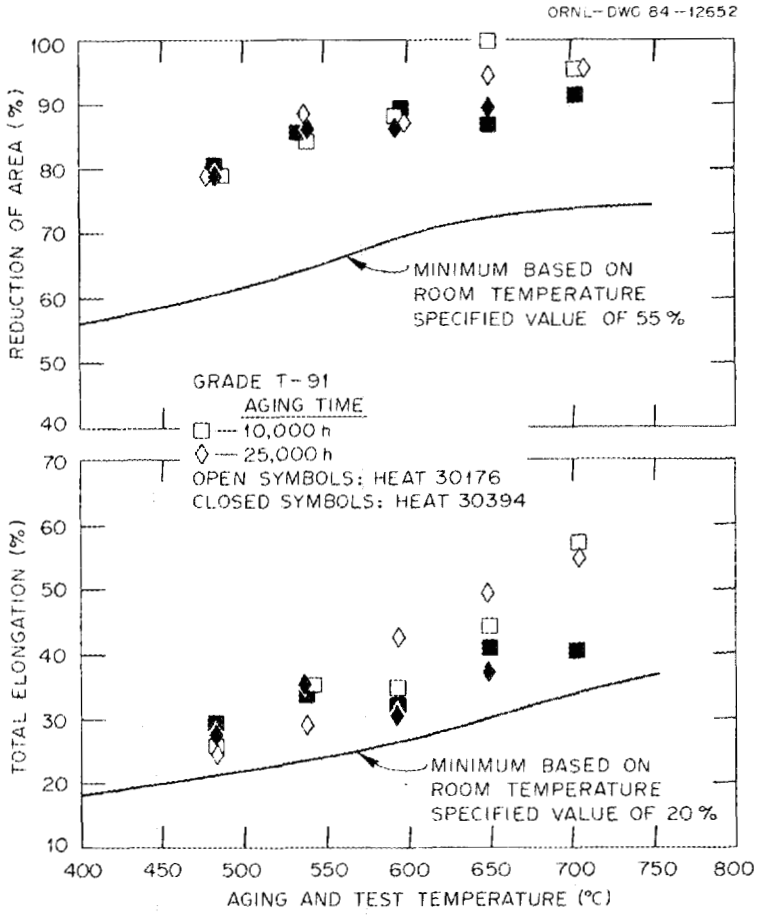


Fig. 84. Testing at aging temperatures after 25,000 h aging showed minimal change in total elongation and reduction of area.

Transmission electron micrographs from thin-foil specimens of heat 30176 of modified 9Cr-1Mo steel are shown in the as-tempered condition and after aging for 25,000 h at 538 to 704°C in Fig. 86. Up to 593°C, the sub-boundary structure within the prior austenitic grains remains fairly constant relative to the as-tempered starting material after 25,000 h. However, at 649°C some of the cell structure coarsened slightly as the elongated former martensite laths become polygonal. At 704°C the subgrains were noticeably larger. At higher magnifications and after aging at 538 and 593°C, the intracell dislocation density was higher in the as-tempered structure, particularly at 538°C. However, at both 650 and 704°C, recovery produced fairly dislocation-free subgrains. Boundary

Table 13. Creep data on specimens aged for 5000 to 25,000 h at temperatures in the range of 482 to 649°C for heats 30176 and 30394

(Creep tests were conducted at the aging temperature and a common temperature of 593°C; all specimens were normalized and tempered before thermal aging)

Test	Aging		Test temperature (°C)	Stress (MPa)	Rupture time (h)	Minimum creep rate (%/h)	Strain (%)		Reduction of area (%)
	Temperature (°C)	Time (h)					Loading	Fracture	
				Heat 30176					
24357	538	25,000	538	207	14,732.7[a]		0.20		
24359	538	25,000	538	186	14,635.7[a]		0.14		
24058	593	25,000	593	172	738	5.0 E-3	0.16	20.89	81.22
24061	593	25,000	593	145	6,421.7	5.8 E-4	0.18	20.11	77.27
24079	593	25,000	593	110	20,062.4[a]		0.10		
24080	649	25,000	649	110	100.4	6.0 E-2	0.18	41.58	90.24
24084	649	25,000	649	76	4,175.2	8.8 E-4	0.11	26.60	76.94
				Heat 30394					
24351	538	25,000	538	207	14,897.3[a]				
24069	593	25,000	593	172	408	1.1 E-2	0.23	24.92	81.35
24072	593	25,000	593	145	4089	9.0 E-4	0.11	22.47	79.54
24073	593	25,000	593	110	20,375.7[a]		0.51		
24298	649	25,000	649	110	341.6	1.25 E-2	0.25	29.80	85.21
24299	649	25,000	649	76	6,619.5	5.8 E-4	0.13	21.48	55.77
22676	482	5,000	593	193	675.9	5.9 E-3	0.0	25.58	83.70
22670	538	5,000	593	193	335.2	1.0 E-2	0.0	20.74	81.49
22663	593	5,000	593	193	97.3	5.8 E-2	0.0	28.33	84.79
22660	649	5,000	593	193	62.9	1.01 E-1	0.18	32.75	85.31
22793	482	5,000	593	159	3,404.6	1.1 E-3	0.19	29.21	82.13
22671	538	5,000	593	159	3,072.6	9.8 E-4	0.15	24.82	82.60
22664	593	5,000	593	159	995.2	3.9 E-3	0.20	21.26	80.31
22674	649	5,000	593	159	604.6	9.6 E-3	0.20	27.33	83.74
22659	649	5,000	593	159	609.0	1.0 E-2	0.17	29.68	84.94

[a]Tests are in progress.

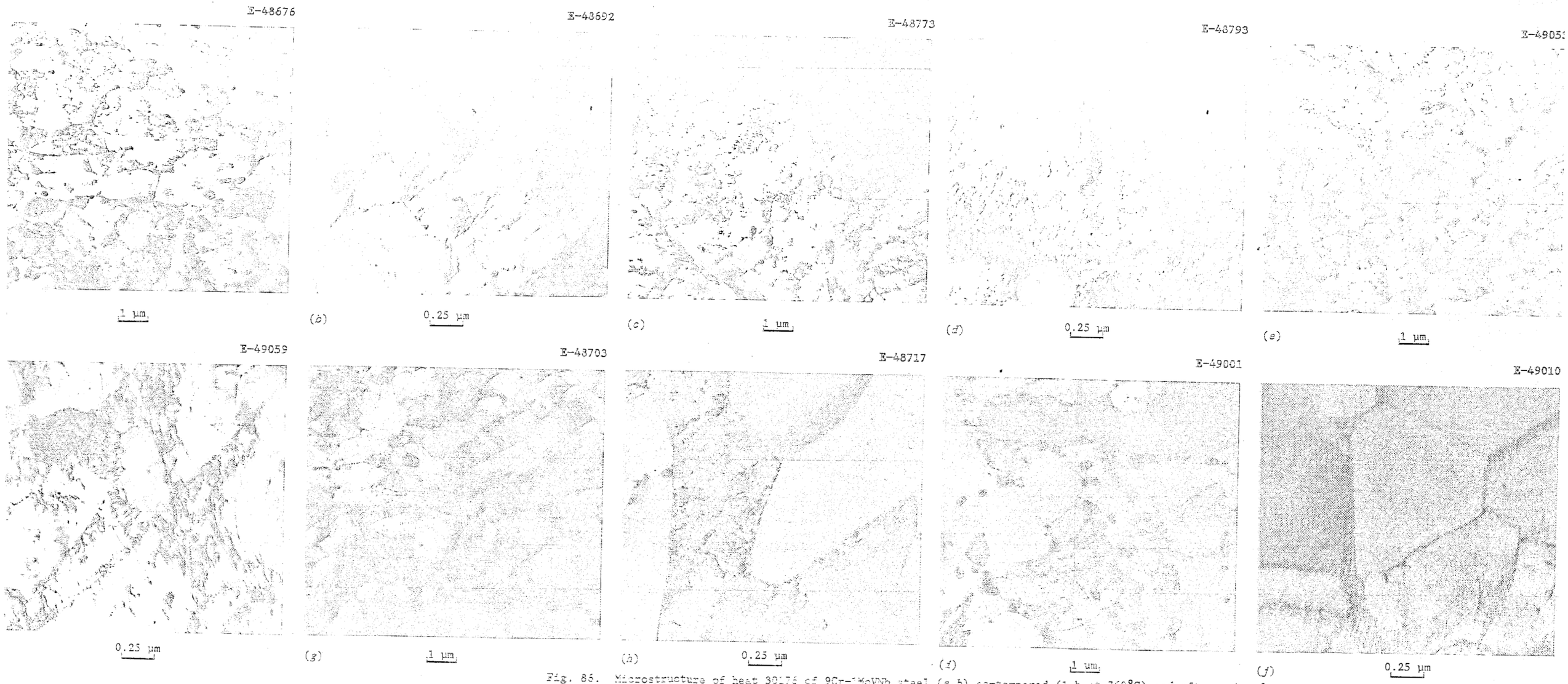


Fig. 86. Microstructure of heat 30176 of 9Cr-1MoVNb steel (*a*,*b*) as-tempered (1 h at 760°C) and after aging for 25,000 h at (*c*,*d*) 538°C, (*e*,*f*) 593°C, (*g*,*h*) 650°C, and (*i*,*j*) 704°C. Low magnification shows overall subboundary structure and higher magnification shows intra-cell dislocation structure; both pictures are included for each sample, as indicated.

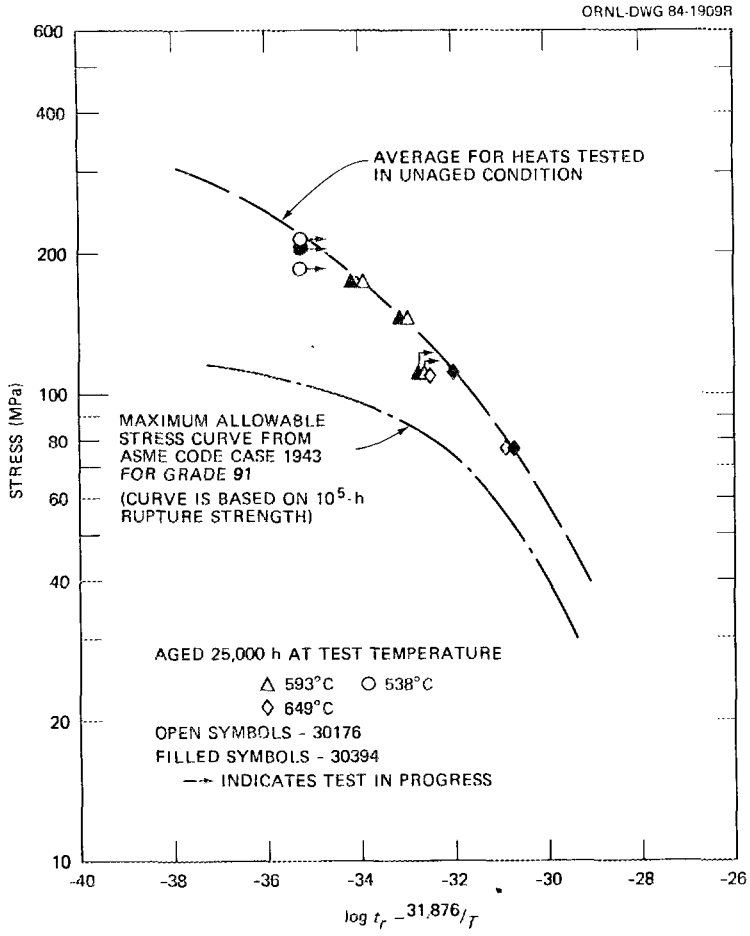


Fig. 85. Effect of aging for 25,000 h on creep-rupture life of modified 9Cr-1Mo steel.

and dislocation structures showed little effect of aging times of 10,000 to 25,000 h at 538 and 649°C in heat 30176 and little or no heat-to-heat variation. The following are the major conclusions from the microstructural evaluation of thermally aged material.

1. The grain and subboundary structure of modified 9Cr-1Mo steel remains stable during long-term aging to 593°C but begins to recover and coarsen at 650°C and above. Dislocation density within cells increased with aging times at 538 and 593°C but then recovered at 650°C and above, relative to the as-tempered dislocation structure.
2. The as-tempered carbide structure ($M_{23}C_6$ + MC) remained stable during aging at 650°C and below but coarsened at 704°C.

3. New precipitates formed during aging at temperatures below 650°C and were found to be superimposed upon the as-tempered precipitate structure. Laves phase formed at 482 to 593°C and was a maximum at ~538°C after 25,000 h. Much more Laves phase formed in the higher silicon heat (30394); however, it only formed after 25,000 h at 650°C.
4. A high concentration of very fine, intracellular precipitates occurred along the dislocation network at 482 to 593°C. Most of these were needles of vanadium carbide, but a few particles of an unidentified Fe-Cr phase were also found in heat 30176 at 538°C after 25,000 h. A small amount of slightly coarser chromium-rich rods, believed to be an M_2X type phase, were also found in heat 30176 after 25,000 h at 482°C.
5. The various phases were compositionally quite distinct. They generally contained Cr, V, or Nb as their major constituent, except for the Fe-Cr and Laves phases. Only Laves phase contained substantial enrichments of P, Si, and Mo; Laves phase contained more silicon when formed in the higher silicon heat of steel (30394). Only $M_{23}C_6$ and VC showed composition sensitivity to temperature; at 704°C, $M_{23}C_6$ became richer in niobium at the expense of chromium. At 482 to 538°C, the VC phase composition was size sensitive; tiny VC needles had more chromium and less vanadium than did larger particles.

3.3.4 Radiation Effects

Considerable effort has been devoted to determining the effects of radiation on high chromium (~9-12%) ferritic steels since it was first reported in 1973 that a ferritic steel exhibited significantly less swelling[74] than austenitic stainless steels (e.g., types 304 and 316). Data from subsequent irradiations revealed that at the same temperature and irradiation conditions (neutron spectra and fluence), high chromium ferritic steels swell approximately an order of magnitude less than austenitic stainless steels.[75,76] Information obtained to date indicates that as a group, ferritic/martensitic alloys are the most swelling resistant of all of the alloy systems irradiated, including the refractory metals.[77] A comparison of the swelling behavior of ferritic and austenitic stainless steels is shown in Fig. 87 (ref. 76).

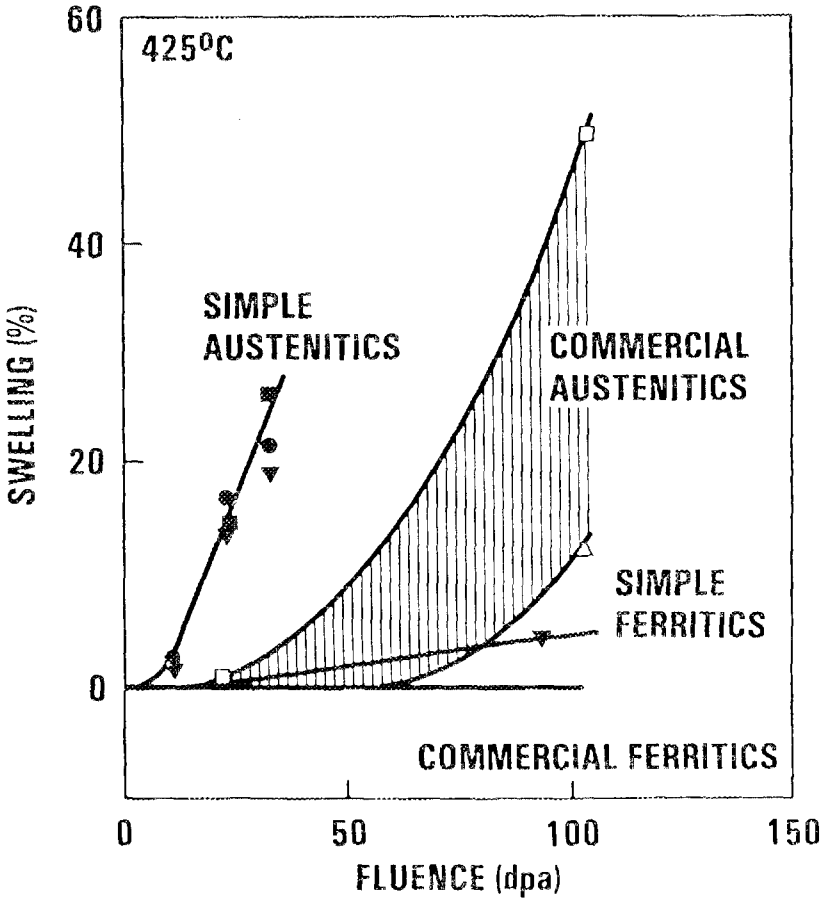


Fig. 87. Comparison of swelling behavior of commercial and simple ferritic steels to that of commercial and simple austenitic steels. *Source*: T. A. Lechtenberg, "Irradiation Effects in Ferritic Steels," pp. 109–16 in *Alloy Development for Irradiation Performance Semiannual Progress Report for Period Ending March 31, 1985*, DOE/ER-0045/14, U.S. Department of Energy, Office of Fusion Energy, July 1985.

Studies of the effects of radiation on modified 9Cr-1Mo steel (ASTM designation T-91) and a similar 12Cr-1Mo ferritic alloy (HT-9) have been conducted under the LMR Fuels and Materials Program and the Fusion Energy (FE) First Wall and Blanket Materials Program. Emphasis in the LMR program has been on the effects of elevated-temperature (~400–600°C) irradiation in the Experimental Breeder Reactor (EBR-II and the Fast Flux Test Facility (FFTF) on the swelling, microstructural, tensile, and fracture toughness properties at the high displacements per atom (dpa) required for fuel cladding and duct applications. Emphasis in the FE program has been on the effects of neutronically produced helium combined with high dpa on swelling, microstructure, tensile properties, and DBTT.

The limited data for T-91 will be compared with those for HT-9 because studies of the effects of radiation on the swelling and mechanical properties of HT-9 have been a major effort for more than a decade in the LMR program and for more than 6 years in the FE program.

HT-9 nominally contains 12% Cr and higher C, Ni, and Si than T-91. Higher C, Ni, and Si result in a somewhat different radiation response for HT-9 than that observed for T-91. The range of chemical compositions for T-91 and HT-9 for which there is information on the effects of radiation is given in Table 14.

Table 14. Ranges of chemical compositions of T-91 and HT-9 irradiated in EBR-II

Element	T-91 (%)	HT-9 (%)	Element	T-91 (%)	HT-9 (%)
Cr	8.47–9.00	11.99–12	Co	0.017	0.017
C	0.086–0.090	0.21	Cu	0.03	0.05
Mo	0.88–0.98	0.93–1.01	Al	0.004–0.014	0.030
Mn	0.36–0.38	0.45–0.55	B	0.0006–<0.001	<0.001
Ni	0.09–0.11	0.43–0.52	Sn	0.003–0.003	0.002
Si	0.08–0.19	0.18–0.41	N	0.050–0.054	0.020
Ti	<0.001–0.001	0.003	O	0.007–0.008	0.005
V	0.19–0.24	0.27–0.38	As	<0.001	
Nb	0.06–0.18	0.018	Zr	<0.001	
W	<0.001–0.01	0.51–0.54	Pb	<0.001	
P	0.011–0.012	0.011	Sb	<0.001	
S	0.003–0.004	0.004			

3.3.4.1 Microstructure and Swelling

T-91 and HT-9 were irradiated together in EBR-II to 2.5×10^{22} neutrons/cm^2 (E > 0.1 MeV) at 400, 450, 500, and 550°C.[78] This irradiation produced ~12 dpa in these alloys.

Irradiations of HT-9 at 400°C resulted in the development of a high density of homogeneously distributed, equiaxed precipitates (28-nm diam) of G phase, a nickel-rich silicide. No voids were detected. No voids or swelling have been detected in HT-9 after irradiation to more than 60 dpa at 400 to 550°C in EBR-II.[79] Suppression of void formation and swelling in HT-9 is attributed to the formation of G phase during irradiation.[80]

For T-91, irradiation at 400°C resulted in a high density of dislocation loops and tangles, small rod-shaped precipitates of Cr_2C, and a low density of faceted voids. Cr_2C rather than G phase forms in T-91 because of its lower nickel content compared with HT-9. The low density of voids resulted in negligible swelling for the fluence attained, but their presence indicates that swelling will likely increase with increasing fluence.[79] The presence of voids in T-91 is considered significant because voids are rarely seen in ferritic steels at fluences as low as 2.5×10^{22} neutrons/cm^2. However, because of the inherently low swelling rates of ferritic steels, swelling will still be significantly less than that for an austenitic steel irradiated under the same conditions.

At 450°C irradiation of HT-9 resulted in the production of dislocations and a fine precipitate structure between the very widely spaced martensite lath boundaries. The densities of dislocations and precipitate were much lower than after irradiation at 400°C. No voids were present in T-91 after irradiation at 450°C and above; however, there were more precipitation of MC and recovery of dislocations.

At 500 and 550°C irradiation of HT-9 resulted in the coarsening of $M_{23}C_6$ at martensite lath and prior austenitic grain boundaries and additional precipitation within the subgrains. Irradiation of T-91 at 500 and 550°C resulted in recovery of the preirradiation dislocation structure and precipitation of MC and a phosphide phase. The MC had varying concentrations of V, Cr, and Nb. The phosphide-rich phase was rich in Fe, Cr, and Mo; had a lath morphology; and was heterogeneously distributed. In addition, MnS and Fe_2Mo (Laves phases) were present, and some of the phosphide precipitates were in bands. Because of these bands, the investigators suggest that the material may not have been well homogenized before irradiation, a fact that, if so, could explain the presence of voids at a fluence of 2.5×10^{22} neutrons/cm^2 ($E > 0.1$ MeV).

High Flux Isotope Reactor (HFIR) irradiation to 36 dpa at 300, 500, and 600°C resulted in a negligible amount of swelling in N&T T-91 (ref. 81). Only a few cavities, ~4 to 6 nm in diameter, were observed. The cavities were preferentially located at dislocations and lath boundaries of the tempered martensite. There was no change in the precipitate morphology and distribution after irradiations at 300 and 600°C. After irradiation

at 500°C, a few large, elongated precipitates that have not been observed in unirradiated material or material irradiated at 300, 400, or 600°C were present.

Irradiation of T-91 at the intermediate temperature of 400°C resulted in a well-developed cavity structure that was homogeneously distributed throughout the microstructure. The cavities were 4 to 30 nm in diameter and had an average size of 15 nm, which is three times larger than the cavities formed by irradiation at 300, 500, and 600°C. These results are consistent with those for T-91 irradiated in EBR-II where voids were observed only after irradiation at 400°C.

Although no swelling or cavities had been observed after irradiation of HT-9 in EBR-II, swelling and cavities were observed after irradiation in HFIR. During irradiation in HFIR, the helium produced by the interaction of thermal neutrons with nickel in the alloys enhances cavity formation. The temperature dependence of the cavity microstructures and swelling for HT-9 and T-91 were similar, but although the T-91 used in the HFIR irradiations had only one-third the nickel content of the HT-9, it had a swelling rate three times as large as that observed for HT-9. As mentioned earlier, the formation of G phase in HT-9 results in very low swelling rates.

3.3.4.2 Toughness Properties

Several irradiations were conducted in EBR-II to characterize the fracture toughness properties of T-91 and HT-9 (refs. 82-87). Figure 88 shows the effect of irradiating T-91 in EBR-II to 3.8 to 4.7 × 10^{22} neutrons/cm^2 (E > 0.1 MeV) (~20-25 dpa) over the temperature range 384 to 425°C.[82] Irradiation raised the DBTT ~80°C (from about −68 to 10°C), as measured by the 41-J transition temperature, which is the temperature at which 41 J are required to fracture the sample in a Charpy-impact test. The use of the 41-J transition temperature as an index for DBTT is admittedly arbitrary; however, it has been used rather extensively in reporting the results of Charpy-impact testing in the LMR and FE irradiation programs. Therefore, it is used in this report as well. Also shown in Fig. 88 is the small reduction (15%) in the USE from ~200 to 170 J.

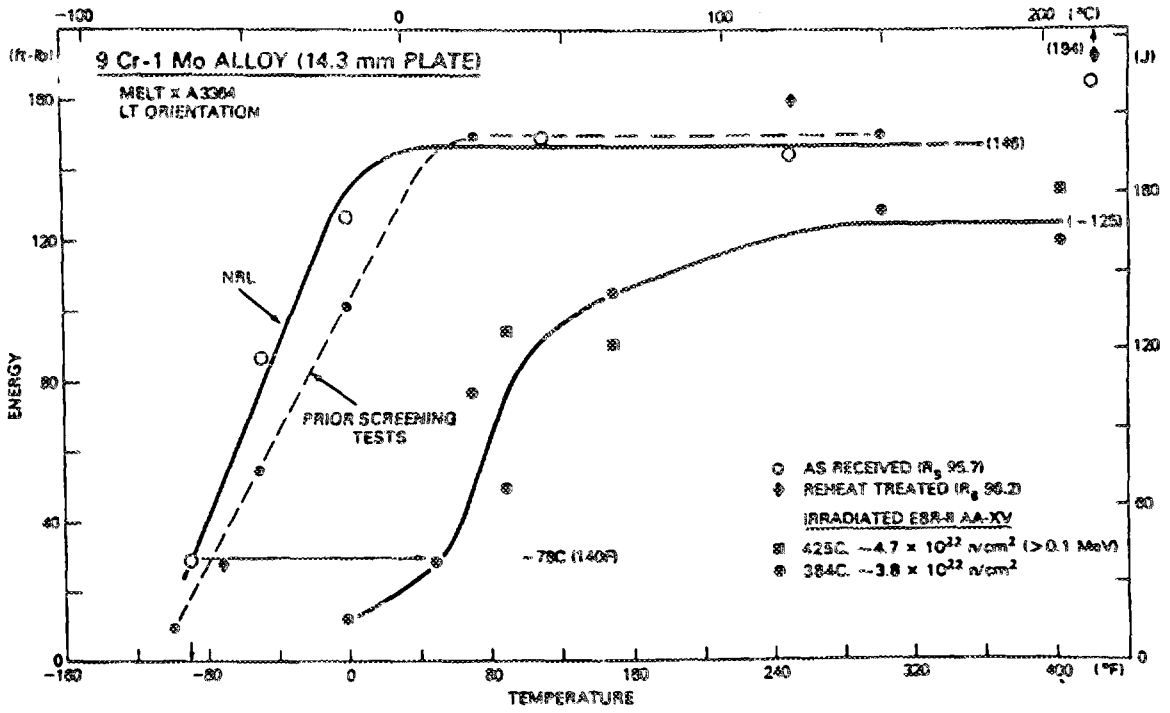


Fig. 88. Effect of fast neutron irradiation at 20 to 25 dpa on the DBTT and upper-shelf energy for T-91 (Heat XA3364). *Source*: J. R. Hawthorne, "Postirradiation Fracture Resistance Determinations for Ferritic Alloys from the AA-XV Experiment," pp. 171-78 in *National Cladding/Duct Material Development Quarterly Technical Progress Letter, July-August-September 1981*, HEDL-TC-160-30, Hanford Engineering Development Laboratory, Richland, Wash., 1981.

Similar irradiation of HT-9 [5.1-5.9 × 10^{22} neutrons/cm^2 (E > 0.1 MeV) (~25-30 dpa) at 391 and 398°C] resulted in the DBTT being increased by ~155°C (from −25 to ~130°C) and the USE being reduced significantly (~50%) from 170 to about 80 J. The increase in DBTT and decrease in USE for HT-9 is much greater than that for T-91. Irradiation at temperatures below ~550°C also resulted in significantly larger increases in room temperature yield strength and hardness for HT-9 than for T-91. These larger increases are attributed to the formation in HT-9 of G phase which does not form in T-91 (ref. 83). Therefore, while beneficial in suppressing void formation and swelling, G phase is detrimental to fracture toughness and DBTT properties. Table 15 summarizes the information on fracture toughness and hardness changes for T-91 and HT-9 as a result of the irradiation histories stated at the beginning of this section.

In other irradiation tests in EBR-II at 390°C to about 13 dpa, the DBTT of T-91 was increased by about 55°C from approximately −30 to 25°C; however, irradiation at 450, 500, and 550°C resulted in little (~5°C) or

Table 15. Effect of irradiation on Charpy-impact fracture toughness and hardness properties of T-91 and HT-9

Alloy/melt	ϕ ($\times 10^{22}$ neutrons/cm^2)[a]	Irradiation temperature[b] (°C)	C_v 41-J temperature (°C)			C_v energy at ~232°C (J)			$K_J \sqrt{m}$ at ~232°C, irradiation (MPa)	Hardness range (R_c)
			Initial	Irradiation	Increase	Initial	Irradiation	Decrease		
HT-9/91354 (33-mm rod)	5.1	391	−26	141	167	170	83	87	181	29.5–31.5
	5.9	398	−26	118	144	170	76	94	175	33.5–34.5
HT-9/91353 (12.1-mm bar)	3.5	432	−18	71	89	129	85	43	209	27.0–29.5
	4.7	425	−18	82	100	129	~85	43	205	29.0–30.5
9Cr-1Mo/XA 3364 (14.3-mm plate)	3.8	384	~ −68	10	~78	≥199	170	>29	231	20.0–21.0[c]
	4.7	425	~ −68	~10	~78	≥199	~184	>15		23.0–26.5

[a]Neutrons/cm^2, E > 0 MeV.

[b]Current estimate.

[c]One specimen had R_c 25.

Source: J. R. Hawthorne, "Postirradiation Fracture Resistance Determinations for Ferritic Alloys from the AA-XV Experiment," pp. 171–78 in *National Cladding/Duct Material Development Quart. Technical Prog. Letter*, July–September 1981, HEDL-TC-160-30, Hanford Engineering Development Laboratory, Richland, Wash., Table 4, p. 176.

no change in the DBTT. Also, for these irradiations, the higher the irradiation temperature, the lower the decrease in the USE. These data are summarized in Fig. 89.

For HT-9, irradiation at 390, 450, 500, and 550°C resulted in shifts in DBTT of ~125, 25, 30, and 55°C, respectively (Fig. 89). The DBTT of HT-9 after irradiation at 390°C is about 135°C; that for the T-91 is over 100°C lower. The shift in DBTT for HT-9 does not have a simple irradiation temperature dependence; the largest shift resulted from irradiation at the lowest temperature (390°C), and the second largest shift resulted from irradiation at the highest irradiation temperature (550°C). Irradiation at 450 and 500°C resulted in intermediate shifts in DBTT. The same irradiation temperature relationship was observed for the decrease in USE. For irradiation at 390°C, the postirradiation USE for HT-9 is only one-half that of T-91 (~23-24 and 46-48 J/cm^2, respectively). Figure 90 summarizes the effect of this irradiation on the room-temperature hardness of T-91 and HT-9. The higher hardness of HT-9 relative to T-91 is associated with the lower toughness properties of HT-9.

Data were also obtained on the effect of irradiation from 390 to 425°C on the DBTT for T-91 and HT-9 irradiated in EBR-II to higher fluences (35-45 dpa).[84,85] Figure 91 shows that for T-91 irradiation at 415 to 425 the DBTT is ~25°C (at 41 J) [Fig. 91(*a*)] and the shift in DBTT from the unirradiated condition (~ -80°C) apparently has saturated because the DBTT has shifted only 15°C from the value of 10°C at 20 to 25 dpa (ref. 82) and remains constant at 25°C from about 35 to 45 dpa. Also, the decrease in USE appears to have saturated at about 150 J/cm^2, which is only slightly lower than the 170 J/cm^2 reported at 20 to 25 dpa (ref. 82).

For HT-9 (Fig. 91), DBTTs demonstrate a continual increase to 50, 85, 120, and 165°C for irradiation at 425, 405, 400, and 390°C, respectively. There is no evidence of saturation in DBTT from the unirradiated value of -25°C (ref. 82) or of the decrease in USE from the unirradiated value of ~170 J for irradiation to 45 dpa at these temperatures. Although Fig. 91 shows data for irradiation of T-91 at 415, 420, and 425°C and at 390, 400, 405, and 425°C for HT-9, these data should be considered as one temperature for T-91 (420 ±5°C) and, at most, two temperatures (400 ±10°C and ~425°C) for HT-9.

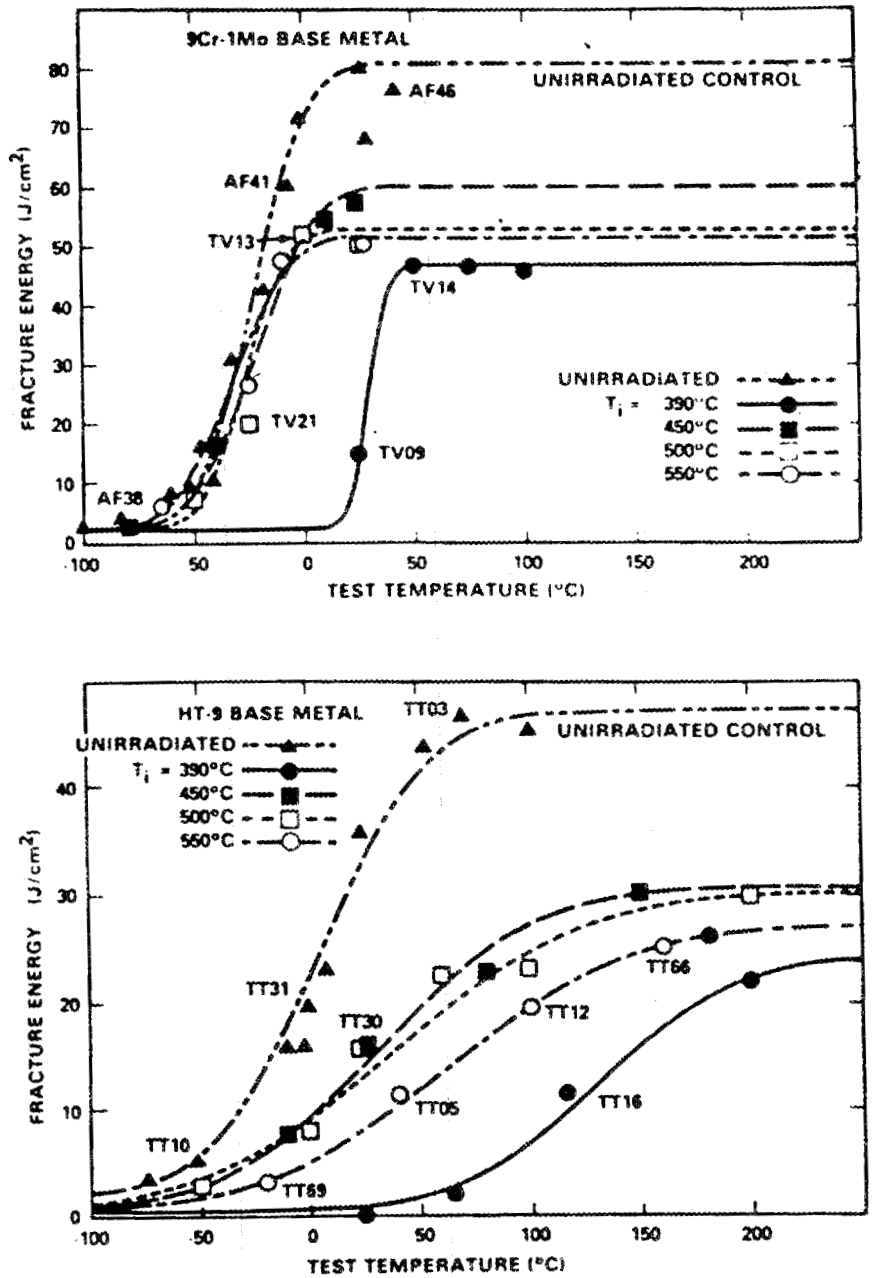


Fig. 89. Effect of irradiation temperature on DBTT and USE for T-91 and HT-9 irradiated to 13 dpa in EBR-II: (*a*) T-91, (*b*) HT-9. *Source*: W. L. Hu and D. S. Gelles, "Miniature Charpy Impact Test Results for the Irradiated Ferritic Alloys HT-9 and Modified 9Cr-1Mo," pp. 631–45 in *Proceedings of Topical Conference on Ferritic Alloys for Use in Nuclear Energy Technologies, Snowbird, Utah, June 19–23, 1983*, ed. J. W. Davis and D. J. Michel, The Metallurgical Society of the American Institute of Mining, Metallurgical, and Petroleum Engineers, New York, September 1984.

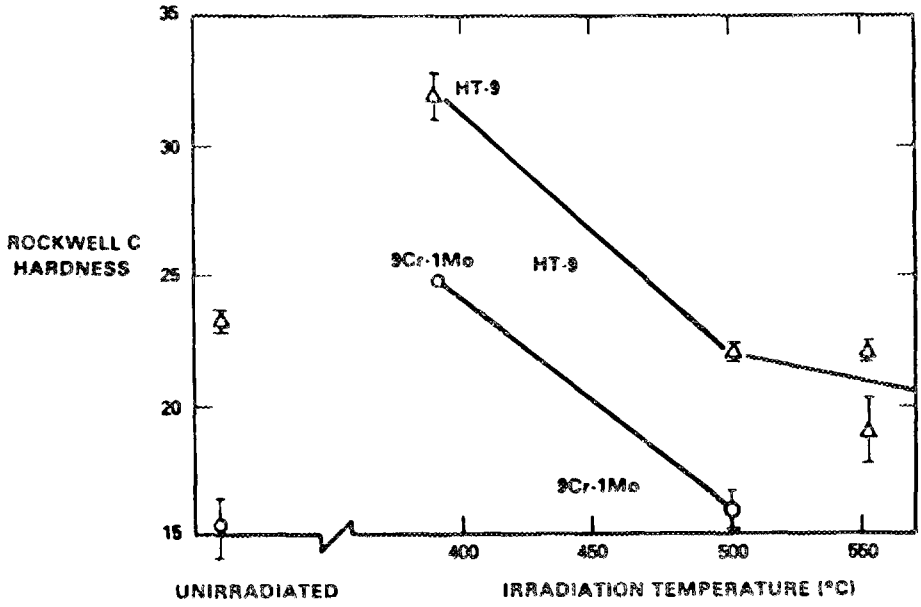
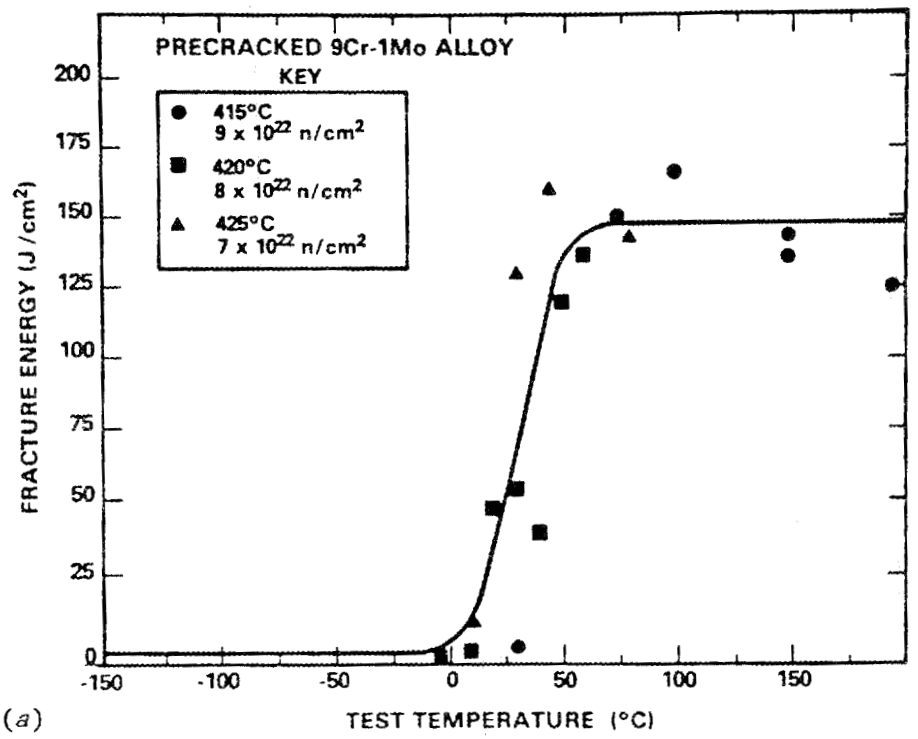


Fig. 90. Room-temperature hardness as function of irradiation temperature for T-91 and HT-9 irradiated to ~13 dpa in EBR-II. *Source*: W. L. Hu and D. S. Gelles, "Miniature Charpy Impact Test Results for the Irradiated Ferritic Alloys HT-9 and Modified 9Cr-1Mo," pp. 631–45 in *Proceedings of Topical Conference on Ferritic Alloys for Use in Nuclear Energy Technologies, Snowbird, Utah, June 19–23, 1983*, ed. J. W. Davis and D. J. Michel, The Metallurgical Society of the American Institute of Mining, Metallurgical, and Petroleum Engineers, New York, September 1984.

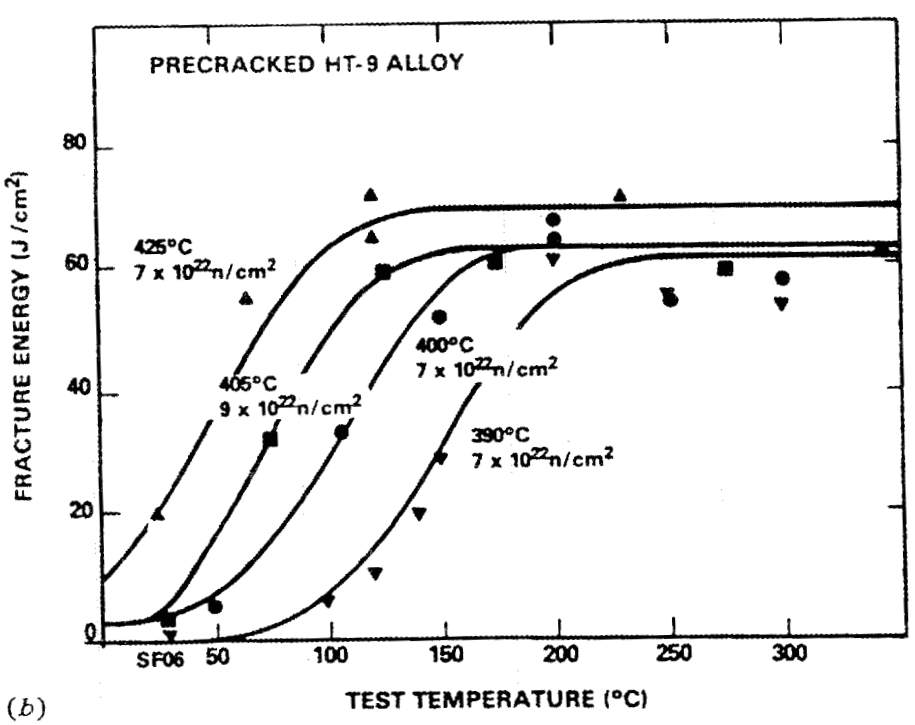
Fracture toughness (J) vs crack length (Δa) values for T-91 and HT-9 after irradiation to ~30 dpa at 410°C in EBR-II are shown in Fig. 92 (ref. 86). Test data are available for room temperature (25°C), the temperature of the FFTF reactor vessel at full power operation (427°C), and the refueling temperature (232°C). Figure 93 shows J_{IC} vs test temperature, and Table 16 provides data for J_{IC} and the tearing modulus (T) as a function of test temperature for these irradiation parameters.

These data are consistent with that presented earlier in this section. T-91 has a higher fracture toughness and is more resistant to crack propagation than HT-9 irradiated under the same or similar conditions (primarily temperature and dpa). As was observed for the irradiations to ~25 to 30 dpa, the formation of G phase in HT-9 results in fracture toughness properties significantly inferior to those for T-91 irradiated under the same conditions.

Irradiation of T-91 and HT-9 in a mixed spectrum (thermal and fast neutrons) such as HFIR results in larger reductions in tensile ductility



(*a*)



(*b*)

Fig. 91. Effect of irradiation temperature on DBTT and USE for T-91 and HT-9 irradiated to 35 to 45 dpa in EBR-II: (*a*) T-91, (*b*) HT-9. *Source*: D. S. Gelles and W. L. Hu, "Fractographic Examination of Charpy Specimens Irradiated in the AA-XV Phase II Experiment," pp. 64–76 in *National Cladding/Duct Material Development Quarterly Technical Progress Letter, April–May–June 1985*, HEDL-TC-160-45, Hanford Engineering Development Laboratory, Richland, Wash., 1985.

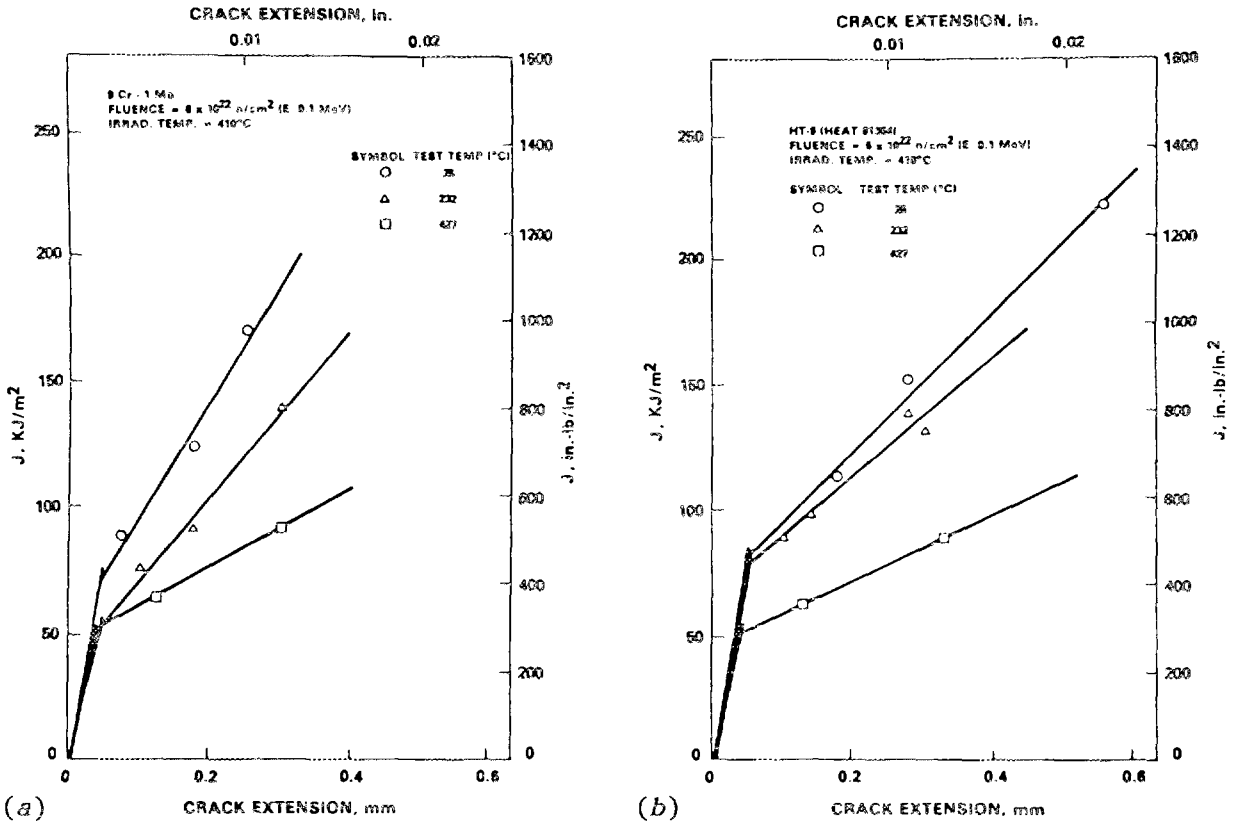


Fig. 92. J vs crack extension at 25, 232, and 427°C for T-91 and HT-9 irradiated at 410°C to 30 dpa in EBR-II: (*a*) T-91, (*b*) HT-9. *Source*: F. H. Huang, "Fracture Resistance of Irradiated HT-9 and 9Cr-1Mo," pp. 367-74 in *National Cladding/Duct Material Development Quarterly Technical Progress Letter, April-May-June 1981*, HEDL-TC-160-29, Hanford Engineering Development Laboratory, Richland, Wash., 1981.

and fracture toughness properties than does irradiation in fast neutron reactors such as EBR-II (refs. 86-92). The helium produced via thermal and epithermal neutron reactions with nickel in the alloys degrades mechanical properties more rapidly and severely than displacement damage alone, which is produced by irradiation in an LMR. For the data reported in refs. 87 and 88, the irradiation in HFIR was at the reactor coolant temperature of 50°C. At this temperature, almost all of the defect structures produced by irradiation are retained. In the previously described irradiations at 390 to 550°C in EBR-II, significant amounts of thermally induced recovery of irradiation-induced defect structures occur at the irradiation temperature. The combination of high defect density with helium effects results in large reductions in ductility and toughness properties.

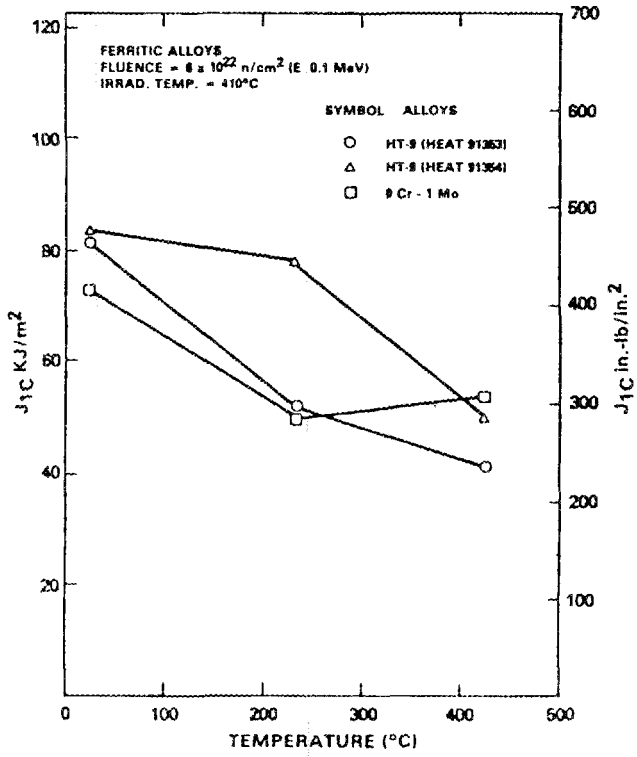


Fig. 93. Temperature dependence of J_{IC} for HT-9 and T-91 irradiated at 410°C to 30 dpa in EBR-II. *Source*: F. H. Huang, "Fracture Resistance of Irradiated HT-9 and 9Cr-1Mo," pp. 367–74 in *National Cladding/Duct Material Development Quarterly Technical Progress Letter, April–May–June 1981*, HEDL-TC-160-29, Hanford Engineering Development Laboratory, Richland, Wash., 1981.

Table 16. Fracture toughness for HT-9 and 9Cr-1Mo irradiated to ~30 dpa at 410°C

Alloy	Test temperature (°C)	J_{Ic} KJ/m²	J_{Ic} (in.-lb/in.²)	T
HT-9	25	83.6	477.5	87
(Heat 91354)	232	77.8	444.7	82
	427	50.0	285.3	60
9Cr-1Mo	25	73.1	417.5	181
(XA33641)	232	50.0	285.8	143
	427	53.9	308.0	60

Source: F. H. Huang, "Fracture Resistance of Irradiated HT-9 and 9Cr-1Mo," pp. 367–74 in *National Cladding/Duct Material Development Quarterly Technical Progress Letter*, HEDL-TC-160-29, Hanford Engineering Development Laboratory, Richland, Wash., table 2, p. 372.

Table 17 provides data for J_{IC} vs T for T-91 and HT-9 irradiated to 10 dpa at ~50°C in HFIR.[87] At 93 and 205°C, where data exist for both alloys, J_{IC} is significantly lower for T-91 than for HT-9. Also, the increase in DBTT was 90°C for T-91 and only 30°C for HT-9; the hardness of HT-9 increased 15%, and T-91 increased 25%. These relative results are opposite to those observed after irradiation of these two alloys in EBR-II where fracture toughness degradation was much more severe for T-91 than for HT-9. These results are shown schematically in Figs. 94 and 95.

From overall application considerations, T-91 should provide better long-term performance than HT-9 under LMR irradiation conditions for temperatures up to about 550°C. Although the swelling rates of T-91 are larger than those for HT-9, they are still low relative to swelling rates for the currently used austenitic steels and are acceptable for any proposed LMR or fusion reactor structures and components. With respect to reactor operation, inspections, maintenance, and surveillance criteria and/or requirements, the significantly smaller changes in DBTT, USE, and hardness because of irradiation of T-91 relative to HT-9 should far outweigh the larger increase in swelling.

Table 17. Fracture toughness of HT-9 and T-91 irradiated to 10 dpa at 50°C in HFIR

Alloy	Test temperature (°C)	J_{IC} (KJ/m^2)	Tearing modulus (T)
HT-9	25	63.3	
HT-9	93	52.0	33
HT-9	205	56.7	28
9Cr-1Mo	93	32.9	23
9Cr-1Mo	205	35.2	16
9Cr-1Mo	450	30.8	59

Source: F. H. Huang, "The Fracture Toughness of Ferritic Alloys Irradiated in HFIR," pp. 180–84 in *ADIP Semiannual Progress Report for Period Ending September 30, 1983*, DOE/ER-0045/11, U.S. Department of Energy, 1983, Table 7.17.1, p. 182.

