

Austenite to Martensite Phase Transformation in 9 Cr 1 Mo Steel Weld Metal

M. L. Santella, S. S. Babu, R. W. Swindeman, E. D. Specht, G. E. Ice and P. Zschack**

*Metals & Ceramic Division, Oak Ridge National Laboratory, Oak Ridge, TN

**Frederick Seitz Materials Research Laboratory, University of Illinois at Urbana-Champaign, Urbana, Illinois

Introduction

In the USA in the early 1980's, tubing of 9Cr-1Mo-V (P91) steel was introduced into the superheaters of power boilers as a replacement for 300 series austenitic stainless steels [1]. Because P91 steel is a high-strength alloy in which austenite normally transforms completely to martensite during air cooling [2], the specification of preheat and post weld heat treating (PWHT) conditions is an important practical matter [3,4]. One aspect of this issue is the handling of weldments between completion of welding and PWHT. Weldments that are cooled to room temperature before PWHT will transform more completely to martensite than weldments that are maintained at or above minimum preheat temperature prior to PWHT. Consequently, weldments that are cooled to room temperature are less likely to contain untempered martensite after PWHT. However, maintaining preheat temperature before PWHT is essential for minimizing the probability of hydrogen cracking in weld heat-affected zones. In this research, martensite formation in a P91 base metal and weld deposit was investigated with experimental research and theoretical calculations.

Procedure

The submerged arc weld was made using Thermanit MTS3 welding filler metal (AWS A5.28, ER90S-B9) and Marathon 543 flux, both products of Böhler Thyssen. The weldment was given a PWHT of 8 hours at 774°C (1425°F) before being supplied for testing and analysis. Specimens from normalized-and-tempered plate identified as alloy 30383 [see Table 1] were used to establish baseline transformation behavior for P91 plate. The phase transformation behavior was investigated while subjected to following heat-treatment following: Austenitizing for 2 min at 1040°C (1904°F); Cooling to either room temperature or 200°C (392°F); Tempering for 15 min at 740°C (1364°F); Cooling to room temperature at 6 °C/min. The Gleeble specimens used for this study were 6.35 mm (0.25 in) diameter x 108 mm (4.25 in) long rods. In the weld deposit, the rods were cut transverse to the welding direction from near the weld surface. The in-situ transformation behavior of was also analyzed using the UNICAT X-33 beam line at the Advanced Photon Source. X-rays of wavelength, $\lambda = 0.041328$ nm, were provided for diffraction while specimens were heated through thermal cycles identical to those used for the dilation analysis. The x-rays were incident at a 7.5° glancing angle, and a 2-D image of the diffracted x-rays was recorded every 8 sec.

Table 1. Alloy chemical compositions (wt.%)

ID/heat	C	Mn	Si	Cr	Mo	Ni	Nb	V	N	Al
Weld	0.099	0.61	0.17	9.33	0.97	0.80	0.043	0.23	0.051	0.022
30383	0.083	0.46	0.41	8.46	1.02	0.09	0.072	0.198	0.051	0.002

Results and Discussion

Examples of the specimen dilation results are shown in Figs. 1a 30383 base metal, and in Fig. 1b for the weld deposit. The martensite reaction in heat 30383 appeared to be completed before the tempering portion of the treatment because no other phase transformation was detected during tempering treatment. The martensite start temperature, M_s , was 393-

395°C (739-743°F). In contrast, the transformation of the weld deposit appears to be incomplete [see Fig. 1b]. A secondary martensite reaction was detected in the weld deposit after tempering. The primary M_S for the weld deposit was 348-350°C (658-662°F). The secondary M_S was near 400°C (752°F). These results show that the weld deposit is likely to contain some untempered martensite after welding and PWHT. In-situ diffraction studies were used to confirm this behavior. The diffraction results are shown in Fig. 2, where the sample temperature histories are plotted along with the variations of diffracted intensities. Figure 2 clearly shows that both the 30383 base metal alloy and the weld deposit are fully austenitized at 1040°C. The disappearance of the austenite (200) reflection [see Figure 2a] confirms that the transformation of austenite to martensite is completed in the base metal by cooling to 200°C. In the weld metal sample [see Fig. 2b], the austenite coexists with martensite below the M_S temperatures. The results also showed that this austenite coexists even during tempering and transforms during cooling from tempering treatment. An alloy having the weld deposit composition was further analyzed using thermodynamic software (ThermoCalc™). The variations of C, Cr, Ni, and Mn with fraction solid were calculated assuming the nonequilibrium (Scheil) conditions of equilibrium at the liquid-solid interface, no solid diffusion, and complete diffusion in the liquid. The calculations showed the possibility of microsegregation in the weld metal region [5]. As a result, the transformation behavior to martensite will not be uniform throughout its microstructure, but it will depend on local chemical composition.

