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Time-dependent interfacial failure in metallic alloys

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Abstract

Time-dependent intergranular brittle fracture has now been studied experimentally in a number of alloy systems, and the generic features are becoming clear. Mobile surface-adsorbed elements are caused to diffuse inward along grain boundaries under the influence of a tensile stress, and this can lead to sub-critical crack growth by decohesion. Oxygen is found to play this role in nickel-base superalloys and intermetallics, as well as in a precipitation-strengthened Cu–Be alloy. Crack-growth rates lie in the range 10^{-7}–10^{-4} m sec^{-1}. The same kind of cracking is found in steels treated so that free sulfur is able to segregate to the surface, as well as in Cu-Sn alloys, in which the embrittling element is surface-segregated Sn. The latter has been studied in bicrystals, and the importance of the variation in diffusivity with grain boundary structure has been documented. Hydrogen-induced cracking is a special case of an extremely mobile embrittling element and is responsible for much of the brittleness found in intermetallics. The effect of boron in retarding brittle behavior in Ni_3Al has been shown to result partly from its interaction with hydrogen. This is a prime example of how segregated solutes can be used to ameliorate the tendency for diffusion-controlled brittle fracture. © 1999 Published by Elsevier Science S.A. All rights reserved.

Keywords: Crack-growth; Oxygen; Metallic alloys; Time-dependent intergranular brittle fracture

1. Nickel-base superalloys

A critical technological case of interfacial failure is that found in nickel-base superalloys, which are the primary materials used in gas turbines for aircraft engines and power generation. These alloys have an FCC matrix strengthened by ordered precipitates, and the alloys reach their maximum strength at temperatures around 700°C. They are subject to intergranular crack growth in air at these elevated temperatures, and this can be a life-limiting factor in components like turbine discs. The problem has been extensively studied in the context of cyclic loading, in which a hold-time is interjected between the loading and unloading parts of the cycle. In cyclic-loading at room temperature, crack growth occurs by the successive blunting and sharpening of a crack as a tensile load is applied repeatedly. The rate of crack growth is usually expressed as the variation of crack length per cycle, da/dN, which would not normally vary with a hold time. What is actually observed [1] at elevated temperatures in tests in air is an increase in da/dN with hold time, as indicated in Fig. 1a. There is a transition in behavior from pure cycle dependence to pure time dependence as the cycle period in increased (Fig. 1b). The latter type of cracking is characterized by an intergranular fracture mode. The tendency for this time-dependent behavior varies widely from one alloy to another, and the rate of crack growth increases strongly as temperature is increased and as the frequency is reduced.

The effect of a hold time is important, because this loading pattern is common in gas turbines as part of the normal duty cycle. In addition to the variability among alloys of different compositions, there are large effects in a given alloy from variations in heat treatment, with attendant variations in microstructure. The physical basis for such variability is still far from understood, but it is certain that the environmental effect is central to the phenomenon.

Recent results [2] from static loading of pre-cracked specimens of alloy IN718 in pure bending under constant-displacement at 650°C in air and oxygen at various pressures have been consistent with the results from cyclic loading. Fig. 2 shows the decrease in load with time that results from crack growth and the arrest of

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the cracking when the environment is switched from oxygen to vacuum; the cracking resumed when the oxygen was re-admitted to the chamber. The brittle nature of the intergranular fracture is also shown. The effect of oxygen pressure on the cracking rate, Fig. 3, is consistent with that reported by Andrieu et al. [3] for cyclic loading of IN718. The cracking apparently begins essentially immediately when oxygen is admitted to the chamber after loading in vacuum. Fig. 4a shows one of a number of patches (covering about 15% of the pre-crack front) that formed during two minutes of loading in oxygen at 650°C. This intergranular crack, with room-temperature fatigue cracks to the left and right, travelled about 120 μm in 2 min, for an average velocity of about 1 μm s^{-1}. The portion at a higher magnification in Fig. 4b shows what appear to be crack-arrest marks spaced a few tenths of a micrometer apart. This is commonly observed in polycrystalline materials that have cracked from dynamic embrittlement [4–6].