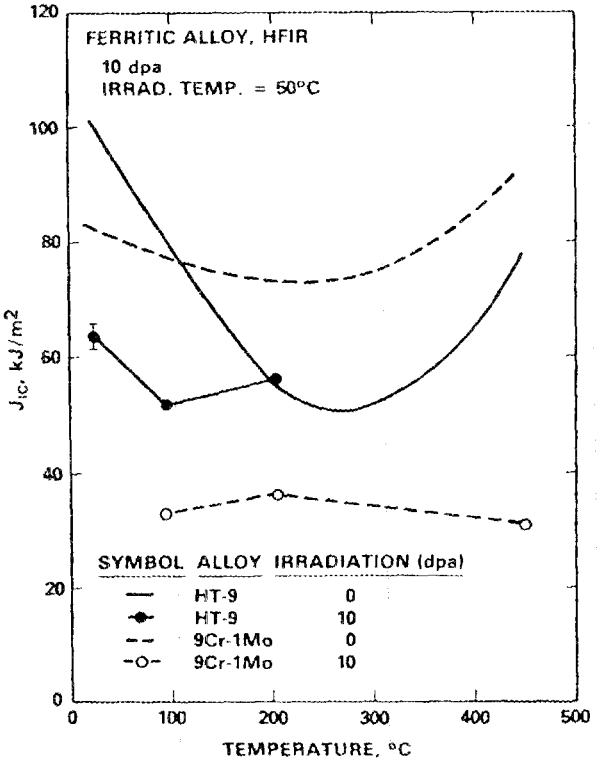


Fig. 94. Temperature dependence of fracture toughness (J_{IC}) of T-91 and HT-9 in unirradiated condition and after irradiation to 10 dpa at 50°C in HFIR. *Source*: F. H. Huang, "The Fracture Toughness of Ferritic Alloys Irradiated in HFIR," pp. 180–84 in *Alloy Development for Irradiation Performance Semiannual Progress Report for Period Ending September 30, 1983, DOE/ER-0045/11, U.S. Department of Energy, Office of Fusion Energy, March 1984.*

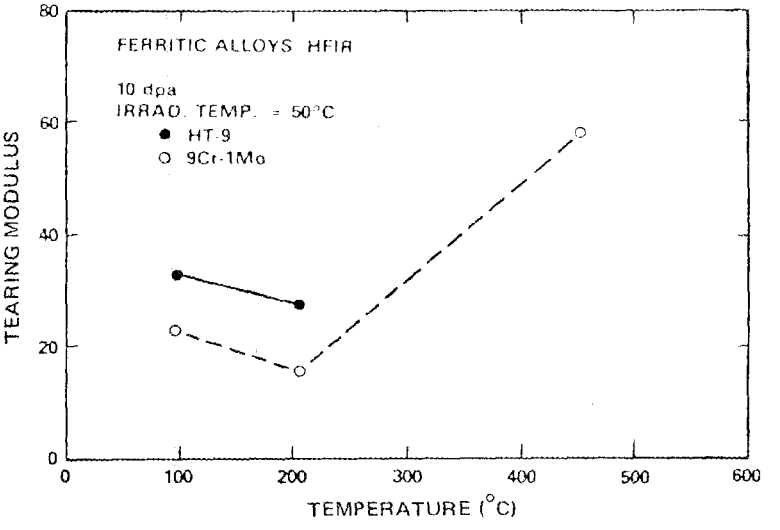


Fig. 95. Temperature dependence of tearing modulus for T-91 and HT-9 irradiated to 10 dpa at 50°C in HFIR. *Source*: F. H. Huang, "The Fracture Toughness of Ferritic Alloys Irradiated in HFIR," pp. 180–84 in *Alloy Development for Irradiation Performance Semiannual Progress Report for Period Ending September 30, 1983*, DOE/ER-0045/11, U.S. Department of Energy, Office of Fusion Energy, March 1984.

3.3.4.3 Tensile Properties

Figures 96 and 97 show the strength and ductility of T-91 after irradiations to ~25 dpa at about 400, 450, and 550°C in EBR-II.[93] Included in Figs. 96 and 97 are lines for data on HT-9 in the N&T and cold-worked conditions from similar irradiations in EBR-II.[94] The lines in Figs. 96 and 97 are based on very limited data, especially for T-91; however, for the three irradiation temperatures used, the strength and ductility of T-91 are significantly lower than for HT-9. As shown in Table 18 (ref. 95), the tensile properties of the unirradiated alloys were approximately equal for temperatures up to 600°C. The only exception is total elongation, where the values for HT-9 at 550 and 600°C are almost twice those for T-91. Therefore, tensile properties shown in Figs. 96 and 97 for 400, 450, and 550°C are a result of the irradiation and are not due to large differences in preirradiation values. For most irradiations at temperatures below about 0.4 T_m (T_m = melting temperature in Kelvin), at which helium is not produced, the material with the lower strength properties usually has higher ductility properties. The higher yield strength of HT-9 is most probably due to formation of G phase during irradiation in EBR-II. Results similar to those shown in Figs. 96 and 97 were also observed after irradiation of T-91 and HT-9 in HFIR (Fig. 98 and Table 19).[89,90] Table 20 presents data on the ratios of irradiated to unirradiated yield and ultimate tensile strengths for T-91 and HT-9 for test temperatures of 25 and 300°C. For these data, HT-9 is strengthened a little more than T-91, especially yield strength at 25°C (Fig. 97). The reason for the larger increase in yield strength of HT-9 is not known. The irradiation temperature of 55°C is too low for formation of G phase. The helium contents are low (10-26 appm), and data from other irradiations indicate that helium has a negligible effect on hardness and strength properties at these low concentrations and temperatures where helium is immobile.[87-91]

Figure 99 shows strain vs applied stress at 25°C for T-91 and HT-9 in the unirradiated condition and for material irradiated to ~9 dpa at 50°C in HFIR.[89] There are no obvious differences in the unirradiated curves or the irradiated curves that would indicate a difference in the irradiation

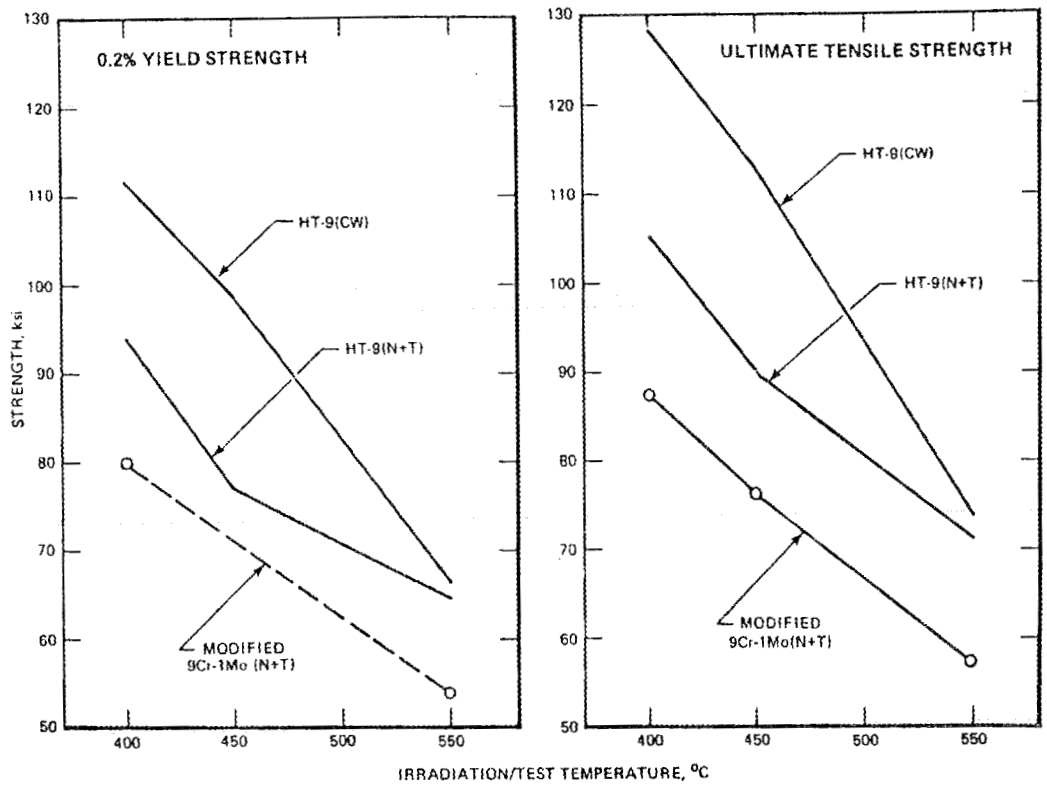


Fig. 96. Yield and ultimate tensile strengths of T-91 and HT-9 after irradiation to ~25 dpa in EBR-II at 400, 450, and 550°C. Test temperature was the same as the irradiation temperature for each specimen. *Source*, figure: T. Lauritzen, R. E. Blood, and S. Vaidyanathau, "Some Effects of Irradiation on the Tensile Properties of Modified 9Cr-1Mo," pp. 183-90 in *National Cladding/Duct Material Development Quarterly Technical Progress Letter, January–February–March 1984*, HEDL-TC-160-40, Hanford Engineering Development Laboratory, Richland, Wash., 1984. *Source*, data for HT-9 lines: T. Lauritzen, R. E. Blood, and S. Vaidyanathau, "Mechanical Properties of Reactor-Irradiated HT-9: Effect of Thermomechanical Treatments," pp. 13-25 in *National Cladding/Duct Material Development Quarterly Technical Progress Letter, October–November–December 1983*, HEDL-TC-160-39, Hanford Engineering Development Laboratory, Richland, Wash., 1984.

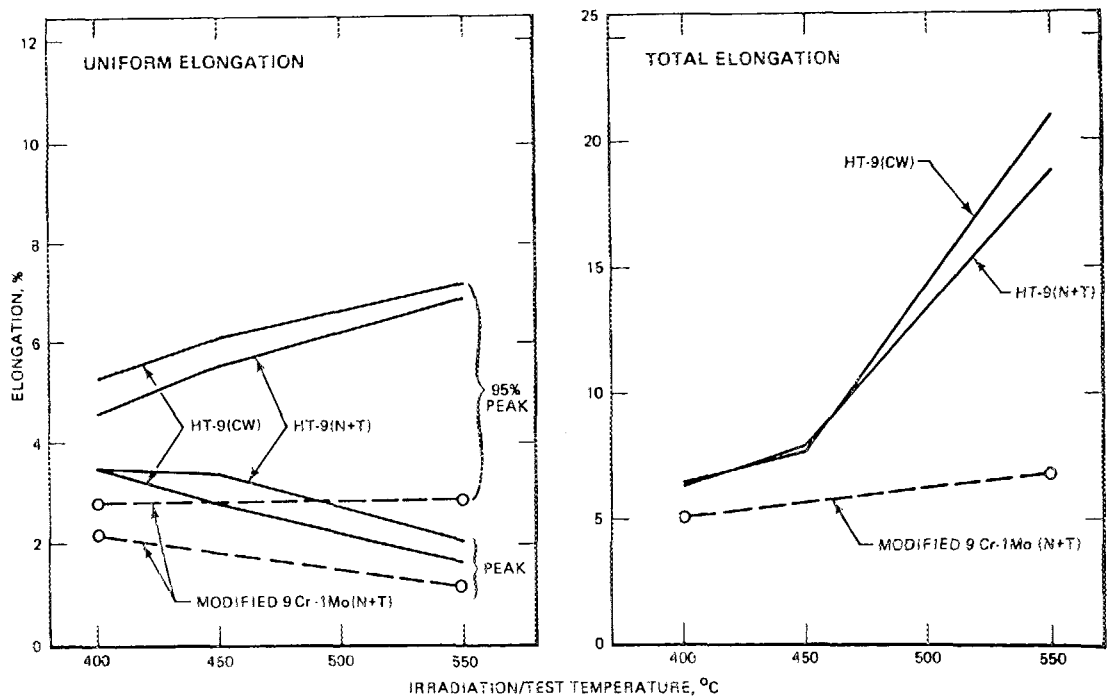


Fig. 97. Uniform and total elongation of T-91 and HT-9 after irradiation to ~25 dpa in EBR-II at 400, 450, and 550°C. Test temperature was same as irradiation temperature for each specimen. *Source*, figure: T. Lauritzen, R. E. Blood, and S. Vaidyanathau, "Some Effects of Irradiation on the Tensile Properties of Modified 9Cr-1Mo," pp. 183-90 in *National Cladding/Duct Material Development Quarterly Technical Progress Letter, January–February–March 1984*, HEDL-TC-160-40, Hanford Engineering Development Laboratory, Richland, Wash., 1984. *Source*, data for HT-9 lines: T. Lauritzen, R. E. Blood, and S. Vaidyanathau, "Mechanical Properties of Reactor-Irradiated HT-9: Effect of Thermomechanical Treatments," pp. 13-25 in *National Cladding/Duct Material Development Quarterly Technical Progress Letter, October–November–December 1983*, HEDL-TC-160-39, Hanford Engineering Development Laboratory, Richland, Wash., 1984.

Table 18. Tensile properties of unirradiated T-91 and HT-9 from room temperature to 650°C

Test temperature (°C)	Yield strength (MPa)	Yield strength (ksi)	Ultimate strength (MPa)	Ultimate strength (ksi)	Elongation (%) Uniform	Elongation (%) Total	Reduction in area (%)
			T-91[a]				
RT[b]	575.7	83.5	681.9	98.9	7.8	10.5	74.4
232	535.7	77.7	608.1	88.2	4.6	7.9	70.8
400	499.2	72.4	561.9	81.5	3.3	6.9	68.1
450	499.2	72.4	558.5	81.0	3.8	7.9	63.4
500	468.8	68.0	513.0	74.4	6.0	9.6	71.4
550	448.8	65.1	481.3	69.8	4.0	8.6	70.3
600	350.2	50.8	362.7	52.6	5.2	11.0	82.5
			HT-9[c]				
RT	606.7	88.0	759.1	110.1	9.7	11.9	57.4
232	547.4	79.4	678.4	98.4	6.1	8.2	58.6
400	504.0	73.1	598.5	86.8	5.1	8.1	58.7
450	508.1	73.7	610.2	88.5	5.7	8.4	59.5
500	484.0	70.2	561.2	81.4	7.8	11.4	60.8
550	418.5	60.7	459.2	66.6	6.7	18.6	74.0
600	339.9	49.3	361.3	52.4	5.9	21.2	84.5
650	207.5	30.1	235.8	34.2	8.6	19.8	89.8

[a]Alloy condition: T-91 austenitized and tempered [1038°C/5 min, room air cooled (RAC) + 760°C/1 h, RAC], heat 91887.

[b]Room temperature.

[c]Alloy condition: HT-9 austenitized and tempered (1038°C/5 min, RAC + 760°C/30 min, RAC), heat 91354.

Source: T. Lauritzen et al., "Preirradiation Tensile Properties of AA XIV Ferritic Alloys," pp. 291–315 in *National Cladding/Duct Material Development Quarterly Technical Progress Letter, April–May–June 1985*, HEDL-TC-160-45, Hanford Engineering Development Laboratory, Richland, Wash., 1985. (T-91 data from Table 4, p. 301; HT-9 data from Table 2, p. 295.)

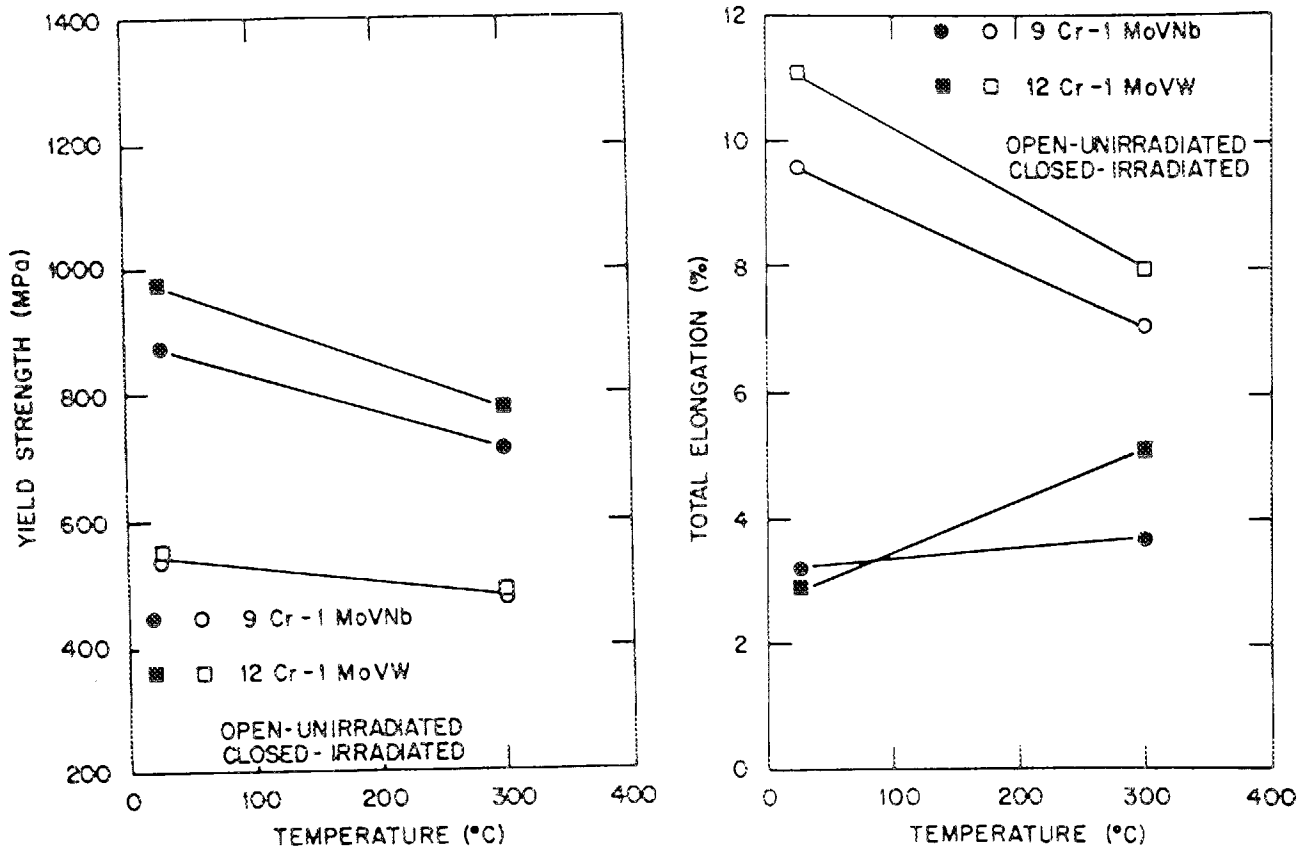


Fig. 98. Effect of irradiation in HFIR to ~9 dpa at 50°C on 0.2% yield strength and total elongation of T-91 and HT-9 at 25 and 300°C. *Source*: R. L. Klueh, J. M. Vitek, and M. L. Grossbeck, "Nickel-Doped Ferritic (Martensitic) Steels for Fusion Reactor Irradiation Studies: Tempering Behavior and Unirradiated and Irradiated Tensile Properties," pp. 648–64 in *Effects of Radiation on Materials*, ASTM-STP-782, American Society for Testing and Materials, Philadelphia, 1982.

Table 19. Unirradiated and irradiated tensile properties of T-91 and HT-9

Displacement level[a] (dpa)	Helium concentration[b] (at. ppm)	Test temperature (°C)	Strength (MPa)		Elongation (%)	
			Yield	Ultimate	Uniform	Total
		T-91 (N&T)[c]				
		25	541	656	5.1	9.6
9.3	11	25	878	878	0.2	3.2
		300	483	581	3.6	7.1
7.6	10	300	716	716	0.2	3.7
		HT-9 (N&T)[c]				
		25	553	759	8.1	11.2
9.3	26	25	980	992	0.4	2.9
		300	483	652	5.1	8.0
9.1	18	300	783	815	1.8	5.1

[a]Irradiation was in HFIR at ~50°C.

[b]Calculated level of helium from ^{58}Ni and ^{10}B.

[c]N&T = normalized and tempered.

Source: R. L. Klueh and J. M. Vitek, "Tensile Behavior of Three Commercial Ferritic Steels After Low-Temperature Irradiation," pp. 615–22 in *Proceedings of Topical Conference on Ferritic Alloys for Use in Nuclear Energy Technologies, Snowbird, Utah, June 19–23, 1983*, ed. J. W. Davis and D. J. Michel, The Metallurgical Society of the American Institute of Mining, Metallurgical, and Petroleum Engineers, Warrendale, Pa., September 1984, Table II, p. 616.

Table 20. Relative strengthening of T-91 and HT-9 by irradiation in HFIR to 8 to 9 dpa at 50°C

Alloy	Relative strengthening at each test temperature			
	25°C		300°C	
	R_Y[a]	R_U[b]	R_Y[a]	R_U[b]
T-91	1.6	1.3	1.5	1.2
HT-9	1.8	1.3	1.6	1.3

[a]R_Y = ratio of irradiated to unirradiated yield strength.

[b]R_U = ratio of irradiated to unirradiated ultimate tensile strength.

Source: R. L. Klueh and J. M. Vitek, "Tensile Behavior of Three Commercial Ferritic Steels After Low-Temperature Irradiation," pp. 615–22 in *Proceedings of Topical Conference on Ferritic Alloys for Use in Nuclear Energy Technologies, Snowbird, Utah, June 19–23, 1983*, ed. J. W. Davis and D. J. Michel, The Metallurgical Society of the American Institute of Mining, Metallurgical, and Petroleum Engineers, Warrendale, Pa., September 1984, Table III, p. 619.

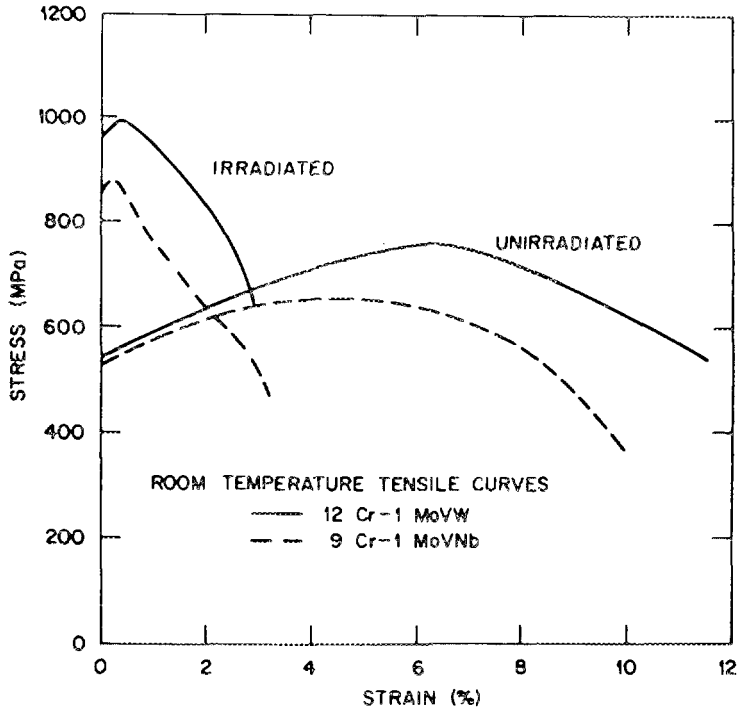


Fig. 99. Comparison of the strain vs stress curves for T-91 (9Cr-1MoVNb) and HT-9 (12Cr-1MoVW) at 25°C in the unirradiated condition and after irradiation to ~9 dpa at 50°C in HFIR. *Source*: R. L. Klueh, J. M. Vitek, and M. L. Grossbeck, "Nickel-Doped Ferritic (Martensitic) Steels for Fusion Reactor Irradiation Studies: Tempering Behavior and Unirradiated and Irradiated Tensile Properties," pp. 648–64 in *Effects of Radiation on Materials*, ASTM-STP-782, American Society for Testing and Materials, Philadelphia, 1982.

response of the two alloys. However, strain as a function of applied stress at 300 to 550°C would be far more valuable in providing an understanding of the irradiation response of the two alloys.

Additional information on the effects of irradiation in EBR-II on the tensile properties of T-91 are provided in Figs. 100 and 101 and Table 21 (ref. 91). Figure 100 shows the 0.2% yield and ultimate tensile stresses as a function of temperature from 400 to 550°C for T-91 in the unirradiated condition; after aging for 5000 h at 400, 450, 500, and 550°C; and after irradiation to 10 to 12 dpa at 390, 450, 500, and 550°C in EBR-II. For each specimen the tensile test was conducted at the same temperature as the aging temperature or the irradiation temperature. Figure 100 shows the uniform and total elongations for these specimens. Determination of the effects of aging at the temperatures at which the specimens were irradiated and for times relevant to the time of irradiation

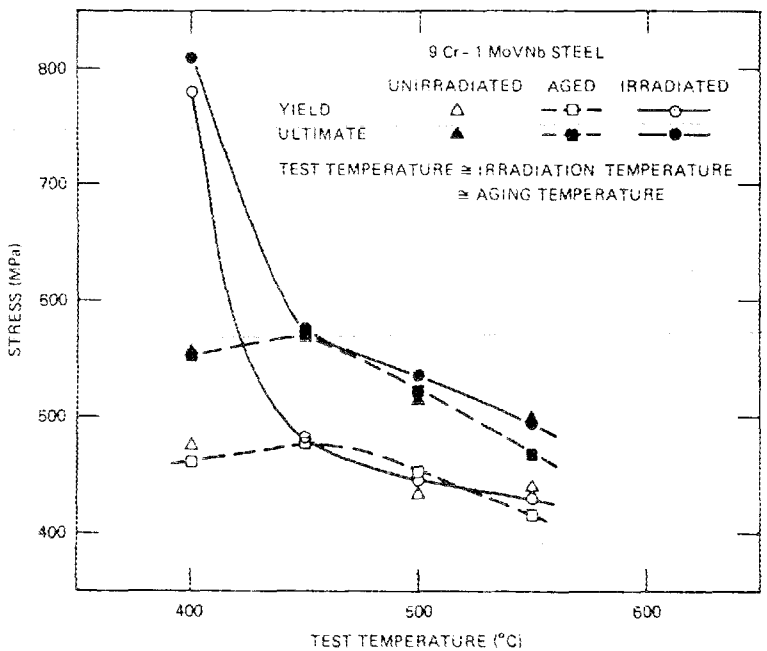


Fig. 100. Effects of aging for 5000 h at 400, 450, 500, and 550°C or irradiation in EBR-II to 10 to 12 dpa at these temperatures on 0.2% yield strength and ultimate tensile strength at aging or irradiation temperature. *Source*: R. L. Klueh and J. M. Vitek, "Elevated-Temperature Tensile Properties of Irradiated 9Cr-1MoVNb Steel," *J. Nucl. Mater.* **132**, 27-31 (1985).

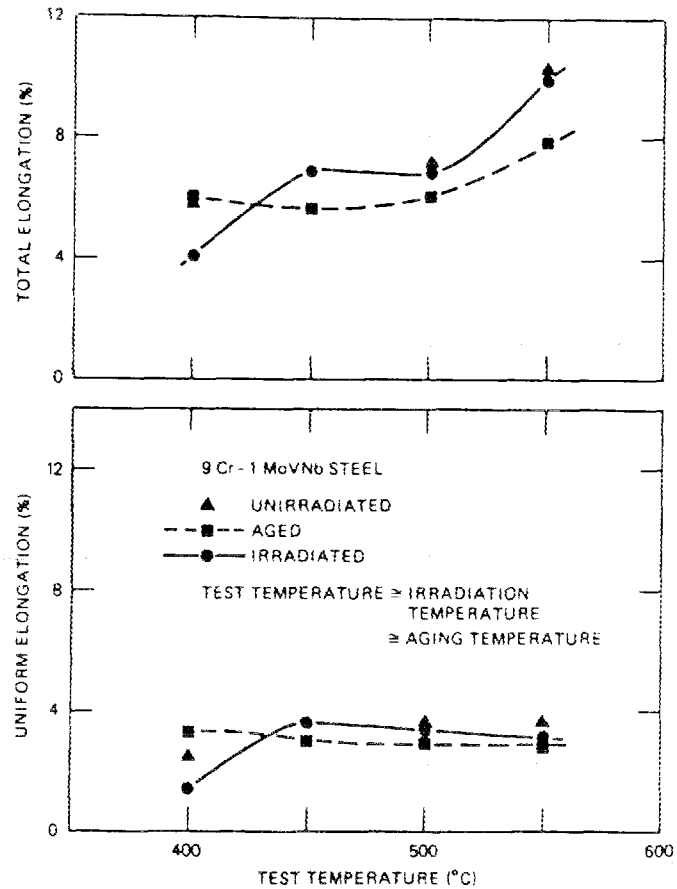


Fig. 101. Effects of aging for 5000 h at 400, 450, 500, and 550°C or irradiation in EBR-II to 10 to 12 dpa at these temperatures on uniform and total elongation at aging or irradiation temperature. *Source*: R. L. Klueh and J. M. Vitek, "Elevated-Temperature Tensile Properties of Irradiated 9Cr-1MoVNb Steel," *J. Nucl. Mater.* **132**, 27-31 (1985).

Table 21. Effect of aging for 5000 h or irradiation to 10 to 12 dpa in EBR-II on the tensile properties of T-91

Condition	Temperature (°C)		Strength (MPa)		Elongation (%)	
	Aged/ irradiated	Test	Yield	Ultimate	Uniform	Total
As heat treated		22	547	697	7.3	11.9
Irradiated	390	22	881	933	3.6	7.0
	500	22	558	712	7.9	12.4
	550	22	544	697	6.1	10.0
As heat treated		400	474	553	2.5	5.9
Irradiated	390	400	781	808	1.4	4.1
Aged 5000 h	400	400	460	555	3.3	6.0
Irradiated	450	450	480	575	3.6	6.9
Aged 5000 h	450	450	473	570	3.0	5.6
As heat treated		500	438	517	3.6	7.2
Irradiated	500	500	445	536	3.3	6.8
Aged 5000 h	500	500	451	522	2.9	6.1
As heat treated		550	440	497	3.6	10.3
Irradiated	550	550	429	495	3.1	9.9
Aged 5000 h	550	550	414	467	2.9	7.8

Source: R. L. Klueh and J. M. Vitek, "Elevated-Temperature Tensile Properties of Irradiated 9Cr-1MoVNb Steel," *J. Nucl. Mater.* **132**, 27-31 (1985), Tables 2 and 3, p. 29, and Table 4, p. 30.

permits the effect of time at the irradiation temperature to be separated from the effects of irradiation. Figures 100 and 101 illustrate that 5000 h at the irradiation temperature had no effect on the tensile properties of T-91 at the four irradiation temperatures. Figure 100 shows that the strength properties at 400°C have been increased by 65 (σ_y) and 46% (σ_{UTS}) by irradiation at 390°C; Fig. 101 shows that the ductility properties have been reduced by 44 (ϵ_u) and 31% (ϵ_T). However, irradiation at 450, 500, and 550°C had no effect on the strength or ductility properties at these test temperatures.

The strength and ductility properties after irradiation are in excellent agreement with the microstructural observations reported for T-91 under the swelling and microstructure portion of this section.

Irradiation at 400°C resulted in the formation of a high density of dislocation loops and tangles, small rod-shaped precipitates of CrC_2, and a low density of faceted voids.[78] The dislocation loops and tangles and precipitates increase the strength properties and decrease the ductility properties. Irradiation of T-91 at 450°C and above results in precipitation of MC and recovery of the preirradiation dislocation structures, and no irradiation-induced dislocation structure or voids are observed after irradiation at these temperatures. The lack of any irradiation-induced defect structures at 450, 500, and 550°C coincides with the unchanged tensile properties after irradiation at these temperatures. After irradiation in EBR-II, the DBTT USE, fracture toughness, crack growth rate, and tearing modulus values for T-91 were superior to those for HT-9. These parameters include a measure of ductility; therefore, the lower tensile ductility values of T-91 relative to HT-9 are not obviously reflected in fracture toughness properties.

It appears that more test data, especially over the range of 1 to 10 dpa, will have to be obtained to provide the information required to determine the mechanism(s) responsible for the large differences in fracture toughness and tensile properties of T-91 and HT-9 after irradiation in EBR-II vs HFIR. Significantly more data are required to evaluate the technological importance of the effects of irradiation on the tensile properties of these alloys for fusion reactor and advanced fusion reactor applications.

3.3.4.4 Creep

There are no data for the effects of radiation on creep of T-91. However, combined with the observations on the effects of irradiation on microstructures of T-91 at 390 to 550°C, data for the effects of radiation on creep of HT-9, 9Cr-2Mo, 2.25Cr-1Mo, and type 316 stainless steel provide a rather sound basis for estimating the effects of irradiation on creep of T-91.

Pressurized tube creep specimens of HT-9, 9Cr-2Mo, 2.25Cr-1Mo steels, and 20% cold-worked type 316 stainless steel were irradiated in EBR-II to ~14 dpa at temperatures of 445, 505, and 570°C.[96] Each alloy was irradiated with midwall hoop stresses of 0, 50, 75, and 100 MPa at each temperature. After irradiation (4500 h) at 505 and 570°C, total creep

strains for all three ferritic alloys were the same as the thermal creep strains for all three applied stress levels, and creep strains for 9Cr-2Mo were lower than those for the other alloys. For irradiation at 445°C, the total creep strains for the irradiated specimens were slightly larger than those caused by thermal creep alone; this is the expected behavior for irradiation creep. For most alloy systems, irradiation creep becomes measurable at temperatures just below those where thermal creep becomes so small that it is not measurable in tests of a few hundred hours duration at stresses reasonably below the yield stress. For all stresses and temperatures employed, the total creep strains for 9Cr-2Mo and HT-9 are significantly less than those for the 20% cold-worked type 316 stainless steel. For irradiation at 570°C and a stress of 100 MPa, the total creep strain for the 2.25Cr-1Mo was approximately ten times that for 9Cr-2Mo and HT-9 and about five times that for the 20% cold-worked type 316 stainless steel.

Through the use of the equation for minimum creep rate [Eq. (3) for T-91 provided in ref. 97] and the temperatures and stresses used for the EBR-II irradiation, creep strains in T-91 were calculated to be lower than those reported for HT-9 in ref. 96. The yield strength of T-91 is slightly lower than that for HT-9 at 445 and 505°C and slightly higher at 570°C (Table 18), and their creep strengths are similar at temperatures near 570°C. Also, the calculation of creep strain for T-91 [using Eq. (3) of ref. 97] does not include primary creep strain. Considering the differences in yield strength and the exclusion of primary creep strain in calculating total creep strain for T-91, the calculated creep strains for T-91 are in reasonable agreement with those reported after irradiations of HT-9.

As stated earlier in this section, T-91 irradiation at temperatures between 450 and 550°C results in no accumulation of radiation-induced defect structures. The combination of no observed changes in the microstructure of T-91 for irradiation between 450 and 550°C, combined with the similarity of yield and creep strengths of T-91 and HT-9 at these temperatures, indicates that fast reactor irradiation will have no significant effect on the creep rates of T-91 at temperatures between ~450 and 550°C, the principal temperature range of application for T-91 in LMR systems.

As is the case for the other ferritic alloys described here, at temperatures just below 450°C, irradiation creep will contribute to the total creep strain at high levels of applied stress.

3.3.4.5 Effects of Helium

For fusion reactor applications the effects on mechanical properties of atomic displacements combined with neutronically produced helium must be determined. Irradiation conducted in nuclear reactors that have high fluxes [$>5 \times 10^{13}$ neutrons/($cm^2 \cdot s$)] of both low-energy neutrons [3.8×10^{-21} J (0.024 eV) $< E < 1.6 \times 10^{-16}$ J (1 keV)] and fast neutrons ($E >$ 0.1 MeV) can produce helium contents and atomic displacements relevant to fusion reactor operations in alloys that contain nickel. The helium is produced by low-energy neutron captures in nickel, and the atomic displacements are produced by the fast neutrons. Helium production during irradiation is a direct function of nickel in the alloy. Irradiations of T-91 and HT-9 with standard nominal nickel contents of 0.1% and 0.4 to 0.5%, respectively, and with up to 2.30% additional nickel to increase helium production within the alloys, together with type 316 stainless steel, have been conducted up to ~5 or 6 dpa at 55°C in HFIR. Because of its high nickel content (~12%), type 316 stainless steel produced copious amounts of helium (~234 appm) during irradiation in HFIR to this displacements-per-atom level. This helium results in large losses of strength and ductility in type 316 stainless steel when tensile tested at temperatures above ~500°C. Embrittlement of the steel by helium increases with increasing temperature above about 500°C because of an agglomeration of helium atoms into bubbles and an intergranular alloy failure with no measurable plastic strain at temperatures above 650°C.

To determine the susceptibility of T-91 and HT-9 to helium embrittlement, the alloys were tensile tested at 700°C after irradiation. Irradiation resulted in helium concentrations from about 10 to 50 appm in these two alloys. As shown in Table 22, the ferritic alloys exhibit no evidence of the helium embrittlement at 700°C that occurs in austenitic stainless steels, such as types 304 and 316. Also, the strain vs applied stress curves for T-91 and HT-9 are the same before and after irradiation. Note that for type 316 stainless steels (Table 22) total elongation has been reduced from 8.5 to 0.3% after irradiation. Except for two test

Table 22. Tensile properties of unirradiated and irradiated 9Cr-1Mo, 12Cr-1Mo, and 20% cold-worked type 316 stainless steel at 700°C[a]

Steel	Fluence >0.1 MeV ($\times 10^{25}$ neutrons/m^2)	Displacement level (dpa)	Helium concentration[b] (at. ppm)	Strength (MPa)		Elongation (%)	
				Yield	Ultimate	Uniform	Total
12Cr-1MoVW	0			110	142	3.1	19.6
	6.3	5.2	13	103	123	2.2	20.5
12Cr-1MoVW-1Ni	0			93	121	2.5	19.0
	8.9	6.4	28	95	105	1.2	22.9
12Cr-1MoVW-2Ni	0			139	163	2.1	25[c]
	8.9	6.5	49	128	140	1.9	25.5
12Cr-1MoVW-2Ni (adjusted)	0			181	212	1.7	17.3
	6.3	5.1	36	123	153	1.3	21.0
9Cr-1MoVNb	0			105	131	2.5	16.8
	6.3	5.1	8				>10[d]
9Cr-1MoVNb-2Ni	0			148	171	1.8	19.1
	8.9	6.4	47	119	131	1.2	25.9
9Cr-1MoVNb-2Ni (adjusted)	0			181	192	0.6	14.4
	6.3	5.2	36	194	203	0.6	11.5
CW 316 SS	0			368	400	1.5	8.5
	8.9	6.6	234	330	341	0.3	0.3

[a]Specimens irradiated in HFIR at ~55°C.

[b]Calculated level of helium from ^{58}Ni and ^{10}B.

[c]Crosshead stopped for a short time during test; the total elongation was estimated by extrapolation of the early portion of the load-elongation curve.

[d]Data for this test were accidentally not recorded. An estimate of the elongation was obtained by fitting the fractured sample together and measuring the length (in a hot cell); thus, this is a very rough estimate for the elongation.

Source: R. L. Klueh and J. M. Vitek, "The Resistance of 9Cr-1MoVNb and 2.25Cr-1MoVW Steels to Helium Embrittlement," *J. Nucl. Mater.* 117, 295-302 (1983), Table 2, p. 297.

specimens, which still exhibited more than 10% total elongation, the ferritic alloys had ~20% total elongation, and the values were higher after irradiation than before irradiation.

Although the available data are for temperatures and helium concentrations much higher than those anticipated for the use of T-91 in an LMR, these results indicate that T-91 has a high resistance to irradiation-induced helium embrittlement.

3.3.5 Sodium Effects

3.3.5.1 Overview

Little experimental data presently exist on the effects of exposure to sodium on the mechanical properties of T-91. However, it is indicated that short-term (5-6 × 10^3-h) exposure to sodium at 538°C does not affect the strength of T-91, but exposure at 593°C can result in a significant reduction in strength.[99-103] These results appear to support observations that indicate that in a bimetallic system with austenitic stainless steel, T-91 loses nitrogen when exposed to sodium at temperatures above 550°C, resulting in a decrease in strength properties.[104] Calculations for T-91 in a sodium system containing austenitic stainless steel (0.05°C) indicate that it will carburize at temperatures up to ~600°C and decarburize at temperatures above 600°C;[104,105] the measured decarburization rate over the temperature range of 600 to 650°C is reported to be a factor of two higher than those reported for other ferritic steels of comparable composition.[106] This results in large strength decreases at temperatures above 600°C. For this reason, T-91 is no longer under consideration for LMR fuel cladding applications. These results are not considered to be too significant for other LMR structural applications, which are at temperatures of 550°C and below. T-91 was never considered for long-term structural applications at temperatures of 600 to 650°C. The alloy should perform well for fuel duct and out-of-core structures and components where sodium temperatures are 550°C or lower.

3.3.5.2 Creep-Rupture

Seven creep-rupture tests in flowing sodium have been conducted on heat number 30394 of T-91.[99,100] Six tests were conducted on base metal,

three each at 538 and 593°C. One test on a specimen of all weld metal was conducted at 593°C; however, the test was terminated before rupture, and no information was obtained on the effects of sodium on creep-rupture properties.

Two of the six tests on base metal were terminated before rupture. For the other four tests the data for creep-rupture life vs applied stress lie above the average curve for five heats of T-91 tested in air at 538 and 593°C. However, all of the data (sodium or air) for heat 30394 lie above the average curve for the five heats used to compile the reference line in air (Fig. 3.2 of ref. 100). The data for the four tests in sodium lie on the line drawn through the data for the tests in air for heat 30394. Based on these short-term data (230–3722 h), sodium has no apparent major effect on the creep rupture life of T-91.

However, this statement is not in complete agreement with the data in Table 3.2 of ref. 100 that show creep-rupture life in sodium to be one-fourth of that in air (356 vs 1403 h, respectively). In addition, for tests conducted at 593°C and 138 MPa in high-purity argon on specimens that were in vacuum or exposed to sodium for 5000 h at 593°C, the measured creep-rupture lives were 5800 and 2970 h, respectively.[103] The reported creep-rupture life in air at 593°C and 138 MPa for the same heat of material (30394) was ~20,000 h.[103] Hence, aging reduced creep-rupture life by approximately a factor of three (3.45), and sodium exposure reduced creep-rupture life by almost a factor of seven (6.73). Also, this trend has been observed for shorter-term tests at 172 and 207 MPa.[103] Because T-91 does not decarburize in sodium at temperatures below 600°C,[104,105] the decrease in creep-rupture life by sodium pre-exposure at 593°C was attributed to a loss of nitrogen from the alloy.[104,107] However, the largest decrease in creep-rupture life occurred as a result of aging in vacuum where the driving force and mechanism for nitrogen removal from the alloy is significantly less than for exposure in hot sodium. Variations in creep-rupture life of the magnitude described here (i.e., factors of 3–7 or more for rupture times of a few hundred to a few thousand hours) are often observed in structural alloys; however, the variations are usually random and not all in one direction. For the tests reported here, the variations are not random, but all are toward decreasing creep-rupture

life. It is apparent from the data obtained at both ANL and WAESD that additional testing is required to establish the effects of sodium on the creep-rupture life of T-91.

Creep-rupture data for T-91 are consistent with those reported for standard 9Cr-1Mo steel and 12Cr-1MoNiV steel for temperatures below 550°C.[108-110] For tests conducted in 5-m/s flowing sodium at 525°C with an applied stress of 225 MPa, creep-rupture lives in air ranged from 2204 to 3026 h. For tests conducted on the same heat of material in sodium at the same test temperature and applied stress conditions, creep-rupture times ranged from 1862 to 3523 h. Data for test times in sodium significantly longer than those obtained to date are required to determine the effects of sodium on the creep-rupture life of standard 9Cr-1Mo steel and T-91. However, because there are no current plans to use this alloy above 525°C because of its strength limitations, such tests are not considered to be high priority. Testing of a 12Cr-1MoNiV steel in static sodium at temperatures up to 500°C and for rupture times up to 8000 h had virtually no effect on creep-rupture strength or ductility.[110]

3.3.5.3 Fatigue

The low-cycle fatigue behavior of T-91 in flowing sodium at 538 and 593°C has been determined for material in the N&T condition and for material exposed to flowing sodium before testing.[111] For the tests conducted ($<2 \times 10^5$ cycles) the fatigue life (N_f) of T-91 in sodium is five to seven times greater than the fatigue life in air at 538 and 593°C. In addition, there is no temperature dependence of the fatigue life at these temperatures, and exposure to flowing sodium at 538°C for 6000 h before testing had no effect on this behavior.

These results are consistent with the observations by Chopra et al.,[104] thus indicating that exposure of T-91 to sodium results in nitrogen loss for temperatures above 550°C and no loss of nitrogen for temperatures below 550°C. The results are also consistent with the data from the ANL creep-rupture tests at 593°C on material exposed to sodium at 593°C. Exposure to sodium at 593°C is suggested to result in nitrogen loss from the alloy and an associated decrease in strength. No decrease in fatigue properties was observed for testing at 593°C because the tests are completed in <2 d and insignificant nitrogen loss, if any, occurs during this short time.

The observed reduction in creep-rupture life when tested in sodium at 538°C by WAESD is not consistent with the calculations of Chopra et al.,[104] especially because the material in the WAESD test was exposed to sodium for only 356 h. Little, if any, nitrogen loss could occur during this time at 538°C. As stated earlier, additional testing in sodium is required to quantitatively assess the effects of sodium on the elevated-temperature mechanical properties of T-91.

3.4 JOINING BEHAVIOR

3.4.1 Weldability

The weldability of modified 9Cr-1Mo steel has been investigated by numerous organizations. Studies have included the evaluation of several welding processes, the determination of preheat and PWHTs, cold- and hot-cracking susceptibility, and reheat cracking response; filler wire and electrode development; and the characterization of weldment mechanical properties.

One hundred fifty-seven weldments of modified 9Cr-1Mo steel have been prepared at ORNL. Of this total, 91 were produced with the GTA process, 32 with the shielded metal-arc (SMA) process, and 34 with the submerged arc (SA) process. The material thickness has ranged from 3 to 50 mm, and the product forms included tubing, plate, and pipe. Dissimilar metal weldments were made between modified 9Cr-1Mo and (1) standard 9Cr-1Mo, (2) 2.25Cr-1Mo, and (3) stainless steels of various compositions. The filler metals for producing these weldments have included (1) standard composition 9Cr-1Mo, (2) modified 9Cr-1Mo, (3) 2.25Cr-1Mo, and (4) ERNiCr-3. Extensive testing and mechanical property characterization have been conducted on the various weldments, and the results are included below.

The hot-cracking sensitivity of the modified 9Cr-1Mo alloy was evaluated in Tigamajig tests of 11 experimental heats.[112] The results of this work indicated that hot-cracking should not be a problem with the range of compositions being investigated, a fact proven to be true because no hot-cracking has been encountered during welding, either in the laboratory or by commercial fabricators.

The Y-groove restraint test was used to evaluate the relative weldability of the steel and to examine preheat temperature. In tests at ORNL, weld metal cracking was observed in test plates with no preheat and those preheated to 93°C. No cracking was observed in a test plate preheated to 204°C before welding. Grobner and Wada[113] reported a preheat level of 150°C was sufficient in this test to prevent cracking.

Lundin and coworkers[114] at UT conducted research on the welding metallurgy of the heat-affected zone (HAZ) of modified 9Cr-1Mo and other Cr-Mo steel alloy weldments. Stress-relief cracking (SRC) and hydrogen assisted cracking (HAC) susceptibility of the steels were assessed. The HAZ microstructure in the modified 9Cr alloy was concluded to be independent of cooling rates normally encountered during welding; the SRC susceptibility of the modified 9Cr-1Mo was concluded to be less than 2.25Cr-1Mo steel and HAC susceptibility about the same. No major weldability concerns from SRC or HAC were identified.

The numerous weldments produced and test results of the various evaluations have all indicated there are no major concerns with the weldability of modified 9Cr-1Mo steel.

3.4.2 Hardness

Hardness measurements were often used during the modified 9Cr-1Mo welding investigations to indicate changes in the microstructure and properties of the various weldment regions.[115] A typical microstructure produced during the welding of this N&T steel is shown in Fig. 102. The various regions shown in this figure are W (weld zone), TZ [transformed zone, which is the region exposed to temperatures above AC_1 (about 840°C)], and TMPZ (tempered zone, which is exposed to temperatures below the AC_1 and above the tempering temperature). Typical microhardness measurements across these regions show variations (Fig. 103) that were made on a SA weld in modified 9Cr-1Mo steel produced at Struthers Wells. This weldment was made with standard 9Cr-1Mo filler metal and received a PWHT of 732°C for 2 h before shipping to ORNL. Sections of this weldment were subsequently heat treated at the same temperature an additional 8, 18, 38, and 78 h to make total times of 10, 20, 40, and 80 h. For comparison, a SA weldment that was similar, except that the weld metal was of the modified 9Cr-1Mo composition, was made at ORNL.

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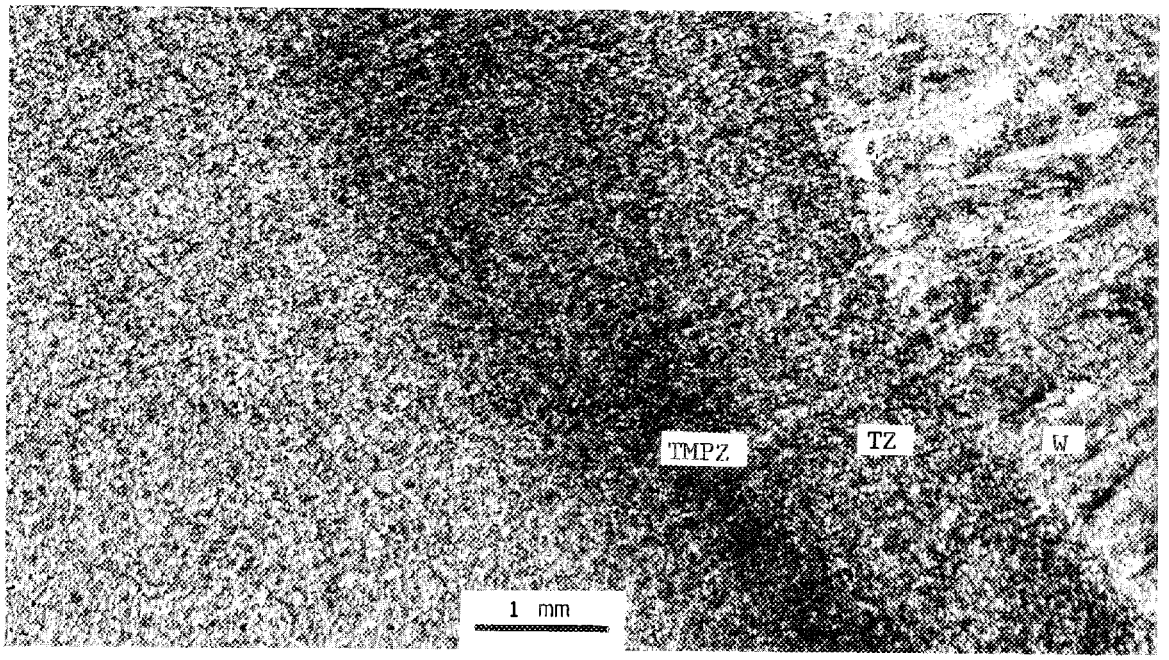

Fig. 102. Typical microstructure produced during welding of normalized and tempered Cr-Mo steels. Micrograph shows the weld (W) and two regions of the heat-affected zone: TZ (transformed zone), and TMPZ (tempered zone).

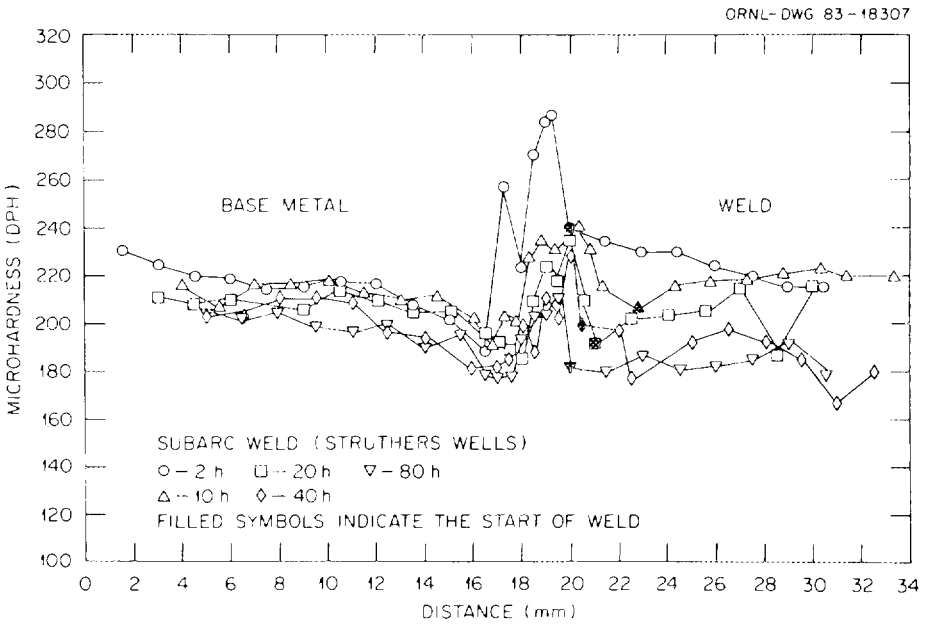

Fig. 103. Microhardness traverse across submerged arc weldment produced at Struthers Wells. Base plate was modified 9Cr-1Mo, and filler wire was standard 9Cr-1Mo. Postweld heat treatments were for 2, 10, 20, 40, and 80 h at 732°C.