Conclusions

The phase transformations that occur in the 9 Cr 1 Mo base metal and weld metal region was investigated with in-situ time-resolved X-ray diffraction, Gleeble™ thermal simulation, and thermodynamic modeling. Both Gleeble and diffraction studies from weld metal samples showed the presence of retained austenite due to incomplete martensite formation from austenite. This retained austenite at room temperature then transforms to fresh martensite during cooling from post weld heat treatment temperature. This transformation behavior was not observed in the base metal samples. The origin of such transformation in welds is attributed to microsegregation that occurs during weld solidification.

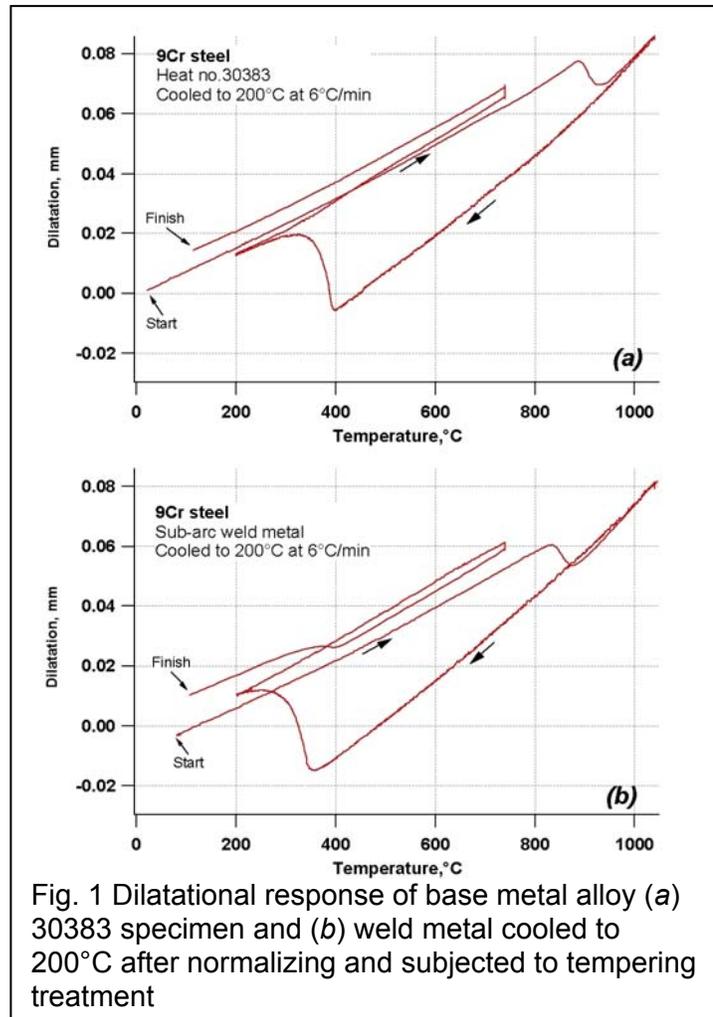
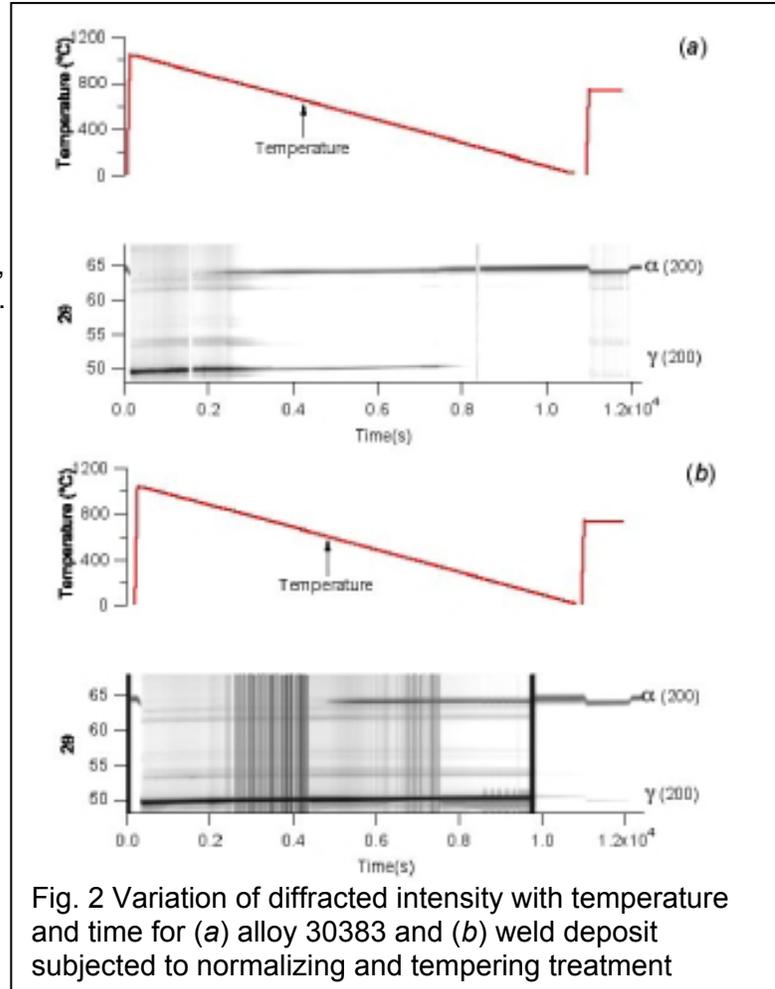