Increasing the PWHT from 2 to 80 h produces changes in the microhardness of the various regions. The very stable base metal shows only a small decrease in hardness. The TMPZ shows slightly more decrease with increasing PWHT. The TZ, which was partially tempered at 732°C, tempers further with time. The W shows the greatest decrease in hardness with additional PWHT time. This observation suggests that if long PWHTs are necessary, the modified composition filler wire should be used because it remains harder than the base metal.

Microhardness traverses for weldments in modified 9Cr-1Mo, standard 9Cr-1Mo, 2.25Cr-1Mo, and HT9 (12Cr-1MoVW alloy) show that all of the alloys contain a TMPZ as a result of welding. Differences in microhardness between the base metal and TMPZ is about the same for modified 9Cr-1Mo and the other alloys. Because of its possible effect on the mechanical properties of modified 9Cr-1Mo steel, two solutions are proposed. Normalizing and tempering at 1040 and 760°C, respectively, after PWHT eliminates the hardness variations and offers a means of achieving uniform properties.[115] However, a NTT heat treatment after welding is not usually possible or practical, but it may be used for certain applications. We caution that NTT should only be applied to weldments made with the modified composition filler metal. Standard composition weld is softer than modified composition weld metal and tempers much faster. Therefore, to avoid overtempering that significantly weakens the weld metal, NTT should not be applied to standard composition weld metal.

A second solution is to select a lower tempering temperature. The overtempered region is believed to originate from heating of the base material to near the AC_1 during welding. To increase the initial hardness of the base material, hardness variations can be minimized by tempering below 760°C before welding. Results indicated that tempering at 621°C followed by PWHT at 760°C essentially eliminated the hardness variations across the weldment produced. However, at a PWHT of 732°C, the overtempered region remained.

3.4.3 Filler Wire Composition, Fluxes, and Electrodes

Weldability studies of the 9Cr-1Mo steels have included the GTA, SA, and SMA welding processes and have addressed filler materials that produced weld deposits of both standard and modified composition of 9Cr-1Mo.

At ORNL several hundred feet of weld have been made with a little over one-half being GTA welds and the remainder being SA and SMA welds. The majority of these welds were made to provide material for subsequent chemical, metallographic, or mechanical property evaluation, but in every case the general welding characteristics were also assessed from a user's standpoint. Essentially, no unusual weldability problems were encountered making GTA welds with 9Cr-1Mo materials, but this was not the case for SA or SMA welding.[115]

3.4.3.1 Gas Tungsten-Arc Welding

Typical data from tensile and Charpy V-notch impact testing of a GTA weld deposit are given in Tables 23 and 24. The Charpy V-notch absorbed energy values for the weld deposit are excellent, particularly in light of its high tensile strength. Based on past experience, this filler wire should have no significant problem meeting a minimum value of 68 J at room temperature when deposited with the GTA process and appropriately postweld heat treated.

Table 23. Data from all-weld-metal tensile test of United States Welding heat 21648

0.2% Yield strength, MPa	671
Ultimate tensile strength, MPa	803
Elongation in 50 mm, %	20
Reduction of area, %	65

Table 24. Data from all-weld-metal Charpy V-notch test of United States Welding heat 21648

Test temperature (°C)	Absorbed energy (J)	Lateral expansion (mm)	Fracture appearance (% shear)
−73	8	0.06	3
−46	8	0.14	5
−18	97	1.08	45
−1	73	0.96	27
10	104	1.23	50
23	217	2.10	100
66	235	2.32	100
121	191	1.93	100

Results of a welding procedure tensile test as specified in Sect. IX of the ASME Code are presented in Table 25. The specimen fractured in the base metal away from the HAZ. Side-bend specimens passed with no visible defects.

The tests for welding material certification and procedure qualification illustrate that when the GTA process is used, the weldability of 9Cr-1Mo steel is very good.

Table 25. Tensile data from procedure qualification of GTA weld made with United States Welding heat 21648

0.2% Yield strength, MPa	638
Ultimate tensile strength, MPa	768
Elongation in 50 mm, %	18
Reduction of area, %	67

3.4.3.2 Submerged-Arc Welding

Many SA welds have been made on 9Cr-1Mo modified plate using standard 9Cr-1Mo filler wire and OP-76, a neutral flux made by Oerlikon. This combination of materials has excellent weldability and mechanical properties. Because OP-76 behaved so well with standard 9Cr-1Mo filler wire, it was used to make the initial SA welds with the modified composition. However, the combination of OP-76 flux and modified 9Cr-1Mo wire resulted in significant problems. Fused OP-76 adheres tenaciously to SA welds made with modified filler wire so that each weld bead must be surface ground to properly descale the weldment. The same behavior was obtained for both direct- and alternating-current operation.

Problems were also encountered using Oerlikon INS-22 flux.[116] Of the fluxes evaluated by ORNL, the one showing the most promise of good weldability with a modified 9Cr-1Mo electrode was IND-24. Tensile data from an all-weld-metal specimen obtained at a nominal strain rate of 9.3×10^{-5}/s are given in Table 26. All of the tensile data exceed the minimums specified by the ASME Code. Data from Charpy V-notch impact testing of the SA weld deposit are given in Table 27. The absorbed energy values are very

Table 26. Data from all-weld-metal tensile test of SA deposit made with modified 9Cr-1Mo electrode and IND-24 flux

0.2% Yield strength, MPa	560
Ultimate tensile strength, MPa	693
Elongation in 50 mm, %	21
Reduction of area, %	64

Table 27. Data from all-weld-metal Charpy V-notch test of SA weld deposit

Test temperature (°C)	Absorbed energy (J)	Test temperature (°C)	Absorbed energy (J)
−46	8	21	29
−18	8	38	51
21	50	66	125
21	26	107	174
21	18	149	175
21	14	260	171

good, although the desired minimum of 68-J absorbed energy could not be met. Results of the weldment tensile test are presented in Table 28. The specimen fractured in the base metal away from the HAZ. Side-bend specimens passed the procedure qualification test with only one specimen showing a visible defect that probably was caused by a slag inclusion.

Table 28. Tensile data from procedure qualification of SA weld made with modified 9Cr-1Mo electrode and IND-24 flux

0.2% Yield strength, MPa	563
Ultimate tensile strength, MPa	677
Elongation in 50 mm, %	16
Reduction of area, %	66

Provided that a suitable flux is chosen, results to date suggest there are no major weldability problems associated with SA welding with modified 9Cr-1Mo wire. Even so, flux selection continues to be a concern. All of the fluxes evaluated thus far at ORNL have not performed very well when welding with ac. For welding thick components (over 25 mm), ac is sometimes preferred over dc because it provides much more stable operation in the presence of the magnetic fields generated during welding. In addition, a reproducible means of obtaining the desired minimum of 68-J absorbed energy at room temperature has not yet been identified.

3.4.3.3 Shielded Metal-Arc Welding

The development of coated electrodes for SMA welding the modified 9Cr-1Mo steel was conducted primarily at CE's Metallurgical and Materials Laboratory. Emphasis was on improving the Charpy V-notch impact strength of the SMA weld deposit while maintaining sufficient high-temperature strength and carbide stability. In the early phases of the program, 191 batches of electrodes were made from 127 different coating formulations to evaluate a wide combination of chemical compositions and fabrication conditions. These electrodes were formulated by introducing the alloying elements via the electrode coating. The results of this early work showed the effect of the various elements on the Charpy V-notch properties. Boron and zirconium were found to be very detrimental to impact strength. Tungsten had little noticeable effect on the impact strength of the weld metal, and 0.05% nitrogen added to the 9Cr-1Mo composition produced only a slight decrease in impact strength. Vanadium additions of 0.17% with 0.08% carbon gave an additional shift in the transition temperature. Niobium additions >0.05% were detrimental to the Charpy V-notch properties. These findings and others established a data base for the final development of an electrode with suitable Charpy V-notch properties.

Later work on electrode development used the modified 9Cr-1Mo alloy composition for the core wire. A weld rod coating formulation with no alloying elements in the coating was designed. The Charpy V-notch properties at room temperature are excellent for this electrode based on five tests that were above the 68-J goal.[115] Electrodes were also produced for other fabricators who requested specific compositions for their applications.

This work has shown that the weldability of modified 9Cr-1Mo is greatly improved by the use of alloyed core wire compared with electrodes that introduce alloying elements through the coating. There has been more consistent alloy transfer from the wire than from the coatings. Welding personnel have also reported improved clarity of the welding arc and better slag removal characteristics with the newer electrodes. A highly basic slag system was developed for the coating to give good control of the weld puddle and yet have low silicon in the weld deposit. Before a moisture resistant coating binder was used, slag fluidity and control were problems in electrodes with <0.35% silicon in the deposit. The current electrodes now have a typical silicon content of 0.15 to 0.25% in the undiluted weld metal.

Charpy-impact toughness of the SMA weld is influenced more by small changes in carbon and niobium than by other elements in the deposit. A carbon range of 0.090 to 0.125% and a niobium range of 0.018 to 0.030% have given the best room-temperature Charpy V-notch results. As carbon is decreased or niobium is increased, the toughness decreases in the SMA welds. A recommended weld deposit chemical analysis for modified 9Cr-1Mo SMA welding is listed in Table 29.

Table 29. Recommended chemical composition range for 9Cr-1Mo, V, and Nb SMA weld metal

Element	Composition (wt %)		Element	Composition (wt %)	
	Minimum	Maximum		Minimum	Maximum
C	0.090	0.125	Cr	8.0	9.5
Mn	0.35	1.0	Mo	0.85	1.05
P		0.02	V	0.18	0.25
S		0.01	Nb	0.018	0.030
Si	0.15	0.35	Cu		0.2
Ni		0.40	N	0.035	0.055

3.4.4 Commercial Welds

Significant contributions have been made to the welding development of modified 9Cr-1Mo steel by commercial fabricators. Their work has included both the production of weldments by various processes and some basic weldability investigations.

The AMAX Materials Research Center conducted Y-groove tests on standard and modified 9Cr-1Mo steel and compared the results with those of HT-9. A detailed microstructural analysis of weld metal, HAZ, and base metal of 2.25Cr-1Mo and HT-9 was also conducted and reported.[117] It was concluded that compared with the base metal, a region of slightly lower hardness in the HAZ develops in all four alloys. The lower hardness is believed to be caused by carbide coarsening due to overtempering of a narrow region.

Struthers Wells Corporation produced two SA weldments in 51-mm-thick plate and four SMA weldments in 25-mm-thick plates. The SA weldments were made with standard composition, ER505, 9Cr-1Mo filler wire and Oerlikon OP-76 flux. Standard composition 9Cr-1Mo electrodes were used for the SMA welds. A preheat of 288°C with a maximum interpass temperature of 400°C was used. Weldments from both processes received a PWHT of 2 h at 732°C. A plate of each weldment was subjected to testing and passed the requirements of Sect. IX of the ASME Code. The remainder of the weldments were used for mechanical property testing at ORNL.

Leighton Industries prepared GTA welds in 76-mm-OD by 13-mm-wall (3 × 0.5-in.) tubes of modified 9Cr-1Mo steel. The welds were made with modified 9Cr-1Mo filler wire and ERNiCr-3 wire. These welds were not PWHT and were used in long-term creep testing at ORNL. Leighton Industries also prepared several welds in 51-mm-OD (2-in.) tubes for installation in one of the American Electric Power's (AEP's) steam plants. These welds were between modified 9Cr-1Mo steel and modified 9Cr-1Mo or type 304H stainless steel. All of these tubes have been operating in AEP's steam plant since April 1981.

Rockwell International Corporation (RI) prepared 14 internal-bore tube-to-tube welds in 25-mm-OD (1-in.) tubing. Welds were examined in accordance with requirements of ASME Code, Sects. III and IX. RI concluded that quality welds are readily made in modified 9Cr-1Mo steel by the autogeneous GTA welding process.

CE made more than 100 tube welds. These included joints between modified 9Cr-1Mo and modified 9Cr-1Mo, standard 9Cr-1Mo, 2.25Cr-1Mo, type 321 stainless steel, and type 347 stainless steel. SMA electrodes were also developed at CE and have been produced in relatively large quantities. In addition, CE conducted creep tests on full-section tubular specimens.

CBI Industries (formerly known as Chicago Bridge & Iron Co.) produced a SA weld in 203-mm-thick plate of modified 9Cr-1Mo steel. This weldment was made with a 200°C preheat temperature and was welded with standard 9Cr-1Mo filler wire and Oerlikon OP-76 flux. A total of 145 passes were deposited to complete this weld, which received a 732°C PWHT for 6 h. This weldment was sectioned and distributed for testing at ORNL, CE, Georgia Institute of Technology, and the Central Electricity Generating Board (CEGB) in the United Kingdom.

Babcock and Wilcox Company (B&W), Nooter Corporation, and Agecroft (England) conducted welding and testing on modified 9Cr-1Mo tubing. Agecroft prepared welds between modified 9Cr-1Mo and 2.25Cr-1Mo tubing for installation at its steam plant.

The participation of these and other fabricators has been extremely valuable in bringing industrial expertise to the program. Additionally, this work has provided experience to the industry while producing the weldments needed for testing and characterization in the laboratory.

3.4.5 Weldment Properties

Extensive mechanical property testing — including tensile, creep, Charpy, fatigue, and creep-fatigue — has been conducted on weldments during the 9Cr-1Mo development program.

3.4.5.1 Tensile Properties

According to ASME rules, the qualification of 9Cr-1Mo welds requires that they meet the room-temperature minimum properties for the base metal. The base metal minimum values for yield and ultimate tensile strengths are 414 and 585 MPa. The minimum value of reduction of area is 55%. To determine if weldments meet the minimum strength properties, data were normalized by dividing weldment properties by base metal minimum values and plotting the ratio as a function of test temperature (Fig. 104). Unnormalized reductions of area values are plotted in Fig. 105. These figures show the following:

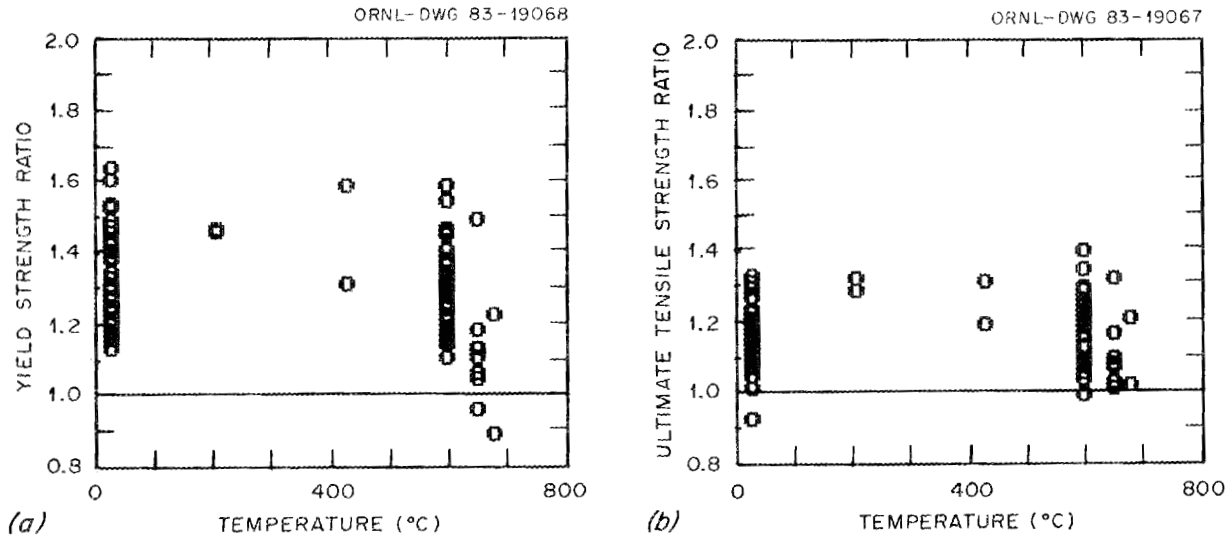

Fig. 104. Strength ratios (weldment-to-base metal minimum) for weldments of modified 9Cr-1Mo steel. Weldment data are for both standard and modified 9Cr-1Mo filler wire and for all three welding processes (gas tungsten-arc, shielded metal-arc, and submerged-arc). Unity line representing equal strength of base metal and weldment is included. (*a*) Yield strength, (*b*) ultimate tensile strength.

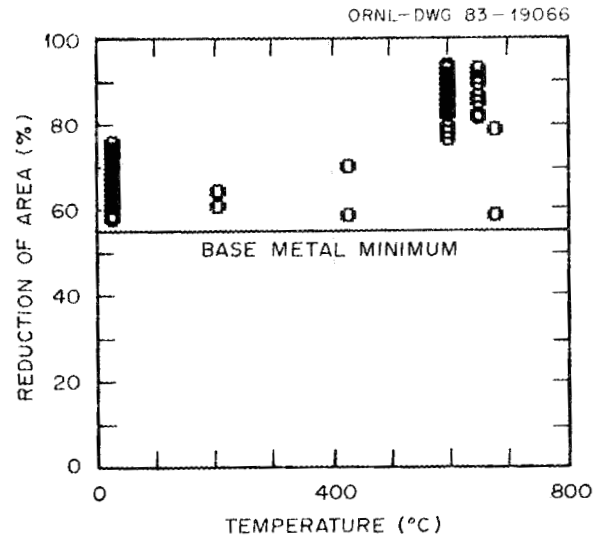

Fig. 105. Reduction of area as function of test temperature. Weldment data are for both standard and modified 9Cr-1Mo filler wire and for all three welding processes (gas tungsten-arc, shielded metal-arc, and submerged-arc). Base metal minimum value line of 55% at room temperature is included.

1. Yield strengths [Fig. 104(*a*)] of all weldments made by GTA, SMA, and SA processes exceed the minimum values for the base metal to 600°C. The only two points that fall below the minimum were data from tests at very high test temperatures.
2. Ultimate tensile strength [Fig. 104(*b*)] for all but two weldments meet or exceed the base metal minimum value.
3. Reductions of area values for all weldments exceed the minimum value of 55% for the base metal for all test temperatures.

Based on these results, modified 9Cr-1Mo weldments were concluded to have no problem meeting the base metal minimum properties.

3.4.5.2 Creep Properties

Creep data for all welding processes and test temperatures have been combined by normalizing to base metal average or minimum properties. Weldment stress ratios based on the base metal average are plotted in Fig. 106. The lines shown are the equal strength ratio and the line representing the basis for allowable stresses in *ASME Boiler and Pressure Vessel Code*, Sect. VIII. This figure shows that the basis used for determining the allowable stresses for ASME Code, Sect. VIII, is met by the creep-rupture data of modified 9Cr-1Mo weldments.

Figure 107 is a similar stress-rupture plot but is based on normalization by base metal minimum properties. An alternative allowable stress criterion for Sect. VIII (0.80 times minimum stress to rupture) and the S_t criteria for Code Case N-47 (0.67 times minimum stress to rupture) are also included in this figure. The criterion of 0.80 times minimum stress to rupture is met by the creep-rupture data of modified 9Cr-1Mo weldments. This figure also shows that the 0.67-times-minimum-stress-to-rupture basis for S_t values in ASME Code Case N-47 is quite conservative for the weldment data.

Figures 108 through 110 show the stress ratio plots (based on base metal average) for the GTA, SMA, and SA processes. These plots show that all three welding processes are equally satisfactory. The plots also show that the stress ratio was not affected by the use of either standard or modified filler wire.

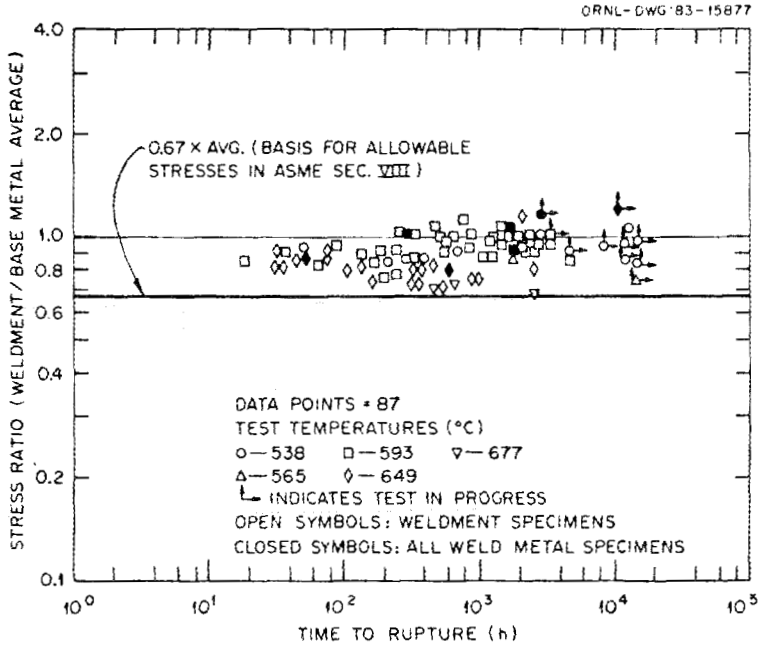

Fig. 106. Stress ratio (weldment-to-base metal average) as function of time to rupture. Weldment data are for both standard and modified 9Cr-1Mo filler wire and for all three welding processes (gas tungsten-arc, shielded metal-arc, and submerged-arc). Data are for test temperatures of 538, 565, 593, 649, and 677°C. Stress ratios of unity, representing equal strengths, and 0.67, representing the ASME Sect. VIII criteria for allowable stresses, are also included.

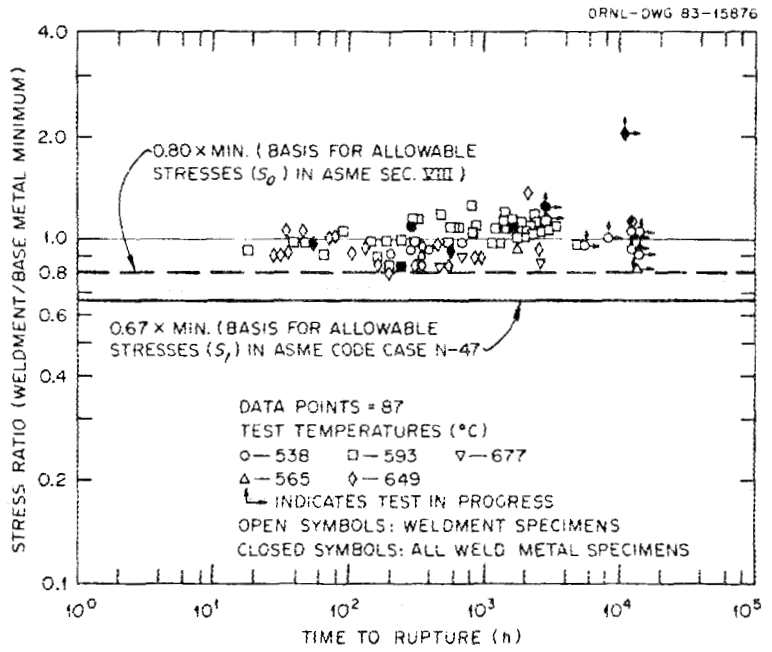

Fig. 107. Stress ratio (weldment-to-base metal minimum) as function of time to rupture. Weldment data are for both standard and modified 9Cr-1Mo filler wire and for all three welding processes (gas tungsten-arc, shielded metal-arc, and submerged-arc). Data are for test temperatures of 538, 565, 593, 649, and 677°C. Stress ratios of unity, representing equal strength; 0.80, criterion for allowable stresses in ASME Sect. VIII; and 0.67, the criterion for allowable stresses in ASME Code Case N-47, are also included.

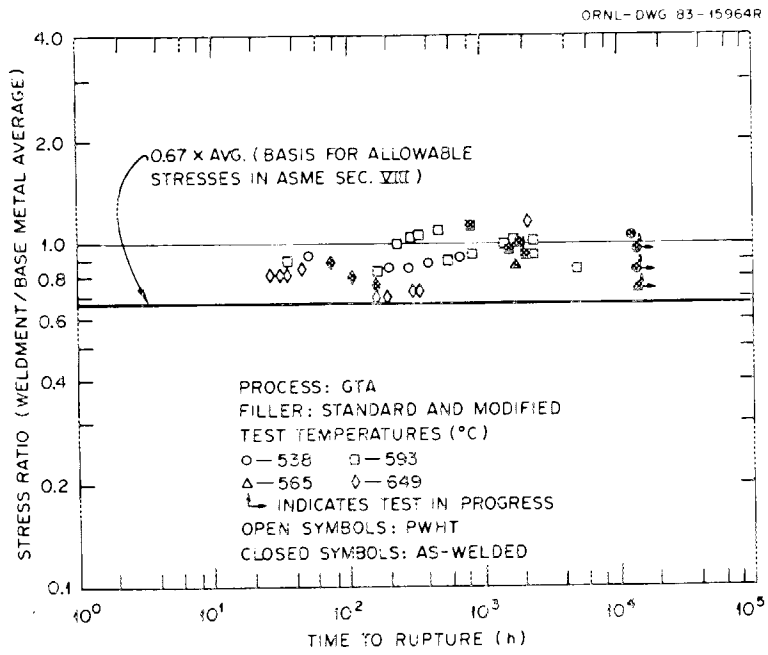

Fig. 108. Stress ratio (weldment-to-base metal average) as function of time to rupture for gas tungsten-arc weldments of both standard and modified 9Cr-1Mo filler wire compositions. Data are for test temperatures of 538, 565, 593, and 649°C. Stress ratios of unity, representing equal strength, and 0.67, the criterion for allowable stresses in ASME Sect. VIII, are also included.

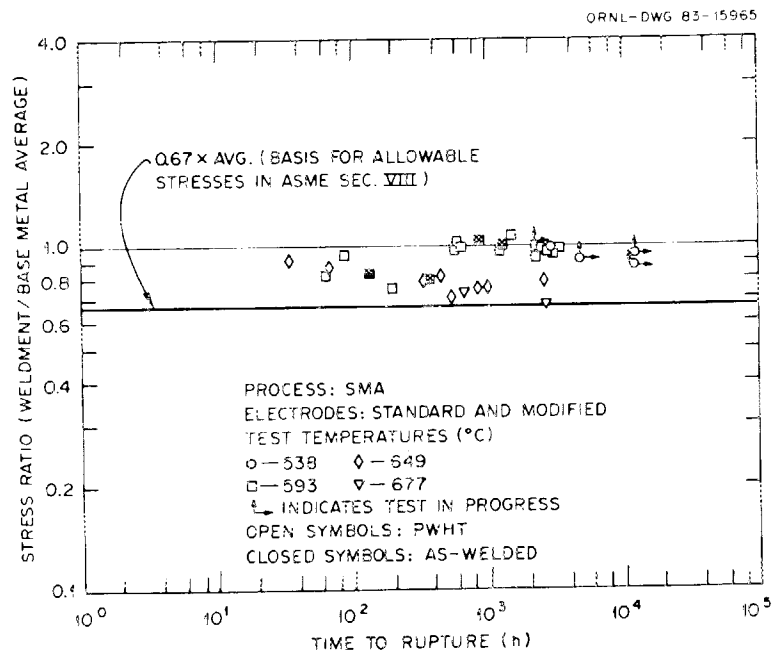

Fig. 109. Stress ratio (weldment-to-base metal average) as function of time to rupture for shielded metal-arc weldments of both standard and modified 9Cr-1Mo filler wire compositions. Data are for test temperatures of 538, 593, 649, and 677°C. Stress ratios of unity, representing equal strength, and 0.67, the criterion for allowable stresses in ASME Sect. VIII, are also included.

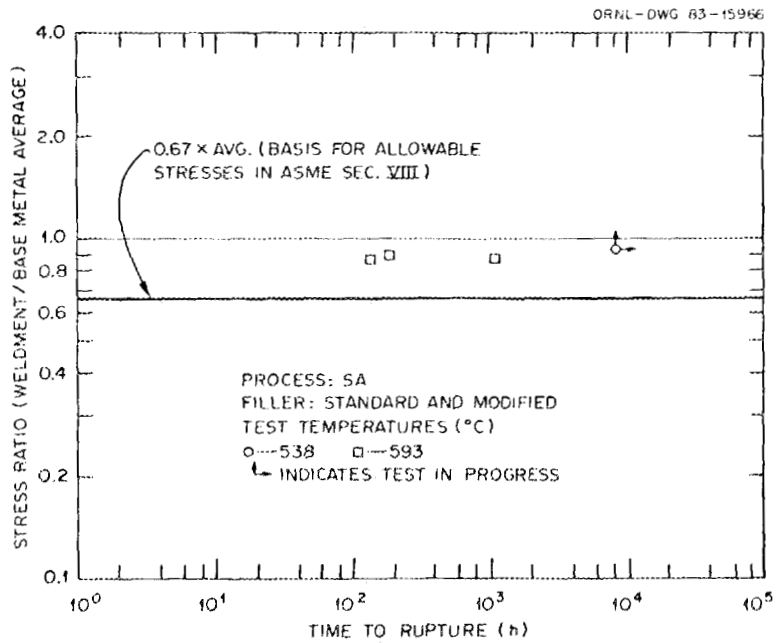

Fig. 110. Stress ratio (weldment-to-base metal average) as function of time to rupture for submerged-arc weldments of both standard and modified 9Cr-1Mo filler wire compositions. Data are for test temperatures of 538 and 593°C. Stress ratios of unity, representing equal strength, and 0.67, the criterion for allowable stresses in ASME Sect. VIII, are also included.

Data taken from Klueh and Canonico[118] show that if carbon in the filler wire is not specified, data for the 2.25Cr-1Mo weldment specimens can fall below the line of equal strength. Data on the weldment strength ratio of modified 9Cr-1Mo steel weldments appear to behave very similarly to those observed for 2.25Cr-1Mo steel weldments.

3.4.5.3 Charpy-Impact Data

Charpy-impact data on welds have been obtained for the standard and modified filler and electrode compositions and for various PWHT conditions. Figure 111 shows Charpy-impact data on GTA welds tested in the as-welded condition. Curves are drawn through the minimum and maximum energy data points. The energy curves show the range of transition temperatures at 68 J (50 ft-lb) to be from 25 to 125°C. After a 732°C PWHT, the transition temperature of GTA welds decreases to 50°C and below, and

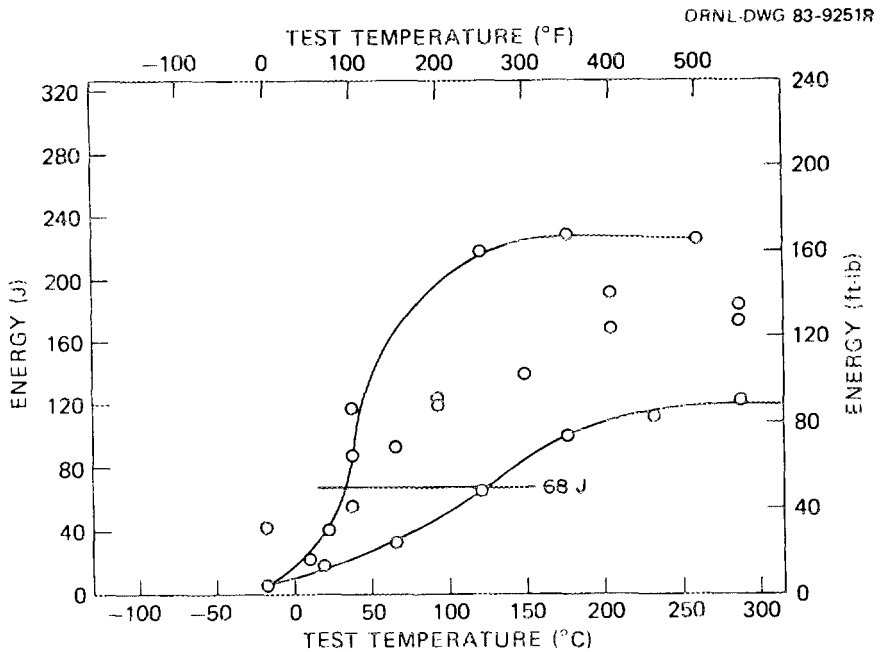

Fig. 111. Charpy impact data on gas tungsten-arc welds of modified 9Cr-1Mo. Filler wire was modified 9Cr-1Mo, and welds were not postweld heat treated. Curves through data show upper and lower bounds.

the USE increases above 160 J. Compared with modified 9Cr-1Mo filler wire, the standard wire gives ~40°C lower transition temperature but similar USEs.

The Charpy-impact energy curves for SMA welds of modified and standard composition are shown in Figs. 112 and 113, respectively. Comparison of these two figures shows that the Charpy-impact properties of the SMA weld with standard electrodes are significantly better than those with modified electrodes. The 68-J transition temperature for the standard electrodes is 50°C or below. The reasons for less toughness of modified electrodes compared with the standard electrodes are not clear.

The Charpy-impact energy curves for SA welds are shown in Fig. 114. The filler wire used was of standard composition. The 68-J transition temperature for these welds is 5°C or below, and the USE is at least 140 J. Figure 115 compares the Charpy-impact energy curve for the base metal, weld metal, and the HAZ for the SA weld made in 203-mm-thick plate of modified 9Cr-1Mo steel by CBI Industries. The 68-J transition temperature for the base, weld metal, and HAZ of this weld are −35, 10, and −50°C,

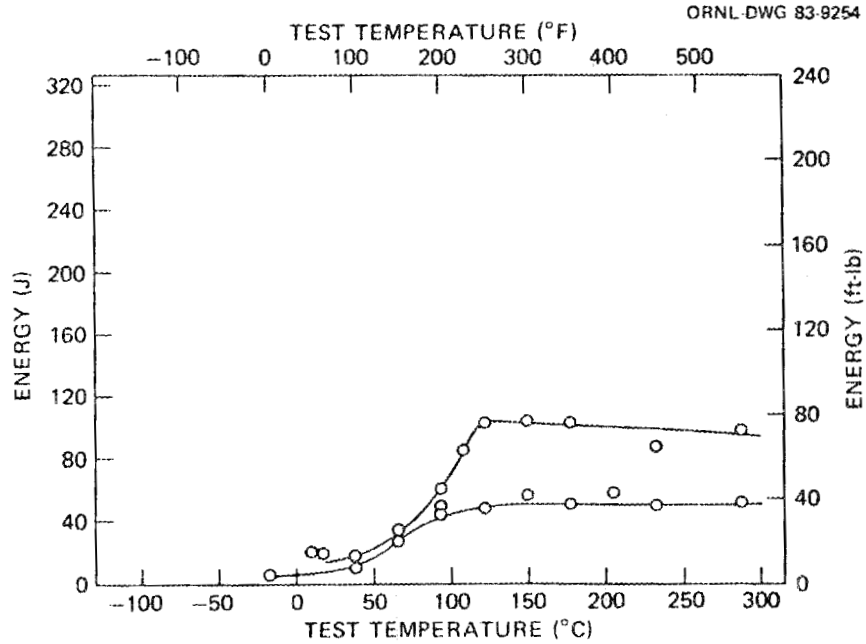

Fig. 112. Charpy impact data on shielded metal-arc welds of modified 9Cr-1Mo. Filler wire was modified 9Cr-1Mo, and welds were postweld heat treated 1 h at 732°C. Curves through data show upper and lower bounds.

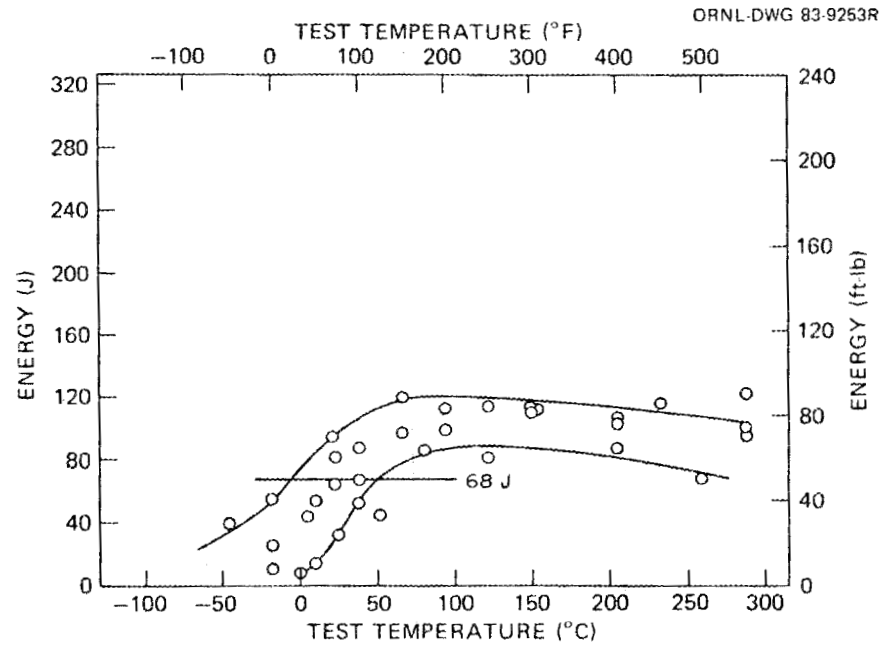

Fig. 113. Charpy impact data on shielded metal-arc welds of modified 9Cr-1Mo. Filler wire was standard 9Cr-1Mo, and welds were postweld heat treated 1 h at 732°C. Curves through data show upper and lower bounds.

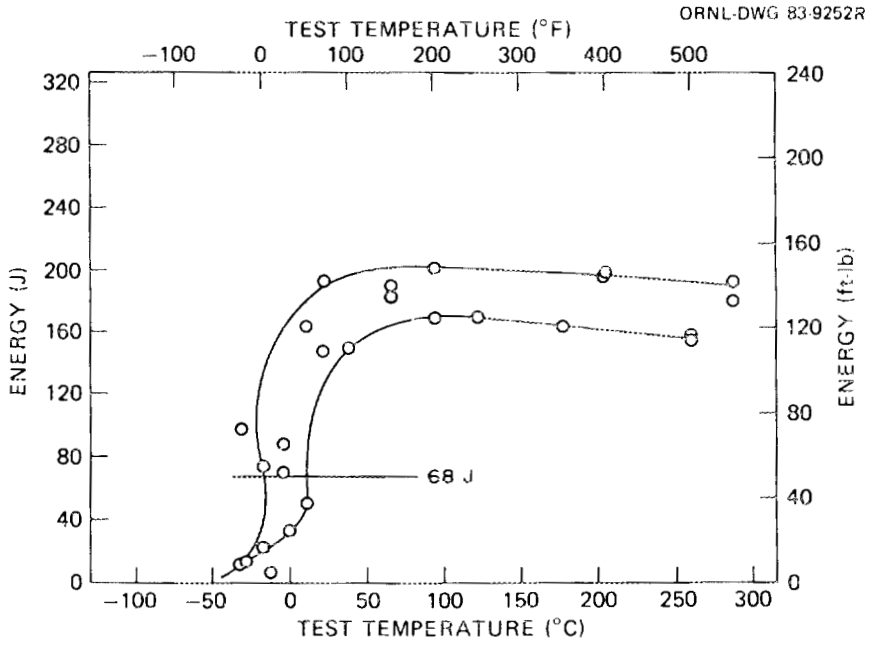

Fig. 114. Charpy impact data on submerged-arc welds of modified 9Cr-1Mo. Filler wire was standard 9Cr-1Mo, and welds were postweld heat treated 1 h at 732°C. Curves through data show upper and lower bounds.

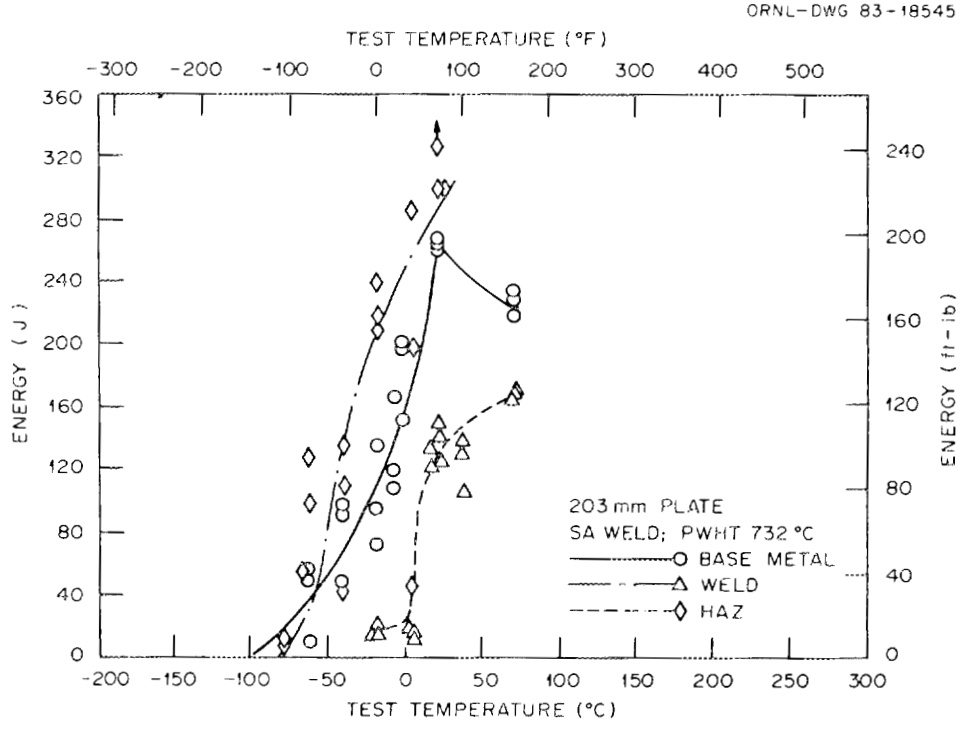

Fig. 115. Comparison of Charpy impact data of base metal, heat-affected zone, and weld metal of submerged-arc weld in 203-mm-thick plate of modified 9Cr-1Mo steel. Weld metal is of standard 9Cr-1Mo composition, and weld was postweld heat treated 6 h at 732°C.

respectively. Comparison of Figs. 113 and 114 shows that increasing the weld thickness from 50 to 203 mm does not affect the Charpy-impact properties significantly. Similar results would be expected if the weld thickness were scaled up for the other welding processes.

3.4.5.4 Fatigue and Creep-Fatigue

Low-cycle, continuous, fully reversed fatigue tests have been performed on uniform gage specimens taken from several positions within a weldment.[119] Tests were then run at two strain ranges (0.45 and 1.0%) and at temperatures of 25 and 593°C. The weldment, consisting of a "V" groove butt weld and made between two pieces of 203-mm-thick plate, was prepared using the SA process and standard composition 9Cr-1Mo weld filler metal. The weldment was tempered at 760°C for 4.5 h before specimen fabrication. Specimens were fabricated with their major axis transverse to the weldment so that the gage section (25.4 mm long and 6.35 mm diam) was either all base (wrought) material; all as-deposited weld metal; or a combination of both base, HAZ, and weld metal. Low-cycle fatigue life was highest for the base metal and lowest for the all-weld-metal specimens for all test conditions evaluated. The ratio of base metal life to weld material life was ~2:1. Specimens containing the HAZ and fusion line had fatigue lives intermediate between all base or weld metal specimens. All material showed cyclic softening with the base tending to soften more than the all-weld metal.

No results have been reported from creep-fatigue studies conducted on weldments or weld metal.

3.4.6 Thermal Aging Effects

The effects of long-time thermal aging on weldment properties have not been determined at this time. Weldments have been thermally aged to 10,000 h at 482 and 649°C, but no characterization has been completed. Most of the long-term thermal-aging-effects data will be obtained from the weldments currently in service in various fossil-fired steam plants and other applications.

3.4.7 Transition Joint Welds and Properties

Transition joint welds between modified 9Cr-1Mo steel and austenitic stainless steel were prepared for a pipe test in the Sodium Components Test Loop (SCTL) at the Energy Technology Engineering Center (ETEC). The test article consisted of modified 9Cr-1Mo steel pipe 232 mm in diam with a 12.7-mm wall thickness and a length of 610 mm. The pipe was safe-ended with type 304L stainless steel spool pieces 152 mm long on each end. The joint between modified 9Cr-1Mo and type 304 was made by GTA welding with ERNiCr-3 filler wire. The details of the fabrication and inspection have been reported.[120] No unacceptable defects were present. The 9Cr-1Mo spool piece was installed in the sodium loop and performed satisfactorily for the life of the test.

Mechanical property testing was performed on an additional weldment of the same materials produced by the identical procedures. Tensile testing was conducted on specimens from the 9Cr-1Mo pipe, the 304L stainless steel stock used for the safe-ends, and the dissimilar metal weldments. Yield and ultimate tensile strength data on modified 9Cr-1Mo/ERNiCr-3/304L, modified 9Cr-1Mo base metal, modified 9Cr-1Mo/ERNiCr-3/316, and modified 9Cr-1Mo/ERNiCr-3/304L aged for 2000 h at 510°C show the following:

1. Both 0.2% yield and ultimate tensile strengths of modified 9Cr-1Mo/ERNiCr-3/304L and modified 9Cr-1Mo/ERNiCr-3/316 stainless steel joints were lower than the corresponding values for the 9Cr-1Mo base metal over the entire test temperature range. However, the failures were generally in the stainless steel base metal.
2. Thermal aging for 2000 h at 510°C produced no change in the yield and ultimate tensile strengths of modified 9Cr-1Mo/ERNiCr-3/304L specimens.
3. The total elongation values of weldment specimens were generally higher than those of the modified 9Cr-1Mo steel base metal.

Creep data on transition joints of modified 9Cr-1Mo/ERNiCr-3/304L and modified 9Cr-1Mo/ERNiCr-3/316 are presented in Table 30. These data were compared with modified 9Cr-1Mo base metal data. At 510°C rupture life of

Table 30. Creep data on modified 9Cr-1Mo steel transition joint specimens with ERNiCr-3 weld metal

Test	Temperature (°C)	Stress (MPa)	Rupture time (h)
Modified 9Cr-1Mo/ERNiCr-3/304L			
23718	510	276	8,046
23756	593	172	1,368
23769	593	124	5,014
23733	593	97	14,042
23759	649	76	1,092
Modified 9Cr-1Mo/ERNiCr-3/316			
23762	593	172	838
23684	593	145	2,304

transition joint specimens equals the average rupture life of modified 9Cr-1Mo base metal. At 593 and 649°C the rupture life of the transition joints equals or exceeds the rupture life obtained from a plot of stress to rupture vs the average time to rupture minus twice the standard error of estimate (SEE) for the base metal.[120] Also, a comparison of creep data on modified 9Cr-1Mo transition joints with similar data for 2.25Cr-1Mo/ERNiCr-3 joints showed that rupture life of modified 9Cr-1Mo/ERNICr-3 joints is at least an order of magnitude longer than 2.25Cr-1Mo/ERNICr-3 joints and that creep rupture strength is double.[120] Additional long-term creep tests should be performed to validate these observations.

3.4.8 Needs

The GTA, SA, and SMA processes have been used to join a variety of product forms and thicknesses of modified 9Cr-1Mo steel, and welding has been performed satisfactorily by numerous fabricators. However, there are two areas that should be addressed in the future. The first is to characterize the effects of long-term thermal aging on weldment properties. Material currently in service may provide this information if proper evaluations are conducted. The second area in which further work is needed is

the selection of a flux for use in SA welding with alternating current, which is compatible with the modified 9Cr-1Mo filler composition. Slag removal from the weld bead and low Charpy-impact properties have been problems with commercial fluxes tried to date.

3.5 STRUCTURAL DESIGN METHODOLOGY

LMR components present unique structural design requirements. First, because of the good heat transfer characteristics of the coolant and the relatively large temperature rise through the reactor core, reactor power changes can result in significant thermal transient loadings on system components. Second, the components operate at temperatures where creep effects and time-dependent failure mechanisms are significant. For the austenitic stainless and ferritic steels used in LMRs, these effects are implicitly assumed by the *ASME Boiler and Pressure Vessel Code*, Sect. III, to occur upon significant exposure to temperatures above 427°C (800°F) and 371°C (700°F), respectively.[121]

In the late 1960s it was recognized in the United States that the low-temperature structural design methodology developed and used for light-water reactors would not be adequate for LMR systems. Consequently, an effort was mounted to develop a high-temperature structural design methodology that explicitly accounts for the effects of nonlinear material deformation and time-dependent damage mechanisms and failure modes. That methodology is now reasonably well established for the austenitic stainless steels and, to a somewhat lesser extent, for 2.25Cr-1Mo steel. Criteria for guarding against structural failures have been developed and, for Class 1 nuclear components, are given in ASME high-temperature Code Case N-47.[122] Likewise, inelastic design analysis methods, as well as some simplified methods based on elastic calculations, have been developed, and these are specified in DOE design guidelines.[123] Although known shortcomings and needs remain,[124] the methodology has been used in the United States in the design of the FFTF and the Clinch River Breeder Reactor Plant and in the development of high-temperature components and facilities. The task with modified 9Cr-1Mo steel is to adopt, to the extent possible, this existing methodology, while at the same time recognizing and accounting for the unique behavioral features of the alloy.

As background, it will be useful at this point to review the potential structural effects that a typical repeated LMR thermal transient loading can have on a simple component – a straight pipe. The situation is depicted in Fig. 116. Assume that the sodium temperature drops 167°C (300°F) in several seconds (line a–b in *b*), rises slowly (line b–c in *b*), and then is constant for several hundred hours of normal operation (line c–d in *b*); then the cycle is repeated. The predicted response is shown in Fig. 116(*c*). Starting with point a, the inner surface of the pipe first yields in tension as it contracts; as the outer surface subsequently begins to cool, however, the inner surface stress reverses sign and goes into compression. At b, the wall is uniformly at the lower temperature; slow heating from b to c causes the residual stress to decrease because of the decreasing yield condition with increasing temperature. At c, a compressive residual stress remains that relaxes during the subsequent hold period. The response to subsequent cycles is depicted in the figure by dashed lines.

The behavior shown in Fig. 116 illustrates three key potential LMR failure modes. First, creep-rupture damage is accumulated during the hold

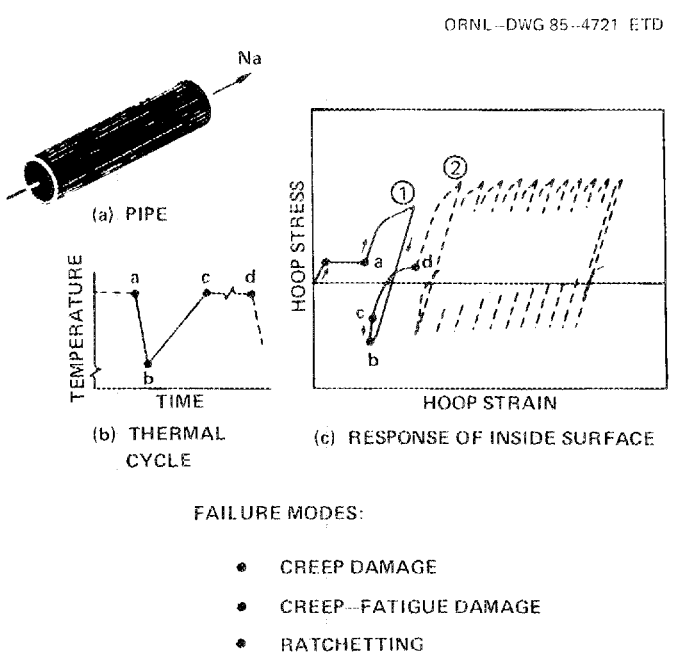

Fig. 116. Schematic representation of effects of repeated thermal transient loadings.

periods; second, the plastic cycles introduce fatigue damage that interacts with the creep to produce a creep-fatigue interaction damage mode; and third, ratchetting occurs, introducing the potential for failure as a result of excessive deformation. Thus, the LMR structural designer must have criteria that guard against failure by time-dependent cracking or rupture and by excessive deformation, and he must be assured that enough residual strength remains at the end of a 40- or 50-year life to withstand an accident. Because testing cannot be carried out for the 50-year life of a plant, the designer must rely on design by analysis, which requires that analysis methods be available for predicting, or conservatively estimating, the inelastic response. It is the task of the high-temperature structural design technology to provide valid design analysis methods, rules, and criteria.

Plant failure experience indicates that concerns about the damaging effects of thermal transient loadings and creep are justified. Just two examples will be mentioned here. The first one is the French Phenix reactor that, although extremely successful overall, was initially plagued with several problems of structural cracking.[125,126] These included cracking at welds joining piping tees and valves to piping runs and cracking in intermediate heat exchanger (IHX) welds. In each case, the cause was identified as fatigue or creep-fatigue caused by the repeated thermal transient loadings. It has been stated that these problems "cost the power station many months of unavailability and constituted a considerable loss of earning power."[126]

A second example is the Eddystone coal-fired generating plant operated by Philadelphia Electric Company in the United States. Unit 1 of that station went into operation in 1960 and has operated for most of the intervening period with steam temperatures of ~621°C (1150°F). In 1964 cracking was detected in the type 316 stainless steel main steam line junction headers, and in 1982-1983 extensive and widespread cracking was detected in the main steam lines themselves, which were also of 316 stainless steel.[127] The plant was shut down, and much of the steam piping has been replaced. Again, the cause was diagnosed as classical creep-fatigue (dominated by creep damage) as a result of repeated thermal transient loadings (exactly the situation that was illustrated in Fig. 116).

ASME Code Case N-47 was developed with the failure modes illustrated by Fig. 116 in mind. The Case has two general categories of limits. The first consists of primary stress limits placed on elastically calculated load-controlled quantities, such as the internal pressure stresses. The second consists of strain, creep-fatigue, and buckling limits. In meeting these latter limits, either inelastic analysis predictions or conservative estimates based on elastic analysis predictions must be used.

The primary stress limits are based on, in addition to the time-independent yield and ultimate strengths used at low temperatures, the minimum stress to cause rupture, the minimum stress to initiate tertiary creep, and the stress to produce 1% strain, all in a specified time t. The resulting allowable stresses (S_{mt} values) that are provided for types 304 and 316 stainless steels, 2.25Cr-1Mo steel, and alloy 800H are, thus, time and temperature dependent; the allowable loads depend on the duration of the loading.

The strain limits in Code Case N-47 are not necessarily related quantitatively to specific failure modes. Rather, they are aimed primarily at limiting the accumulation of ratchetting strains and ensuring the applicability of the creep-fatigue assessment procedures. A number of very conservative checks, using elastically calculated quantities, are provided for satisfying the strain limits and avoiding inelastic analysis when the loading conditions are not too severe.

The creep-fatigue rules in Code Case N-47 are, by far, the most frequently limiting criteria, and they are perhaps also the most questionable. Along with effective quantities, a linear time- and cycle-fraction damage accumulation rule, expressed as

$$\sum_j (n/N_d)_j + \sum_k (\Delta t/T_d)_k \leq D \ ,$$

is used by the Code to represent multiaxial stresses and strains. The time fraction represents the time Δt at a given condition (stress and temperature) divided by the allowable time at that condition. The cycle fraction represents the number of cycles n at a given condition (strain range and temperature) divided by the allowable number of cycles at that

condition.* The quantity D, which is ≤1, comes from an interaction diagram that is experimentally determined from creep-fatigue tests. Both the quantities N_d, which comes from a design fatigue curve, and T_D, which comes from a minimum creep-rupture curve, have design margins built into them. Although inelastic analyses are generally required for a creep-fatigue evaluation, procedures are given for estimating the creep and fatigue damage from elastically calculated quantities; use of these elastic analysis procedures is limited, however, to relatively mild loading conditions.

The background provided in the previous paragraphs will help in understanding the scope of the structural design methodology task for modified 9Cr-1Mo steel. The existing methodology is the result of years of development, focused primarily on types 304 and 316 stainless steel. Extensive uniaxial and multiaxial exploratory materials tests were carried out as a basis for developing (1) failure criteria and (2) constitutive equations (mathematical descriptions of the inelastic response of a structural alloy to complex time-varying loadings) and inelastic design analysis guidelines. Numerous high-temperature structural tests were performed to confirm the validity of the criteria and analysis methods for LMR-type loadings.

For 2.25Cr-1Mo steel a smaller but still significant structural design methodology development effort was required. The approach was to adopt, to the extent possible, the methodology framework developed for the austenitic stainless steels, modifying it only where required by unique behavioral features of 2.25Cr-1Mo steel. This same approach is being followed for modified 9Cr-1Mo steel, and while the full development effort has only recently gotten under way, significant progress has been made.

The remaining subsections summarize what has been accomplished, present key results, and outline what yet remains to be done to establish a viable high-temperature structural design methodology for modified 9Cr-1Mo steel. In the following subsection, the results of an early analytical

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*Recently, the Code Subgroup on Elevated Temperature Structural Design has approved the adoption of weldment reduction factors for both the primary stress limits and the N_d and T_j values in the creep-fatigue rules of Case N-47.

assessment of the performance characteristics of modified 9Cr-1Mo steel relative to those of other alloys in a typical sodium reactor thermal transient environment are presented. This early assessment established the inherent superior resistance of modified 9Cr-1Mo steel to such loadings. Following that is a description of the progress made toward developing constitutive equations for modified 9Cr-1Mo steel. The same is then done for failure behavior modeling. The design criteria anticipated to be appropriate for modified 9Cr-1Mo steel are described, and, finally, the key role that a few confirmatory structural tests must play is discussed.

3.5.1 Early Evaluation of Modified 9Cr-1Mo Steel in LMR Environment

An analytical assessment of the performance characteristics of modified 9Cr-1Mo steel relative to those of other alloys in a typical sodium reactor thermal transient environment was performed relatively early in the 9Cr-1Mo steel alloy development program.[128] Although the study was based on sparse preliminary materials properties data and was carried out before a thorough assessment of the design methodology that would be required for modified 9Cr-1Mo steel, the results do illustrate, at least qualitatively, the inherent superiority of 9Cr-1Mo steel. Thus, they are summarized here.

The following alloys were considered in the study:

1. type 304 stainless steel,
2. 2.25Cr-1Mo steel,
3. alloy 800,
4. standard commercial 9Cr-1Mo (England), and
5. U.S. modified 9Cr-1Mo steel.

In each case, a pipe with a mean diameter of 406 mm (16 in.) was subjected to an internal pressure of 3.78 MPa (548 psi).* The internal sodium temperature histogram is shown in Fig. 117. The pipe wall thickness for each alloy was selected based on limiting the primary stress to the S_{mt} allowable corresponding to 100,000 h. The S_{mt} values and wall thicknesses used are given in Table 31.

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*The pressure chosen is high for an LMR, but because other primary loads were ignored, the resulting primary stresses are typical.

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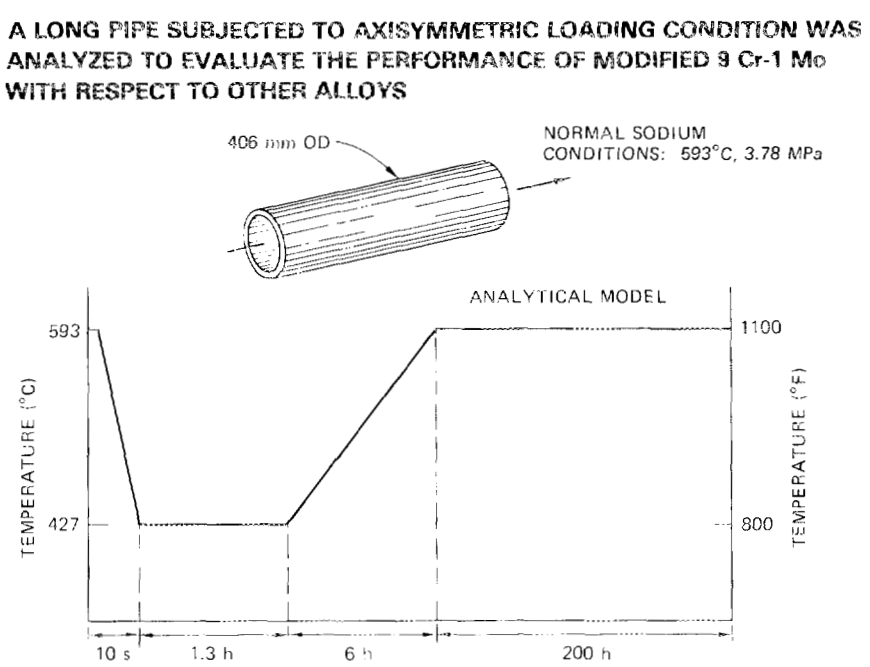

Fig. 117. Temperature histogram.

Table 31. The S_{mt} value for 100,000 h at 593°C (1100°F) and wall thickness for five alloys

Material	S_{mt}		Wall thickness	
	(MPa)	(ksi)	(mm)	(in.)
Alloy 800H	80.7	11.7	9.5	0.37
Modified 9Cr-1Mo	57.2	8.3	13.4	0.53
Type 304 stainless steel	46.9	6.8	16.4	0.65
2.25Cr-1Mo	22.8	3.3	33.8	1.3
British 9Cr-1Mo	18.4	2.7	41.7	1.6

In each case, a detailed inelastic design-type analysis was carried out. Thirty cycles of the histogram were analyzed.

The data for the modified and British steels were put in rational form for use in the analysis. Fatigue and creep-rupture curves were developed by using the ASME Code procedures. Examination of exploratory uniaxial cyclic stress-strain curves for modified 9Cr-1Mo steel showed that bilinear stress-strain relations in the constitutive equations that

are currently in use for type 304 stainless steel and 2.25Cr-1Mo steel are not amenable to use for modified 9Cr-1Mo steel. Therefore, a nonlinear plasticity model[129] was applied to the modified 9Cr-1Mo and 9Cr-1Mo (standard British) steels. The data for alloy 800 were obtained from Yang and Palmer[130] and ASME Code Case N-47. The data for type 304 stainless steel and 2.25Cr-1Mo steel were obtained from the *Nuclear Systems Materials Handbook* (NSMH)[131] and ASME Code Case N-47. The fatigue data for the 2.25Cr-1Mo steel were obtained from Booker et al.[132,133]

The material properties for both the modified 9Cr-1Mo steel and the British commercial 9Cr-1Mo steel were supplied by Booker and Sikka.[134] The same elastic and physical properties were used for both 9Cr-1Mo steels, but the yield stress for the British commercial steel was 15% lower than for the modified steel.

The analyses used the pipe wall model in the inelastic finite-element analysis computer program PLACRE.[135] The constitutive equations specified in the 1974 version of RDT standard F 9-5T (ref. 136) were used, except that the so-called α-reset procedure was introduced (see ref. 123) and, as previously mentioned, a nonlinear plasticity model, rather than the usual bilinear one, was used for the 9Cr-1Mo steels.

Results of the analyses are compared in Figs. 118 through 120. The cumulative ratchet strain is plotted against time in Fig. 118. Modified 9Cr-1Mo steel exhibited less thermal ratchetting than any of the materials except the British 9Cr-1Mo. The cumulative creep damage vs time is plotted in Fig. 119. The modified 9Cr-1Mo accumulated the least creep damage of any of the alloys. The cumulative fatigue damage vs time is plotted in Fig. 120. Again, the modified 9Cr-1Mo steel proved to be the superior alloy.

There are several reasons for the good showing of modified 9Cr-1Mo steel. Primarily, however, the reason is that the relatively high strength (yield and creep) of modified 9Cr-1Mo is coupled with the relatively low coefficients of thermal expansion and the high thermal conductivities of the ferritic steels. As a result, the required wall thickness of modified 9Cr-1Mo steel is relatively thin, the thermal loads are relatively small, and the resistance to creep damage is relatively high. In terms of LMR plants, the inherent resistance of modified 9Cr-1Mo steel to thermal expansion loads means substantial capital cost savings.

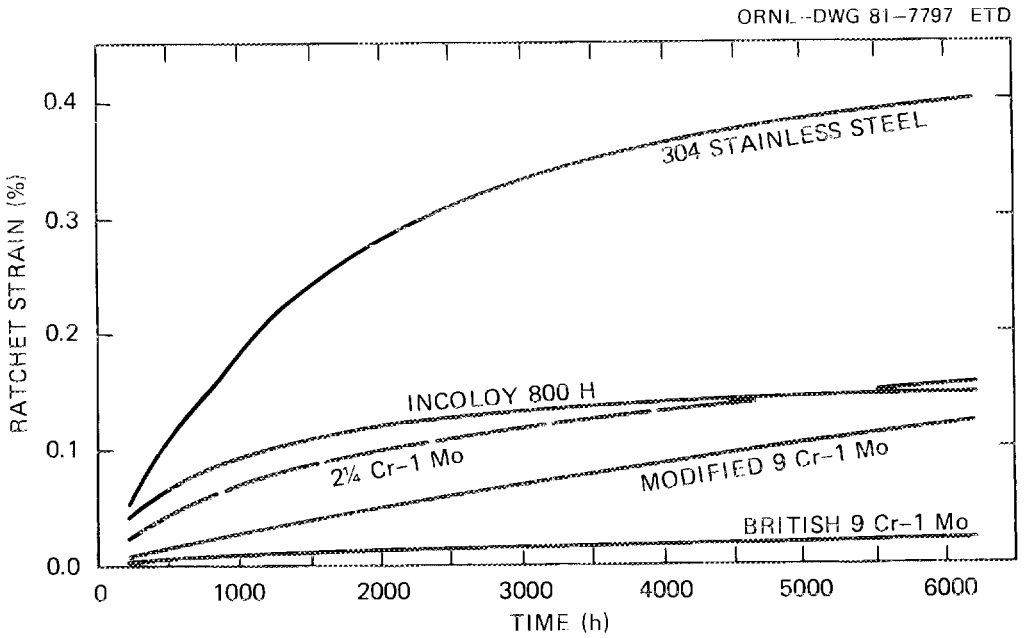

Fig. 118. Cumulative ratchet strain.

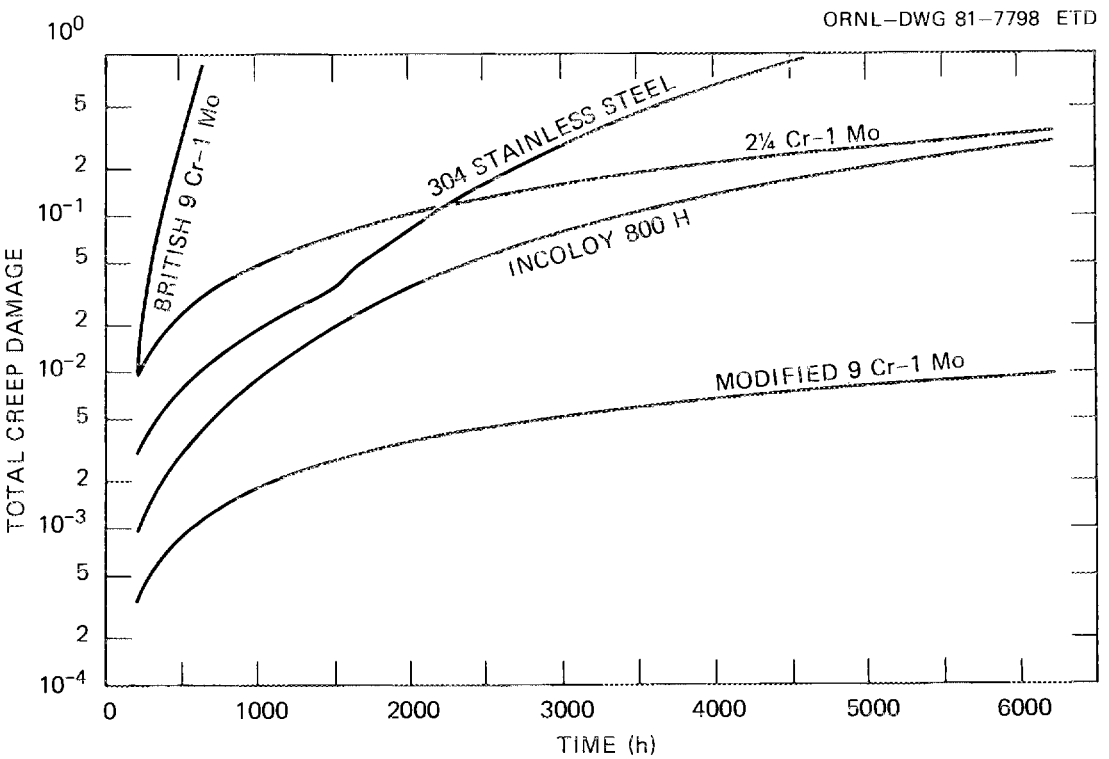

Fig. 119. Cumulative creep damage.

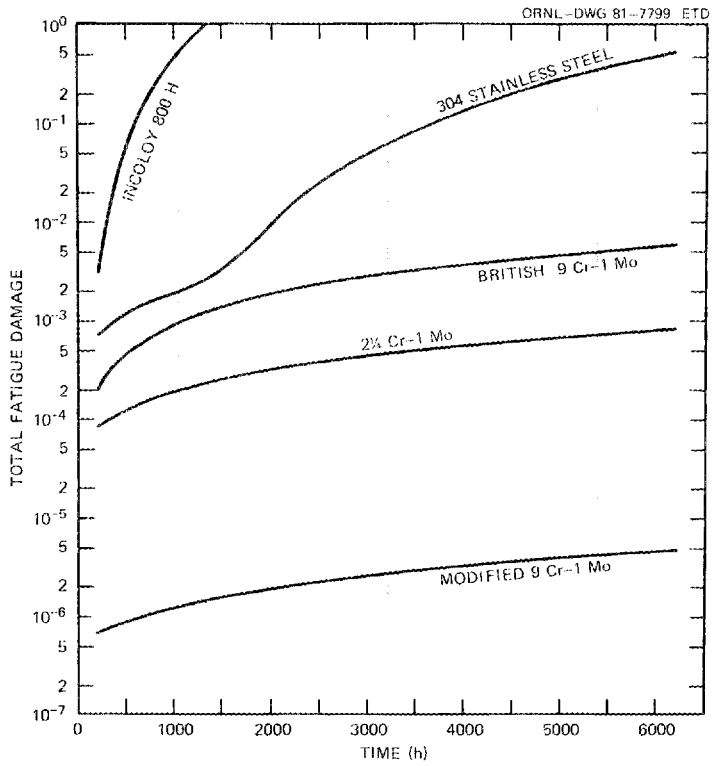

Fig. 120. Cumulative fatigue damage.

3.5.2 Constitutive Behavior Modeling

In the mechanics of continuous media, material behavior is described by constitutive equations or mathematical relationships between components of stress and strain. Depending on the material behavior being modeled, such equations often involve temperature and time or time derivatives. The nuclear standard NE F 9-5T (ref. 122) currently provides constitutive equations for 304 and 316 stainless steel and 2.25Cr-1Mo steel. These equations were developed for use in detailed inelastic analyses of structural components for nuclear reactors designed for operation at elevated temperatures where material behavior is significantly time dependent. Using the finite-element method, many detailed inelastic analyses have been performed for actual component geometries under expected service conditions.

The cost of detailed inelastic analyses is such that various simplified methods of analysis are used wherever possible in routine design evaluations. These analysis methods are less realistic and are, therefore,

used with more-conservative design margins. The constitutive equations in NE F 9-5T have played a role in the development of simplified analysis methods. Many detailed inelastic analyses have been performed on representative component geometries under conditions generally more severe than those expected in service. The results of these analyses were used to verify the adequacy of simplified methods of analysis.

The constitutive equations in NE F 9-5T are based on classical concepts of time-independent plasticity and time-dependent creep. Separate equations are provided for plastic and creep strains, although on the microscale there may be no physical basis for making such a distinction. There is some coupling between the equations to partially account for certain aspects of material behavior that, in the context of this approach, is referred to as creep-plasticity interactions.

To assess the validity of the current NE F 9-5T approach for modeling the constitutive behavior of modified 9Cr-1Mo steel, a series of exploratory tests was conducted on material in the postweld heat treated condition. As described by Swindeman,[137-141] a considerable number of tests were conducted under uniaxial stress to provide information on:

1. effects of strain rate and temperature on the yield strength and flow stress,
2. effects of strain rate and temperature on the cyclic hardening response,
3. effects of accumulated time or strain on the constant-stress creep behavior at several temperatures,
4. effects of simple changes in temperature or stress on the subsequent creep behavior,
5. effects of cyclic and reversed stresses on the creep responses,
6. behavior under creep-recovery conditions,
7. behavior under relaxation conditions,
8. behavior under nonisothermal plasticity conditions,
9. effect of plastic strain pulses on relaxation, and
10. behavior under interspersed creep-plasticity conditions.

Two tests were also conducted at room temperature under combined axial and torsional stress to characterize the multiaxial yield and hardening

behavior. As described by Ellis,[142,143] four biaxial yield loci were determined under proportional stress conditions and five under nonproportional conditions.

Examples are presented in the next subsection of the results of the exploratory uniaxial and biaxial tests. Based on these test results, the adaptability of the current NE F 9-5T constitutive framework to modified 9Cr-1Mo steel was assessed and found to be inadequate. The results of that assessment are summarized in the following subsection. A nonlinear viscoplastic model, unified in the sense that no distinction is made between creep and plasticity, was concluded to be required for this material. The final subsection provides initial results of an effort to develop such a model.

3.5.2.1 Exploratory Test Results

The influence of temperature and strain rate on monotonic, uniaxial stress-strain behavior[141] of modified 9Cr-1Mo is illustrated in Figs. 121 through 123. Relative to constitutive behavior modeling, three features of the stress-strain behavior at 450°C (842°F) and above are noteworthy. Compared with current NE F 9-5T materials, the influence of strain rate is high, the ratio of yield strength to ultimate strength is high, and the strain to ultimate strength is low.

Cyclic stress-strain behavior[144] of modified 9Cr-1Mo is illustrated in Fig. 124. The influence of strain rate is again pronounced, and cyclic softening is exhibited. In contrast, 304 and 316 stainless steels cyclically harden, and 2.25Cr-1Mo steel first hardens and then softens, with the hardening generally greater than the softening.

Constant load, strain-time behavior[144] is illustrated in Fig. 125. Relatively large changes in strain rate take place during the primary stage of creep, and although not obvious in log-log plots such as Fig. 125, the secondary stage of creep begins at rather low strains. A number of tests were conducted with stepwise changing stresses. The results of one such test[142] are shown in Fig. 126.

A few creep tests in which the load was cycled at regular intervals were performed. The results of two such tests[141] are compared with a constant load test in Fig. 127. As indicated in the figure, the stress was ramped from 276 MPa (40 ksi) to −276 MPa in 10 s and immediately ramped

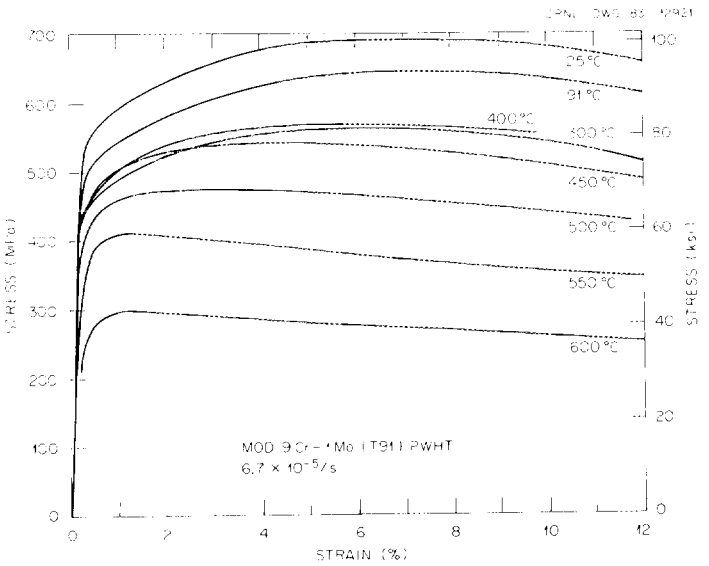

Fig. 121. Influence of temperature on stress-strain behavior.

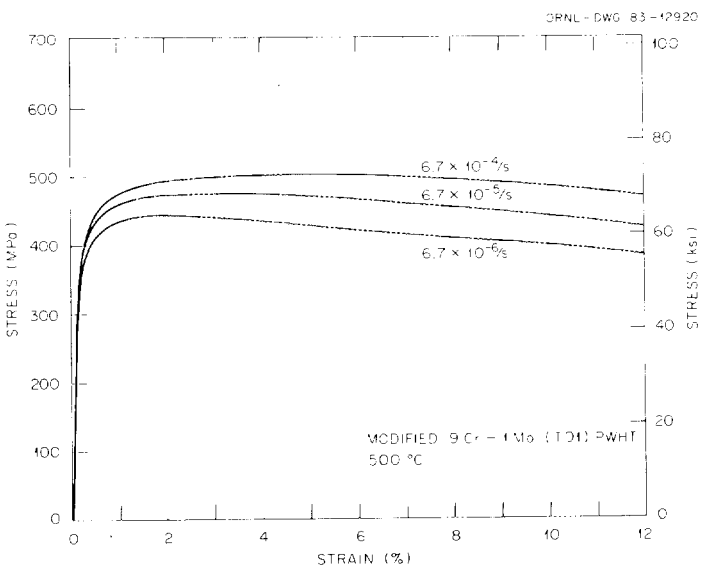

Fig. 122. Influence of strain rate on stress-strain behavior at 500°C (932°F).

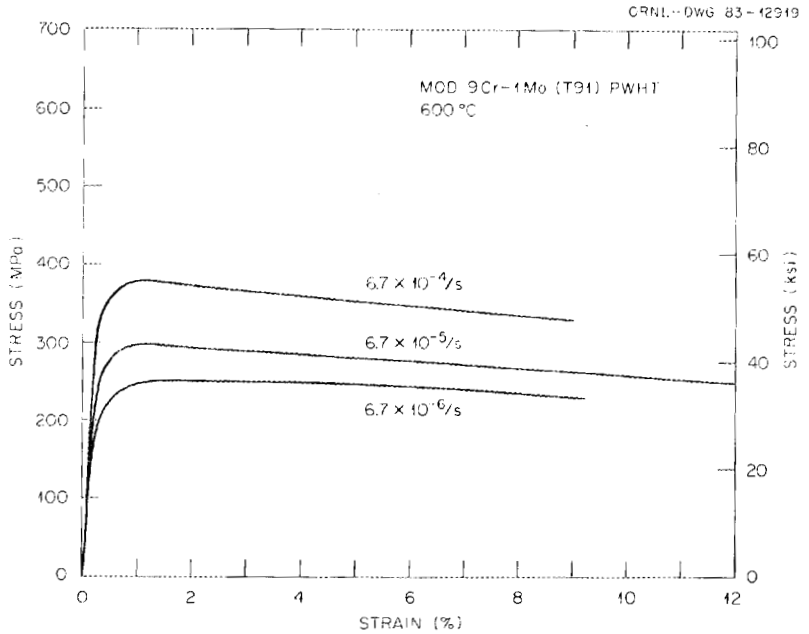

Fig. 123. Influence of strain rate on stress-strain behavior at 600°C (1112°F).

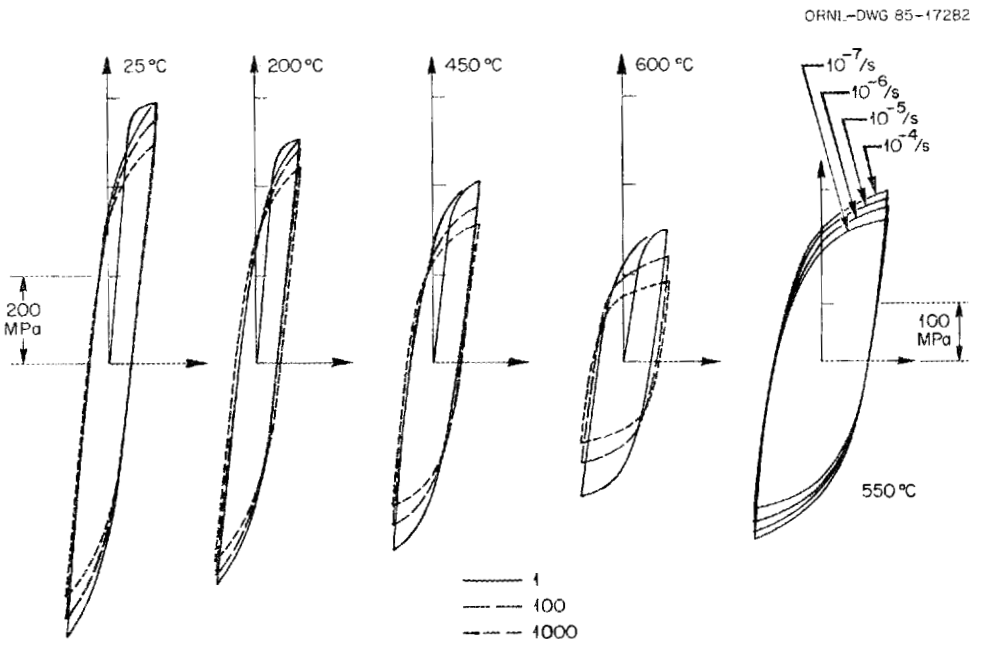

Fig. 124. Cyclic stress-strain behavior at a strain range of ~1%.

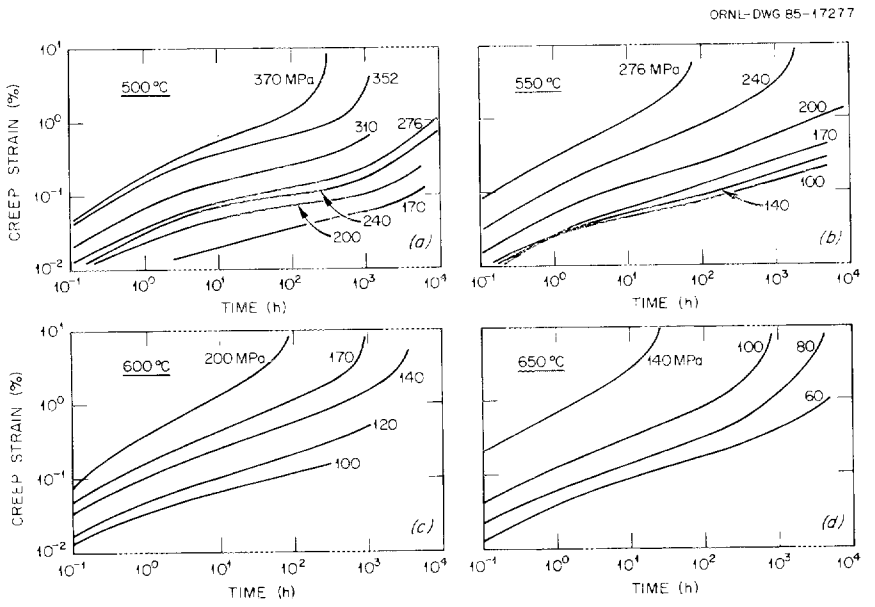

Fig. 125. Constant-load creep behavior.

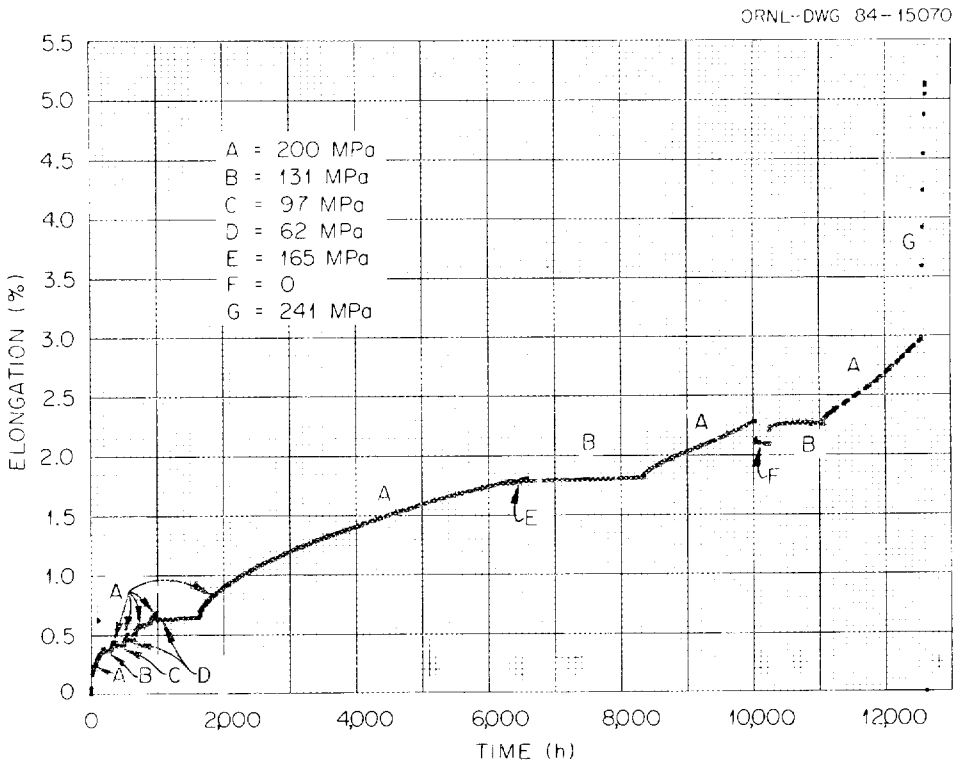

Fig. 126. Creep under stepwise changing stress.

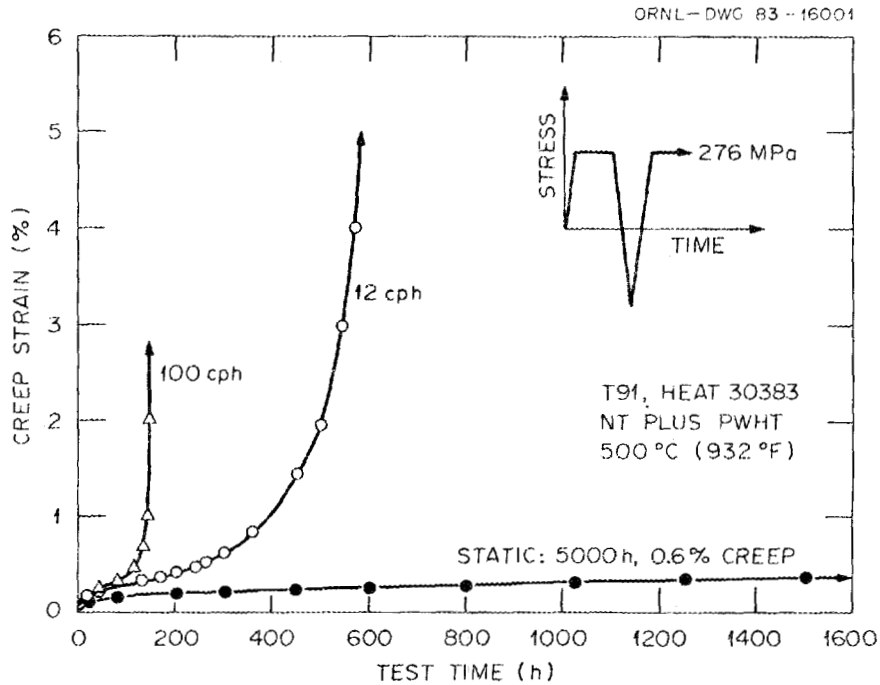

Fig. 127. Effect of short-time stress reversals on creep.

back to 276 MPa in 10 s. Including the constant load test, this was done 0, 12, and 100 times per hour, resulting in successively larger creep rates.

Results of three stress relaxation tests[138] are shown in Fig. 128. Each test included two successive 100-h periods of relaxation at constant strain from the same starting stress. Three different starting stresses and temperatures were used in the three different tests. In each test, the stress at the end of the second relaxation period was greater than the stress at the end of the first.

As explained earlier, rather significant thermal transients can occur during operation of liquid-metal cooled reactors. Three kinds of cyclic testing were performed to investigate the behavior of modified 9Cr-1Mo under variable temperature conditions. Examples of the results obtained in each kind of test are provided in Figs. 129 through 132.

To obtain the stress-strain curves[140] shown in Fig. 129, a specimen was subjected to blocks of constant-amplitude strain cycles at constant temperatures of 150°C (302°F) and 550°C (1022°F). A 1-h hold period at maximum compressive strain was included at the higher temperature.

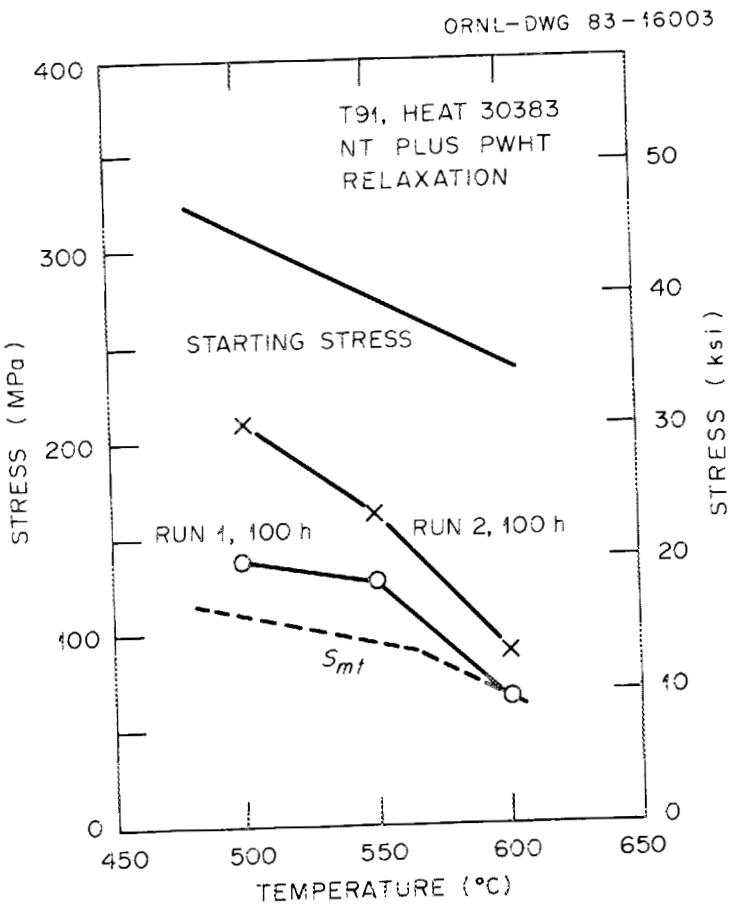

Fig. 128. Stress-relaxation behavior for three starting stresses and temperatures. Estimated allowable stress S_{mt} for commercial heats is shown for comparison.

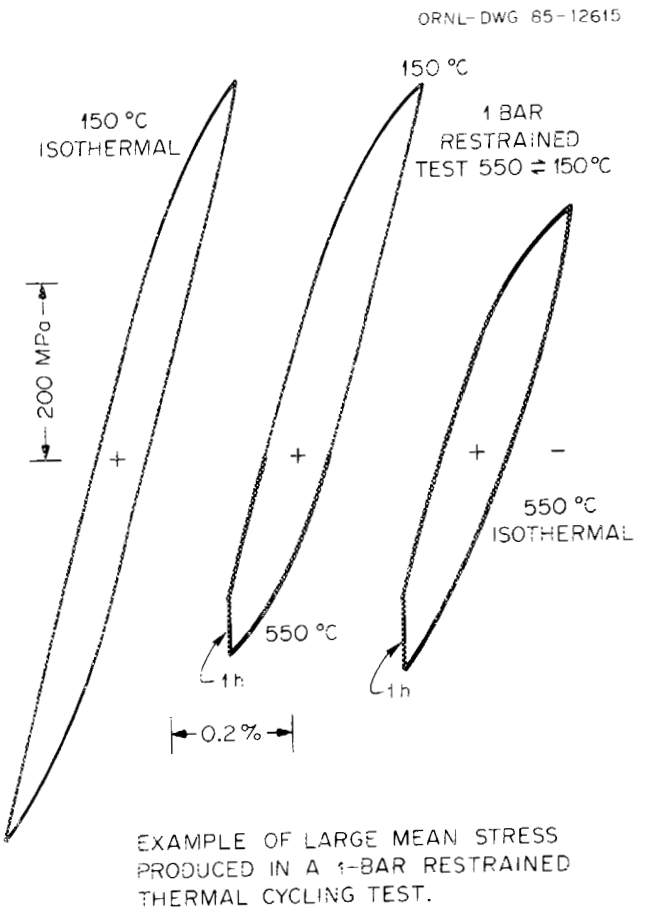

Fig. 129. Comparison of isothermal stress-strain loops with one in which tensile and compressive portions were obtained at different temperatures.

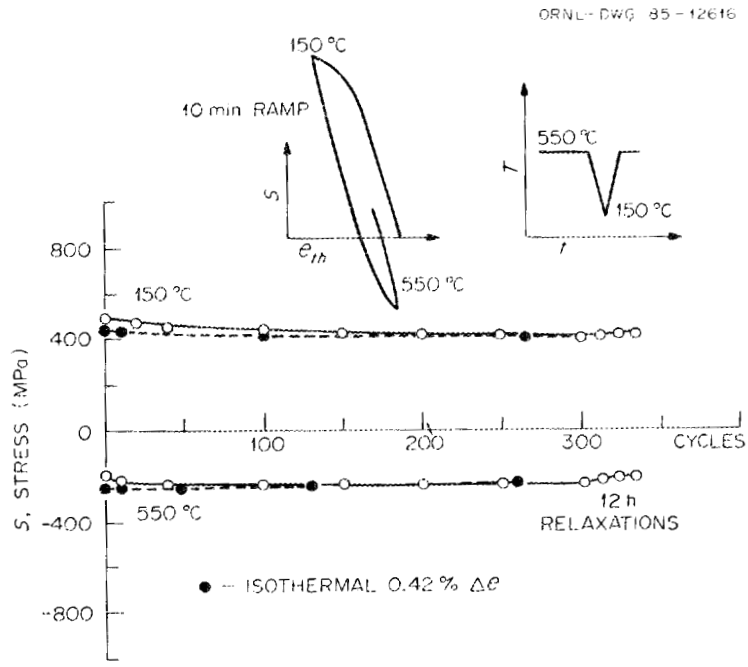

Fig. 130. Extreme stresses in restrained thermal cycling test compared with those in isothermal strain cycling test.

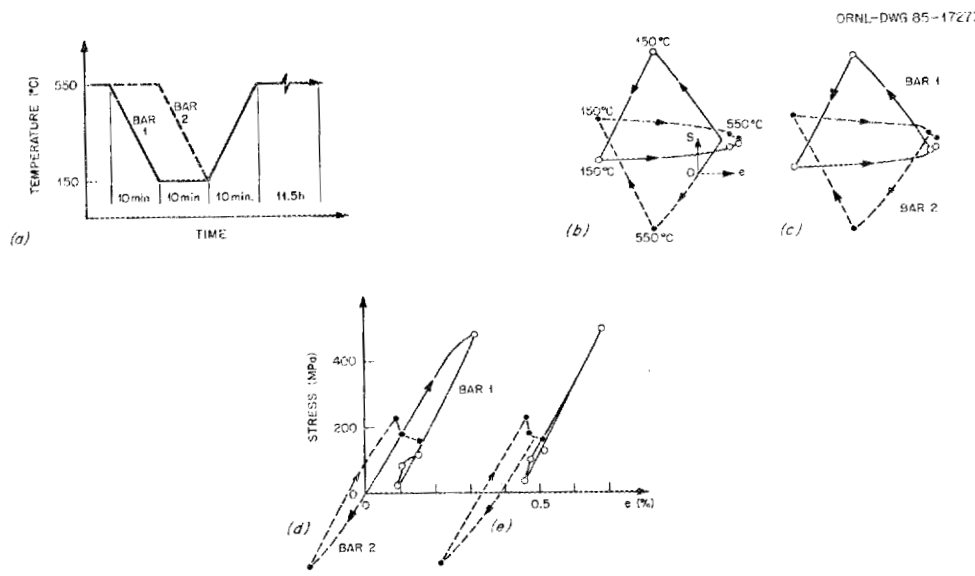

Fig. 131. Results of two-bar creep ratchetting test: (*a*) temperature vs time, (*b*) stress vs strain for both bars in first cycle, (*c*) stress vs strain for both bars in 30th cycle, (*d*) stress vs mechanical strain for both bars in first cycle, and (*e*) stress vs mechanical strain for both bars in 30th cycle.

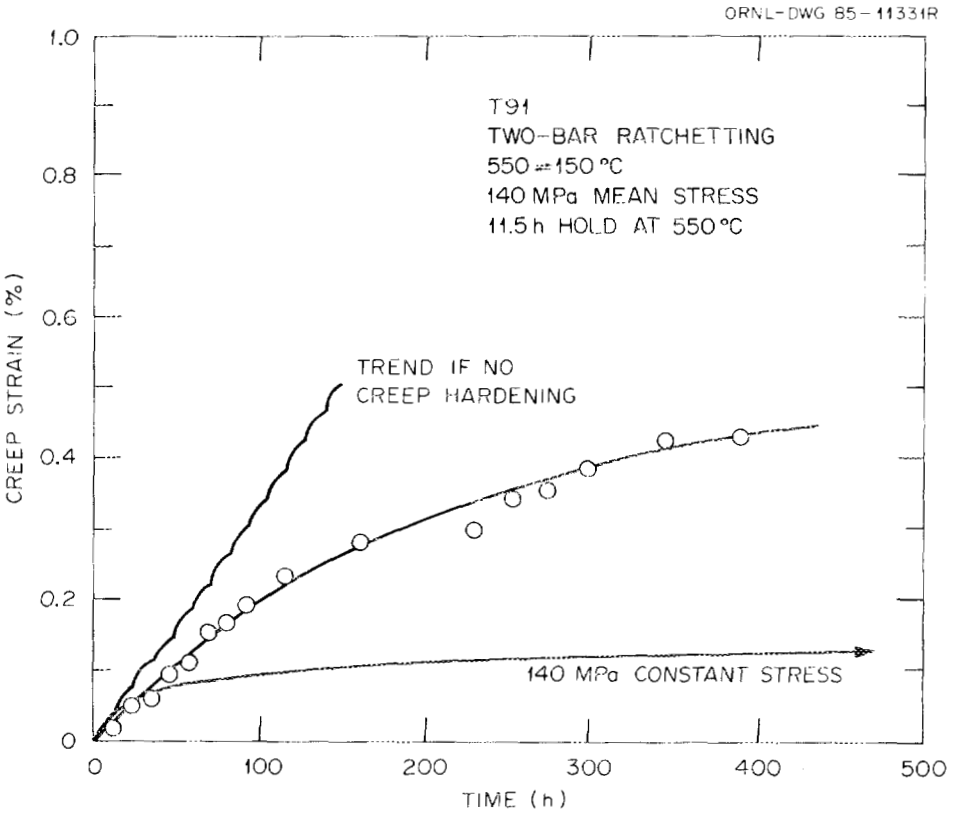

Fig. 132. Comparison of creep strain vs time in a two-bar ratchetting test with estimates based on assumptions of no creep hardening and full creep hardening.

Finally, the specimen was subjected to cycles that were a composite of the other two. The tensile stress portion of the cycle was at 150°C, and the compressive stress portion was at 550°C. Mechanical straining was interrupted, and zero stress was maintained while the temperature was changing. These results suggest that in certain circumstances isothermal test data can be used to approximate behavior under variable temperature conditions.

For Fig. 130 a specimen was restrained[141] with fixed gage length, and the temperature was cycled between 550 and 150°C. One inset in the figure shows stress vs thermal strain for the first cycle, and another inset shows temperature vs time for each cycle after the 300th when a 12-h hold was introduced at 550°C. As shown in the figure, the extremes of the stress cycle approached those obtained in isothermal tests at a comparable cyclic strain amplitude.

Figure 131 shows results[141] from a two-bar thermal ratchetting test conducted in the manner described by Stentz.[144] In this test, two uniaxial specimens were repeatedly subjected to conditions that simulate those at the inner and outer surfaces of a piping component in a LMR during abrupt shutdown and restart. The electronic control systems of two testing machines were linked together to keep the average of the stresses in the two specimens constant and the strains equal while the temperatures were cycled out-of-phase. As indicated in Fig. 131, the specimens were initially loaded to the same stress, 140 MPa (20 ksi), at 550°C. The temperature of the first specimen was then ramped down to 150°C in 10 min. Next, the temperature of the second specimen was reduced in the same manner. The temperatures of both specimens were then ramped back up to 550°C in 10 min and held there for the remainder of the 12-h temperature cycle. The figure also shows stress in each specimen vs total strain for the 1st and 30th cycles and stress in each specimen vs mechanical strain for the same cycles. The inelastic strain increased during every cycle, with almost all of the increase occurring while the temperatures were held constant. In Fig. 132 the accumulation[141] of inelastic strain vs time is compared with a curve from a constant load creep test at 140 MPa (20.3 ksi) and with a curve constructed with segments similar to the first 11.5 h of the constant load curve. Agreement of the ratchetting curve with the constant-load curve would be expected if stress reversals had no effect on creep hardening, and agreement with the segmentally constructed curve would be expected if stress reversals removed all creep hardening. The ratchetting curve lies between these two extremes, as expected.

Two thin-walled tubular specimens were tested at room temperature under combined axial and torsional stresses to determine the position, size, and shape of biaxial yield loci. A 25-microstrain offset was used to define yield. One specimen was used to determine initial and subsequent yield loci along a radial path in stress space, and another was used to determine loci along a nonradial path.[143] The process is illustrated in Fig. 133, where the circles represent von Mises idealizations of the yield loci. The experimentally determined loci are compared in Figs. 134 through 140 with translated von Mises circles about the same size as the initial

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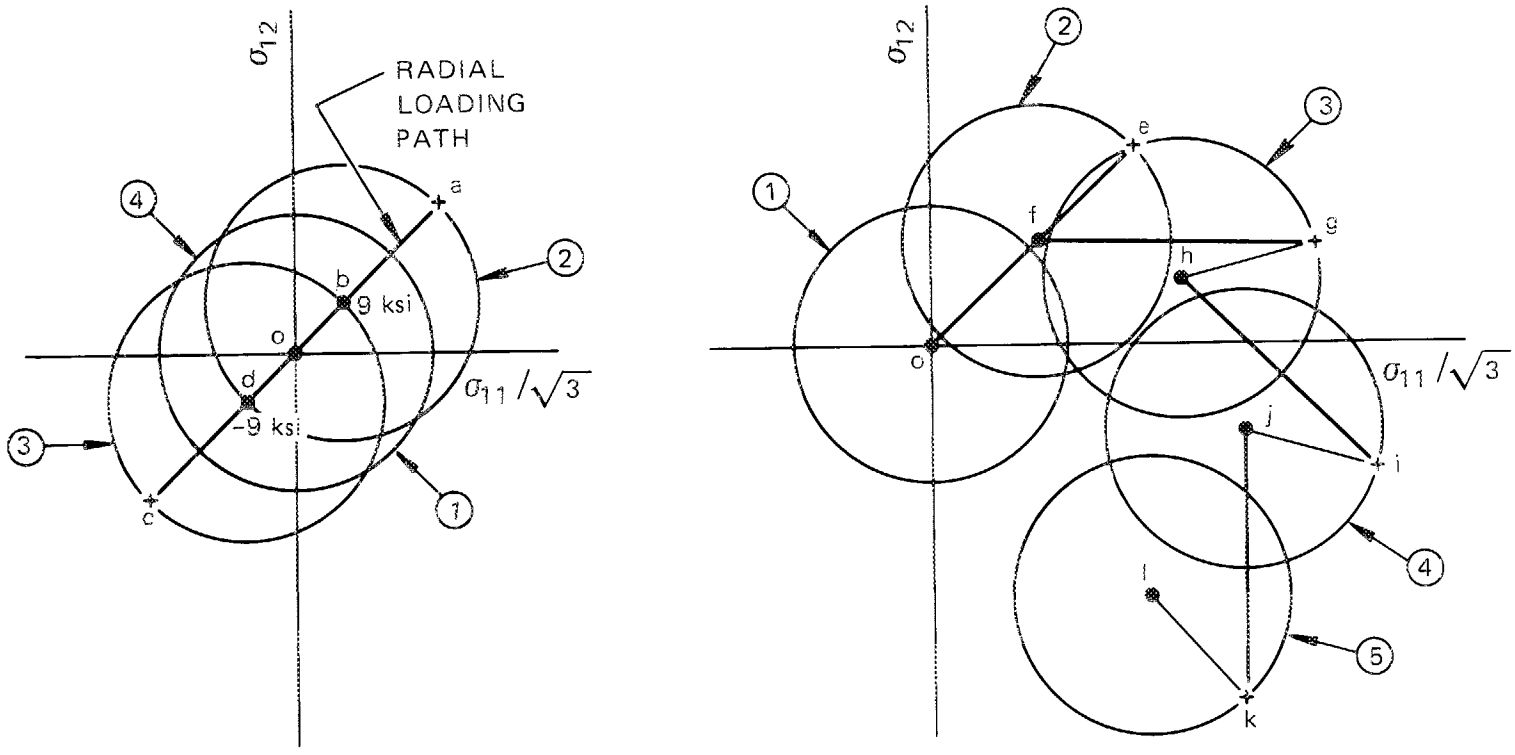


(a) RADIAL LOADING TEST

(b) NON RADIAL LOADING TEST

○ IDEALIZED Nth YIELD LOCUS

● CENTER OF IDEALIZED YIELD LOCUS

Fig. 133. Loading paths used in yield locus investigations at room temperature.