Fig. 1 Dilatational response of base metal alloy (a) 30383 specimen and (b) weld metal cooled to 200°C after normalizing and subjected to tempering treatment

Acknowledgements

This research was cosponsored by the Office of Fossil Energy, Advanced Research Materials Program, and by the Office of Industrial Technologies Program, Industrial Materials for the Future Program, both of which are in the U.S. Dept. of Energy. Work done at Oak Ridge National Laboratory is performed under Contract DE-AC05-00OR22725 with UT-Battelle, LLC. The UNICAT facility at the Advanced Photon Source (APS) is supported by the University of Illinois at Urbana-Champaign, Materials Research Laboratory (U.S. DOE, the State of Illinois-IBHE-HECA, and the NSF), the Oak Ridge National Laboratory (U.S. DOE under contract with UT-Battelle LLC), the National Institute of Standards and Technology (U.S. Department of Commerce) and UOP LLC. The APS is supported by the U.S. DOE, Basic Energy Sciences, Office of Science under contract No. W-31-109-ENG-38.



References

1. F. V. Ellis, J. F. Henry, and B. W. Roberts, "Welding, Fabrication, and Service Experience with Modified 9Cr-1Mo Steel," pp. 55-63 in *New Alloys for Pressure Vessels and Piping*, PVP Volume 201, American Society of Mechanical Engineers, NY, 1990.
2. T. Wada, *The Continuous Cooling Transformation Diagram and Tempering Response of 9Cr-1Mo-V-Nb Steels*, J-4672, Climax Molybdenum Company of Michigan, Ann Arbor, MI, 1981.
3. V. K. Sikka, C. T. Ward, and K. C. Thomas, "Modified 9Cr-1Mo Steel – An Improved Alloy for Steam Generator Application," pp. 65-84 in *Ferritic Steels for High-Temperature Applications*, edited by A. K. Khare, American Society for Metals, Metals Park, OH, 1983.
4. G. C. Bodine, C. Chakravarti, C. M. Owens, B. W. Roberts, D. M. Vandergriff, and C. T. Ward, *A Program for the Development of Advanced Ferritic Alloys for LMFBR Structural Application*, ORNL/Sub-4291/1, TR-MCD-015, Oak Ridge National Laboratory, 1977.
5. N. Saunders and A. P. Miodownik, *Calphad (Calculation of Phase Diagrams): A Comprehensive Guide*, Pergamon Materials Series, Volume 1, Elsevier Science Inc., New York, 1998.