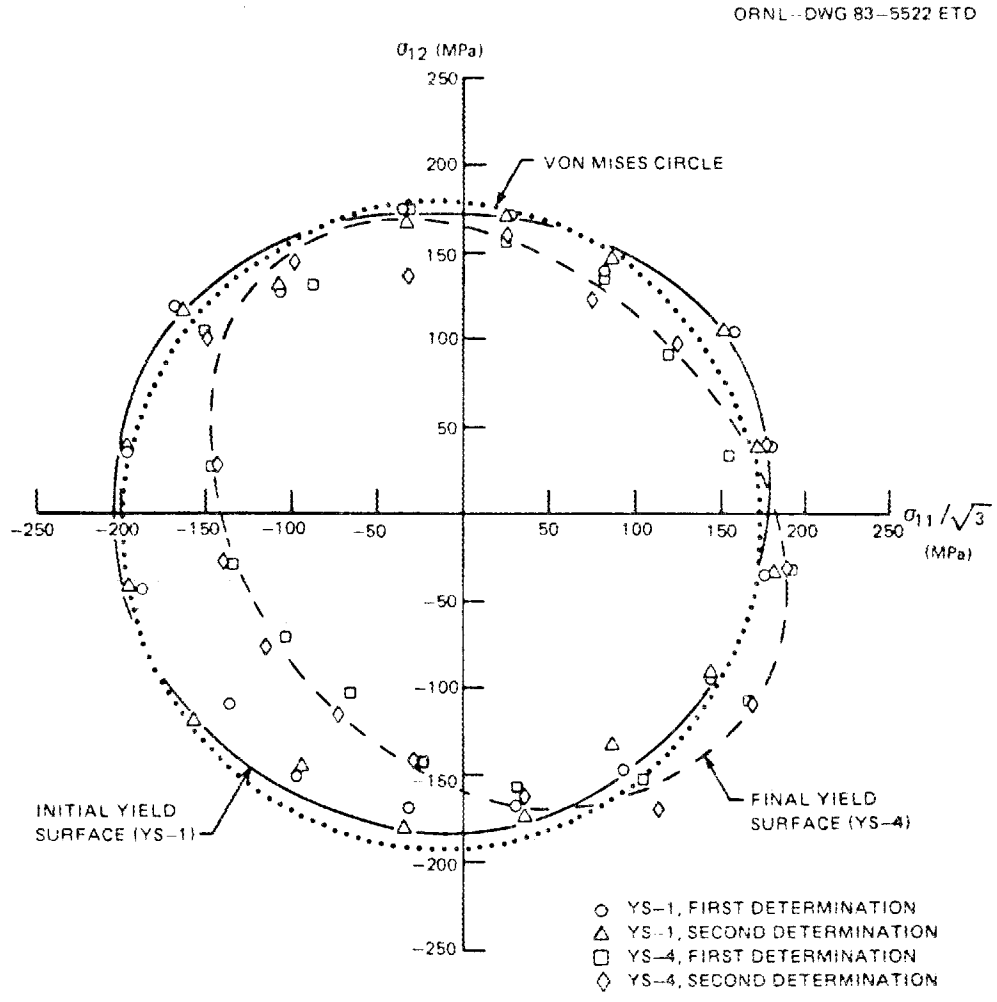


Fig. 134. Comparison of initial yield locus, YS-1, with final yield locus, YS-4, for radial loading test series.

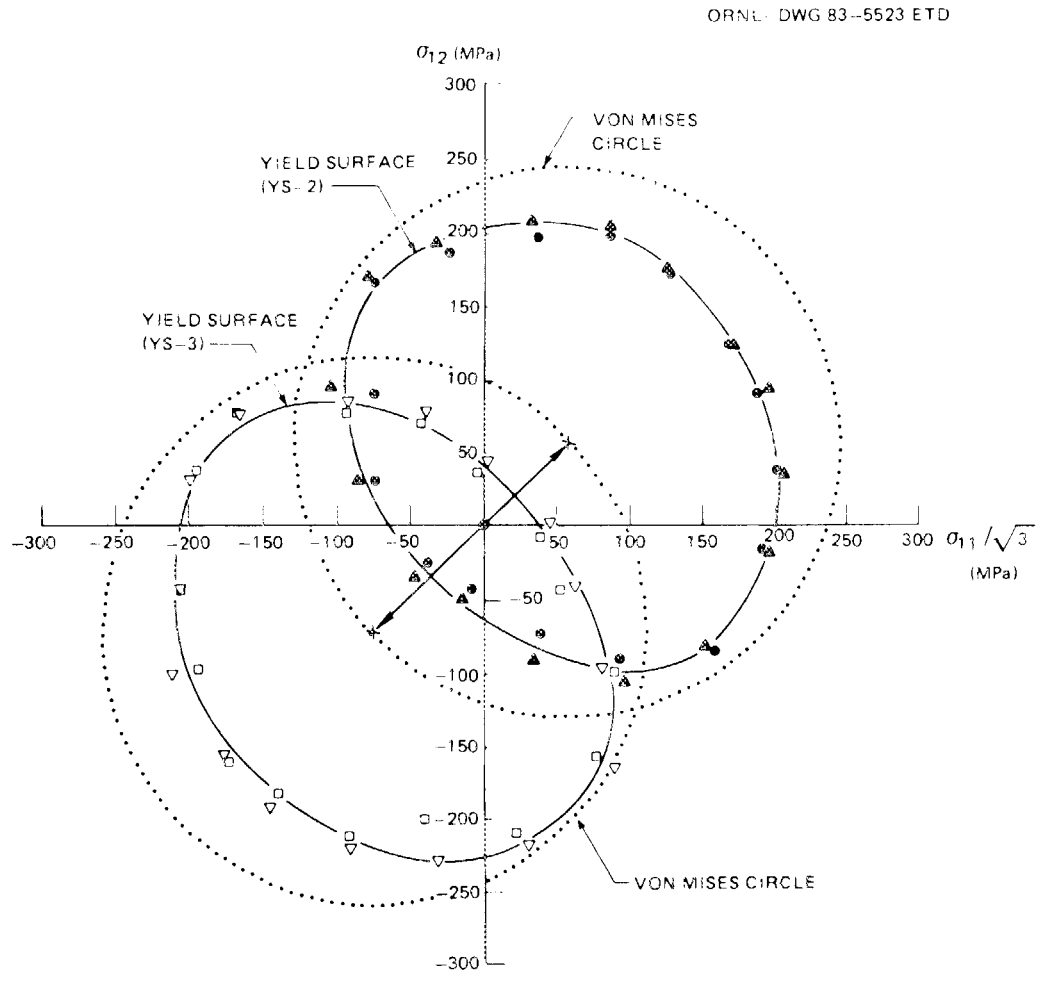


Fig. 135. Yield behavior exhibited after radial loading to about 40% beyond the initial yield locus, YS-1.

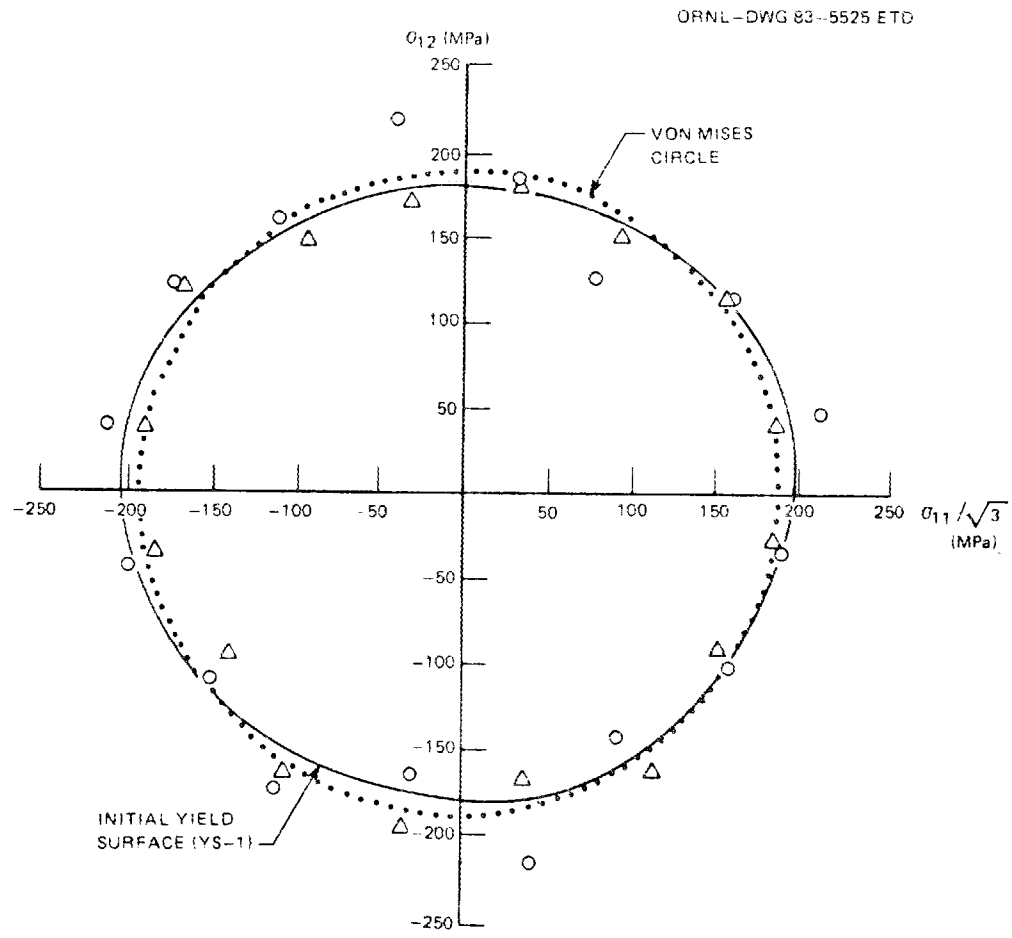


Fig. 136. Initial yield locus YS-1 for the nonradial loading test series.

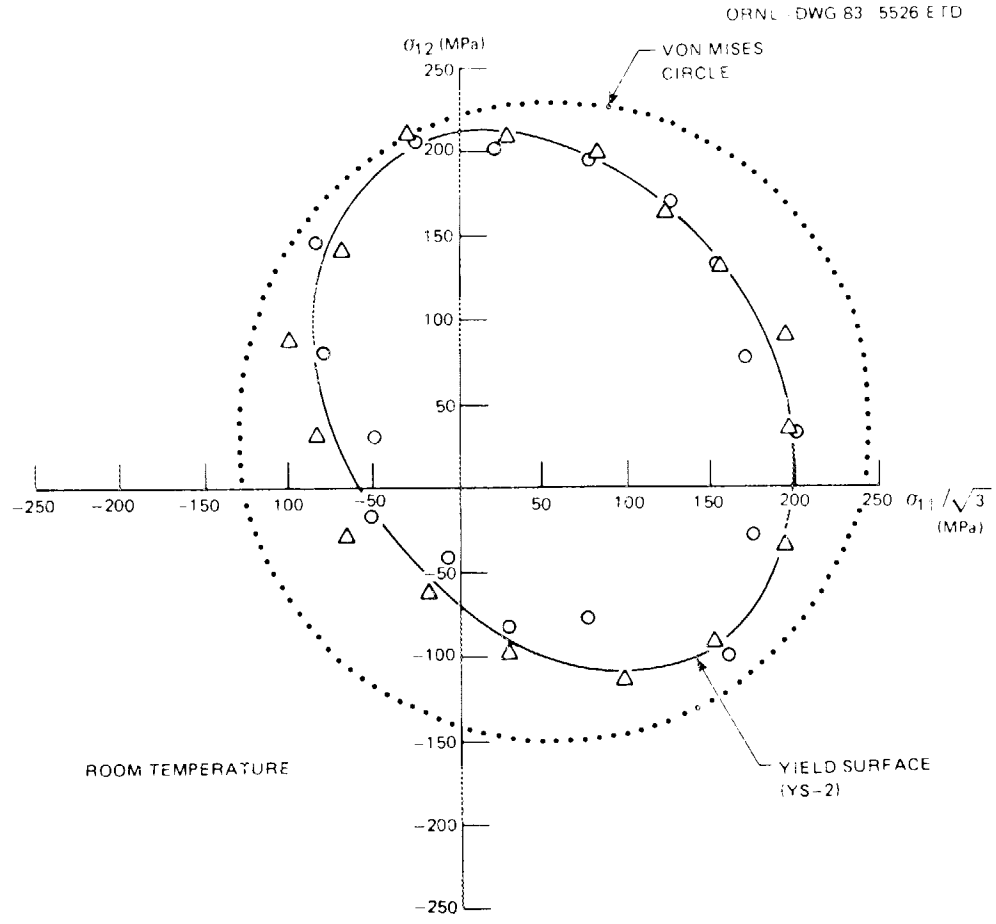


Fig. 137. Yield locus YS-2 exhibited after inelastic, tension-torsion loading from point o to point e [Fig. 18(*b*)].

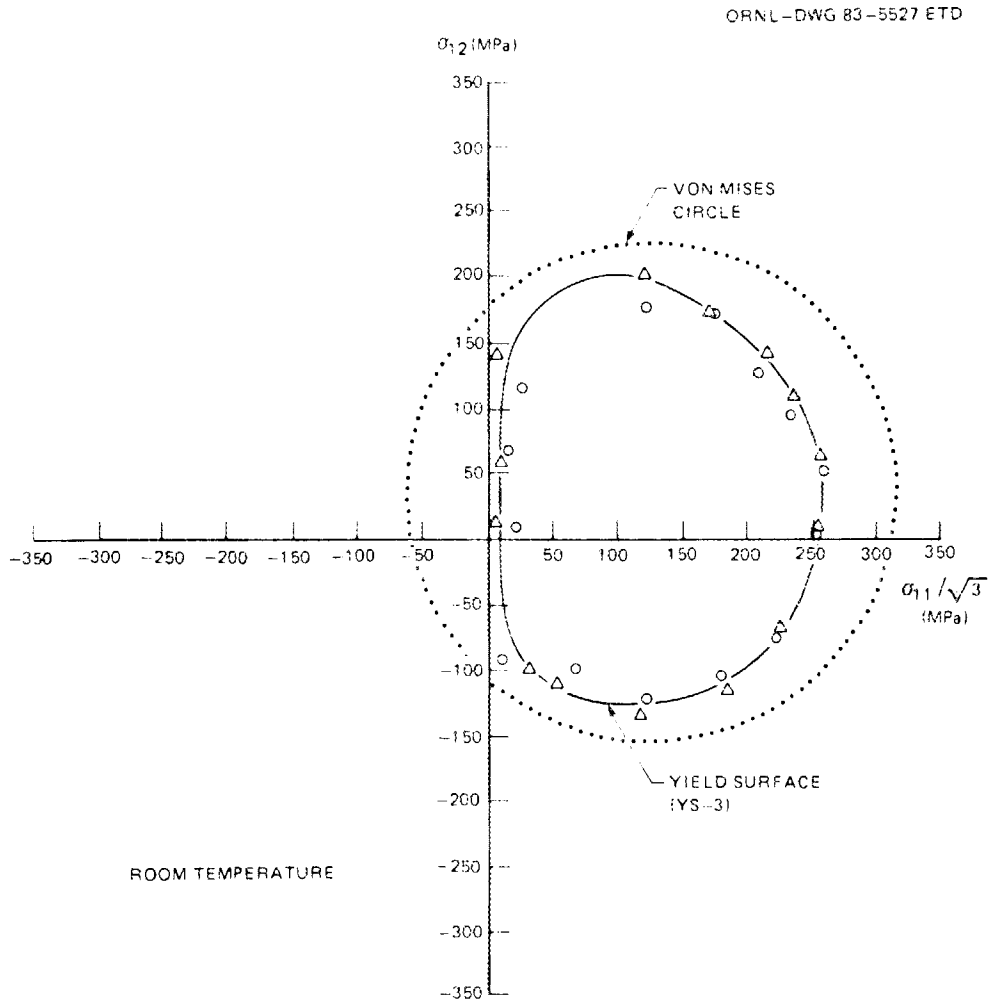


Fig. 138. Yield locus YS-3 exhibited after inelastically tension-torsion loading from point f to point g [Fig. 18(*b*)].

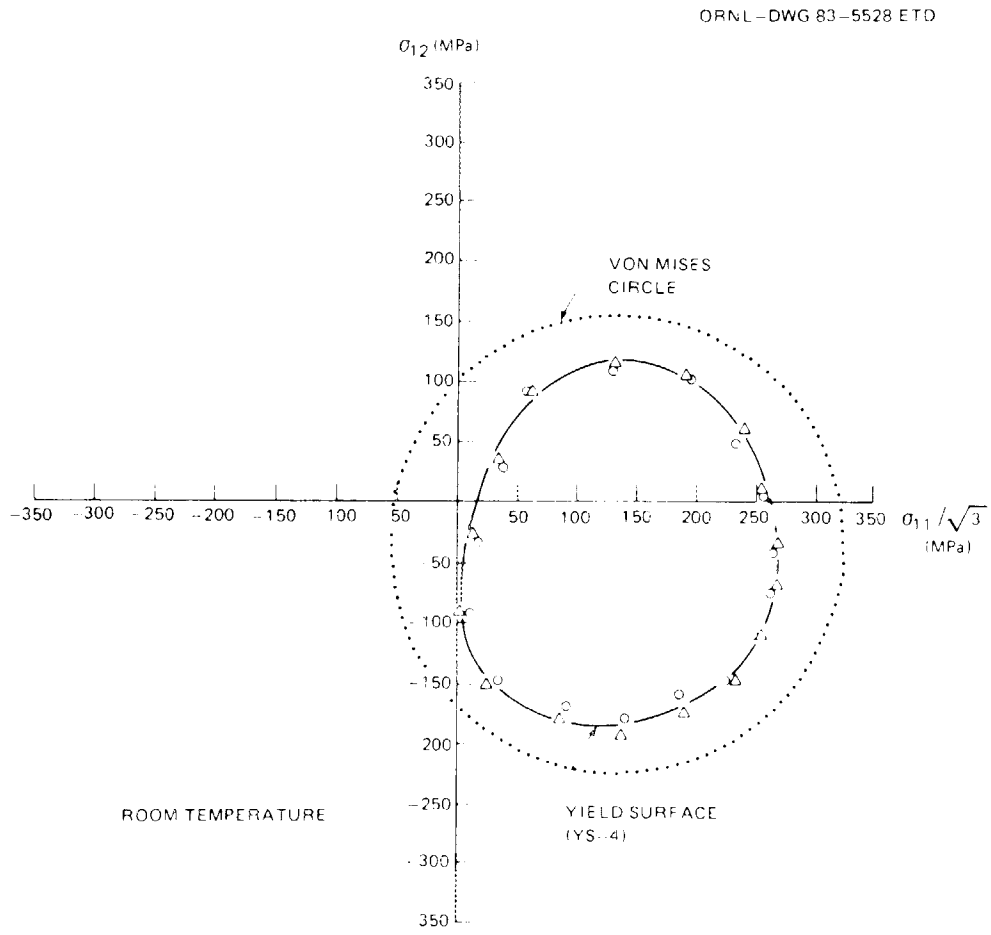


Fig. 139. Yield locus YS-4 exhibited after inelastically tension-torsion loading from point h to point i [Fig. 18(*b*)].

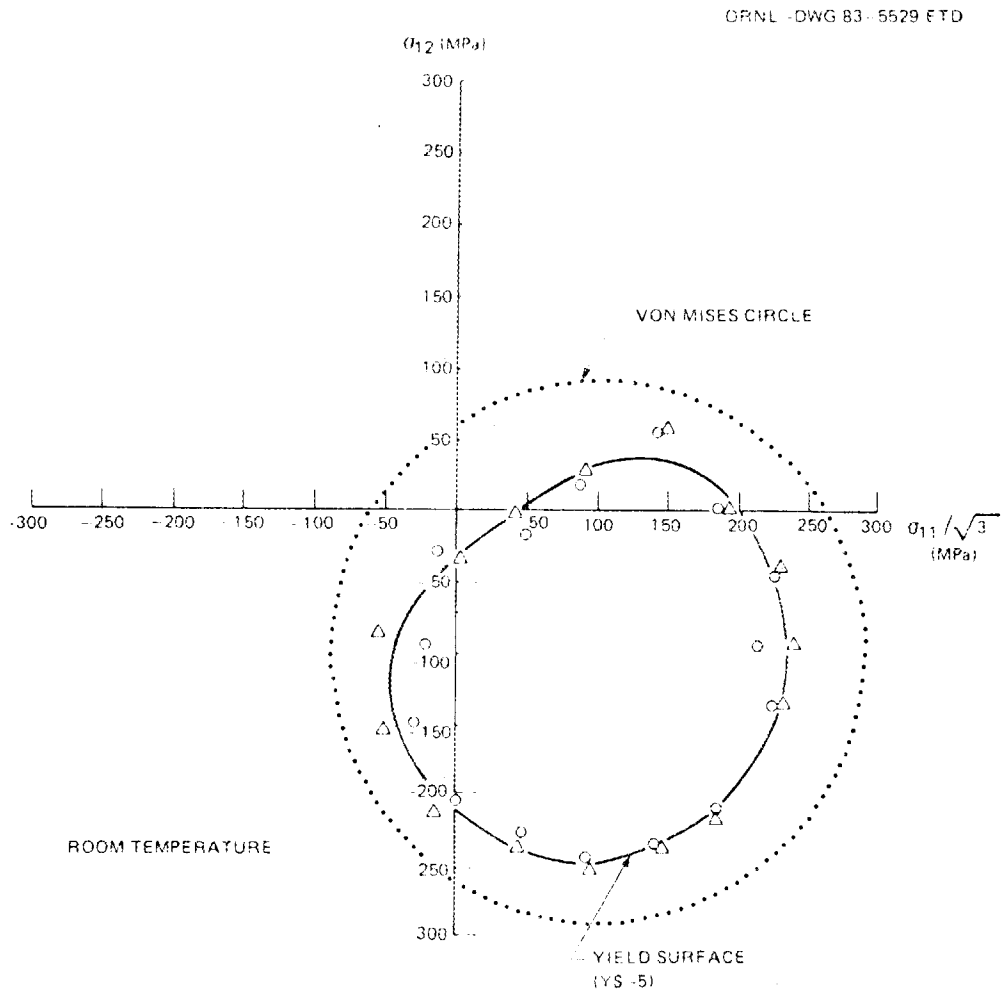


Fig. 140. Yield locus YS-5 exhibited after inelastically tension-torsion loading from point j to point k [Fig. 18(*b*)].

yield locus. These results indicate that (1) the von Mises yield criterion can be used as a basis for modeling initial yield behavior; (2) subsequent yield behavior is more kinematic than isotropic in nature (i.e., the yield locus tends to translate rather than change in size with plastic straining); and (3) nonuniform softening (shrinking of the yield locus) takes place, leading to an elliptically shaped yield locus that is smaller in the direction of loading than perpendicular to the direction of loading.

The results of the exploratory uniaxial and biaxial tests described previously were used as the basis for an assessment of the applicability of the NE F 9-5T constitutive framework to modified 9Cr-1Mo steel. This assessment is summarized in Sect. 3.5.2.2.

3.5.2.2 Applicability of NE F 9-5T Constitutive Framework

The applicability of the NE standard F 9-5T constitutive framework to modified 9Cr-1Mo steel was assessed by Clinard,[145,146] based on the results of the exploratory tests described in Sect. 3.5.2.1. The assessment is summarized in the following paragraphs and consists of brief statements concerning certain features of the behavior of modified 9Cr-1Mo and corresponding features of the NE standard F 9-5T constitutive models.

For constant strain-range and temperature, the shapes of uniaxial cyclic stress-strain curves for modified 9Cr-1Mo (see Fig. 124) resemble those of 2.25Cr-1Mo. The shape of the curves for neither material is notably bilinear. The bilinear character of the NE F 9-5T plasticity model is a known shortcoming but probably only slightly more for modified 9Cr-1Mo than for other materials.

Modified 9Cr-1Mo exhibits a pronounced Bauschinger effect. The kinematic hardening model of NE F 9-5T is capable of representing this behavior rather well.

Significant isotropic softening is apparent in cyclic tests of modified 9Cr-1Mo (see Fig. 124). Recent modifications to NE F 9-5T provide an improved isotropic hardening law for yield stress that will accommodate continuous changes as a result of cyclic softening. Models for approximating the softening through abrupt jumps in yield stress are also available and are probably adequate for describing isotropic softening of modified 9Cr-1Mo.

In design analyses based on NE F 9-5T procedures, a single value of bilinear plastic modulus is selected for all times and all locations in a structure. NE F 9-5T provides procedures for determining this value based on an a priori estimate of maximum strain (or strain range). The virgin bilinear yield stress is also determined in this process, but subsequent values of the yield stress also depend on the isotropic hardening rules invoked. Problems arise when selecting a single representative maximum strain is difficult and/or when the bilinear plastic properties depend greatly on that selection. Using the NE F 9-5T procedures for austenitic stainless steels, bilinear properties were determined for modified 9Cr-1Mo at a number of temperatures and a strain rate of $6.7 \times 10^{-5}\ s^{-1}$. Values of bilinear yield strength and plastic modulus vs strain are plotted in Figs. 141 and 142, respectively. The strain dependence of bilinear properties is much greater for modified 9Cr-1Mo than for current NE F 9-5T materials.

The monotonic and cyclic stress-strain behavior of modified 9Cr-1Mo is greatly dependent on strain rate (see Figs. 127-129). The rate dependence of 2.25Cr-1Mo is much less, and those of 304 and 316 are practically zero. The NE F 9-5T plasticity models have no rate dependence, which is believed to be an important shortcoming for applicability to modified 9Cr-1Mo.

Results of preliminary room-temperature tests indicate that the von Mises multiaxial yield criterion of NE F 9-5T is probably adequate for modified 9Cr-1Mo. Additional experimental and analytical studies are needed before further assessment of multiaxial behavior is possible.

The shape of constant-load creep curves for modified 9Cr-1Mo is not unusual. Considerable creep occurs at stresses lower than the plastic yield stress, however. The austenitic stainless steels, on the other hand, exhibit very little creep at stresses below yield. At stresses higher than yield, creep rates for modified 9Cr-1Mo are of the order of plastic strain rates, making it difficult to distinguish between creep and plastic strains.

Creep-test data for modified 9Cr-1Mo show that recoverable (visco-elastic) strains, as in NE F 9-5T, can probably be ignored. However, significant thermal recovery of creep properties, which is also ignored in NE F 9-5T, occurs with time.

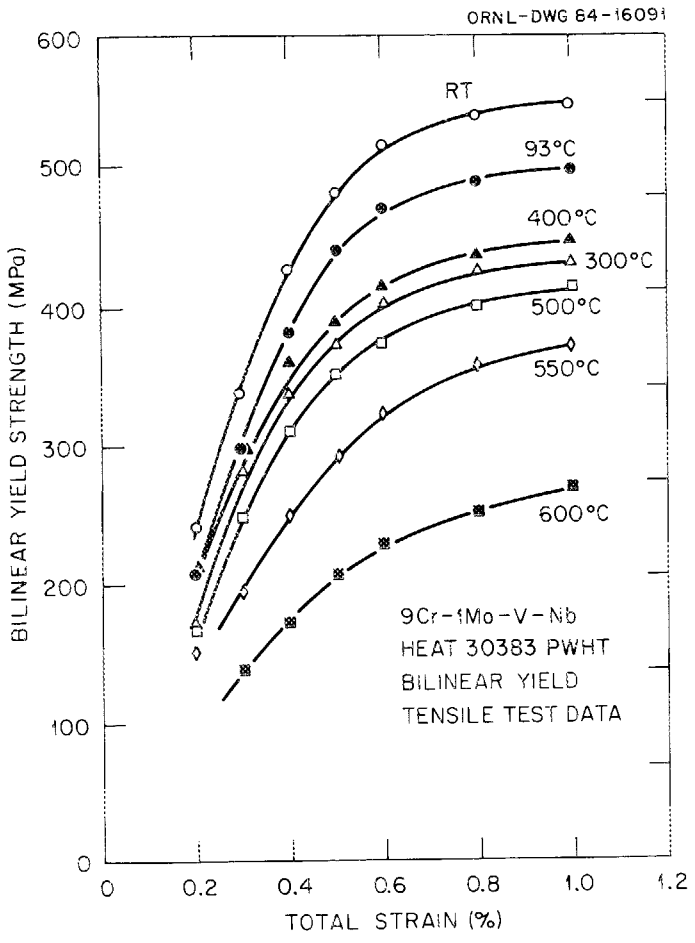


Fig. 141. Bilinear yield strength dependence on strain and temperature.

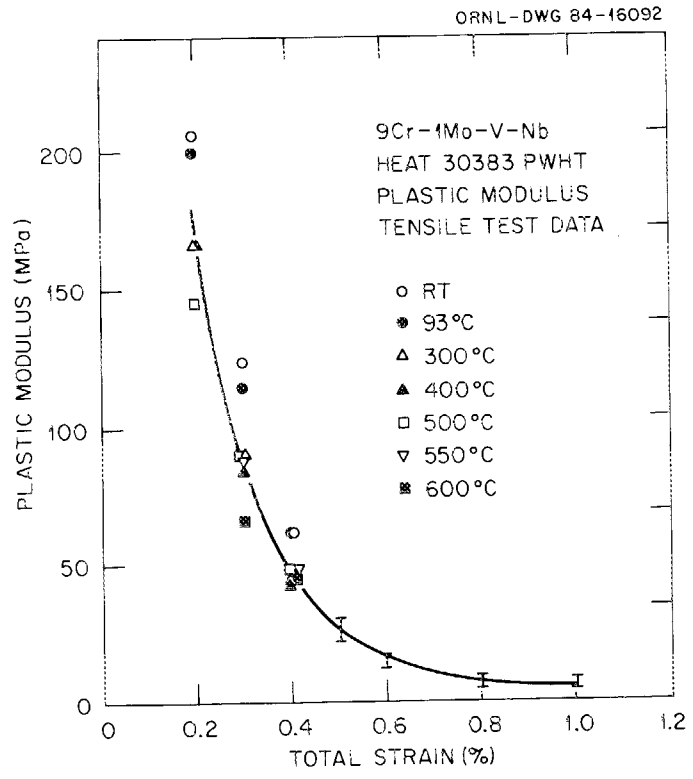


Fig. 142. Plastic modulus dependence on strain.

Modified 9Cr–1Mo exhibits a strong Bauschinger effect in cyclic creep tests. The NE F 9-5T auxiliary hardening rules for stress reversals in creep probably adequately account for this behavior.

The high strain-rate dependence of flow stress and high creep rates observed for modified 9Cr–1Mo indicate that a viscoplastic constitutive model is required. Standard NE F 9-5T does not include such models. When creep and plastic strain rates are quite different in magnitude, such as for the austenitic stainless steels, the separation of creep and plastic behavior in NE F 9-5T (with only mild ad hoc interaction rules) is an adequate basis for a constitutive model. The adequacy of current NE F 9-5T creep-plasticity interaction rules is doubtful for a viscoplastic material such as modified 9Cr–1Mo.

From these observations, the present NE F 9-5T constitutive framework was concluded to be inadequate for representing modified 9Cr–1Mo steel. The material appears to require a nonlinear viscoplastic model, such as the one developed by Robinson[147,148] of ORNL for 2.25Cr–1Mo steel, which is unified because no distinction is made between creep and plastic strains. Initial results of an effort to adapt this model to modified 9Cr–1Mo are presented in Sect. 3.5.2.3.

3.5.2.3 Development of Unified Constitutive Model

Certain shortcomings of the current NE F 9-5T constitutive equations led Robinson[147,148] to develop a nonlinear viscoplastic model for 2.25Cr–1Mo steel. Certain features of this ORNL model make it seem ideal for modified 9Cr–1Mo as well. The model is rate dependent and unified because no distinction is made between creep and plastic strains. Thus, there is no need for creep-plasticity interaction rules, thus making the equations simple in form. Thermal recovery of properties is included naturally, and the representation of uniaxial stress-strain and strain-time curves is inherently nonlinear.

Written in the usual tensor notation, the model consists of two coupled differential equations for the rates of change of the components of inelastic strain (the flow law) and internal stress (the evolutionary law). The equations have been incorporated into a finite element computer code and used in detailed inelastic analyses of structural test articles.

The unified model described previously is being adapted to represent the behavior of modified 9Cr-1Mo steel. The form of the equations has been modified to incorporate strain softening in the monotonic and cyclic stress-strain behavior. The values of certain parameters have been adjusted to provide better agreement with selected data. The results of this initial effort are illustrated in Figs. 143 through 145.

A numerical procedure, developed to obtain optimum values of parameters in the model based on the method of least-squares, makes use of existing computer programs for solving nonlinear least-squares problems and for numerically integrating differential equations. Data from a variety of tests can be processed at one time, and parameters that minimize the overall sum of squares of differences between observed and estimated values of inelastic strain can be calculated.

Interim and final unified constitutive models will be developed for modified 9Cr-1Mo. These will be used for detailed inelastic analyses of structural components fabricated from this material and for validating simplified methods of analysis of such components.

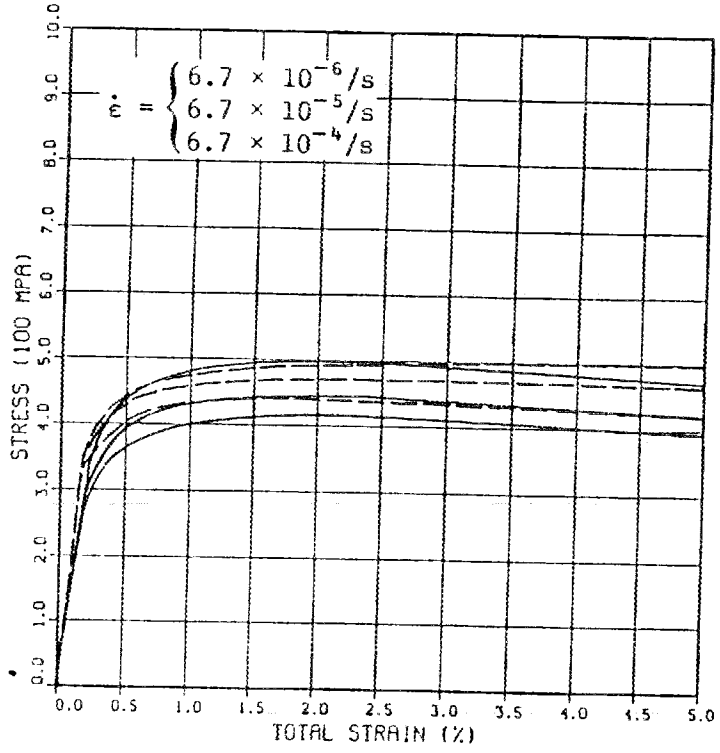
3.5.3 Failure Behavior Modeling

This section is divided into three areas: (1) an overview of the ingredients for time-dependent failure rules, (2) experiments and assessments related to the damage accumulation rules, and (3) similar experiments and assessments relating to the multiaxial strength rules.

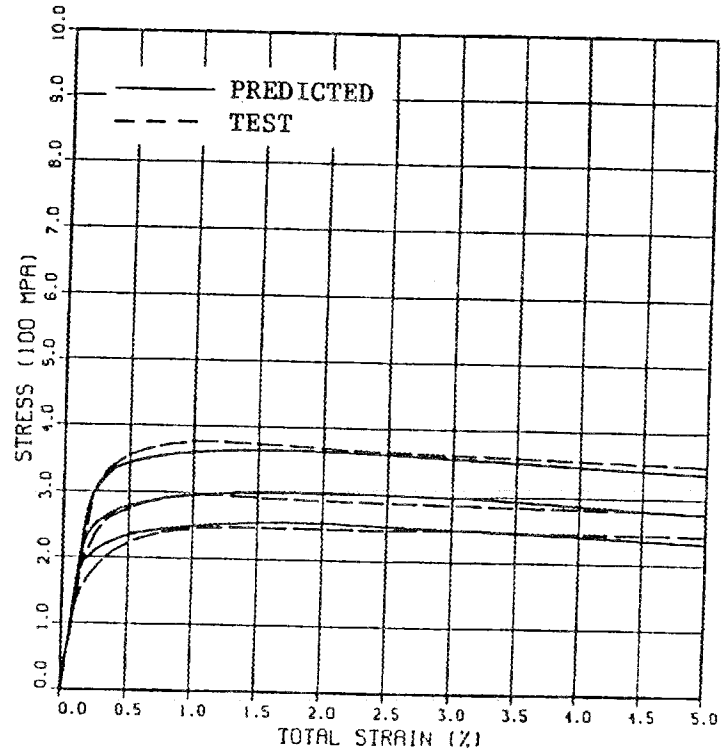
3.5.3.1 Overview

Rules and criteria to guard against creep-rupture failure modes are included in ASME Code Case N-47 (ref. 122) under "Load-Controlled Stress Limits for Levels A through D Service Limits." Rules and criteria to guard against time-dependent fatigue or creep-fatigue failure modes are included in the Code Case N-47, Appendix T, rules for "Strain and Deformation Limits" and are applicable to service levels A, B, and C.

Failure prediction models typically uncouple the "damage accumulation criterion" from the stress state effect ("multiaxial strength criterion"). Thus, two independent criteria are required for predicting failure.



(*a*) 500°C



(*b*) 600°C

Fig. 143. Influence of strain rate and temperature on estimated and observed stress-strain behavior.

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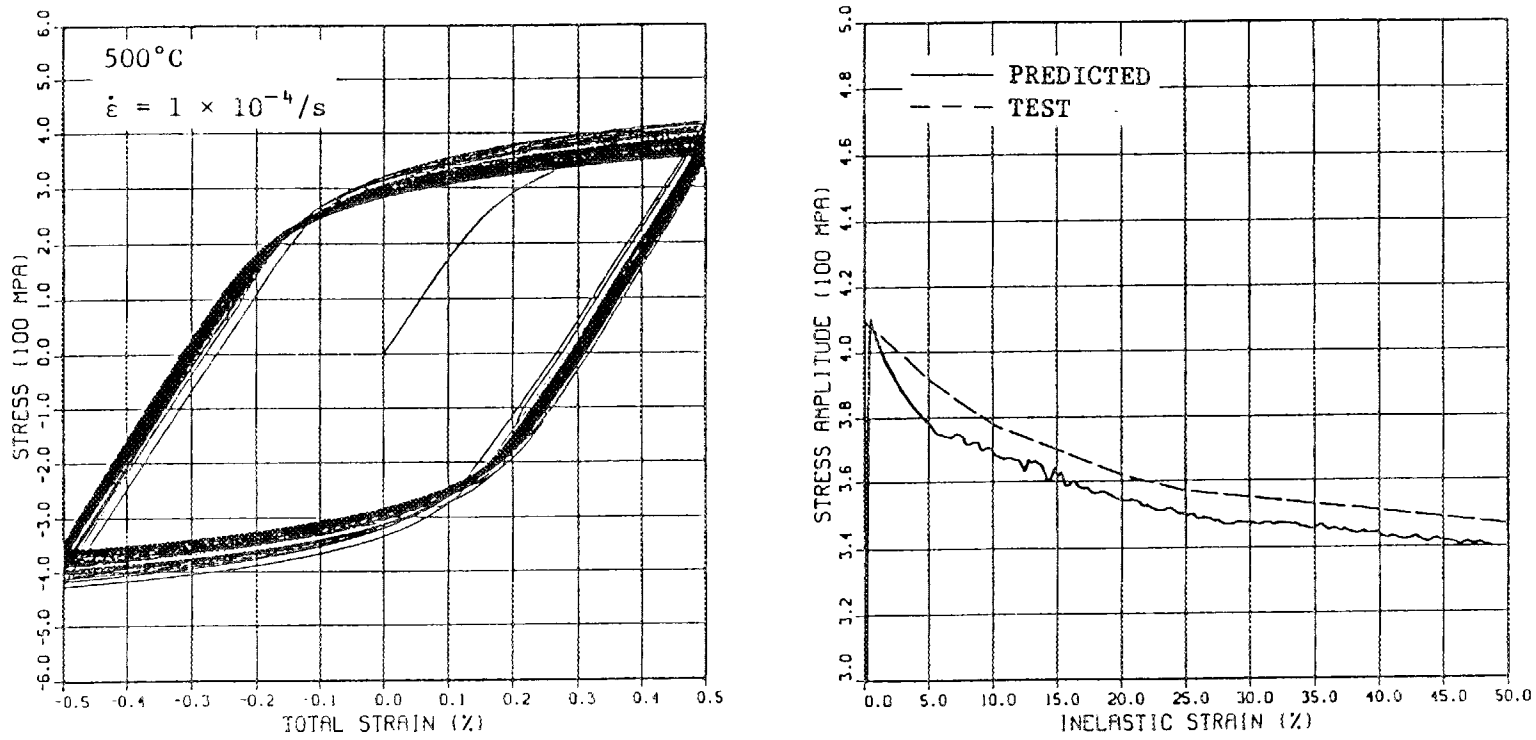

Fig. 144. Cyclic stress-strain and strain softening behavior.

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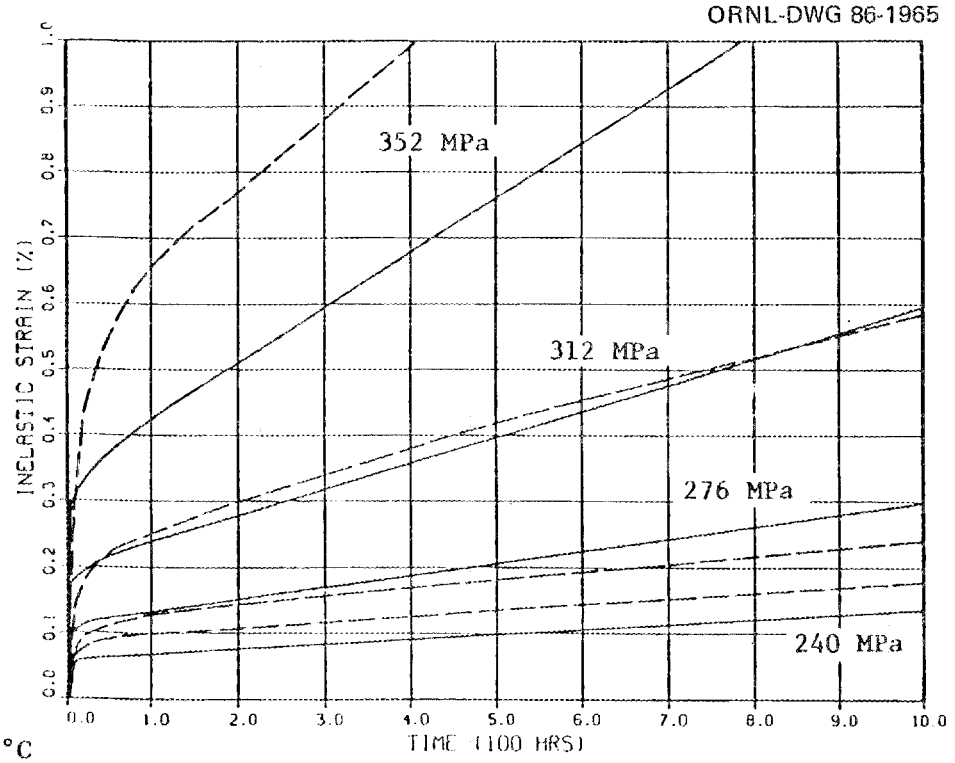
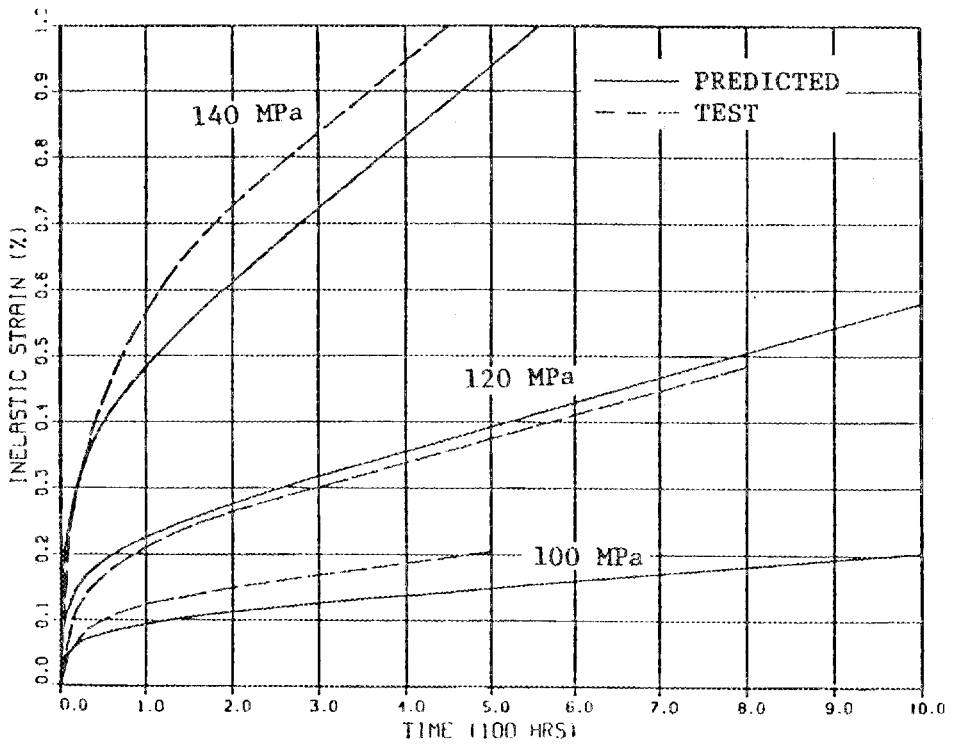


(*a*) 500°C



(*b*) 600°C

Fig. 145. Estimated and observed creep behavior.

For computing failure under load-controlled conditions or for using elastic calculations, the Tresca (maximum shear stress) strength theory is used:

$$S = \max\left(|\sigma_1 - \sigma_2|, |\sigma_2 - \sigma_3|, |\sigma_3 - \sigma_1|\right), \tag{6}$$

where max denotes the maximum value for the three quantities inside the brackets. Thus, to compute damage accumulation, Eq. (1) reduces the multiaxial stress state to an "equivalent uniaxial stress state." For load-controlled conditions, the associated damage rule is based on the linear time-fraction damage accumulation theory, expressed by

$$D = \sum_i \Delta t_i/(T_R)_i \leq B, \tag{7}$$

where, in the absence of safety margins,

D = creep damage,
Δt_i = total duration of a specific loading i,
$(T_R)_i$ = creep-rupture life under the specific loading, i,
B = allowable damage (theoretically 1.0).

As previously noted, rules for guarding against the inelastic creep-fatigue failure mode are incorporated in Appendix T of Code Case N-47. These rules are applicable to cyclic strain-controlled design conditions. When used, inelastic calculations incorporate the von Mises strength theory, expressible as

$$\sigma_e = \sqrt{[(\sigma_1 - \sigma_2)^2 + (\sigma_2 - \sigma_3)^2 + (\sigma_3 - \sigma_1)^2]/2}, \tag{8}$$

where σ_1, σ_2, and σ_3 denote the principal stresses.

Damage accumulation is based on the linear time and cycle-fraction model, which can be expressed as

$$D = \sum_{j=1}^{P} \frac{n}{N_f}_j + \sum_{k=1}^{q} \frac{\Delta t}{T_R}_k, \tag{9}$$

where (neglecting safety margins)

D = total creep-fatigue damage,

n = number of applied cycles of loading condition j,

N_f = number of fatigue cycles to failure under loading condition j,

P = total number of different applied loading cycles,

q = number of time intervals with a unique stress-temperature combination,

$(\Delta t)_k$ = total duration of a specific loading k,

$(T_R)_k$ = creep-rupture life under specific loading k.

The allowable damage is based on the degree of creep-fatigue interaction and is determined using the D-diagram. Figure 146 contains the Code Case N-47 D-diagram for the austenitics (304 and 316 stainless steel) and alloy 800H.

Modified 9Cr-1Mo steel has not been incorporated into Code Case N-47; thus, no rules exist for this alloy. The purpose of this task is to assess the applicability of the existing rules to modified 9Cr-1Mo steel and to

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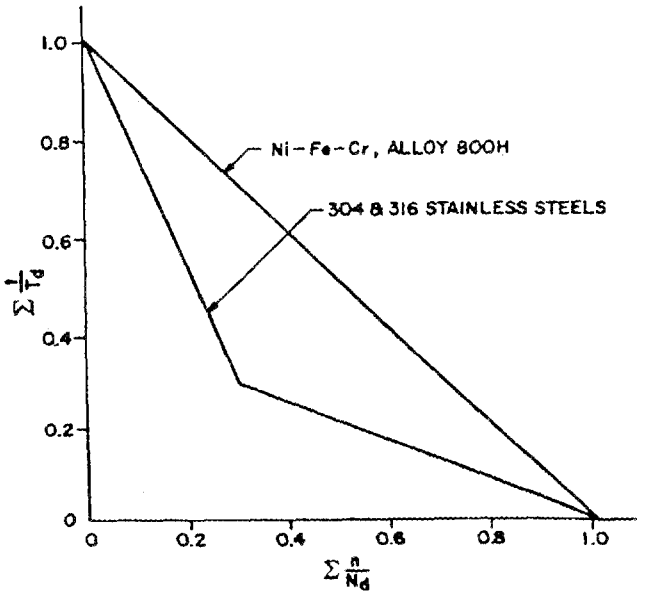

Fig. 146. Creep-fatigue damage envelope.

modify them as necessary for interim use with this alloy. To date, only a few selected scoping tests have been conducted as a basis for criteria assessments.

It was decided early in the failure criteria studies to select one heat of material and one heat treatment representative of material for service and to use these throughout the exploratory testing program. Heat 30383, along with the following heat-treatment procedure, was selected:

1. normalize: heat to 1038°C (1900°F), hold for 1 h, and air cool to room temperature;
2. temper: heat to 760°C (1400°F), hold for 1 h, and air cool to room temperature; and
3. PWHT: heat to 732°C (1350°F), hold for 42 h, and air cool to room temperature.

3.5.3.2 Damage Accumulation

A matrix of creep-rupture tests consisting of four baseline constant-load and eight variable-load and variable-temperature (VLT) tests of modified 9Cr-1Mo steel were conducted during FY 1984 (ref. 149) as a basis for a preliminary assessment of the applicability of the linear time-fraction creep-damage accumulation theory to this alloy. All tests were conducted under subcontract by the Joliet Metallurgical Laboratories, Joliet, Illinois.

Test matrix and damage results are summarized in Table 32. The four baseline constant-load test results (tests C1-C4) compare very well with Sikka, Cowgill, and Roberts' (ref. 150) correlation, which represents average behavior for a number of heats of modified 9Cr-1Mo steel. Computed linear time-fraction damages[149] for the eight VLT tests fall in the range $0.453 \leq D \leq 1.760$. Average damage at rupture (based on a logarithmic average) was 1.07, with two standard errors (~95% confidence level) being equivalent to a factor of 2.27 on damage relative to the average value of 1.07. Thus, these preliminary results indicate statistically that an allowable damage of 0.471 is adequate to ensure, at the 95% confidence level, that life observed in an individual VLT test will be equal to or

exceed predicted life. These results are statistically consistent with results obtained in a more extensive (108 tests) VLT test program completed for type 304 stainless steel.[151] This set of eight VLT tests for modified 9Cr–1Mo, therefore, indicates that the linear time-fraction creep-damage accumulation model applies about equally well to modified 9Cr–1Mo steel and type 304 stainless steel for uniaxial VLT conditions.

There have been no creep-fatigue damage accumulation tests of modified 9Cr–1Mo steel conducted in the high-temperature structural design program. Results of some uniaxial creep-fatigue tests were, however, presented previously in this report. These tests tend to show compressive

Table 32. Results of baseline constant-load and variable-load and -temperature creep-rupture damage accumulation tests of modified 9Cr–1Mo steel (heat 30383, N&T plus PWHT)

Test	Stress, temperature, and time for successive stages of test (MPa/°C/h)[a]			Rupture life (h)	ϵ_R[b] (%)	D_f[c]
	1st stage	2nd stage	3rd stage			
C1	260/538/R			1597.8	10.0	
C2	280/538/R			367.3	12.6	
C3	169/593/R			1380.6	12.1	
C4	190/593/R			374.4	12.2	
C5-1	243/538/1695	260/538/R		2557.9	9.3	1.022
C5-2	243/538/1695	260/538/R		2667.8	11.5	1.101
C-6	260/538/766	243/538/R		4606.0	10.6	1.465
C-7	243/538/1695	169/593/R		3696.1	15.1	1.760
C-8	169/593/779	243/538/R		669.0	11.0	0.453
C-9	260/538/168	243/538/1695	260/538/R	2640.3	9.9	1.081
C-10	260/538/598	243/538/1695	260/538/R	3025.8	10.1	1.357
C-11	243/538/1195	260/538/766	243/538/R	2193.9	9.9	0.889

[a]R indicates tested until rupture; 100 MPa = 14.5 ksi; 538°C = 1000°F; 593°C = 1100°F.

[b]Approximate creep strain at rupture.

[c]Linear time-fraction damage based on baseline correlations for heat 30383.

holds to be more damaging than equivalent tensile holds and also to show a significant mean stress buildup during the course of a test. Creep-fatigue failures in air have shown surface oxidation, although not nearly as extensive as in 2.25Cr-1Mo steel. The failures have been transgranular with no evidence of cavitation. As in the case of 2.25Cr-1Mo steel these characteristics suggest that environment may be playing a role in determining the damage and failure mechanisms in modified 9Cr-1Mo steel. For example, as depicted in Fig. 147, cyclic oxide cracking at the surface in short-term laboratory tests could lead to surface-dominated transgranular failures before sufficient grain boundary cavitation resulted in intergranular failure. Obviously, additional studies are required to understand creep-fatigue failure in modified 9Cr-1Mo steel well enough to establish a sufficiently realistic creep-fatigue damage model for reliably predicting long-term failures. Current recommendations are to (1) conduct a number of short-term creep-fatigue tests under inert conditions to determine if the cavitation-based intergranular failure mode results and provide data for modeling and (2) conduct a selected few long-term tests

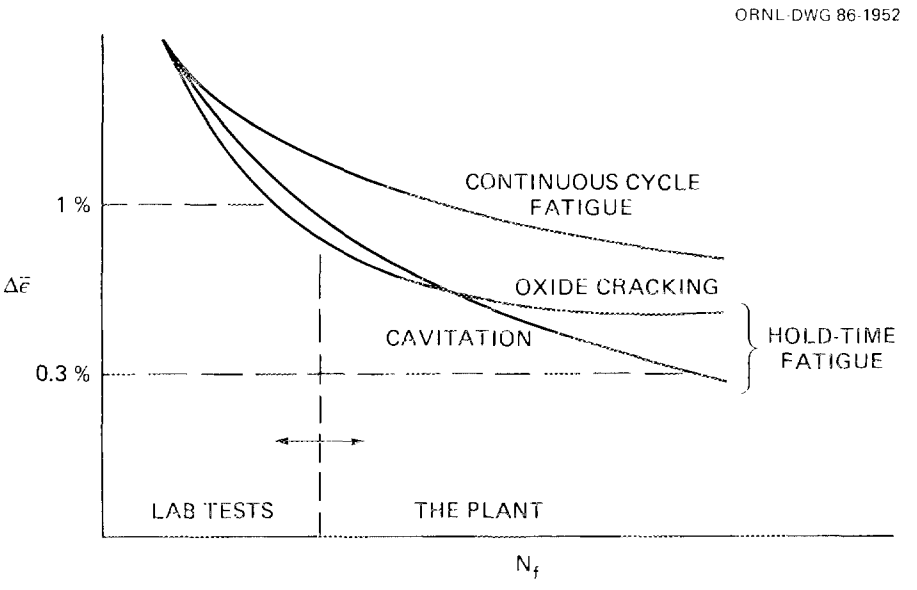

Fig. 147. Depiction of two potential failure mechanisms for modified 9Cr-1Mo steel. (Note: Figure represents idealization and not observed data trend.)

in air at low strain ranges (below oxide cracking strain) to determine if cavitation-based intergranular failures occur, thus validating the more extensive short-term data base generated under inert conditions.

As an interim measure, criteria will have to be generated based upon the available uniaxial creep-fatigue data. The approach to be taken in the development of an interim damage model will use the linear time and cycle-fraction damage theory. It may be possible to incorporate the mean stress effect in the baseline fatigue curve. Both the mean stress-corrected fatigue curve and a D-diagram will need to be developed for modified 9Cr-1Mo steel.

3.5.3.3 Multiaxial Strength Theories

Both fatigue and creep-strength criteria are required as elements of failure criteria. A total of four multiaxial fatigue tests have been completed as a minimum basis for a preliminary assessment of fatigue strength criteria. Two multiaxial creep tests have been completed as a bare minimum basis for assessment of creep-strength criteria.

Results of the four continuous cycling, multiaxial fatigue tests conducted at 538°C (1000°F) are compared with baseline uniaxial data in Fig. 148 on the basis of the von Mises total equivalent strain-range criterion. The von Mises equivalent strain can be expressed as

$$\epsilon_e = \frac{\sqrt{2}}{3}\sqrt{(\epsilon_1 - \epsilon_2)^2 + (\epsilon_2 - \epsilon_3)^2 + (\epsilon_3 - \epsilon_1)^2} \ , \qquad (10)$$

where ϵ_e = equivalent strain and ϵ_1, ϵ_2, ϵ_3 = principal strains.

The multiaxial data compare quite well with the uniaxial data (Fig. 148). On the basis of these results, the preliminary indication is that the von Mises equivalent strain criterion is adequate for fatigue of modified 9Cr-1Mo steel.

Two biaxial creep-rupture tests of tubular specimens of modified 9Cr-1Mo steel were conducted at 538°C (1000°F), one under internal pressure loading and one under pure torsional loading. As previously noted,

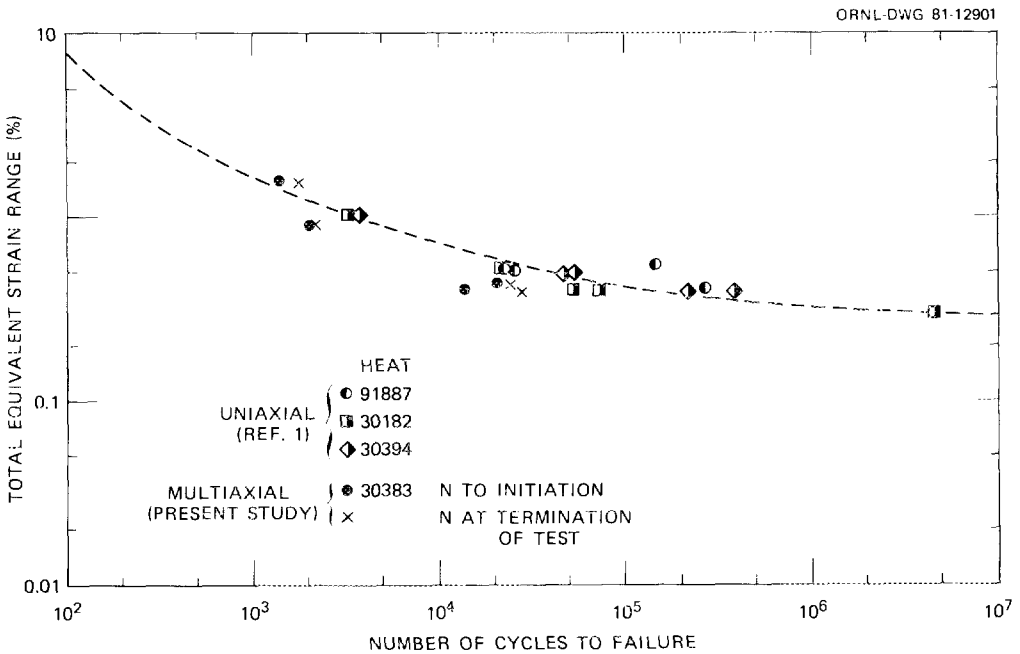

Fig. 148. Comparison of multiaxial with uniaxial low-cycle fatigue data for modified 9Cr–1Mo steel at test temperature of 538°C (1000°F) and equivalent strain rate of 4×10^{-3}/s.

these tests provided the basis for a preliminary assessment of creep-rupture strength criteria. Results of the two tests are summarized in Fig. 149, which shows the two data points relative to isochronous creep-rupture curves based on the classical strength criteria of von Mises, Tresca, maximum principal stress, and a new ORNL-developed criterion.

The analogous isochronous stress-rupture plot incorporating the available data for types 304 and 316 stainless steel and Inconel is shown in Fig. 150 for comparison. Figures 149 and 150 show the results for modified 9Cr–1Mo steel to exhibit the same basic trend as the data for 304 and 316 stainless steel and Inconel. None of the classical criteria is accurate in the second and fourth quadrants; only the new ORNL criterion does an adequate job. The new criterion has been approved for Code Case N-47 use, but only for the austenitic alloys. The new strength theory can be expressed as

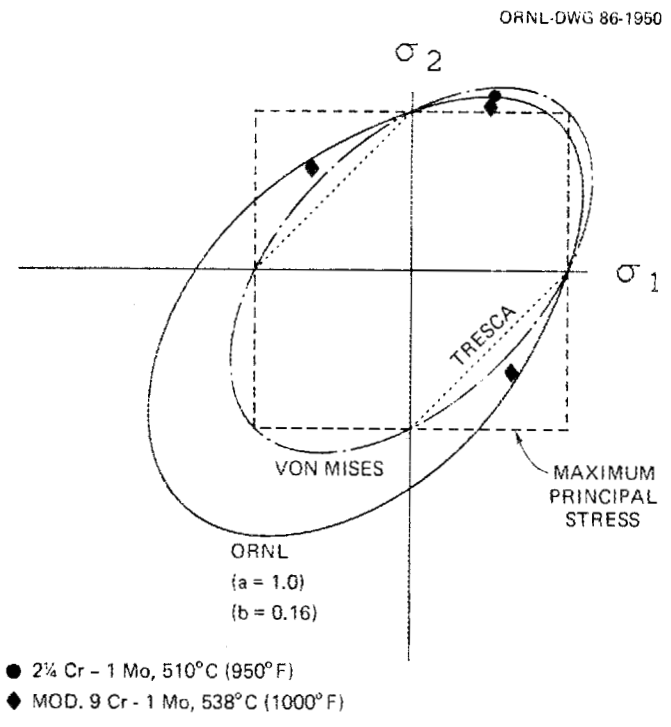

Fig. 149. Plane stress isochronous stress-rupture contours showing relative agreement of available modified 9Cr-1Mo and 2.25Cr-1Mo data with classical strength criteria and new ORNL criterion.

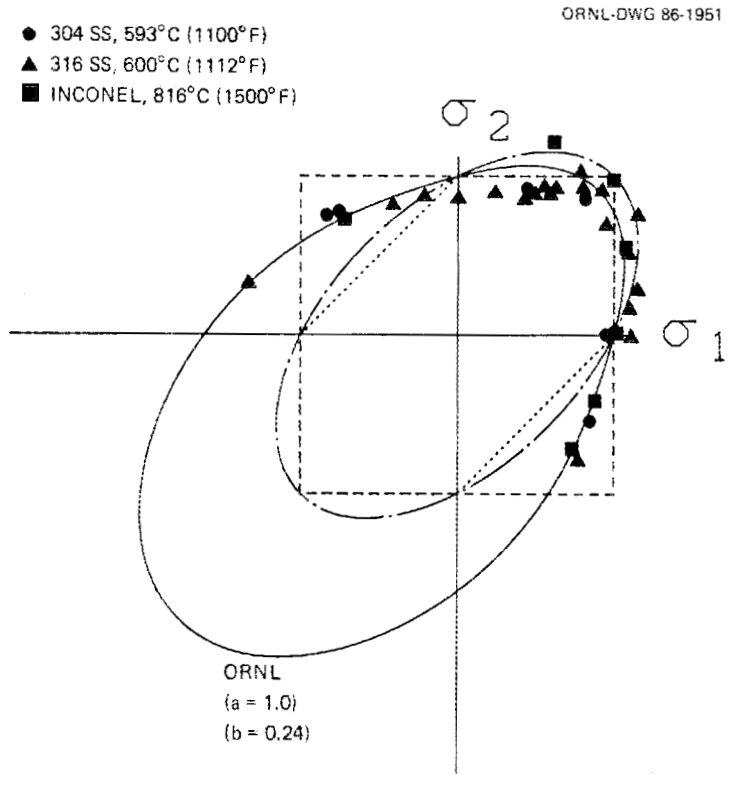

Fig. 150. Plane stress isochronous stress-rupture contours showing relative agreement of available data for types 304 and 316 stainless steel and Inconel with classical strength criteria and the new ORNL criterion.

$$\sigma_e = \frac{3}{2} S_1 \left(\frac{2\bar{\sigma}}{3S_1}\right)^a \exp b \left(\frac{J_1}{S_s} - 1\right) , \tag{11}$$

where "a" and "b" are material constants and

J_1 = first stress invariant

$= \sigma_1 + \sigma_2 + \sigma_3$,

$\bar{\sigma}$ = von Mises equivalent stress

$= \sqrt{(\sigma_1 - \sigma_2)^2 + (\sigma_2 - \sigma_3)^2 + (\sigma_3 - \sigma_1)^2}$,

S_1 = maximum deviatoric stress

$= \sigma_1 - J_1/3$,

S_s = a stress invariant

$= \sqrt{\sigma_1^2 + \sigma_2^2 + \sigma_3^2}$.

For best-life predictions, constants a and b can be material specific; however, a universal value of a = 1.0 with the value of b being alloy type specific (b = 0.24 for the austenitic alloys and b = 0.16 for the ferritic alloys) still gives significantly better creep-rupture life predictions than the classical strength criteria of von Mises, Tresca, or maximum principal stress. Taking a = 1 as a universal value reduces the new strength criterion to the form $\sigma_e = \bar{\sigma} \exp[b(J_1/S_s - 1)]$, approved for Code Case N47 use with the austenitics in conducting inelastic creep-fatigue analyses. The new criterion distinguishes between creep-fatigue life under tensile ($J_1 > 0$) vs compressive ($J_1 < 0$) stress states, with tensile states being the more damaging (for $b > 0$). This same equation form with b = 0.16 is the recommended form for use with the ferritic alloys; however, it has not been adopted for Code use with the ferritics.

3.5.4 Design Criteria

ASME Boiler and Pressure Vessel Code Case N-47 for Class 1 elevated-temperature nuclear components has been described and discussed in previous subsections. A companion Case, N-253 (ref. 152), provides elevated temperature rules for Class 2 and 3 nuclear components. However, Case N-253 has no creep-fatigue or ratchetting rules; rather, it refers the designer to the rules in the Class 1 Code Case N-47. In past loop-type LMR designs, such as the Clinch River Breeder Reactor Plant, the reactor

vessel and primary loop were designed as Class 1 nuclear components; the steam generators were Class 2 nuclear components. More recent LMR designs have sought to limit nuclear safety functions to the primary system; the remainder of the nuclear steam supply system would be nonsafety related and, therefore, could be built to industrial codes and standards. This means that *ASME Boiler and Pressure Vessel Code*, Sect. VIII, Division 1 (ref. 153), would be used for vessels and ASME Power Piping Code ANSI B31.1 (ref. 154) would be used for piping.

Thus, depending on the plant design and the extent of the intended use of modified 9Cr–1Mo steel in the plant, any one or more of four different codes would be used. Currently, modified 9Cr–1Mo has been accepted for use only in Sect. VIII, Division 1; it must be incorporated into the other three as well. Of equal importance, whether nuclear or non-nuclear grade, rules and criteria for use of modified 9Cr–1Mo steel in LMR components must be developed. These considerations mean that the unique LMR loading conditions and failure modes that were discussed previously must be addressed. For nuclear grade components, modified 9Cr–1Mo steel must be adopted into the Case N-47/N-253 framework. For non-nuclear grade, supplemental code cases or owner design criteria must be established.* Either way, the criteria are expected to be similar, except that in the case of non-nuclear components, a greater reliance will be placed on simplified elastic design assessment procedures. In the remaining paragraphs of this subsection, the special 9Cr–1Mo rules and criteria needed for LMR applications and the interim forms that these rules and criteria may take are outlined.

First, it should be emphasized again that the cyclic creep-fatigue and ratchetting behavior of modified 9Cr–1Mo steel has not yet been adequately explored and is not fully understood. However, there are behavioral features of modified 9Cr–1Mo that will likely have to be given special consideration in developing ratchetting and creep-fatigue rules and criteria. These include the following.

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*ASME Sect. VIII and ANSI B31.1 non-nuclear design codes were established for the steady state operating conditions typical of fossil power plants. Short-term thermal transient loadings, seismic events, and sodium water reactions are treated as if the loads lasted 100,000 h; thus, creep-rupture failure is limiting even when creep damage is not an issue. On the other hand, creep-fatigue damage, which is the key failure mode in most LMR situations, is not addressed.

1. Strain softening. Current Code Case N-47 rules are predicted on materials that exhibit strain-hardening behavior; this behavior provides some margin of forgiveness. If an area of a component begins to accumulate plastic strain, the resulting local hardening tends to shift further strain accumulation away from the problem area. With a strain-softening material such as modified 9Cr-1Mo steel in the N&T condition, there could be a synergism leading to increasing strain accumulation.
2. Low creep strains at onset of tertiary creep. For modified 9Cr-1Mo this strain level is around 2% vs 5 to 6% for 2.25Cr-1Mo and 10 to 17% for 316 stainless steel. This behavior, in combination with, or possibly as a result of, the strain-softening behavior, could make strain concentrations and elastic followup a more significant design consideration.
3. Rate dependence of yielding and plastic flow. Modified 9Cr-1Mo exhibits several times more rate dependence than 2.25Cr-1Mo and the austenitic stainless steels. Because several of the shakedown and ratchetting procedures currently in Code Case N-47 depend on yield strength, special consideration may need to be given to this rate dependence.

The effects of these behavioral features on criteria have not yet been determined.

For limiting deformations, retention of the current 1% strain limit, adoption of an improved simplified ratchetting prediction procedure, and development of a more stringent treatment of elastic follow-up are recommended. The recommended simplified ratchetting procedure, which would augment the O'Donnell-Porowski procedure in Code Case N-47 and which would, along with the O'Donnell-Porowski procedure, serve as the sole ratchetting check in rules for industrial grade LMR components, has recently been developed by Sartory.[155] Whereas the O'Donnell-Porowski procedure is conservative and, thus, can only be used for linear temperature gradients, the Sartory modification considers peak temperatures and is, thus, generally conservative. The procedure has been shown to be conservative for a number of austenitic stainless steel axisymmetric geometries. An effort is currently under way to establish the method's conservatism for modified 9Cr-1Mo steel. Using the unified constitutive

equations, we are performing a number of detailed inelastic ratchetting analyses and comparing the results with strain predictions derived from the simplified approach, which uses only the results of elastic analyses.

With regard to elastic follow-up, efforts are under way to develop elastic follow-up rules, especially in piping. However, a modified 9Cr-1Mo piping elastic follow-up test is needed to assess the significance of strain softening on the piping elastic follow-up problem. To assess the softening problem in nonpiping components, a number of detailed inelastic analyses of typical discontinuity geometries are being carried out. These analyses will reveal any unusual ramification requiring special attention.

As is the case for other materials (the austenitic stainless steels, 2.25Cr-1Mo steel, and alloy 800H), the linear damage rule currently used in Code Case N-47 does not adequately describe the time-dependent fatigue processes observed in modified 9Cr-1Mo. Nonetheless, it is recognized that, at least in the near term, the linear damage rule must be adapted to modified 9Cr-1Mo LMR components. This is particularly true for nonsafety components where simplified creep-fatigue rules based on elastic analyses will be used; the only viable procedures currently available are based on the linear damage rules.

For the creep damage portion of the creep-fatigue rule, the multiaxial strength theory proposed in Sect. 3.5.3.3 should be used. For fatigue, the von Mises theory should be used, as in Code Case N-47.

As was previously stated, the mean stress effect observed in some 9Cr-1Mo hold-time fatigue tests may be incorporated into the baseline fatigue curves. Efforts are under way to explore this possibility. Likewise, efforts are under way to establish a D-diagram from available creep-fatigue data. Finally, elastic creep-fatigue rules will be checked against the results of detailed inelastic creep-fatigue predictions (for representative problems) to ensure their conservative applicability to modified 9Cr-1Mo steel.

One final area where special rules will be required, no matter what code is used, is that of weldments. Data presented earlier in this report show that the creep rupture of weld metal can be less than that of modified 9Cr-1Mo steel base metal. This reduced strength should be used to correspondingly reduce the allowable primary stresses as well as the creep damage portion of the creep-fatigue assessment. Modified 9Cr-1Mo steel

weldment fatigue tests show about the same life reduction as exhibited by other alloys. Therefore, the same life reduction factor (0.5 on the allowable number of fatigue cycles) that was recently adopted by the ASME Code Subgroup on Elevated Temperature Design for Code Case N-47 should be used (ref. 156).

Obviously, a concerted effort will be required to flush out these design criteria thoughts and either to (1) present them to the appropriate Code bodies and foster their adoption or (2) prepare supplementary design rules for LMR components, whichever is appropriate.

3.5.5 Confirmatory Structural Tests

Material modeling for both constitutive equations and failure behavior is based for the most part on laboratory tests that are sufficiently simple that they may be precisely interpreted without the need for difficult two- or three-dimensional structural analysis. Usually these tests involve homogeneous stress states throughout the relevant part of the specimen. The models for constitutive and failure behavior thus developed are then combined with analysis techniques and design rules and criteria to provide a design procedure that must be applied to much more complicated reactor structures and loading histories that can involve inhomogeneous stress states, stress and strain concentrations at welds and notchlike discontinuities, repeated thermal transient and ratchetting loading histograms, and elastic follow-up. Simplified methods are then developed to ease the analysis burden of the designer while maintaining adequate conservatisms. In the NRC licensing review of the Clinch River Breeder Reactor Plant, the conservatism of the design procedures at welds and notchlike geometric stress concentrations were key issues. Even in non-safety-related components, the validity of the overall design procedures when applied to complex geometries and loadings requires confirmation. To resolve these issues and confirm the validity of the design procedures, a few critical confirmatory structural tests have been performed, planned, or proposed using modified 9Cr-1Mo steel.

The first modified 9Cr-1Mo steel confirmatory structural test was a cyclically loaded, simply-supported beam test reported by Gwaltney et al.[157] The specimen was a beam of rectangular cross section: 51.39 mm (2.035 in.)

high, 26.24 mm (1.033 in.) wide, and 0.61 m (24 in.) between supports. It was loaded at 593°C (1100°F) by applying a prescribed cyclic center deflection interspersed with creep hold periods as illustrated in Fig. 151. The resulting center load was measured and compared with an inelastic finite-element analysis by using a nonlinear cyclic plasticity model combined with a cyclic strain-hardening creep model. The experimental-analytical comparison is illustrated in Fig. 152 for the rapid cycling and in Fig. 153 for the hold periods. Overall, the agreement is encouraging.

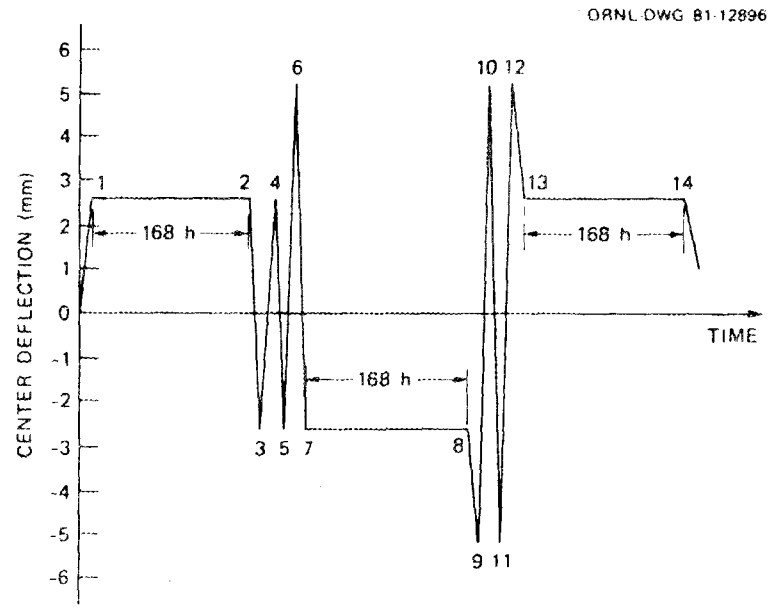

Fig. 151. Applied deflection histograms for center deflection of modified 9Cr-1Mo steel beam.

The cyclic nonlinear plasticity model used in this analysis is not presently included in NE standard F 9-5T, however, and the bilinear plasticity models currently allowed are not considered satisfactory for modified 9Cr-1Mo steel. Because a new constitutive equation will be required in F 9-5T for this material, the ORNL unified equations are considered more promising for a broad range of loading conditions. The beam will be reanalyzed using the equations described under Sect. 3.5.2, constitutive behavior modeling.

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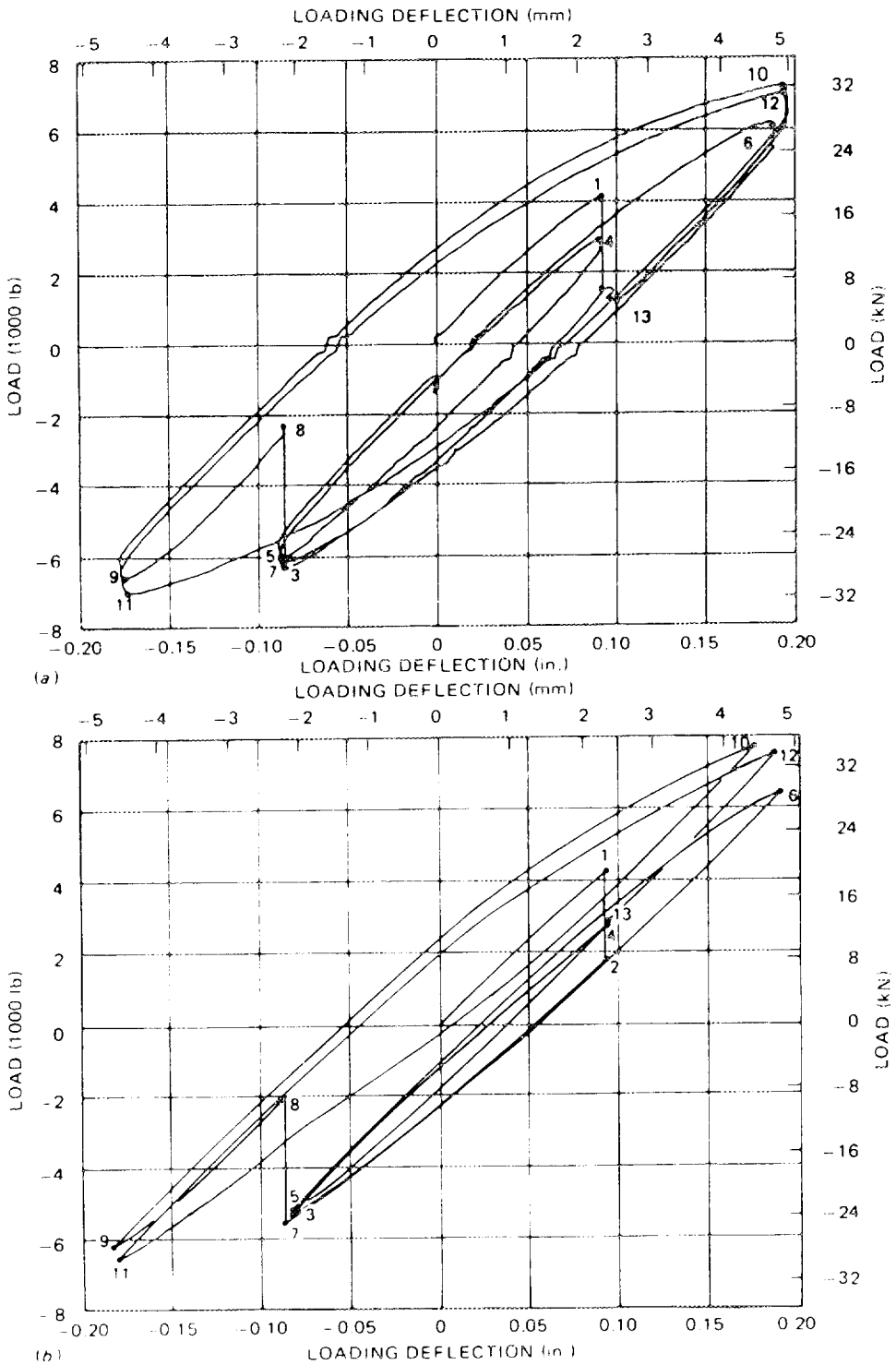


Fig. 152. Comparison of measured and calculated total center load vs deflection for modified 9Cr-1Mo steel beam: (*a*) experimental result, (*b*) analytical result. Numbered points are defined in histogram in Fig. 151.

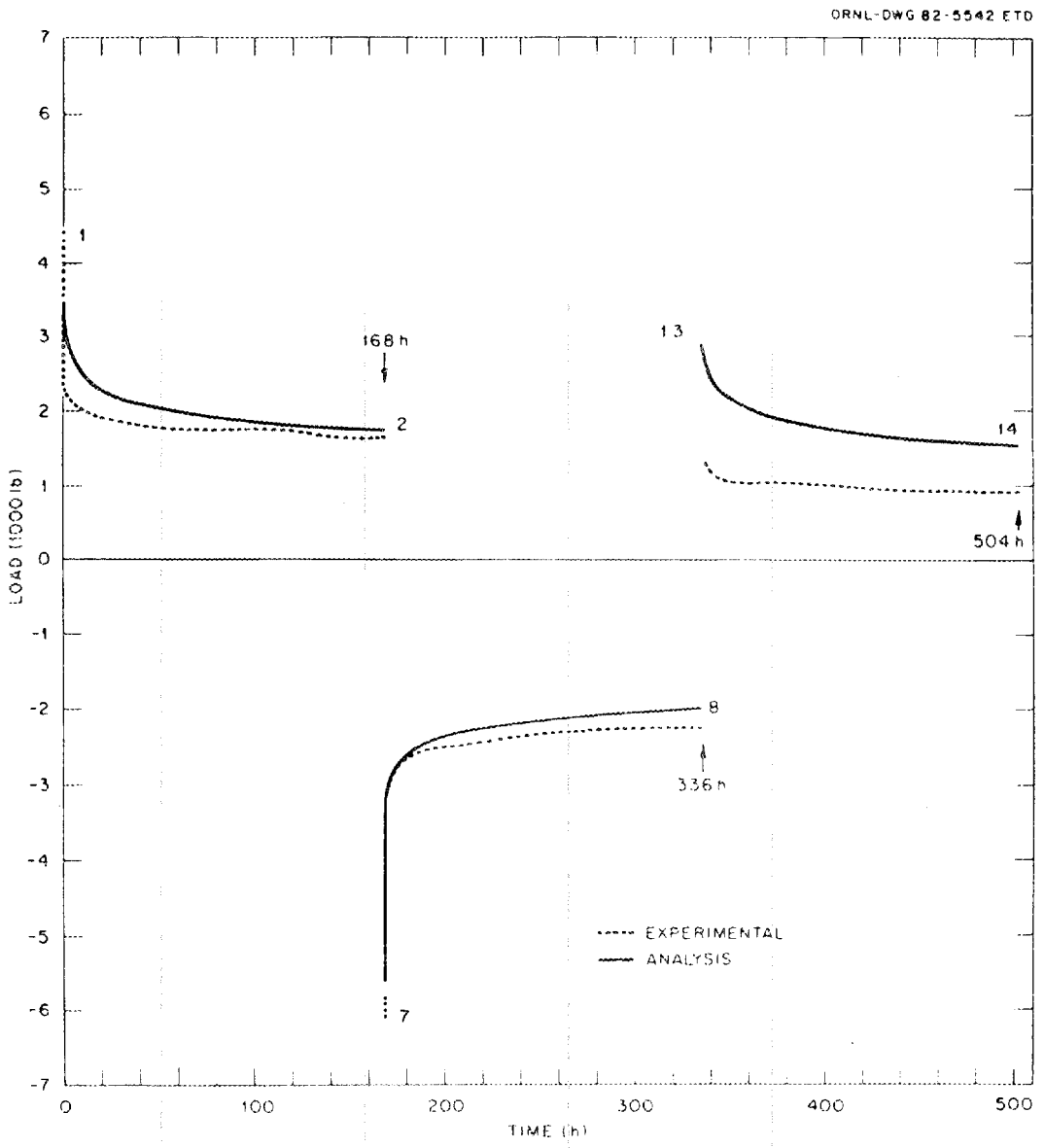


Fig. 153. Comparisons of measured and calculated center time-dependent load vs time for modified 9Cr-1Mo steel beam. Numbered points are defined in histogram in Fig. 151.

Severe cyclic thermal transient loading is especially important in LMR operation and can cause creep-fatigue damage and failure of components. The ORNL Thermal Transient Cyclic Test Facility provides a comparatively inexpensive means of obtaining data for this type of loading. The test facility is depicted in Fig. 154, and the cylindrical specimen and histogram are shown in Fig. 155. Previous tests with cyclically-hardening materials[158] showed the type of severe cracking that could occur and helped

ORNL PHOTO-8098-82A

A 30-kW INDUCTION HEATING FACILITY
PROVIDES THE THERMAL SHOCK
LOADINGS NECESSARY TO FAIL
SPECIMENS IN REASONABLE
CYCLES/TIME

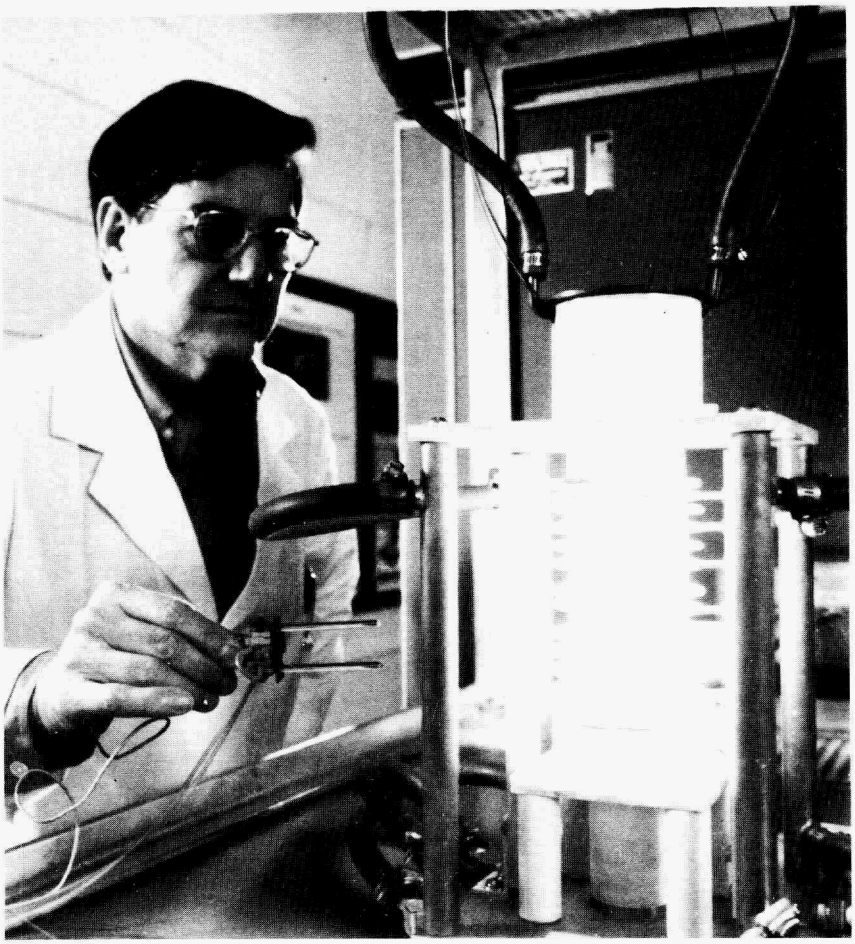


Fig. 154. Type 304 stainless steel thick-walled cylinder specimen after being subjected to thermal upshock on outer surface. Glow of specimen has been accentuated by time exposure.

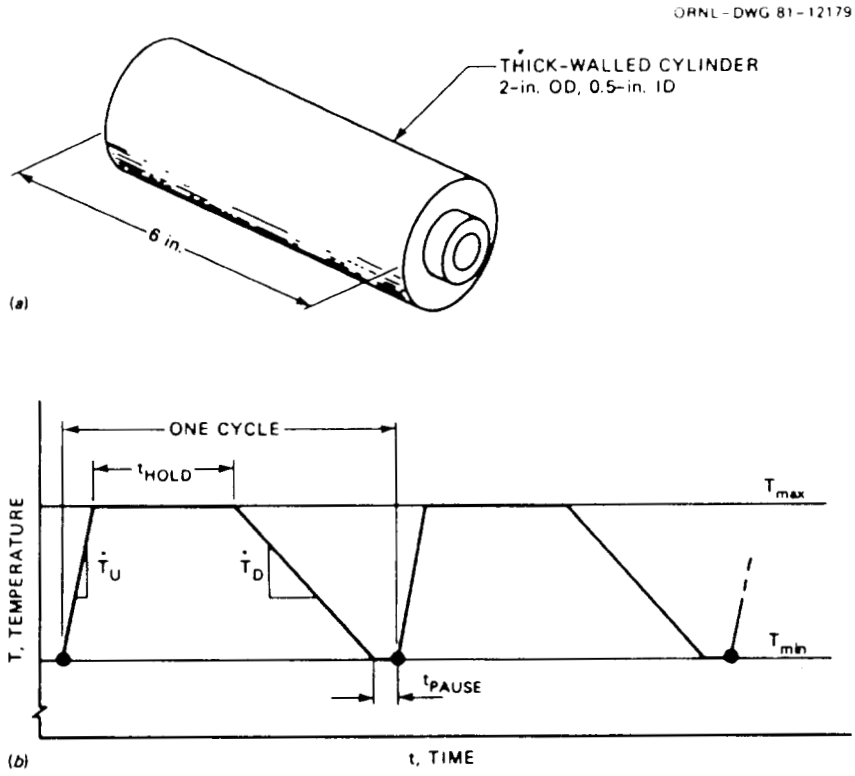


Fig. 155. Thermal creep-fatigue specimen and temperature histogram: (*a*) test specimen (1 in. = 25.4 mm), (*b*) outer-surface temperature histogram.

to confirm design methods. The cyclically softening modified 9Cr-1Mo steel could respond quite differently to this type of loading. Two or three tests are planned to study the failure mechanism and help validate the design methods for this new material. The specimens have been prepared and heat treated.

Although the ORNL Cylinder Thermal Transient Test Facility provides valuable data on failure under thermal cycling, it cannot apply a steady primary load and cannot test under thermal ratchetting conditions. Two thermal ratchetting tests to creep-fatigue failure using unwelded (specimen TT-10) and circumferentially welded (specimen TT-11) modified 9Cr-1Mo steel pipe specimens are planned[159] for testing at ETEC. Both specimens are 1.4 m (54 in.) long with a 0.16-m (6.4-in.) ID and a 19-mm (0.74-in.) wall in the gage length. The welded specimen (see Fig. 156) will have

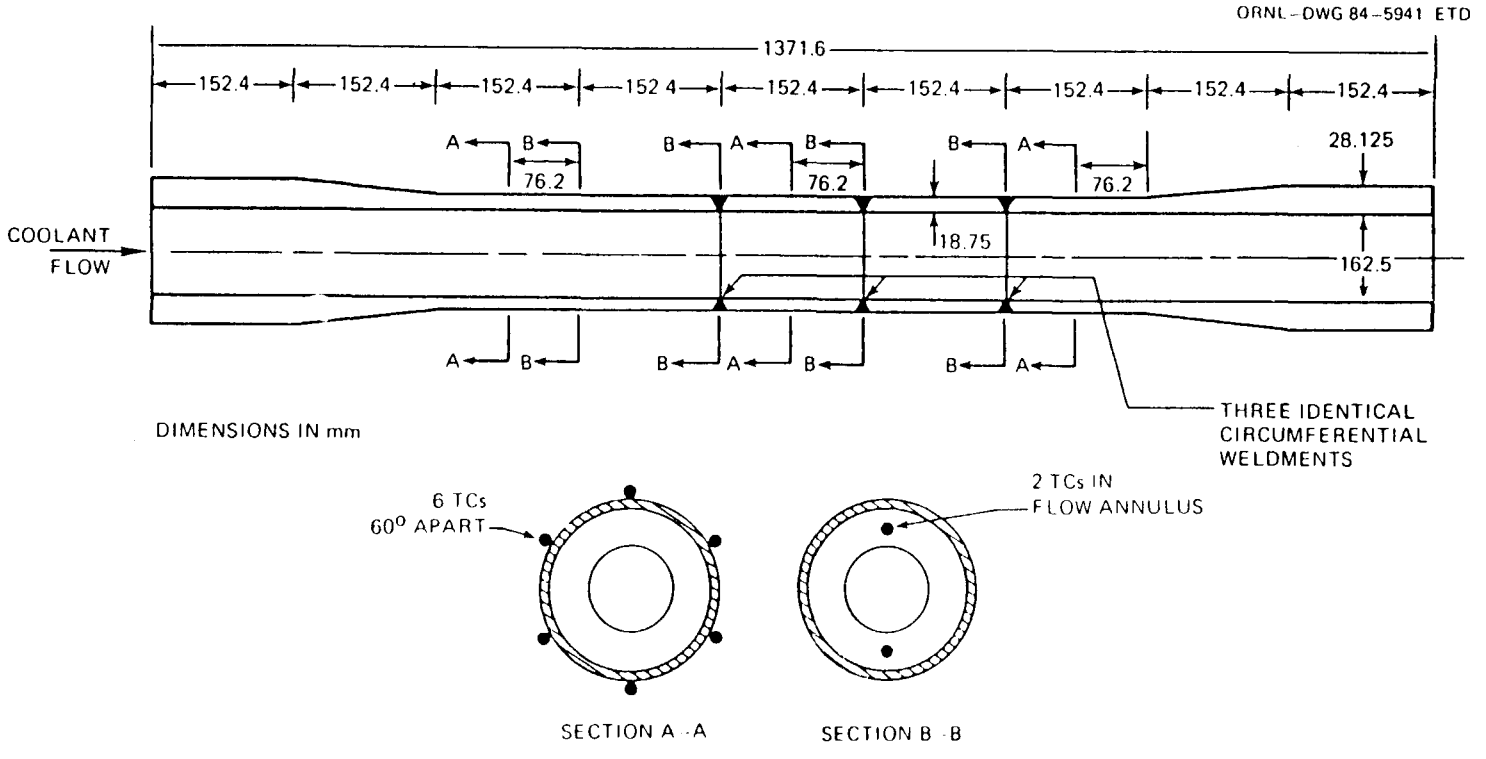


Fig. 156. Modified 9Cr-1Mo steel welded pipe thermal ratchetting specimen TT-11. Dimensions are in mm (25.4 mm = 1 in.).

three identical circumferential welds spaced far enough apart to avoid interactions between adjacent welds. The unwelded specimen has a similar geometry but without welds. Both specimens will be loaded with a steady axial force of 1.1 MN (247,000 lb) and subjected to a severe thermal down-shock at 33°C/s (60°F/s) from 600 to 204°C (1112 to 400°F) and then slowly reheated and held for 16 h at 600°C (1112°F) while creep occurs. These tests will permit validation of the design procedures for predicting strain accumulation and creep-fatigue damage failure under the important thermal ratchetting load conditions. Comparison of the results from the two tests will give a direct indication of the comparative strength of welded and unwelded sections of modified 9Cr-1Mo steel pipe. At present, the test design work at ETEC is essentially complete, and fabrication of the specimens is about to begin.

In elevated-temperature piping systems, elastic follow-up can occur and can increase creep-fatigue damage. The strain softening behavior and low creep strain to the onset of tertiary creep might influence elastic follow-up with modified 9Cr-1Mo steel. The nuclear Code Case states that when simple elastic design methods are used, stress terms with potential for elastic follow-up must be treated as primary stress. This requirement can pose difficulties for the designer. RI is, therefore, planning an elastic follow-up test to be performed at ETEC using a modified 9Cr-1Mo steel piping system. The piping geometry will be the same as used earlier for type 316 stainless steel[160] and will consist of a 0.1-m -diam (4-in.) elbow with an anchored 0.46-m (18-in.) leg on one side and a 3-m (10-ft) leg on the other side. A cyclic prescribed deflection will be applied to the 3-m (10-ft) leg to produce an elastically calculated peak strain of 1%, with hold periods of 100 h at maximum deflection to permit creep. It is expected that this test will demonstrate that the strain concentration caused by elastic follow-up is quite limited and that the Code requirement for treatment of the corresponding stress as primary is excessively conservative. This test is now in the planning stage, and it is expected that a request for Approval-In-Principle will be prepared and that the specimen will be fabricated during this fiscal year.

3.6 CORROSION BEHAVIOR

Ferritic steels have a long history of use in fossil plants. Corrosion by air or water/steam environments is fairly well understood and is reported to be manageable. Compatibility with sodium is an added requirement for LMR applications; however, considerable experience is already available with 2.25Cr-1Mo steel. EBR-II has been operating successfully since 1964 with a 2.25Cr-1Mo steel steam generator. Other operating reactors using 2.25Cr-1Mo steel include PFR (United Kingdom), Phenix (France), BN-350 (Russia), and BOR-60 (Russia). In PFR, leaks in 2.25Cr-1Mo steel tube-to-tubesheet welds that were not given a PWHT were reported[161,162] and were attributed to stress-corrosion cracking. No other unusual corrosion problems involving 2.25Cr-1Mo have been reported.

In general, we expect that modified 9Cr-1Mo steel will behave similarly to 2.25Cr-1Mo, although some differences may occur because of increased chromium concentration and the type and distribution of precipitates in the 9Cr-1Mo steel. Although work has been done to define the corrosion behavior of 9Cr-1Mo steels,[163-167] its corrosion resistance is often compared with 2.25Cr-1Mo steel because of the large successful experience base available with that alloy.

3.6.1 Stress Corrosion

A major advantage of ferritic steels is their resistance to stress-corrosion cracking in chloride or caustic environments. The selection of 2.25Cr-1Mo steel over austenitic stainless steels or alloy 800 in LMR steam generators has generally been because of concern over stress-corrosion cracking. A recent study by General Electric-NSTO concluded that the behavior of modified 9Cr-1Mo steel in caustic is similar to 2.25Cr-1Mo steel. Results of this study can be summarized as follows:

> Modified 9Cr-1Mo plate and standard composition 9Cr-1Mo weld metal exhibit stress corrosion cracking in slow strain rate tests at temperatures to 316°C with caustic concentrations to 10%. At 316°C, ultimate stress is essentially unaffected by caustic; reduction of area values are reduced substantially but still exceed 20%. Degradation of strength values are slightly greater at 232°C in this severe test. Properly heat-treated modified 9Cr-1Mo plate and standard 9Cr-1Mo weld metal are considered to have a low susceptibility to caustic stress corrosion cracking.[168]

Slow strain-rate stress corrosion tests were also recently conducted by GE-NSTO on both 2.25Cr-1Mo steel and modified 9Cr-1Mo steel at 316°C in solutions of 200, 1000, and 2000 ppm chloride as NaCl. Both materials failed by stress-corrosion cracking in 1000 ppm chloride when maintained at either the pitting or protection potential, but they exhibited completely ductile failure in both 200 and 2000 ppm chloride at the open circuit potential.[169] These results indicate that stress corrosion cracking of modified 9Cr-1Mo steel under LMR service conditions is unlikely.

3.6.2 Sodium

Another concern with the use of ferritic steels in LMRs is their susceptibility to decarburization with attendent loss of mechanical strength. In a bimetallic system with stainless and ferritic steels, decarburization can occur from either temperature or compositional driving forces. In an LMR sodium serves as the transport medium as carbon is transferred from the ferritic steel (high carbon activity) to the austenitic stainless steel (low carbon activity). What is most important is what happens to the strength of the ferritic steel as a result of this carbon transfer.

An Intermediate System Mock-Up Loop (ISML) was designed and constructed by GE-NSTO to study carbon transport in a 2.25Cr-1Mo/type 304 stainless steel sodium loop under conditions prototypic of those expected in the Intermediate Heat Transport System of the Clinch River Breeder Reactor.[170] After 30,000 h at 524°C, carbon transport had essentially stopped, and maximum carbon loss from the 2.25Cr-1Mo was 100 to 130 ppm.[171] It was further concluded that, at this temperature, the rate of decarburization of 2.25Cr-1Mo steel was not diffusion controlled; after long times (>1000 h) it appeared to depend upon the decompositon of stable carbides. Ultimate tensile strength decreased by 33% at 524°C, but there was also a 20% reduction at 510°C when material was thermally aged for 300 h. This indicates that in 2.25Cr-1Mo steel the majority of the strength degradation was the result of thermal aging.[172]

The kinetics of carburization/decarburization of 9Cr-1Mo steels were studied in nickel and type 304 stainless steel loops at 500 to 700°C by ANL. The results showed that the 9Cr-1Mo steels were more resistant to carbon transfer than 2.25Cr-1Mo steel.[104] The following are three other conclusions of this study.

1. Conditions of temperature and carbon concentration in sodium for carburization or decarburization of 9Cr-1Mo steel are similar to stainless steels. However, the rate of carbon transfer is slower in 9Cr-1Mo steel.
2. No loss of nitrogen from 9Cr-1Mo steels occurs at temperatures below 550°C.
3. At temperatures >600°C, 9Cr-1Mo steels can decarburize to a surface carbon concentration of <200 ppm depending upon composition and carbide structure. Steels that decarburized to low carbon concentrations contained either M_2X (a metal nitride/carbide phase) or M_6C as the only stable carbide.

Apparently, based on these results with both 2.25Cr-1Mo steel and 9Cr-1Mo steels, below ~540°C (1000°F) no significant changes in properties of modified 9Cr-1Mo steel would be expected as a result of carburization/decarburization reactions in LMR systems. However, additional data on the effects of heat treatment on mechanical properties of sodium exposed and decarburized modified 9Cr-1Mo steel should be obtained.

3.6.3 Steam

The behavior of modified 9Cr-1Mo steel and other alloys in superheated steam at 482 and 538°C was reported by Griess and Maxwell.[173] Their investigation, which terminated after 28,000 h, was conducted in a once-through alloy 617 loop that received steam from the superheater circuit of the Bartow Power Plant of the Florida Power Corporation. Table 33 lists the compositions of all alloys tested. At both temperatures all materials exhibited parabolic oxidation kinetics during the first 8000 h and then subsequently oxidized at constant rates lower than those obtained from extrapolation of the parabolic curves. Modified 9Cr-1Mo steel (Fig. 157) and 2.25Cr-1Mo steel showed little difference in oxidation rate in this study, averaging 0.14 and 0.33 mils/year at 482 and 538°C, respectively, after the first year. At 482°C there was little evidence of exfoliation, but at 538°C exfoliation of almost all of the Cr-Mo steels was noted. In some short-term runs there was some evidence that increasing the silicon content of modified 9Cr-1Mo steel may decrease oxidation in superheated steam (Table 34).

Table 33. Composition of test materials

Alloy	Heat	Note	Composition (wt %)								
			C	Cr	Ni	Mo	Fe	Mn	Si	V	Other
Alloy 800	HH7058	Weight change and welded specimens	0.08	19.77	33.90		43.46				0.44 Ti, 0.46 Al
Alloy 800	HH2521 A	Mill annealed specimens	0.05	20.78	31.39		45.28		0.36		0.46 Ti, 0.38 Al
Alloy 82	NX6537 D	Filler metal	0.02	20.03	72.50		1.16	3.07	0.09		0.39 Ti, 2.51 Nb
Alloy 617	XX01A3US		0.07	20.30	57.35	8.58	1.01	0.05	0.16		11.72 Co, 0.76 A
304 stainless steel	8082800		0.06	18.04	8.90		Bal.	1.84	0.55		
Sandvik HT9	451738	Annealed and ground specimens	0.19	11.4	0.50	1.00	Bal.	0.50	0.26	0.27	0.49 W
	451774	Specimens supplied by GA[a]	0.20	11.2	0.51	0.97	Bal.	0.50	0.32	0.29	0.51 W
	454191	Specimens supplied by GA[a]	0.18	11.3	0.50	0.93	Bal.	0.52	0.35	0.29	0.51 W
Sumitomo 9Cr-1Mo		Specimens supplied by GA[a]	0.04	8.50	0.13	2.06	Bal.	0.56	0.29		
9Cr-1Mo steel	316381 A	Standard air melt	0.08	8.79	0.38	1.00	Bal.	0.34	0.46		
	XA-3177	ESR[b] - supplied by CE[c]	0.09	9.58	0.08	0.80	Bal.	0.44	0.17	0.15	0.46 W, 0.11 Nb
	XA-3178	ESR[b] - supplied by CE[c]	0.10	9.52	0.16	1.55	Bal.	0.50	0.21	0.30	0.14 Nb
	91887	ESR[b] - supplied by CE[c]	0.10	9.22	0.08	1.01	Bal.	0.38	0.06	0.22	0.15 Nb
	31377	Argon-oxygen decarburized	0.05	10.43	0.44	1.02	Bal.	0.56	0.03	0.20	
	31378	Argon-oxygen decarburized	0.07	9.60	0.29	0.86	Bal.	0.44	0.01	0.16	
	31446	Argon-oxygen decarburized	0.06	10.21	0.17	0.96	Bal.	0.61	0.03	0.27	
	F-5349	Argon-oxygen decarburized	0.10	8.75	0.11	0.93	Bal.	0.42	0.37	0.21	
	30383	Argon-oxygen decarburized	0.09	8.29	0.09	1.00	Bal.	0.45	0.43	0.20	0.07 Nb
	Lot 35	Air melt - supplied by CE[c]	0.11	9.22	0.19	0.96	Bal.	0.51	0.71		
	15965	Air melt - supplied by CE[c]	0.11	9.38	0.07	0.99	Bal.	0.59	0.63		
7Cr-1Mo steel	XA-3477	ESR[b]	0.09	7.05	0.05	0.98	Bal.	0.34	0.17		0.14 Nb
	XA-3478	ESR[b]	0.09	7.00	0.05	1.02	Bal.	0.34	0.14		0.37 Nb
5Cr-1Mo	XA-3475	ESR[b]	0.10	5.18	0.16	1.02	Bal.	0.32	0.14		
	XA-3476	ESR[b]	0.09	5.01	0.06	1.01	Bal.	0.34	0.11		0.38 Nb
2.25Cr-1Mo steel	84042	Weight change and welded specimens	0.11	2.01		0.98	Bal.	0.54	0.22		
	X6216	Autogenous weld specimens	0.12	2.19		1.00	Bal.	0.44	0.26		

[a]General Atomic Company, San Diego, Calif.

[b]Electroslag remelt.

[c]Combustion Engineering, Inc., Chattanooga, Tenn.

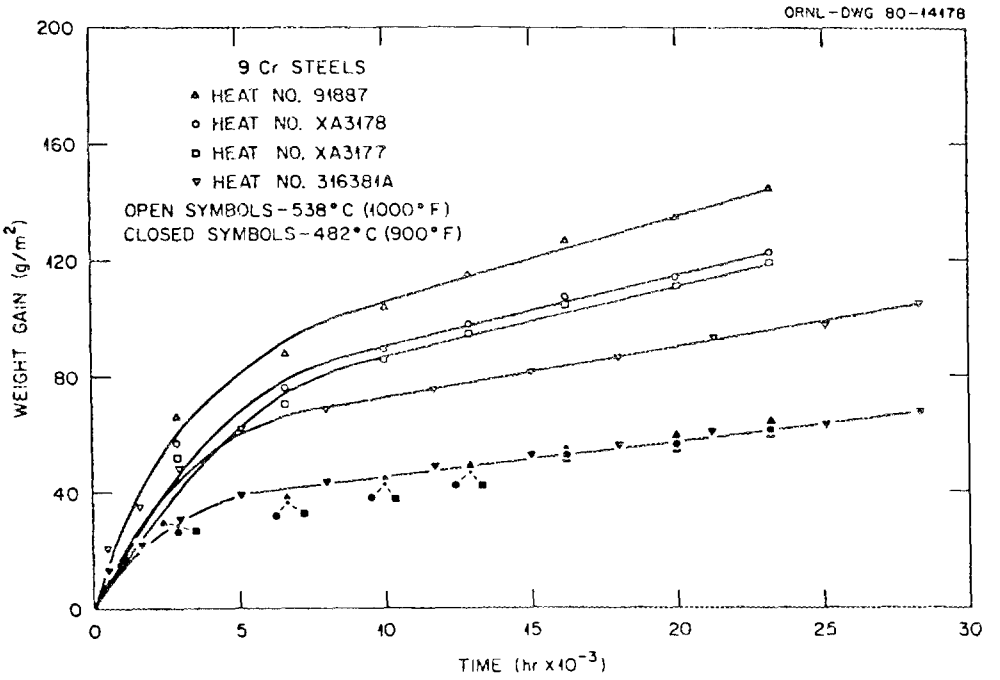


Fig. 157. Weight gains of different heats of 9Cr-1Mo steel in superheated steam.

Table 34. Average weight gains of 9Cr-1Mo steels in superheated steam

Heat	Silicon content (wt %)	Temperature (°C)	Average weight gain (g/m^2)	
			458 h	667 h
30383	0.43	482	9.5	11.2
		538	24.7	30.9
15965	0.63	482	6.8	8.8
		538	23.3	28.5
Lot 35[a]	0.71	482	8.5	10.5
		538	22.3	28.7

[a]British material supplied by Combustion Engineering, Chattanooga, Tenn.

In a later study conducted by Banks et al. at Risley Nuclear Power Development Laboratories (RNL), steam corrosion tests on 2.25Cr-1Mo and 9Cr-1Mo steels to 40,000 h were reported.[174] For times to 1 year, parabolic behavior of both alloys at 450 to 550°C was observed. In longer exposures, oxidation rates became linear as previously reported by Griess.[173] At 550°C the oxide on 2.25Cr-1Mo steel either spalled or showed a higher linear oxidation rate. The 9Cr-1Mo steel showed cyclic protective kinetics (Fig. 158). At low temperatures little difference was noted between 2.25Cr-1Mo and 9Cr-1Mo steels, especially for specimens containing the same silicon contents. However, after 11,000 h at 525°C, there was nearly a 50% increase in weight gain of 2.25Cr-1Mo steel containing 0.35% silicon compared with 9Cr-1Mo containing 0.71% silicon.

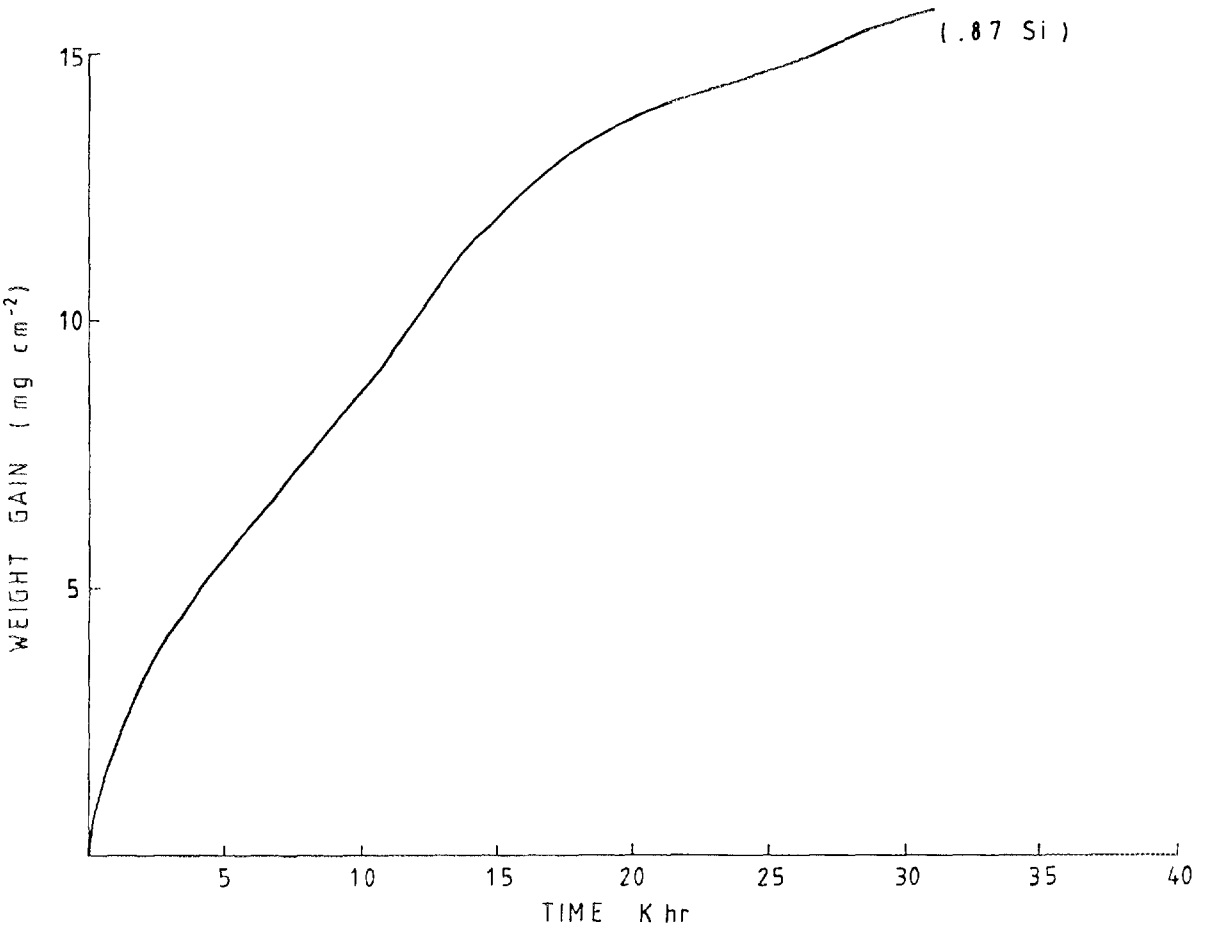


Fig. 158. Oxidation of 9Cr-1Mo specimen E-7 (Cast E) in steam at 823 K and 6.9 MPa. *Source*: J. C. Griess, J. H. DeVan, and W. A. Maxwell, *Oxidation of Heated 2 1/4 Cr-1 Mo Steel Tubing in Flowing Superheated Steam*, ORNL-5373, April 1978.

These results indicate that the steam corrosion behavior of modified 9Cr-1Mo steel should be similar to, but slightly better than, 2.25Cr-1Mo steel. Furthermore, if improvements in steam corrosion resistance were required, the effect of silicon content should be more thoroughly investigated. Also, in LMR steam generators, heat fluxes across tube walls can be considerably higher than those in fossil plants. On tubes of 2.25Cr-1Mo steel in steam with a heat flux of 126 KW/m^2 across the corroding surfaces and temperatures increasing from 510 to 540°C along their heated lengths, oxides were porous, cracked, and poorly adherent to the base metal in contrast with those formed in superheated steam under isothermal conditions.[175] Similar testing should be conducted to determine the effects of heat flux on oxide formation in modified 9Cr-1Mo steel.

3.7 INSPECTION

3.7.1 Techniques

Inspection requirements for nuclear components are determined primarily from the *ASME Boiler and Pressure Vessel Code* rules. The applicable sections include Sect. III for fabrication, Sect. XI for in-service inspection (ISI), and Sect. V for nondestructive examination (NDE) methods. The NDE methods prescribed in Sect. III include liquid penetrant, magnetic particle, radiographic, and ultrasonic examination techniques for various material configurations and welds. Section XI emphasizes ultrasonic ISI of pressure boundary welds and eddy-current techniques for steam generator tubing. Section V contains the general NDE methods for reference by the other sections.

Thus far, only limited efforts have been made to apply these NDE methods to 9Cr-1Mo steel alloy. However, prior application of the methods to other alloys with similar material characteristics (i.e., 2.25Cr-1Mo steel) provide confidence that technology can be adapted, where necessary, to perform necessary examinations of modified 9Cr-1Mo steel components. For example, the liquid penetrant method, which is limited to the examination of clean, metallic surfaces, is affected very little, if any, by the composition of the steel and should be readily applicable to 9Cr-1Mo steel. The magnetic particle method is limited to the examination of clean surfaces

and near-surface regions of ferromagnetic steels. Although no documented evidence is available, the potential differences in magnetic properties of 9Cr-1Mo steel compared with other ferromagnetic steels are not expected to be sufficient to cause problems for magnetic particle examination. On the other hand, volumetric examinations provided by ultrasonic, radiographic, and eddy-current methods could be affected by the composition, microstructure, and other properties of 9Cr-1Mo steel, but these differences are not expected to pose difficult problems. Previous developmental programs in NDE and ISI technology were performed to ensure the availability of adequate techniques for manufacturing examination and ISI of LMR components. For steam generator applications, the emphasis has been on the 2.25Cr-1Mo alloy steel, although some recent activities have now begun to focus on 9Cr-1Mo. With comments about applicability to 9Cr-1Mo steel, the following sections briefly describe some of the past developmental work.

3.7.2 Developments for Heat Exchangers

The major activities for development of NDE techniques for heat exchangers have centered on 2.25Cr-1Mo steel tubing for the Clinch River Breeder Reactor steam generator and the alternate double-wall steam generator designed by Westinghouse. The developments encompassed techniques for both manufacturing examination and in-service inspection.

3.7.2.1 Clinch River Breeder Reactor Steam Generator

For manufacturing examination of the Clinch River Breeder Reactor steam generator, both radiographic and ultrasonic techniques were developed for the tube-to-tubesheet welds. Radiographic development included investigations of both isotopic and X-ray sources, with a microfocus rod-anode X-ray unit being selected as clearly superior.[176,177] Techniques that were capable of finding discontinuities as small as 0.08 mm (0.003 in.) were developed. Techniques and technical support were provided to Atomics International, the contractor responsible for fabrication of the Clinch River Breeder Reactor steam generator. Boreside ultrasonic probes, techniques, and scanning equipment were developed for examination of the tube-to-tubesheet welds,[178-180] and the system was demonstrated to

be capable of detecting cracklike discontinuities with depths of 0.18 mm (0.007 in.). The ultrasonic technique for flaw detection in the tube-to-tubesheet welds was not incorporated into fabrication of the steam generator, but the technology was used as part of the development for ISI. Although the X-ray and ultrasonic techniques were developed for 2.25Cr-1Mo steel, it is anticipated that they will be applicable to 9Cr-1Mo steel. However, this should be confirmed experimentally on appropriate tube-to-tubesheet welds and calibration standards.

For ISI, both eddy-current and ultrasonic methods were investigated. Because of its advantage in speed of application, emphasis was on eddy currents as the primary examination method; ultrasonics was a supplementary examination for the tube-to-tubesheet welds, as noted previously. The ferromagnetic properties of 2.25Cr-1Mo steel limited the extent of penetration of eddy currents through the 2.8-mm (0.109-in.) wall. A direct-current magnetization probe and circuit were incorporated into the eddy-current probe and instrument to reduce the effective magnetic permeability and, thus, allow the eddy currents to penetrate the tube wall. A computer-based multifrequency eddy-current instrument that was capable of detecting wall thickness variations as small as 10% was developed.[181] It is expected that the eddy-current technique would be equally applicable to 9Cr-1Mo tubing in the same size if the electrical and magnetic properties are similar. This should be confirmed on appropriate tubing and calibration standards.

3.7.2.2 Westinghouse Double-Wall-Tube Steam Generator

For manufacturing examination of the double-wall-tube steam generator, radiographic techniques were developed for the 2.25Cr-1Mo tube-to-tubesheet welds by using the microfocus rod-anode X-ray unit that had been used to inspect the Clinch River Breeder Reactor steam generator at Atomics International. For the fillet weld configuration, special masks for adjustment of thickness variation were developed. With section thicknesses up to 9.5 mm (0.375 in.), pores 0.18 mm (0.007 in.) and larger were detectable. Similar techniques and results are expected if the same weldments are 9Cr-1Mo; however, experimental confirmation is recommended.

Both ultrasonic and eddy-current techniques were developed for ISI of the 2.25Cr-1Mo steam generator tubes. The eddy-current technique was again selected as the primary method, with ultrasonics providing supplementary high-resolution examination in selected regions. The system used to examine the Clinch River Breeder Reactor tubing was initially employed for the eddy-current work,[182] but it was determined that boreside direct-current magnetization was inadequate to allow penetration of the eddy currents through the thick double-wall tube [4.6 mm (0.182 in.) thick]. To overcome this problem, feasibility of a pulsed system approach was demonstrated,[183] and a new instrument[184] that uses very short, high-powered pulses on a probe coil to magnetically saturate the tube was developed. The modulated shape of the pulse can be monitored to extract raw data, allowing the suppression of unwanted variables and unambiguous display of the parameters of interest. The system was demonstrated to be sensitive to cracklike flaws and wall thickness changes as small as 10% of the wall.

Modular ultrasonic probes, conceptually similar to those used for the Clinch River Breeder Reactor, were designed and developed to examine the double-wall tube for longitudinal and transverse flaws and to monitor the ultrasonic transmission through the interface between the inner and outer tubes of the double-wall tubing.[185] Flaws ~5% of the wall thickness were demonstrated to be detectable in the inner tube. Similar flaws are detectable in the outer tube if the nature of the interface allows sufficient ultrasonic transmission. Confirmation of the applicability of these techniques to 9Cr-1Mo tubing is needed.

3.7.2.3 Advanced Liquid-Metal Reactors (LMRs)

The IHX for the advanced LMR being designed by RI has modified 9Cr-1Mo steel as the reference material. There is significant economic and technical benefit to be able to perform an ISI on the IHX without having to remove it from the reactor vessel. Development of ultrasonic and eddy-current techniques is under way to allow examination of the IHX in place. A major requirement will be for probes to function at the hot

standby temperature of ~200°C (400°F), preferably with sodium in the bore of the tube with the probe. The initial task is to develop probes that will withstand the temperature. It is anticipated that the inspection techniques and instrumentation systems will be similar to those developed for the Westinghouse steam generator after necessary modifications for the tubular configuration, temperature, and material change to modified 9Cr-1Mo steel. The development of probes should also be beneficial toward meeting needs for ISI of the steam generators.

3.7.3 Developments for Tubular Products and Welds

Only limited applications of NDE techniques have been made to 9Cr-1Mo alloy steel components. A few samples of small-diameter tubing [5.84 by 0.38 mm (0.230 by 0.015 in.)] were fabricated and examined ultrasonically.[18] Good response was obtained from reference flaws only 0.04 mm (0.0015 in.) deep. It was shown that normalized and tempered material with reduced grain size (and associated ultrasonic background noise) provided a more confident ultrasonic examination. In-process anneal on one sample indicated the potential for sufficient reduction in grain size and ultrasonic noise.

A test section of modified 9Cr-1Mo steel pipe[120] [232-mm diam × 12.7-mm wall × 610 mm long (9.1 × 0.5 × 24 in.)] was welded between two spool pieces of type 304L stainless steel [152 mm (6 in.) long] for exposure in the SCTL at the ETEC. The 9Cr-1Mo pipe was examined ultrasonically, and the welds were examined radiographically with no material-dependent difficulties being noted. After delivery to ETEC, the welds were reexamined ultrasonically with no defects being noted.

Earlier breeder reactor designs have incorporated stainless steel piping and welds. These welds have proven to be difficult to inspect ultrasonically (e.g., for ISI) because of the large epitaxial grain growth, and extensive studies have been conducted by many to overcome the problems. It is anticipated that the microstructure of 9Cr-1Mo steel welds will be easier to examine. An important phase of the NDE development at ORNL was the development of ultrasonic signal processing techniques for the quantitative measurement of flaw size.[186] (Conventional NDE provides only

qualitative detection.) The new techniques for determining accurate flaw size should be applicable to 9Cr-1Mo steel, and the signal processing algorithms should be modified for this material.

3.7.4 Summary of Remaining Needs

Based on limited samples, conventional NDE techniques applicable to other steel alloys are apparently equally applicable to 9Cr-1Mo steel with little material-dependent effects. This should be confirmed on modified 9Cr-1Mo steel specimens of appropriate configurations. Ultrasonic signal processing algorithms for quantitative flaw characterization should be confirmed or modified for 9Cr-1Mo. Current activities should continue to develop high-temperature probes, techniques, and equipment for the ISI of modified 9Cr-1Mo tubing for the IHX (and steam generators) of advanced LMRs.

3.8 CODES AND STANDARDS

3.8.1 Specifications (ASTM and ASME, Sect. II)

For commercial applications, alloy specifications must be approved by the ASTM. Inclusion of specifications in the *ASTM Book of Standards* also permits their automatic acceptance by the *American Society for Mechanical Engineers Boiler and Pressure Vessel Code* (ASME Code), Sect. II. The approval of specifications by the ASTM requires information on the acceptable range for each element specified and information on the mechanical properties of the alloy. The various actions leading to publication of specifications are shown in the flow diagram in Fig. 159. A request for approval of tubing (A213) and plate (A387) was initiated in May 1981. Since that time, requests for other product forms have also been made. The current status of approval of various forms in the *ASTM Book of Standards* and ASME Code, Sect. II is presented in Table 35. All of the product forms that were requested for approval have now been accepted in the *ASTM Book of Standards*. Except for the most recently approved specifications, all of those published in the *ASTM Book of Standards* have also been published in the ASME Code, Sect. II.

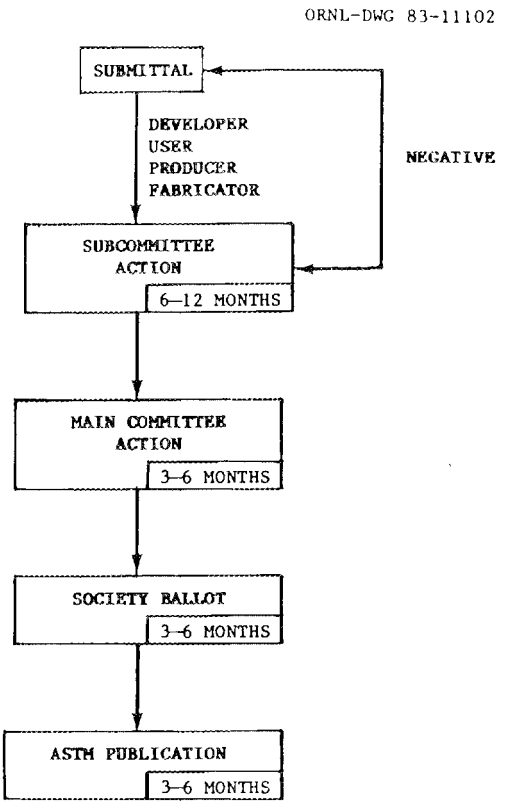


Fig. 159. From 15 to 30 months are required to obtain ASTM approval for each product form of a new alloy.

The only specifications not yet applied for are those for bar product and castings. Bar product has similar composition and properties to product forms already approved and, thus, will likely be readily approved when requested. However, the range of compositions for castings is not fully optimized at present and, therefore, will not be requested for approval until this is completed.

3.8.2 *ASME Boiler and Pressure Vessel Code* Rules

The Boiler and Pressure Vessel Committee of the ASME is responsible for establishing rules of safety governing the design, fabrication, and inspection of boilers, pressure vessels, and nuclear components during construction. These rules cover the establishment of maximum design and operating pressures based on the structural material properties, method of

Table 35. Status of approval of various product forms of modified 9Cr-1Mo steel in ASTM specifications and ASME Code, Sect. II

Specification				ASTM	ASME Code, Sect. II	Comments
Designation		Grade	Description			
ASTM	ASME					
A 182	SA-182	F91	Forged or rolled alloy-steel pipe flanges, forged fittings, and valves and parts for high-temperature service	*1985 Annual Book of ASTM Standards*, Sect. I, Vol. 01.01, pp. 93-106	ASME Code, Sect. II, Part A, Winter 1985 Addenda, issue date Dec. 31, 1985, pp. 171-82	
A 199		T91	Seamless cold-drawn intermediate alloy-steel heat-exchanger and condenser tubes	*1986 Annual Book of ASTM Standards*, Sect. I, Vol. 01.01, pp. 183-91		Not yet in ASME Code
A 200		T91	Seamless intermediate alloy-steel tubes for refinery service	*1985 Annual Book of ASTM Standards*, Sect. I, Vol. 01.01, pp. 136-41		Not a pressure-related application; thus, will not be included in the Code
A 213[a]	SA-213	T91	Seamless ferritic and austenitic alloy-steel boiler, superheater, and heat-exchanger tubes	*1985 Annual Book of ASTM Standards*, Sect. I, Vol. 01.01, pp. 152-59	ASME Code, Sect. II, Part A, Winter 1985 Addenda, issue date Dec. 31, 1985, pp. 223-28.1	
A 234		WP91	Piping fittings of wrought carbon steel and alloy steel for moderate and elevated temperatures	*1985 Annual Book of ASTM Standards*, Sect. I, Vol. 01.01, pp. 174-83		Not yet in ASME Code

Table 35. (continued)

Specification						
Designation		Grade	Description	ASTM	ASME Code, Sect. II	Comments
ASTM	ASME					
A 335	SA-335	91	Seamless ferritic alloy-steel pipe for high-temperature service	*1985 Annual Book of ASTM Standards*, Sect. I, Vol. 01.01, pp. 276-83	ASME Code, Sect. II, Part A, Winter 1985 Addenda, issue date Dec. 31, 1985, pp. 375-81	
A 336		F91	Steel forgings, alloy for pressure and high-temperature parts	*1986 Annual Book of ASTM Standards*, Sect. I, Vol. 01.01, pp. 290-300		Not yet in ASME Code
A 387/ A 387M	SA-387/ SA-387M	91	Pressure vessel plates, alloy steel chromium-molybdenum	*1985 Annual Book of ASTM Standards*, Sect. I, Vol. 01.03	ASME Code, Sect. II, Part A, Summer 1985 Addenda, issue date June 30, 1985, pp. 491-93	
A 826		T91	Austenitic and ferritic stainless steel duct tubes for breeder reactor core components	*1985 Annual Book of ASTM Standards*, Sect. I, Vol. 01.01, pp. 983-89		Not yet in ASME Code

[a]First product to gain acceptance in ASTM and ASME Code, Sect. II.

construction and fabrication, inspection, etc., and are published in a document called the *ASME Boiler and Pressure Vessel Code*. Supplemental Code Cases may also be requested to qualify a particular material or circumstance not covered by that section of the Code. All materials allowed by the various sections of the Code must be in accordance with ASME Materials Specifications (Sect. II) unless otherwise provided by a Code Case. (In general, the requirements of the ASTM Specifications are identical.)

Although modified 9Cr-1Mo steel is being developed for nuclear component applications, we recognized that it would likely first be used for other applications, and, therefore, the strategy employed was to sequentially seek ASME Code approval for modified 9Cr-1Mo steel as follows:

1. Section I: Power Boilers,
2. Section VIII: Pressure Vessels, and
3. Section III: Nuclear Power Plant Components [including Code Case N-47 (ref 122)].

The overall process for securing ASME Code approval is shown for Sect. I in Fig. 160. Note that after submittal of an initial data package to ASME, there must be a request for its approval by a user, producer, etc., before any further action by a subgroup or subcommittee is taken. Thus, although a first data package on modified 9Cr-1Mo steel was submitted in June 1982, consideration by the appropriate subgroups was not initiated until October when B&W requested approval for use of seamless tubing under Sect. I rules. Approval, in the form of Code Case 1943, was granted July 20, 1983. Subsequently (February 14, 1986), Code Case 1992 was approved for seamless and forged pipe, plate, and forgings.

Similar delays were experienced in securing Sect. VIII approval until Nooter Corporation requested the use of modified 9Cr-1Mo seamless pipes and tubes, forged pipe, plate, and forgings in Sect. VIII, Division 1 construction. Code Case 1973 was then approved on February 14, 1985. All three of the above Code Cases specify the same maximum allowable stress values (Table 36).

In 1983 an initial data package was submitted to ASME for inclusion of modified 9Cr-1Mo steel in Sect. III, Code Case N-47. Although no

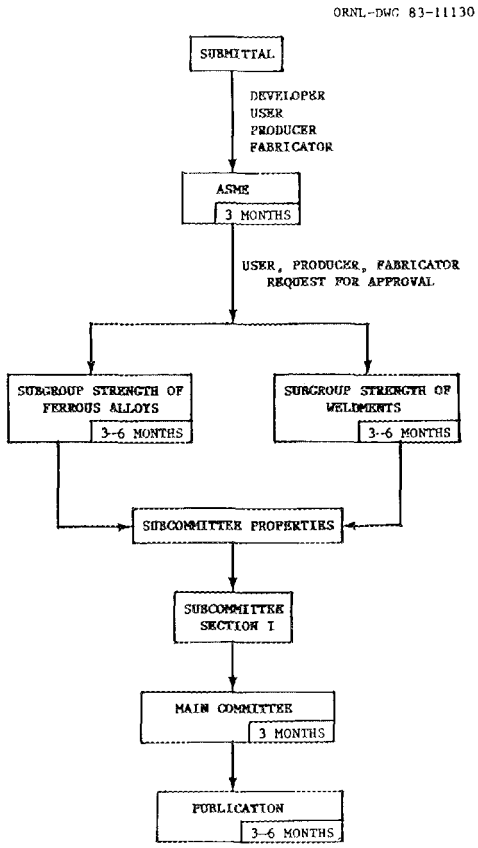


Fig. 160. From 15 to 18 months are required to obtain ASME approval for Sect. I use.

Table 36. Maximum allowable stress values for modified 9Cr-1Mo steel (grade T91)

Metal temperature not exceeding		Maximum allowable stress values		Metal temperature not exceeding		Maximum allowable stress values	
(°C)	(°F)	(MPa)	(ksi)	(°C)	(°F)	(MPa)	(ksi)
<29	-20	146.9	21.3	427	800	128.9	18.7
to 38	to 100			454	850	122.7	17.8
93	200	146.9	21.3	482	900	115.1	16.7
149	300	146.2	21.2	510	950	106.9	15.5
204	400	146.2	21.2	538	1000	198.6	14.3
260	500	145.5	21.1	566	1050	188.9	12.9
316	600	143.4	20.8	593	1100	171.0	10.3
343	650	141.3	20.5	621	1150	148.3	7.0
371	700	137.9	20.0	649	1200	129.6	4.3
399	750	133.8	19.4				

action was taken, a revised data package was submitted in November 1984 that included a more extensive set of tensile, creep, fatigue, and creep-fatigue data. In October 1985 RI sent a letter to the ASME Code describing the planned design basis for the Sodium Advanced Fast Reactor (SAFR) IHX and requested revisions to Cases N-47 and N-253 of Sect. III to permit use of modified 9Cr-1Mo steel. Action by the ASME is pending.

Note that, even in the absence of Sect. III requirements for modified 9Cr-1Mo steel, similar data are needed for the SAFR steam generator and intermediate pump under Sect. VIII rules (see Fig. 161).

3.8.3 *Nuclear Systems Materials Handbook*

The baseline tensile and creep properties of modified 9Cr-1Mo steel have been documented in the NSMH.[131] This handbook includes both summaries of behavior for design use and detailed background documentation describing how those summaries were developed from the raw experimental data. Properties contained in the NSMH include yield strength, tensile strength, and tensile ductility, as well as creep strain-time behavior, creep-rupture strength, time to tertiary creep, and creep-rupture ductility. Also included in the NSMH are pages describing fatigue crack growth behavior.

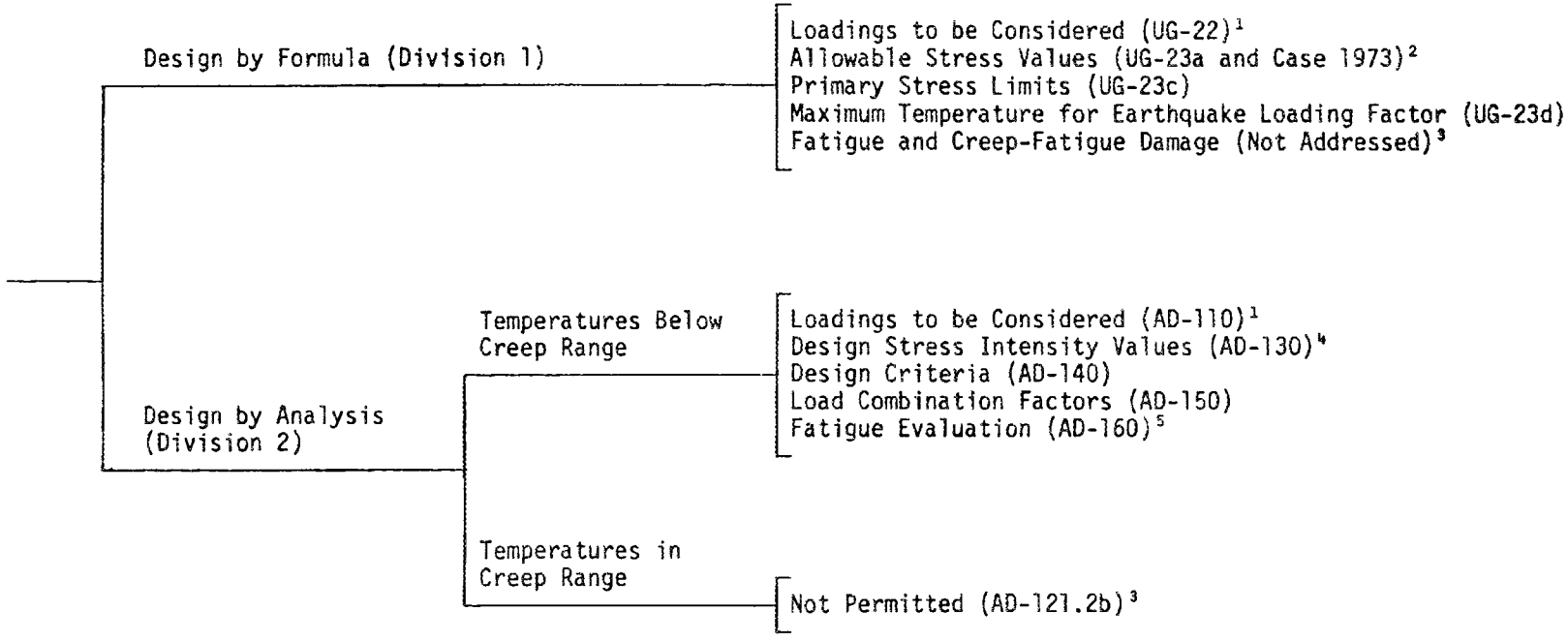
3.9 OPERATING EXPERIENCE

3.9.1 Steam Power Plants

Before a new alloy will be specified for critical nuclear applications, considerable operating experience under actual or near-actual recommended-use conditions are necessary. In addition, data on operating experience are generally required before granting ASME Code approval. Armed with an extensive technical data base, we have persuaded several utilities to allow tubing of modified 9Cr-1Mo steel to be installed in their fossil-fired steam power plants. Details of tube installations at the various plants are summarized in Table 37. The geographical range of utilities involved include those in the United States, the United Kingdom, Canada, and Japan.

Tubes installed in Tennessee Valley Authority (TVA) and AEP plants were from heats produced and fabricated in the United States. Tubes

ORNL-DWG 86-1972



1. Includes Thermal Cycling
2. Based in Part on 100,000 h Creep-Rupture Values (Very Conservative for Short-Time Loadings)
3. Creep-Fatigue Rules (Such as Those in Case N-47) Are Needed
4. Values for Modified 9 Cr–1 Mo Are Needed
5. Fatigue Design Curve for Modified 9 Cr–1 Mo Is Needed

Fig. 161. Design of SAFR steam generator and intermediate pump (ASME B&PV code Sect. VIII).

Table 37. Current status of testing of modified 9Cr-1Mo steel tubes in U.S. and foreign fossil-fired steam power plants

Utility	Plant	Tube location	Operating temperature (°C)	Tubes being replaced	Number of tubes	Date installed	Status
Tennessee Valley Authority	Kingston Steam Plant, Unit 5	Superheater	593	Type 321	8	5/80	O[a]
American Electric Power	Tanners Creek Unit 3	Secondary superheater	592	Type 304	10	4/81	O
Detroit Edison	St. Clair Unit 2	Reheater	593	Type 347	2	2/81	O
Central Electric Generating Board (United Kingdom)	Agecroft Power Station	Superheater	590-620	2.25Cr-1Mo	6	4/82	O
Ontario Hydro (Canada)	Lambton TGS	Reheater Reheater	538 538	Type 304H Standard 9Cr-1Mo	9 9	5/83	O
Ontario Hydro (Canada)	Nanticoke TGS	Secondary superheater	538	2.25Cr-1Mo	11	10/84	O

[a]O = operating.

installed in the United Kingdom were fabricated in the United Kingdom from billets produced in the United States. Tubes for the Ontario Hydro plants in Canada were made in Japan but were welded into specimens for testing in the United States. In most cases, modified 9Cr-1Mo tubes replaced stainless steel tubes in these plants.

The longest operating time achieved is six years for the tubes installed at TVA's Kingston Steam Plant. Photographs in Figs. 162 and 163

Y171284

Fig. 162. Modified 9Cr-1Mo steel tubes before operation in TVA Kingston Steam Power Plant.

CYN-5456

Fig. 163. Modified 9Cr-1Mo tubes after three years of operation at about 593°C in TVA Kingston Steam Plant, Unit 5. Modified 9Cr-1Mo tubes in this figure are those with small transition pieces, easily identified by two adjacent weld seams.

show the tubes of modified 9Cr-1Mo steel before operation and after three years in the Kingston Steam Plant. We plan for some of the tubes to be removed from the Kingston Steam Plant at the next opportunity (1987) for evaluation of steam corrosion, erosion, mechanical properties, and microstructural changes.

Results have also been received from AEP Service Corporation. The first letter report of visual examination of tube specimens after 15,033 h

of service noted that no evidence of tube sagging or swelling existed. Reinspection of these tubes is planned after about three years. No results have yet been received from specimens being exposed in plants in the United Kingdom or Canada.

Ishikawajima-Harima Heavy Industry (IHI) of Japan installed four tubes (38.1-mm OD × 8.0-mm wall) of modified 9Cr-1Mo steel in a Tokuyama SR-type boiler in September 1984. Tubes are now operating at 541°C. One of the tubes, including a weld, was removed after 1 year of operation and is currently being evaluated for changes in properties. Several other steam power plants in Japan also have modified 9Cr-1Mo steel tubes, but exact details of these installations are not known.

3.9.2 EBR-II – Sodium Pipe and Ducts

In 1984 two pipe sections of modified 9Cr-1Mo steel were installed in the EBR-II secondary sodium system. Although the pipe only operates at 454°C, such an installation provides the following benefits:

1. experience in fabricating 305-mm-OD × 9.5-mm-wall pipe;
2. experience in welding modified 9Cr-1Mo to 2.25Cr-1Mo steel, modified 9Cr-1Mo steel, and type 304 stainless steel by personnel at EBR-II;
3. experience in exposing modified 9Cr-1Mo steel to flowing sodium; and
4. evaluation of the transition joint between modified 9Cr-1Mo steel and type 304 stainless steel exposed to thermal cycles.

The pipe for this application was fabricated from 9.5-mm-thick plate by Nooter Corporation.[187] The plate was formed into shape and seam welded into four pipes each 610 mm long. Two pipes were fabricated from the plates that were normalized at 1040°C for 1 h and tempered at 760°C for 1 h. After welding, the pipes had a PWHT at 732°C for 1 h. The other two pipes were fabricated from the plates that were given a normalizing treatment at 1040°C for 1 h followed by a tempering treatment of 621°C for 1 h (half temper). After welding, these pipes had a PWHT at 760°C for 1 h (full temper). The fabrication sequence for each pipe section is given in Table 38. After welding, samples from the pipe were subjected to tensile tests at room temperature, hardness measurements, and Charpy-impact tests at 6°C. Results of these tests by Nooter Corporation are summarized in Table 39. The ultimate tensile strength exceeded the specified minimum of

Table 38. Summary of sequences used during the fabrication of welded pipe at Nooter Corporation[a]

Fabrication step	Pipe temperature (°C)			
	A	B	C	D
Normalize flat plate (for 1 h)	1040	1040	1040	1040
Temper flat plate (for 1 h)	621	621	760	760
Preheat longitudinal seam	204	204	204	204
Weld quench	Slow cool to room temperature	Heat to 316°C, hold for 4 h, and slow cool to room temperature	Heat to 316°C, hold for 4 h, and slow cool to room temperature	Hold heat until postweld heat treatment
Postweld heat treatment	760	760	732	732

[a]One pipe out of those marked A and B and one out of those marked C and D were sent for installation in EBR-II.

Table 39. Summary of tensile, hardness, and Charpy V-notch impact data on welded pipe of modified 9Cr-1Mo steel[a]

(Standard deviations for each property are included)

Room temperature ultimate tensile strength (MPa)	Hardness (BHN)[b]			V-notch impact energy (J) at 0°C		
	Weld	HAZ[c]	BM[d]	Weld	HAz[c]	BM[d]
	Pipes A and B (1/2 temper + 760°C postweld heat treatment)					
684 ± 4 (W)[e]	213 ± 4	225 ± 4	210	27 ± 2	143 ± 27	120 ± 8
	Pipes C and D (760°C temper + 732 °C postweld heat treatment)					
719 ± 1 (BM)[f]	249 ± 4	252	213 ± 4	11 ± 4	130 ± 9	130 ± 14

[a]Data were supplied by Nooter Corporation, the fabricators of the pipe; standard 9Cr-1Mo weld wire was used.

[b]BHN = Brinell hardness number.

[c]HAZ = heat-affected zone.

[d]BM = base metal.

[e](W) = failure location in weld metal.

[f](BM) = failure location in base metal.

585 MPa for pipes made from both the plates (half and full tempered). Note that the half-temper procedure produced essentially the same hardness in the base metal, the HAZ, and the weld metal. The full-temper procedure produced higher hardness in the weld metal and HAZ than the base metal. The half-temper procedure also produced better impact properties in the weld metal and HAZ than the full-temper procedure.

After receiving the pipes at ORNL, flattening tests were conducted on one of each pipe made by using the half-temper and full-temper procedures; results are shown in Fig. 164. One pipe from the half-temper and full-temper procedure was then shipped to ANL-West. A spool piece was fabricated for installation in the secondary sodium yardline to replace the existing transition weld between 2.25Cr-1Mo and type 304 stainless steel.[188] The spool piece incorporated a transition weld between modified 9Cr-1Mo and 304L stainless steel, and a weld between two pieces of modified 9Cr-1Mo pipes. The transition weld was made using Inconel 82 as the filler metal and a buttering technique developed at ORNL.[189]

A partial stress analysis of the transition weld was performed[188] in accordance with the *ASME Boiler and Pressure Vessel Code*, Sect. III.[121] The ANSYS computer program was used for the analysis. The highest stress intensity [197.20 MPa (28.6 ksi)], which was located in the transition between the Inconel and the modified 9Cr-1Mo, was caused by differential thermal expansion (a secondary stress).

The spool piece was installed in the EBR-II secondary sodium yardline during the spring shutdown in 1984 and has operated for over 2 years. No results on the condition of the installed spool piece are presently available.

Although ferritic steels are generally much more resistant to irradiation swelling than austenitic steel, irradiation can cause an increase in their DBTT. Based on some reports of reduced irradiation embrittlement,[73,89,190-195] a decision was made to fabricate several ducts out of modified 9Cr-1Mo steel and evaluate their performance at various locations in EBR-II. ORNL supplied bar stock to ANL-West, which had 25 ducts fabricated. No unusual problems were encountered in the duct fabrication. At present, efforts are under way to install the ducts in EBR-II; however, exact dates for completion of installation of the ducts are not currently available.

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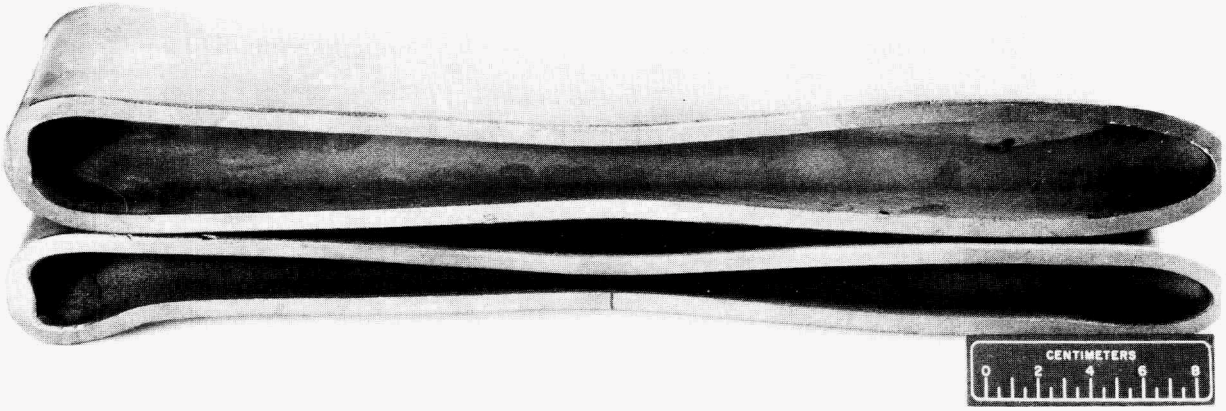

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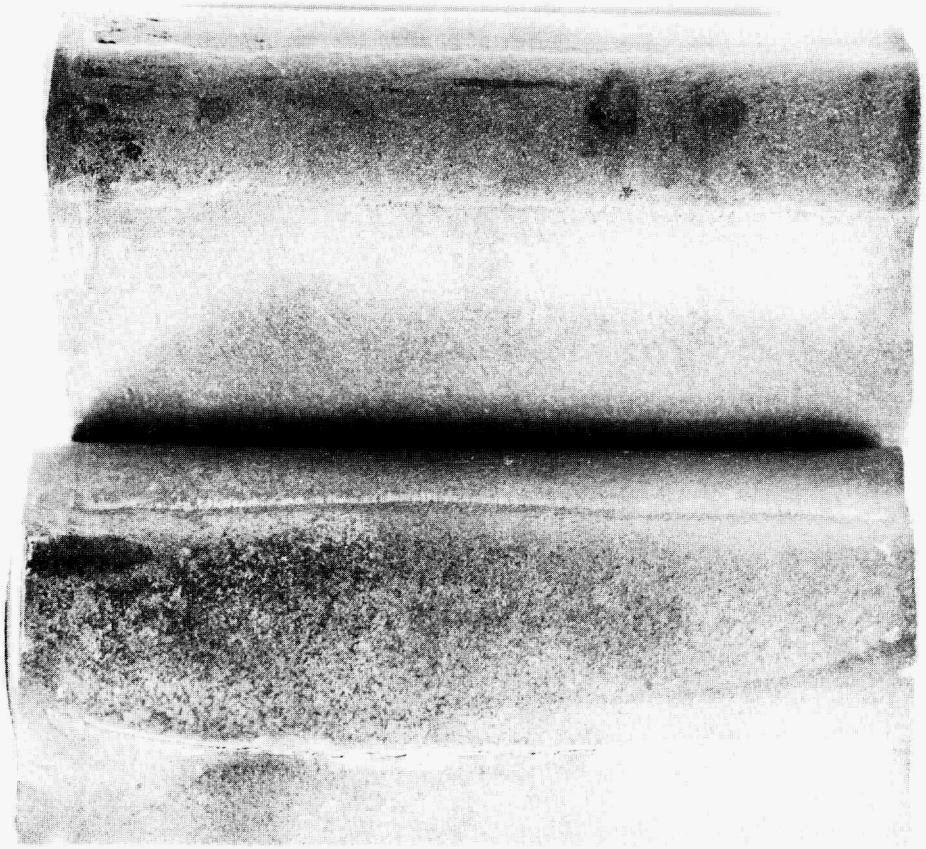

Fig. 164. Photographs showing flattened pipe from the half-temper (bottom) and full-temper (top) procedures. Note that half-temper procedure produced more flattening without any defects.

3.9.3 ETEC – Transition Joint and Large-Diameter Pipe

In late September 1982 a test section of modified 9Cr-1Mo steel pipe was fabricated for incorporation into the SCTL at the ETEC. The 206-mm-ID × 12.7-mm-wall × 914-mm-long spool piece with stainless steel safe ends was then delivered to ETEC in October 1982. Details of fabrication, inspection, shipment, and mechanical properties of the test article are described in a report by Sikka et al.[120] Figure 165 shows the finished test article before shipment to ETEC. Before its shutdown, the test article ran for one year at 510°C. No attempt has been made to date to cut the article out of the loop for postexposure evaluation.

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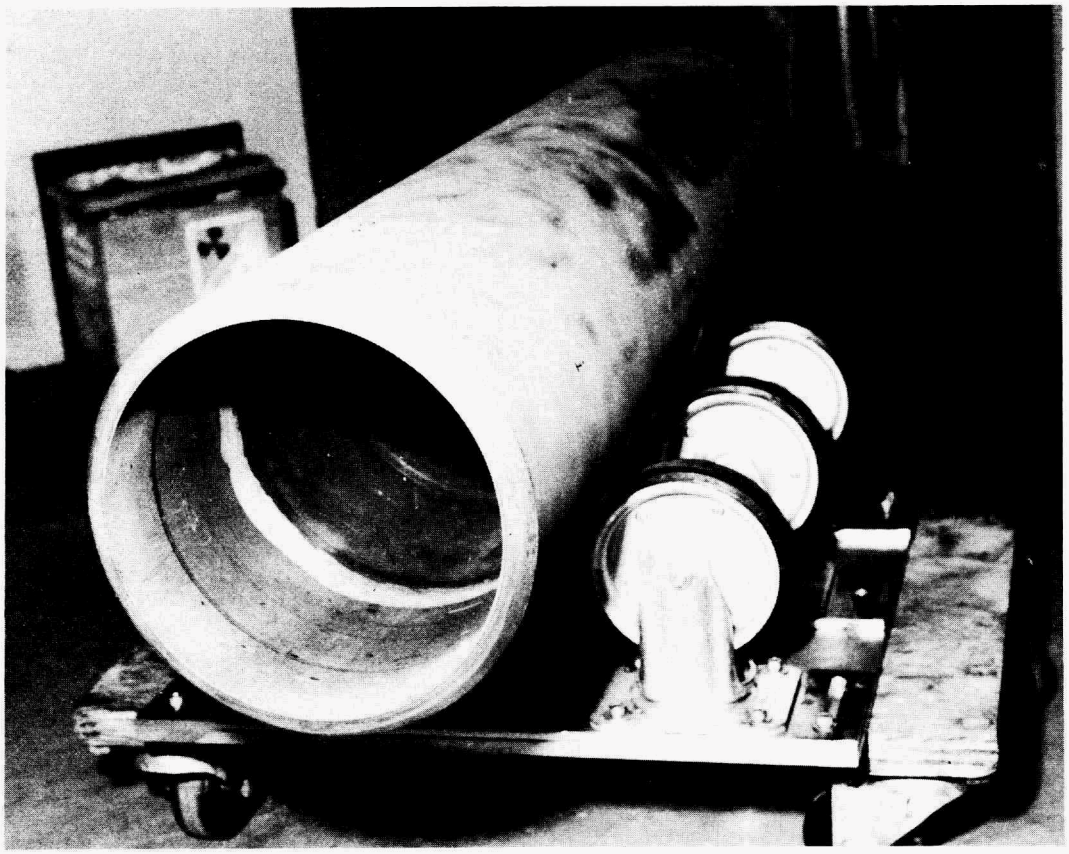

Fig. 165. Finished article of modified 9Cr-1Mo was sand blasted before final inspection. Weld root surfaces were ground smooth by hand.

During 1983 ORNL supplied a large-diameter, thin-wall pipe of modified 9Cr-1Mo steel for installation in the sodium loop at ETEC. The pipe dimensions and relative position in the sodium system are shown in Fig. 166. The main objectives of this task included

1. commercial fabrication of large-diameter, thin-wall, seamless pipe by extrusion of large ingots;
2. machining of as-extruded pipe to finished dimensions;
3. inspection of extruded pipe by commercial techniques;
4. induction bending of large-diameter pipe;
5. transition joint welding a large-diameter pipe; and
6. exposure to flowing sodium in an operating system.

To fabricate the extruded pipe, a 23-ton heat was melted at Electralloy. The heat (14361) was then remelted by the AOD process and bottom poured into four ingots. Two of the four AOD ingots were ESR. Cameron Iron Works extruded one of the ingots to a 9.1-m-long × 388-mm-ID × 15.9-mm-wall pipe. Associated Piping and Engineering Corporation performed

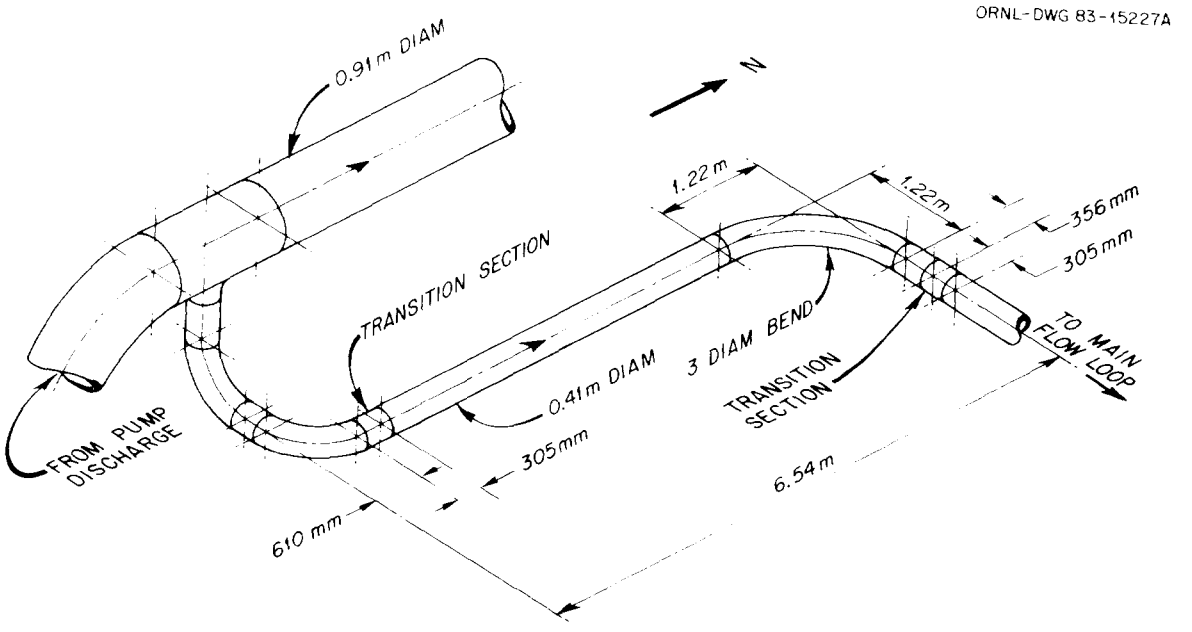

Fig. 166. Pipe scheduled for installation at the Energy Technology Engineering Center.

the induction bending at 1040°C. After bending, the bend area was postweld heat treated at 732°C. The bent pipe was then shipped to ETEC, where safe ends of type 316 stainless steel were added by welding with Inconel 82 as the filler metal. The transition welds were made by the ORNL-developed buttering technique.[189] The pipe has yet to be installed in the sodium line; however, except for gaining operating experience, all of the other objectives of the task were successfully met.

3.9.4 Switch Valves for the Oil Industry

The Oil Well Division of United States Steel Corporation uses ASTM A182 F-9 forgings of standard 9Cr-1Mo steel to make switch valves under the Wilson-Snyder trade name. These valves operate at temperatures up to 510°C; and because the design allowable stresses of modified 9Cr-1Mo steel are nearly twice that of standard 9Cr-1Mo at this temperature, Oil Well was very interested in substituting the modified alloy for standard 9Cr-1Mo. However, the company first wanted to learn more about its fabricability and to gain some actual experience with it. Therefore, ORNL supplied Oil Well with a 76-mm-diam bar of modified 9Cr-1Mo steel that was processed to the required shape and has been operating for the last 2 years. Use of the modified alloy should produce for these valves significantly longer, and more highly reliable, lives. Once these results are verified, Oil Well will likely replace the standard 9Cr-1Mo with modified 9Cr-1Mo steel for switch valves.

3.10 COOPERATIVE PROGRAMS

3.10.1 United States

Although development of modified 9Cr-1Mo steel was undertaken to provide an improved material for LMR applications, it was recognized that extensive technology transfer to U.S. industry would be required before it would be used for a nuclear component application. As a result, cooperative* programs were established at other national laboratories, universities, and private companies, and a technology transfer meeting was

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*Cooperative refers to exchange of materials, results, etc. with ORNL to satisfy the needs of both the LMR program and other programs or organizations.

held.[196] A list of 29 organizations, their funding sources, and a brief description of their work scope are presented in Table 40. The status of this work at various organizations has been compiled and published.[197] Some of the publications from these cooperative programs have been referenced in other sections of this report; however, many other publications[198–206] are available as well.

B&W was the first organization in the United States to publish a product sales letter for modified 9Cr-1Mo steel, a copy of which is reproduced in Fig. 167. It was also the first company in the United States to use modified 9Cr-1Mo steel tubing in a major application for a steam power plant. Approximately 100 tons of tubing was used. B&W currently has studies under way on the weldability of large-diameter heavy-wall pipe [584-mm OD × 101.6-mm wall (23-in. OD × 4-in. wall)] of this alloy.

3.10.2 Foreign

There has been a great deal of foreign industrial interest in modified 9Cr-1Mo steel. A list of the foreign organizations that have been involved in cooperative programs on modified 9Cr-1Mo steel and a description of their work are presented in Table 41.

3.10.2.1 United Kingdom

Tube Investments Chesterfield, Ltd., England, was the first company outside the United States to fabricate boiler tubes out of billets of modified 9Cr-1Mo steel from the United States.[207,208] Tube fabrication was successful, and the company began to accept commercial orders in 1981. Some of those same tubes were later installed in the Agecroft Power Plant of CEGB to obtain commercial operating experience. Figure 168 shows the tubes installed in the Agecroft plant.

CEGB conducted thermal shock tests of both bar and weldment specimens of modified 9Cr-1Mo steel supplied by ORNL, and an assessment report[209] was issued.

3.10.2.2 Japan

Sumitomo Metal Industries, Ltd., and Nippon Kokan K.K. (NKK) have been aggressively melting, fabricating, testing, and marketing modified 9Cr-1Mo steel. Both companies have made several large heats of this

Table 40. List of organizations in the United States involved in cooperative program on modified 9Cr-1Mo steel

Organization	Funding source	Description
Oak Ridge National Laboratory	DOE[a] Breeder and Fossil	ASTM,[b] ASME,[c] and design rules
Combustion Engineering	ORNL subcontract	Fabrication of boiler tubing, welding, and testing
Westinghouse Advanced Reactors Division	ORNL subcontract, company funds	Long-term mechanical properties testing in sodium
Battelle Columbus	ORNL subcontract	Long-term mechanical properties testing
University of Tennessee	DOE Fossil subcontract	Weldability of modified 9Cr-1Mo steel
Hanford Engineering Development Laboratory	DOE Breeder	Base and weld metal fatigue crack growth and fracture toughness
Hanford Engineering Development Laboratory	DOE Fusion	Irradiation effects and fracture toughness of base and weld metal
Oak Ridge National Laboratory	DOE Fusion	Irradiation and helium effects
Argonne National Laboratory	DOE Breeder	Sodium effect on mechanical properties
Naval Research Laboratory	DOE Fusion	Fracture toughness of irradiated material
Ames Laboratory, Iowa State University	University funds	Embrittlement toughness of irradiated material
Climax Molybdenum	Company funds	Transformation diagrams, H_2S, weldability, and microstructure
Babcock and Wilcox Company	Company funds	Chromizing, tube bending welding, heat treatment effects, fabrication of large-diameter pipe, and plant service

Table 40. (continued)

Organization	Funding source	Description
Tennessee Valley Authority		Superheater service
Detroit Edison		Reheater service
American Electric Power		Superheater service
Energy Technology Engineering Center	DOE Breeder	Sodium service at 510°C
University of Connecticut	DOE Basic Energy Sciences	Fatigue and crack growth of base and weldments
Northwestern University	DOE Basic Energy Sciences	Small-angle neutron scattering and creep-fatigue studies
Rockwell International	Company funds	Tube-to-tubesheet welding
Cameron Iron Works	Company funds	Forging of ingots into valve bodies and hot extrusion of large-diameter pipe
TRW, Inc.	Company funds	Forging bar stock for automobile valves
Gulf	Company funds	Resistance to sulfur at 538 to 593°C

[a]U.S. Department of Energy.

[b]American Society for Testing and Materials.

[c]American Society of Mechanical Engineers.

Babcock & Wilcox a McDermott company — **Product Sales Letter**

B&W Croloy 9V
A Modified 9 Cr - 1 Mo Alloy

THE BABCOCK & WILCOX COMPANY • TUBULAR PRODUCTS GROUP • BEAVER FALLS, PA

Introduction

Croloy 9V is a ferritic steel modified with columbium and vanadium, developed to provide improved strength, toughness, fatigue life, oxidation and corrosion resistance at elevated temperatures. The alloy should be of interest to designers of components for coal liquification and gasification, oil and petrochemical refining, and nuclear, fossil, solar and geothermal power generation.

The modified 9Cr-1Mo analysis was developed at Oak Ridge National Laboratory under contract from the U.S. Department of Energy (DOE). Impressive test data indicates the alloy is capable of meeting the requirements of a variety of other applications.

The alloy was developed to provide high temperature strength superior to other ferritic grades and greater or equal to Type 304 up to 1150 F. The higher thermal conductivity, lower thermal expansion and higher yield strength of Croloy 9V results in a thermal stress resistance six times greater than Type 304. The ferritic structure of Croloy 9V is also resistant to such conditions as chloride and caustic stress corrosion cracking and intergranular corrosion, which sometimes affect austenitic stainless steels. The lower chromium and nickel contents conserve strategic alloying elements compared to austenitic stainless steels. Because of the higher elevated temperature strengths, components customarily fabricated from combinations of stainless and ferritic steels can now be fabricated from a single alloy, eliminating problems associated with dissimilar metal transition joints.

Croloy 9V is available from Babcock & Wilcox as tubing, bar or extruded shapes, and in internally ribbed tubing designs.

Specification

Croloy 9V tubing is approved as a construction material under Code Case 1943, Section I of the ASME Boiler and Pressure Vessel Code. It is expected to gain similar approval under Sections III and VIII in the near future. Croloy 9V is included in the 1984 edition of ASTM A213 as Grade T-91. It will be included in ASTM A-335, A-200 and A-199.

Chemical Composition

Babcock & Wilcox Croloy 9V is melted to closely controlled ranges to optimize the elevated temperature high strength characteristics of this grade. The composition is balanced to minimize delta ferrite content which is present in other similar analyses.

TABLE I
CHEMICAL ANALYSIS - WEIGHT PERCENT

C	.08-.12	Mo	.85-1.05
Mn	.30-.60	V	.18- .25
P	.020 Max.	Cb	.06- .10
S	.010 Max.	N	.030-.070
Si	.20-.50	Ni	.40 Max.
Cr	8.00-9.50	Al	.04 Max

Heat Treatment

Extensive testing at Oak Ridge National Laboratories shows optimum mechanical properties are achieved in the normalized and tempered condition. Accordingly, B&W Croloy 9V is supplied normalized at 1900 F minimum and tempered at 1350 F minimum. Products can also be supplied isothermally annealed when requested.

Mechanical Properties

Normalized and tempered B&W Croloy 9V is supplied based on the minimum mechanical properties shown in Table II:

TABLE II - MECHANICAL PROPERTIES

	Minimum	Typical
Tensile Strength PSI (MPa)	85,000 (585)	100,000 (689)
Yield Strength PSI (MPa)	60,000 (415)	75,000 (517)
Elongation* % in 2"	20	38
*Elongation values for strip tests shall be in accordance with Table 3, Column 1 of ASTM A-213		
Hardness-Maximum (BHN)	250 (HRC 25)	207 (HRB 95)

Corrosion and Oxidation Resistance

Because of the higher chromium content, Croloy 9V is more resistant to general corrosion than other heat resisting low alloys. It is also immune to chloride stress corrosion cracking and intergranular corrosion. Oxidation resistance is similar to Croloy 9M (1200 F), although slightly less due to the lower silicon content.

Fabrication

Croloy 9V is readily fabricated by hot and cold bending, swaging and hot upsetting. Hot working operations are performed at 2000 - 2250 F and must be followed by normalizing at 1900 - 2000 F, followed by a 1350 - 1400 F temper for 1 hour. Elevated temperature tensile and creep tests show the grade may be heated as high as 1450 F for extended time periods without degrading the mechanical properties below design minimums. Welding can be accomplished using GTA, SA and SMA processes. GTA wire and SMA electrodes of the modified composition are available. A preheat temperature of 600 F and post weld heat treatment of 1325 - 1375 F are required. An inherent fine grain size and ferromagnetic properties facilitate various nondestructive testing techniques to inspect weld integrity.

Fig. 167. A copy of the Babcock and Wilcox (B&W) product sales letter on modified 9Cr-1Mo alloy developed by ORNL.

Table 41. Foreign organizations involved in the cooperative programs on modified 9Cr-1Mo steel

Country	Organization	Description
England	Tube Investments Chesterfield, Ltd.	Fabricated boiler tubes out of billets supplied by Oak Ridge National Laboratory
	Central Electric Generating Board	Installed boiler tubes fabricated at Tube Investments Chesterfield, Ltd. in Agecroft Power Plant for operating experience. Tubes went into operation during April 1982
		Conducted thermal cycling tests on base and weldment specimens of modified 9Cr-1Mo steel
		Prepared an assessment report on modified 9Cr-1Mo steel
Japan	Sumitomo Metal Industries, Ltd.	Melted the first two large heats and fabricated boiler tubing and heavy-wall tubing for subsequent fabrication. Tubes were supplied to ORNL for testing and evaluation. First heat was made in October 1981
		Melted a 50-ton heat and fabricated boiler tubing and main steam piping. Once again the boiler tubing was supplied to ORNL for installation in Ontario Hydro (Canada) power plants for operating experience. Tubing and main steam piping were also supplied to ORNL for testing and evaluation
		Conducted heat treatment, Charpy-impact, tensile, creep, fatigue, corrosion studies, and weldability on modified 9Cr-1Mo steel. Several reports were presented to ORNL
		Became a supplier of modified 9Cr-1Mo steel to Japanese industry and other countries, including the United States
	Nippon Kokan K.K. (NKK)	Melted a 5-ton heat and fabricated tubes and plates. This heat was extensively studied by NKK for heat treatment Charpy-impact, tensile, creep, and weldability studies. Tubes and plates were also supplied to ORNL for their characterization[a]
		Melted, fabricated, and sold several additional large heats to customers in the United States, Japan, and other countries

Table 41. (continued)

Country	Organization	Description
	Kobe Steel	Developed welding wire and electrodes for modified 9Cr-1Mo steel. A report on the properties of weldments using Kobe wire and electrodes was prepared and presented to ORNL
	Mitsubishi Heavy Industries, Ltd. (MHI)	Very interested in using the modified alloy in a superheater section of an ultra supercritical plant. In support of these uses, MHI has performed extensive welding studies, tensile creep studies, fatigue testing, heat treatment, and microstructural studies. MHI has installed tubing and piping sections in various boilers to get the operating experience on the alloy. The company prepared reports, which have been given to ORNL. MHI is planning to use large quantities of modified 9Cr-1Mo steel in several of the plants they will be building in Japan and other countries
	Ishikawajima-Harima Heavy Industries Company, Ltd. (IHI)	Plans to use modified 9Cr-1Mo steel for various boiler application. One of the areas of special interest to them is cast valve bodies. IHI has spent a significant effort on developing the casting process for modified 9Cr-1Mo valves. It has also performed Charpy-impact, tensile, creep, fatigue, thermal aging, and weldability on cast material
Switzerland	Sulzer Brothers, Ltd.	Interested in gaining experience with modified 9Cr-1Mo steel and comparing data on HT-9 tested in their laboratory. Sulzer has performed tensile, creep, Charpy-impact and thermal aging studies on ORNL-supplied plate and pipe pieces, and it has prepared a report on the work. Long-term creep testing is still in progress
		As a part of the Electric Power Research Institute Program, will also be involved in obtaining and characterizing large castings of the modified alloy

Table 41. (continued)

Country	Organization	Description
Italy	NIRA	Interested in the use of modified 9Cr-1Mo steel for nuclear steam generators. For this possible application, NIRA was interested in knowing more about the material first-hand. NIRA also supported the melting of 40-ton heat at DELTA Cedar in Italy. The heat was subsequently processed into thick-section tube sheets and tubes. Mechanical properties of the tube sheets have been studied. NIRA is currently in the process of making tube-to-tubesheet welds
		Performed initial tube-to-tubesheet welds on ORNL-supplied forging and tubes
France	Vallourec	Conducted a comparative study on the fabrication and mechanical properties of modified 9Cr-1Mo (U.S.), EM12 (French), and X20 (German) tubes for fossil energy boiler applications. Vallourec was pleased with the performance of modified 9Cr-1Mo steel as compared with EM12 and X20. The company's comparative report is presented in this report
	Novatome	Interested in exploring the use of modified 9Cr-1Mo steel for nuclear steam generator applications. Novatome's initial studies involved determining the mechanical properties of an ORNL-supplied forging, which has already been reported. Novatome proposes an extensive test matrix on the ORNL forging. It also plans to buy a portion of the heat, melted and fabricated in France, for its comparative evaluation
Canada	Ontario Hydro	Installed tubes of modified 9Cr-1Mo steel in two of its fossil-fired boilers. Tubes are still operating, and an inspection will be performed at the next scheduled shutdown
India	Indian Institute of Technology	Studied the effect of prior oxidation on mechanical properties of modified 9Cr-1Mo steel as part of a master's thesis

[a]Several reports were presented to ORNL. References to these reports are in the text.

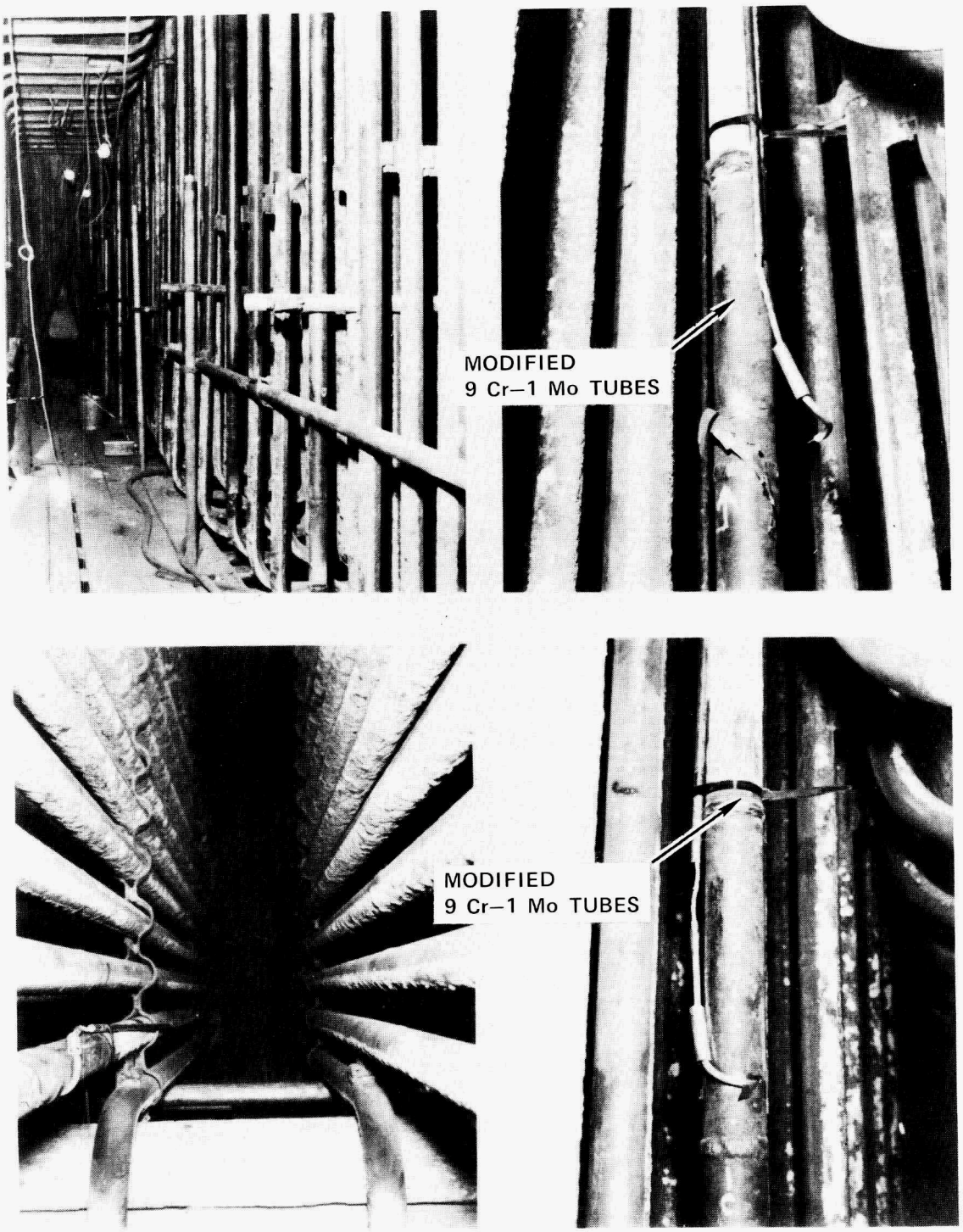

Fig. 168. Modified 9Cr-1Mo steel tubes installed in Agecroft Power Plant of Central Electric Generating Board, United Kingdom.

alloy, tested it extensively,[210-222] and sold it worldwide. Because of these efforts by Sumitomo and NKK, the modified alloy has found applications in the boilers of plants located in Korea, China, the Middle East, Japan, and the United States. These companies are now working to use modified 9Cr-1Mo steel in nuclear steam generator applications in Japan.

Kobe Steel has worked on the development of welding wire and electrodes of modified 9Cr-1Mo steel.[223] Mitsubishi Heavy Industries (MHI), a designer and builder of boilers, has extensively tested[224,225] the alloy and installed it in boilers to gain operating experience. IHI, another designer and builder of boilers in Japan, has produced castings of the alloy and performed extensive testing[22] on them. In the near future they plan to cast large valve bodies of modified 9Cr-1Mo steel and install them in the main steam line of a boiler to obtain operating experience. IHI also has installed several tubes of modified 9Cr-1Mo steel in the superheater section of a boiler. After 1 year IHI removed one tube for post-operation examination, and evaluation is currently in progress.

3.10.2.3 Switzerland

Sulzer Brothers, Ltd., is gaining experience with testing modified 9Cr-1Mo steel and has compared its data with HT-9 material also tested in the Sulzer laboratory. The company also has conducted[226] tensile, creep, and Charpy-impact tests (in the N&T and aged condition) on ORNL-supplied pipe and plate. Sulzer concluded that modified 9Cr-1Mo steel has excellent strength and toughness; however, additional long-term data are needed to determine the accuracy of the extrapolated data.

3.10.2.4 Italy

The Italians are interested in the use of modified 9Cr-1Mo steel for LMR steam generator applications. NIRA melted a 40-ton heat of the alloy, which was subsequently processed into a thick-section forging and tubes. Mechanical properties of the forging have been studied, and NIRA plans additional mechanical properties tests on both the forging and the tubes. NIRA also plans to fabricate and characterize tube-to-tubesheet joints.

3.10.2.5 France

Vallourec of France conducted a comparative study[227] on the fabrication and mechanical properties of modified 9Cr-1Mo, EM12 (French), and X20 (German) tubes for fossil energy boiler applications. Comparative results are presented in Table 42.

Novatome of France is interested in the use of modified 9Cr-1Mo for nuclear steam generators. The company has prepared[228] a test matrix (Table 43) for the evaluation of an ORNL-supplied forging made by National Forge Company. Novatome has now published[229] a paper on the initial characterization of the ORNL-supplied forging. For a comparative evaluation the company also plans to buy a portion of the heat melted and fabricated in France.

3.10.2.6 Canada

Ontario Hydro installed[230] 9 tubes of modified 9Cr-1Mo steel in its Lambton Plant operating at 538°C and 11 tubes in the Nanticoke Plant operating at the same temperature. These tubes were obtained from Sumitomo Metal Industries and were welded in the United States at CE before installation in Canada. No inspection report has yet been received from Ontario Hydro.

3.10.2.7 India

A masters thesis[231] was completed at the Indian Institute of Technology, India, on the effects of prior oxidation on the mechanical properties of modified 9Cr-1Mo alloy. Specimens of modified 9Cr-1Mo alloy were oxidized at 650°C in oxygen at 1 atm. Oxidized specimens were then subjected to tensile, creep, low-cycle fatigue, and creep-fatigue tests. Results showed that (1) tensile properties were not affected, (2) fatigue strength was lowered with increased oxygen content, and (3) creep and creep-fatigue strength increased with increasing oxidation time.

4. SUMMARY AND RECOMMENDATIONS

When established in 1975, the objective of the 9Cr-1Mo steel program was to develop and bring to a state of commercialization an alloy for LMR

Table 42. Comparative evaluations of extruded tubes of T91, EM12, and X20 CrMoV12-1 by Vallourec, Paris, France

ASTM A 213 Criteria	T91	EM12	X20 CrMoV12-1
PROCESS			
Melter	Electralloy and Universal Cyclops	Ugine Aciers, France	Ugine Aciers, France
Commercial material designation		PM 92	Fluginox 61
Heat	10148	33665	35164
Melting practice	AOD[a] + ESR[b]	Electric + AOD[a]	Electric + AOD[a]
Process in accordance with ASTM A 213 § 5	Yes	Yes	Yes
TUBE MANUFACTURE	Hot extruded from forged and turned bar	Hot extruded from rolled and peeled bar	Hot extruded from rolled and peeled bar
	Hot finished seamless tubes	Hot finished seamless tubes	Hot finished seamless tubes
HEAT TREATMENT	Air cooled from 1200°C (2200°F) after extrusion	Air cooled from 1200°C (2200°F) after extrusion	Air cooled from 1200°C (2200°F) after extrusion
	Tempered 2 h at 780°C (1400°F)	Tempered 2 h at 800°C (1470°F)	Tempered 5 h at 780°C (1440°F)
Heat treatment condition in compliance with:	ASTM A 213-83 § 7.1.3 with exception of:	NFA 49213-79 Table 8 with exception of:	DIN 17715 (79) § 7 with exception of:
	no renormalization after hot work	no renormalization after hot work	no renormalization after hot work
	hot work final temperature above 1040°C (1900°F), minimum normalization temperature	hot work final temperature above 1125°C, maximum normalization temperature	hot work final temperature above 1070°C, maximum normalization temperature
HARDNESS			
Measurement:	229 BHN (250)[c]	207 BHN (220)	225 BHN
In accordance with:	ASTM A 213-83 § 13	NFA 49213 (79) § 7.7	Not required

Table 42. (continued)

ASTM A 213 Criteria	T91	EM12	X20 CrMoV12-1
MECHANICAL TESTS			
Flattening			
$H = \frac{(1 + e)\ t}{e + t/D}$ e = 0.08 (ASTM,NF)	Pass Note: D/t < 10	Pass	Pass
[D = OD, t = wt] e = 0.05 (DIN)	Pass Note: t/D > 0.15	Pass	Pass
H (mm) for cracking or full-flattening result	Pass	Pass	Pass
Flaring			
23% ID exp (ASTM) on 60° mandrel	Pass	Pass	Pass
Mini ID exp (%) for cracking	88%	35%	45%
7% OD exp (NF) on 30° mandrel		Pass	
15% ID exp (DIN) on 60° mandrel			Pass
Hydrostatic test			
$p = 220.6\ \frac{t}{\bar{D}}$ (MPa) (mm)	Pass	Pass	Pass
SURFACE CONDITION			
As-extruded surface subsequently submitted to:	Acid deglassing Acid descaling Roll straightening	Acid deglassing Acid descaling Roll straightening	Acid deglassing Acid descaling Roll straightening

Table 42. (continued)

ASTM A 213 Criteria	T91	EM12	X20 CrMoV12-1
Observations:	General corrosion occurred during acid deglassing (insufficient inhibition of bath)	OK	OK
HEAT ANALYSIS			
Heat	10148	33665	35164
Heat analysis certificates are in compliance with:	Table 1, ASTM A 213-83	Table 5, NFA 49213-79	Table 1, DIN 17175 (79)
PRODUCT ANALYSIS			
No product analyses were done, while required by:	ASTM A 213-83 § 11		
ROOM TEMPERATURE TENSILE PROPERTIES			
Sample type			
YS [0.2% (ASTM,NF) / pure (DIN)]	554 MPa (80.3 ksi)	448 MPa (65 ksi)	618 MPa (89.6 ksi)
UTS	702 MPa (101.8 ksi)	647 MPa (94 ksi)	820 MPa (119 ksi)
A% [4D (ASTM) / 5.65 S (DIN, NF)]	29.1%	29.1% 26.6%	22.9% 21.6%
Conclusion: room temperature tensile properties in accordance with:	ASTM A 213-83, Table 3	NFA 49213 (79), Table 6	DIN 17175 (79), Table 5

Table 42. (continued)

ASTM A 213 Criteria	T91		EM12		X20 CrMoV12-1	
TOUGHNESS						
Sample: (longitudinal)	9 × 10 mm		10 × 10 mm		10 × 10 mm	
Test:	KCV[d]		KCV		KCV	
Result:						
Transition temperature TK 50	−40°C (−40°F)		−20°C (−4°F)		−10°C (14°F)	
Ductile level	25 daj/cm²		9.6 daj/cm²		12 daj/cm²	
SURFACE ROUGHNESS						
Longitudinal D[n]:	OD	ID	OD	ID	OD	ID
OD/ID CLA in inches	90.55	106.2	86.6	145.6	114	145
Transverse D[n]:						
OD/ID CLA in inches	122	98.42	102.3	145.6	165.3	78

[a]AOD = argon-oxygen decarburization.

[b]ESR = electroslag remelted.

[c]Numbers in parentheses are maximums specified in appropriate specifications.

[d]French equivalent of Charpy impact test.

Table 43. Test matrix proposed by Novatome for a piece of forged ring[a] supplied by Oak Ridge National Laboratory

Metallurgical state	Tests	Laboratory		
		EDF[b]	CEA[c]	Novatome
AR[d]	Chemical analysis (C, Mn, Si, Ni, Cr, Mo, Cu, Al, S, P, Sn, Co, As, V, N, B, Nb)	X		X
AR	Microstructure			
	Rough estimated		X	X
	Detailed	X		
AR	Hardness			X
AR	Tensile (20, 200, 300, 400, 450, 500, 525, 550°C)		X	
AR	Creep 1000 h, 525°C, five specimens			X
AR	Fatigue endurance limit at 450°C			X
AR	Impact properties KCV[e] at several temperatures	X		X
AR	Fracture mechanics J_{Ic} at room temperature		X	
AR	Stress-corrosion cracking		X	
AR	Performance of plate-tube internal bore welding (if tubes available)			X
AR	Aging heat treatment (10,000 h at 500°C and 550°C; 2,500 h at 600°C)	X		
Aged	Tensile properties + impact properties KCV[e]	X		
AR	Simulation of HAZ in Gleeble machine			X
Simulated heat-affected zone	Hardness			X
	Impact properties			X
	Creep (same test conditions as for "AR" above)			X
AR	Dilatometry (determination AC_1 - AC_3)			X
AR	Simulation of heavy-section thickness: heating 100°C/h up to AC_3 + 50°C, 2 h hold, cooling between 800°C and 500°C at 120°C/h + tempering + postweld heat treatment			X
Simulated heavy-section thickness	Tensile (20°C, 500°C)			X
	Impact properties			X
	Step cooling			X

[a]Modified 9Cr-1Mo steel; forged ring from National Forge Company.

[b]EDF = Electricité de France.

[c]CEA = Centre d'Energie Atomique.

[d]AR = As received.

[e]French equivalent of Charpy impact test.

applications that has better overall properties than standard austenitic stainless steels and other low-alloy ferritic steels. This objective has substantially been achieved inasmuch as modified 9Cr-1Mo steel is presently a commercially available material and the alloy has been specified for use in the IHX and steam generator of an advanced LMR (SAFR). However, even with all that has been accomplished in the last 10 years, there remain some areas in which further work should be done before application in an advanced LMR can be ensured. The status and needs of the various program areas are summarized below.

4.1 ALLOY COMPOSITION

In 1980 final commercial specifications for wrought modified 9Cr-1Mo steel were selected. Subsequent melting, fabrication, and testing activities have not indicated that any change in these specifications is warranted. However, more work is needed to optimize the composition of castings, because they generally show poorer mechanical properties than wrought material. Also, nitrogen appears to be a sensitive element in determining the porosity of sand castings. For improved oxidation resistance, the effect of increased silicon concentration should be fully determined.

4.2 MELTING AND FABRICATION

Commercial procedures for melting and fabricating modified 9Cr-1Mo steel are well established. The properties of 13 different heats totaling >227 tons that were melted for testing and evaluation have been documented. In addition, >600 tons of the alloy have been melted and processed for commercial applications by U.S. and foreign industrial companies. Although the properties of modified 9Cr-1Mo steel are remarkably consistent (due in part to specification limits), some differences based on processing have been noted. For example, in applications where toughness properties are critical, material melted by the ESR process is recommended.

4.3 MECHANICAL PROPERTIES

An extensive mechanical properties data base on modified 9Cr-1Mo steel has been achieved in air environments. Data packages containing

information on tensile, creep, fatigue, and impact properties have been approved by the ASME Code, and baseline tensile and creep properties have been published in the *NSMH*. Additional data needs include long-term creep tests to 10^5 h, tests to determine the effects of heat treatment and heat-to-heat variations on continuous cycling fatigue, and the effect of hold time on fatigue life. Creep-fatigue data are especially needed because little information is available, and ASME design rules must be established for presently anticipated SAFR applications. (An overall test matrix of planned mechanical properties tests can be found in the National Program Plan.[8])

Effects of thermal aging at 482 to 704°C for 25,000 h on Charpy-impact properties were shown to be dependent on the aging temperature. However, tensile and creep properties were not affected at aging and test temperatures up to 649°C. Tensile, creep, fatigue, and creep-fatigue tests are recommended on samples now being aged to 50,000 and 100,000 h.

N&T studies have resulted in the recommendation of an optimum heat treatment for modified 9Cr-1Mo steel.

The swelling rate of modified 9Cr-1Mo steel irradiated in EBR-II is a factor of 50 or more lower than for type 316 stainless steel but is a factor of 3 higher than HT-9. Postirradiation toughness properties of modified 9Cr-1Mo, however, are superior to those of HT-9. On the other hand, after irradiation in a mixed spectrum reactor, toughness and tensile properties of modified 9Cr-1Mo were reduced more than those of HT-9. No data exist on the effects of irradiation on the creep properties of modified 9Cr-1Mo. However, data on HT-9, 9Cr-2Mo, and 2.25Cr-1Mo steel, together with microstructural observations, indicate that irradiation in an LMR between 450 and 550°C should have no significant effect on the creep rate of modified 9Cr-1Mo to 550°C. The available data on helium effects are at temperatures and concentrations much higher than anticipated for use of modified 9Cr-1Mo in an LMR. However, they indicate that the alloy has a high resistance to irradiation-induced helium embrittlement. Overall one may conclude that modified 9Cr-1Mo should have good long-term performance under LMR conditions, but further work should be performed to determine the difference in response between fast reactors and mixed spectrum reactors and to provide data at the low damage levels relevant to out-of-core structures and components.[232]

There has been only limited testing to determine the effects of sodium on the mechanical properties of modified 9Cr-1Mo steel, but the results appear to be encouraging under LMR conditions. Losses of carbon or nitrogen in bimetallic systems with austenitic stainless steels occur only at temperatures >550°C. In addition, short-term exposures to sodium (5000-6000 h) at 538°C were not reported to affect the strength properties of modified 9Cr-1Mo steel significantly. The fatigue life of T-91 was five to seven times greater in sodium than in air at both 538 and 593°C.

Examination of all the reported creep-rupture data in sodium from both ANL and WAESD does, however, indicate a tendency toward shorter life (factor of 6-7) in sodium. Because the variations in rupture life found were not random, further testing in sodium for longer times to better define the effects of sodium on the creep-rupture properties of T-91 for LMR applications is recommended.

4.4 JOINING BEHAVIOR

Studies have shown that the weldability of modified 9Cr-1Mo steel is excellent. These studies have been conducted using the GTA, SMA, and SA processes. Effects of preheat, PWHT, and base metal treatment before welding were all investigated. Welds were made and tested using standard and modified 9Cr-1Mo filler wire, and several flux coatings were tried for SMA electrodes, as well as for the SA welding process. Weldments were subjected to extensive hardness, Charpy-impact, tensile, and creep tests. Microhardness data indicated that welding produces a soft zone in a very narrow tempered region of the base metal near the weld. This also occurs in other heat-treatable steels such as 2.25Cr-1Mo, standard 9Cr-1Mo, and HT-9. A technique whereby the base metal is half tempered before welding was developed to eliminate the soft zone observed during conventional welding.

Tensile properties of weldments are generally similar to that of base metal; however, the creep properties can fall below that of the base metal. For times to 2000 h, weldment creep properties generally differ from the base metal, but beyond that time creep properties begin to approach those of the base metal. Long-term testing is recommended to confirm this effect.

Charpy-impact properties were best for welds made by the GTA process. Additional work is recommended to improve the coatings on the electrodes for SMA welding and fluxes for SA welding. Some optimization of the procedures used for SA welding (e.g., ac vs dc and welding speed) is also needed.

Additional fatigue and creep-fatigue tests are recommended on weldments made by the various welding processes. These should include several different test temperatures and hold times.

4.5 STRUCTURAL DESIGN METHODOLOGY

A program to develop a high-temperature design methodology for LMR applications has been under way since the late 1960s. That methodology has been reasonably well established for the austenitic stainless steels and, to a lesser extent, for 2.25Cr-1Mo steel. For modified 9Cr-1Mo steel the plan is to adopt, to the extent possible, the existing methodology while accounting for any unique features of the alloy. As a first step a preliminary analytical assessment of the structural design characteristics of modified 9Cr-1Mo steel was performed to compare with those of type 304 stainless steel, 2.25Cr-1Mo, alloy 800, and standard 9Cr-1Mo steel. Because of its relatively high yield strength, low coefficient of expansion, and high thermal conductivity, modified 9Cr-1Mo steel was shown to have an advantage in lower wall thickness and thermal loads, and better resistance to creep damage.

Work has continued over the past 2 to 3 years to develop constitutive equations and establish failure behavior. This effort concluded that the present NE F 9-5T constitutive framework is inadequate for modified 9Cr-1Mo steel. The material appears to require a nonlinear viscoplastic model that makes no distinction between creep and plastic strains. A unified model developed by Robinson[147,148] is presently being adapted to represent the behavior of modified 9Cr-1Mo steel. This work should be continued to provide a constitutive model for detailed inelastic analysis of LMR structural components and to validate simplified methods of analysis.

Rules and criteria to guard against creep-rupture and creep-fatigue failure modes are found in ASME Code Case N-47 (ref. 122). Because modified 9Cr-1Mo has not been incorporated into Code Case N-47, no rules

yet exist; to date only a few selected scoping tests have been conducted as a basis for criteria assessments. As an interim measure, criteria can be generated based upon uniaxial data. However, additional multiaxial tests, including creep-fatigue damage accumulation tests of modified 9Cr-1Mo steel, are certainly required to establish LMR design criteria. Behavioral features of modified 9Cr-1Mo that will likely have to be given special consideration in developing ratchetting and creep-fatigue rules include strain softening, low strain at onset of tertiary creep, and rate dependence of yielding and plastic flow.

Material modeling for constitutive equations and failure behavior are evaluated through confirmatory laboratory tests. One such test has been completed,[157] and results are still being analyzed. Two thermal ratchetting tests to creep-fatigue failure using welded and unwelded samples are planned[159] to permit validation of the design procedures for predicting strain accumulation and creep-fatigue damage failure under thermal ratchetting load conditions. Because elastic follow-up can occur in elevated-temperature piping systems, a fact that increases creep-fatigue damage, RI is planning a test at ETEC to demonstrate that, with modified 9Cr-1Mo, strain concentration caused by elastic follow-up is quite limited and that the Code requirement to treat the corresponding stress as primary is excessively conservative.

4.6 CORROSION BEHAVIOR

Studies to establish the corrosion behavior of modified 9Cr-1Mo for LMR applications have not been extensive. Because of its similarity to other ferritic steels such as 2.25Cr-1Mo, however, no corrosion limiting problems are anticipated.

Properly heat-treated, modified 9Cr-1Mo is expected to have low susceptibility to caustic stress-corrosion cracking[168] and behavior similar to 2.25Cr-1Mo steel in chloride environments. Thus stress-corrosion cracking of modified 9Cr-1Mo steel is unlikely under LMR conditions.

No significant changes in properties of modified 9Cr-1Mo steel caused by carburization and decarburization are anticipated, but additional data on the effects of heat treatment and sodium exposure should be obtained.

The steam corrosion behavior of modified 9Cr-1Mo is similar to (and perhaps slightly better than) 2.25Cr-1Mo steel. If improvements in steam corrosion resistance are required, the effect of silicon should be further investigated. Also, the effects of heat flux on oxide formation and spallation should be determined under LMR conditions.

4.7 INSPECTION

Nondestructive testing techniques and procedures have been developed for 2.25Cr-1Mo steel, and we believe the technology can be adapted to perform examinations of modified 9Cr-1Mo steel components. We do, however, recommend that NDE continue as part of the advanced LMR program to confirm the applicability of radiographic, ultrasonic, and eddy-current techniques to the particular system geometries, temperatures, environments, and sensitivities that are required.

4.8 CODES AND STANDARDS

Specifications of modified 9Cr-1Mo have been included in the ASTM book of standards for plate, tube, pipe, and forgings. The same specifications have also been included in Sect. II of *ASME Boiler and Pressure Vessel Code*. The only two products for which specifications are not currently included are bar and cast products. Bar product data are ready, and we recommend that the specifications be submitted for approval. However, casting specifications require some additional work to optimize compositional limitations for mechanical properties. Work should continue to provide the required information for approval of use of castings by ASTM and the ASME Code.

Filler wire compositions and welding requirements have been developed but have not been submitted for approval in the American Welding Society Specifications and in Sect. XI of *ASME Boiler and Pressure Vessel Code*. Cooperative activities with industry should be initiated to obtain these approvals.

Code Cases 1943, 1973, and 1992 have been approved and published for the use of modified 9Cr-1Mo according to the rules of Sects. I and VIII of *ASME Boiler and Pressure Vessel Code*. A data package for the approval of

modified 9Cr-1Mo for nuclear component use according to the rules of Code Case N-47 of Sect. III of *ASME Boiler and Pressure Vessel Code* has been submitted. A user inquiry has been submitted by RI, and action by the Subgroup on Elevated Temperature Design is pending. Before appropriate Code rules are adopted, we expect that additional fatigue and creep-fatigue data will be required as previously discussed under the section on mechanical properties.

4.9 OPERATING EXPERIENCE

Tubes of modified 9Cr-1Mo steel have been operating in three power plants in the United States, one in England, two in Canada, and several in Japan. Tubes in the TVA Kingston Steam Plant, which have been operating since 1980, were visually examined after 3 and 5 years of operation, and no unusual effects were noted. Similar results of visual examinations were received from AEP after 2 years of operation, and no deleterious effects were noted. The recommendation is that plans be prepared to remove one or two tubes from the TVA Kingston Plant to provide data on steam oxidation, erosion, and changes in mechanical properties and microstructure of base metal and weldment regions. After 1 to 2 years of service, the Japanese have already sectioned tubes from several power plants. Cooperative efforts should be continued to obtain their data on the evaluation of these tubes.

Plans should also be prepared to recover the test article from the SCTL at ETEC for postoperation evaluation. This will provide information on the relative performance of base metal, weld metal, and transition joints in flowing sodium at 510°C. Efforts should be continued to support installation of the modified 9Cr-1Mo steel ducts in EBR-II.

4.10 COOPERATIVE PROGRAMS

One of the most successful techniques to promote the commercial application of modified 9Cr-1Mo steel has been technology transfer through cooperative programs with industry, universities, and national laboratories. Contributions of these cooperative programs have been in the form of basic research, applied research (testing of samples), and scale-up to large heats. Other cooperative efforts involving efforts to obtain

operating experience were previously discussed. The Japanese, who have been particularly aggressive in developing modified 9Cr-1Mo steel, have established well-defined procedures for the production of various product forms of modified 9Cr-1Mo steel. They have also developed markets for the alloy in the United States, Japan, Canada, the Middle East, India, and mainland China. Modified 9Cr-1Mo steel is also being tested for steam generator applications in France and Italy. We recommended that the information and progress being made in these programs continue to be followed closely.

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Appendix A

CRITERIA TO ASSESS MODIFIED 9Cr-1Mo STEEL FOR STEAM GENERATOR APPLICATION

The selection of a new (or currently selected) material for steam generator application requires the consideration of a large number of factors.[1-2] A list made by Willby and Walter[1] is shown in Fig. A.1. Spalaris et al.[2] listed the consideration of the following items for the currently proposed steam generator material, 2.25Cr-1Mo steel:

1. melting practice,
2. specifications,
3. heat treatment,
4. mechanical properties,
5. corrosion under departure from nucleate boiling (DNB),
6. corrosion in superheated steam (heat flux),
7. fracture toughness (base, weld metal, and heat-affected zone),
8. carbon transfer (in sodium loop),
9. stress-corrosion cracking (in caustic solution),
10. pitting corrosion,
11. rust problems, and
12. weldability and postweld heat treatment.

Considerations such as (1) propensity of ratchetting and creep damage under transient conditions, (2) inelastic material response (history dependence of material properties) for constitutive equation assessment, and (3) representation of creep-fatigue and creep-rupture damage have been suggested by C. E. Pugh (Oak Ridge National Laboratory, Oak Ridge, Tenn., personal communication to J. R. DiStefano, Oak Ridge National Laboratory, Oak Ridge, Tenn., March 30, 1979).

This appendix presents a list of considerations and the associated criteria to be applied to modified 9Cr-1Mo steel for steam generator applications in Table A.1. The number of heats and the test conditions in the criteria are based on an assessment schedule by the end of FY 1981. The criteria are based on the present knowledge of modified 9Cr-1Mo steel

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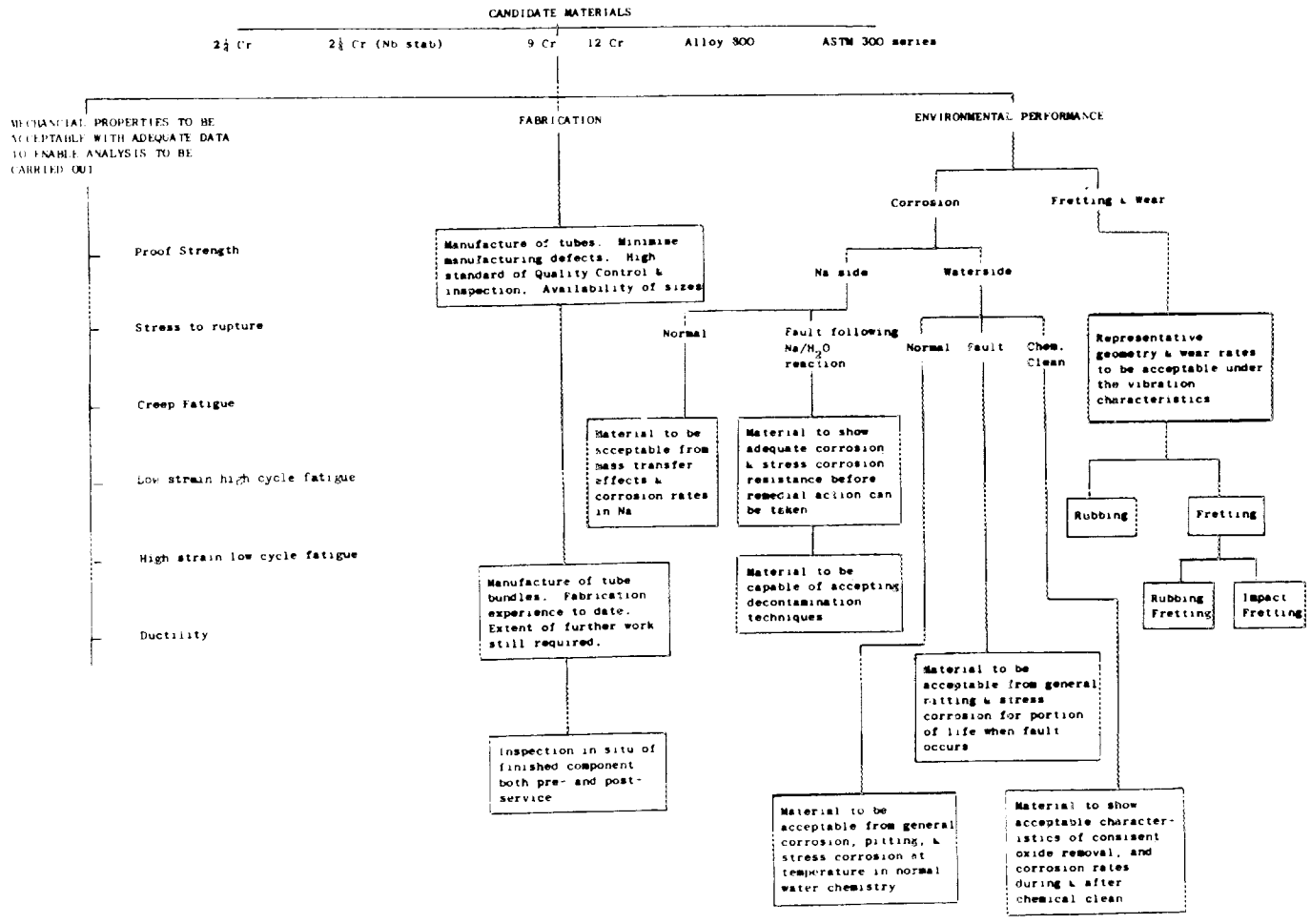

Fig. A.1. Factors to be considered in choice of steam generator tube material. *Source:* C. Willby and J. Walters, "Material Choices for the Commerical Fast Reactor Steam Generators," *Ferritic Steels for Fast Reactor Steam Generators, Proceedings of an International Conference Held by the British Nuclear Energy Society at the Institution of Civil Engineers, London, 30 May–2 June 1977*, vol. 1, ed. S. F. Pugh and E. A. Little, British Nuclear Energy Society, London, 1978, p.47.

Table A.1. Criteria and considerations for steam generator applications of 9Cr-1Mo steel

Considerations	Criteria
Material supply	
Melting	Material can be melted within specifications with controlled additions of Nb = 0.06 to 0.10%, V = 0.18 to 0.25%, and Si = 0.1 to 0.4%, using commercial processes such as argon oxygen deoxidation and electroslag remelt
Fabricability	Material can be processed into tubing and forgings using commercially available processes (material will have to meet the appropriate RDT standard)
Supply	Can get at least two steel mills to remelt this material and two different fabricators to make the tubing and forgings
Mechanical properties	For four commercial heats and several product forms, criteria are as follows
Low-temperature toughness properties	
· Transition temperature	<-12°C
· Upper-shelf energy	≥136 J (≥100 ft-lb)
Short-term mechanical properties	
· 0.2% yield	379 MPa minimum[a,b]
· Ultimate tensile strength	586 MPa minimum[a]
Creep properties	
· Stress to rupture	Stress to rupture values are equal to or better than ASME Code Case minimum curve type 304 stainless steel[c] (comparisons to be made for rupture times approaching 20,000 h for 3 heats and 40,000 h for 1 heat)
· Stress to cause 1% strain in 10^5 h	Stress is equal to or better than minimum values for type 304 stainless steel

Table A.1. (continued)

Considerations	Criteria
· Stress to cause onset of tertiary creep in time t	Time to onset of tertiary creep for modified 9Cr-1Mo steel is expected to be 50% of time to rupture, which is similar to types 304 and 316 stainless steel. This type of behavior will produce stress to cause onset of tertiary creep in time t similar to that observed for type 304
· Ductility	Reduction of area values for this material should exceed 70% for test temperatures in the range of 482–649°C and for test times approaching 40,000 h
Fatigue and creep fatigue properties	
· Continuous cycling fatigue	Cycle life is equal to or better than average for type 304 in the low-cycle fatigue range (>10^4 cycles) and an order of magnitude better than type 304 for the high-cycle fatigue (>10^4 cycles) range
· Hold-time effects (compressive and tensile)	Hold-time effects should be less than observed for type 304 stainless steel
Fracture mechanics properties	
· K_{Ic}, J_{Ic} · da/dN	Crack growth rate is expected to be equal to or lower than the upper bound observed for type 304 stainless steel
Thermophysical properties[b]	For four commercial heats and several product forms
Thermal conductivity	It is a factor of ~1.8 higher compared with types 304 and 316 stainless steel at room temperature
Thermal expansion coefficient	It is expected to be 70% of the value observed for types 304 and 316 stainless steel at 100°C

Table A.1. (continued)

Considerations	Criteria
Heat treatment and welding	
Heat treatment	
· Austenitization temperature/time	A heat treatment should be determined to obtain optimum mechanical properties, corrosion resistance, and ease of fabricability for unusual product forms or components
· Tempering temperature/ time	
· Cooling rate (section thickness)	
Welding procedures and filler metal	When commercial heats are used, the following criteria should be met
· Welding procedure	Material should make acceptable welds by conventional welding procedures (GTA, SMA, SA, GMA, autogenous) in various section thicknesses up to 15.2 cm (6 in.)
· Preweld and postweld heat treatments	Preweld and postweld heat treatment should be determined to prevent possibilities of cold, hot, and stress-relief cracking
· Filler wire composition	Filler metal composition should be optimized to obtain weld metal properties comparable to the base metal (both strength and ductility)
· Bimetallic and transition joints	Should demonstrate successful welding of 9Cr-1Mo to 2.25Cr-1Mo, stainless steels of types 304, 316, 321, 347, alloy 800, etc.
Environmental effects	
Corrosion resistance	Using four commercial heats, the following criteria should be met
· Rusting	No significant rusting for 2.25Cr-1Mo steel expected during fabrication
· Steam corrosion	Material should have a corrosion rate[d] in superheated steam at 482 and 538°C (900 and 1000°F) of ≤0.2 and 0.4 mil/ year,[e] respectively
· Water corrosion	

Table A.1. (continued)

Considerations	Criteria
· Alkaline solutions (OH^-)	Metal loss in 1 M NaOH should be one-fourth of the metal loss observed for 2.25Cr-1Mo steel at 350°C
· Chloride and sulfate concentrates	It should be more resistant than 2.25Cr-1Mo
· Stress-corrosion cracking	Normalized and tempered modified 9Cr-1Mo steel is expected to be virtually immune to stress-corrosion cracking in Cl^- and OH^- solutions
· Sodium effects	It should be more resistant to carbon mass transfer than 2.25Cr-1Mo steel. Data from the equilibration and kinetic experiments have shown that in a liquid-Na environment at temperature <550°C, the modified 9Cr-1Mo provides greater resistance to carbon transfer than 2.25Cr-1Mo alloy. There is essentially no nitrogen loss at temperatures <550°C. At >600°C decarburization can occur to surface C levels of 220 ppm. A limited number of equilibration and kinetics experiments will be conducted on the commercial heats to verify the above-mentioned behavior
Erosion, friction, and wear	The wear rates in sodium for various surface combinations should be $\leq 3 \times 10^{-14}$ to 3×10^{-15} m^3/Nm, which is the same as for 2.25Cr-1Mo steel
Thermal stability (effect of thermal exposure with and without stress on mechanical properties)	For base metal and welds from four commercial heats [exposure 10,000 h at 482, 538, and 593°C (900, 1000, and 1100°F)], criteria are as follows
· Shift in transition temperature	Expected shift will be <10°C
· Shift in upper-shelf energy	Expected shift will be <27 J (<20 ft-lb)

Table A.1. (continued)

Considerations	Criteria
· Creep and rupture strength	Creep and creep-rupture strength decrease by <14 MPa (<2 ksi) for test times up to 1000 h
· Creep ductility	Creep ductility decreases by <10% of the unaged value of 10%
Design	
Propensity of ratchetting and creep damage under transient conditions	Material should be equal to or better than type 304 stainless steel

[a]Room-temperature values specified in Section III of ASME Code Case for forgings of unmodified 9Cr-1Mo steel, to be used in the normalized and tempered condition.

[b]It is anticipated that the combination of high yield strength (σ_y), high thermal conductivity (K), and low thermal expansion coefficient (a) in combination with Young's modulus (E) and Poisson's ratio (ν) will produce a factor of 5 higher thermal stress resistance (M) at 500°C for 9Cr-1Mo steel as compared with type 316 stainless steel. The value of M is given by

$$M = 2\sigma_y K(1 - \nu)/aE \ .$$

[c]These rupture values make the modified 9Cr-1Mo steel seem to be the strongest of the currently known class of 9Cr steels.

[d]These rates are equal to those observed for 2.25Cr-1Mo steel.

[e]EM 12 tubes taken from superheaters after 50,000 h of operation at 590°C showed a total oxide thickness of ~8 mil. Based on corrosion rates listed above for modified 9Cr, the oxide thickness in the same time will be ~5 mil (assuming the corrosion rate to be 0.8 mil/year at 590°C as opposed to 0.4 mil/year at 538°C).

and published data on the 9Cr class of steels. In some instances, criteria are qualitative, and some of these criteria may have to be modified as required by input from the steam generator designers. Furthermore, additional criteria (e.g., irradiation effects) will be required to extend the use of this material to other potential applications, such as reactor internals, intermediate pump, and loop piping (C. A. Anderson, Westinghouse Advanced Reactors Division, Madison, Pa., personal communication to C. L. Stores, Combustion Engineering, C. E. Power Systems, Windsor, Conn., May 1, 1979). Assessment criteria for such applications will be prepared at a later date.

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