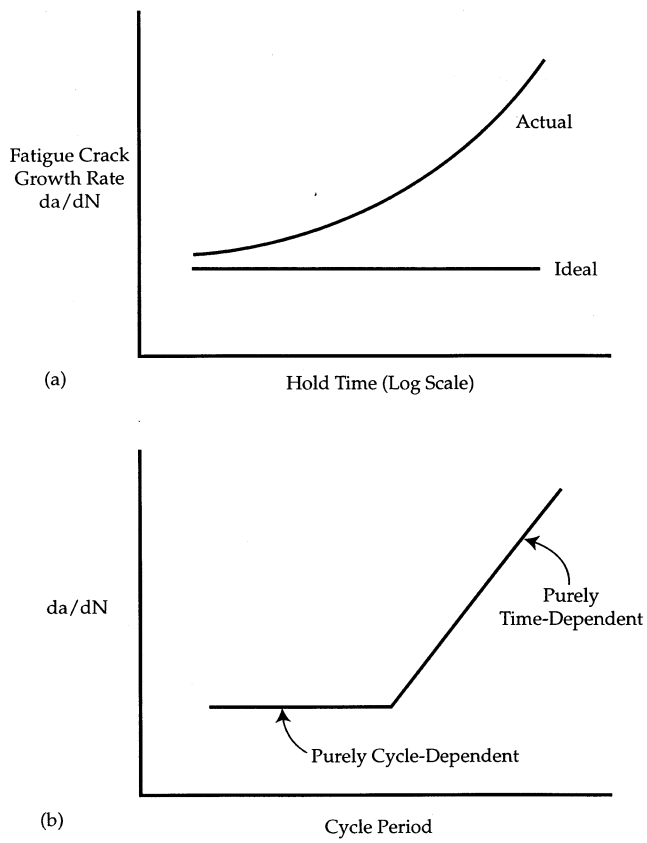

Fig. 1. (a) Effect of hold time on the crack-growth rate in many high-temperature structural materials, compared to the ideal case. (Ref. [1]) (b) Effect of cycle period on a non-ideal material, showing the transition from cycle-dependent to time-dependent behavior.

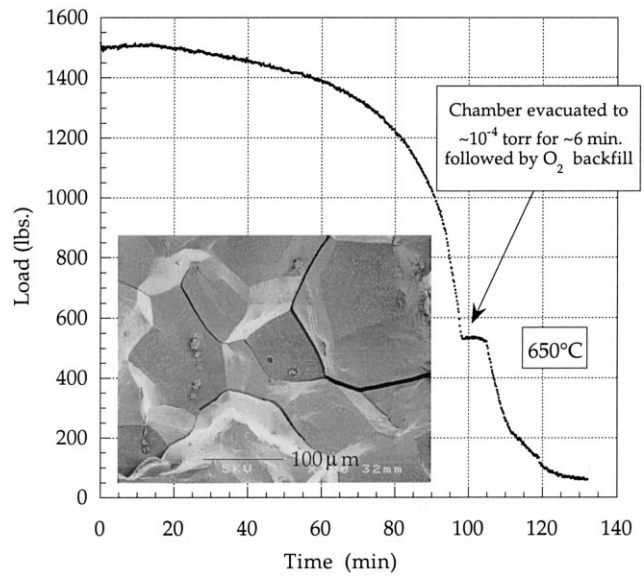

Fig. 2. Load relaxation due to cracking in IN718 in 1atm oxygen in a pre-cracked four-point-bend specimen loaded to a fixed displacement, showing the effect of removal, and then re-admission, of the oxygen to the test chamber. The appearance of the intergranular decohesion is also shown (Ref. [2]).

2. Intermetallics

Oxygen-induced cracking has also been studied in intermetallics, particularly Ni_3Al. Liu and White [7] coined the term 'dynamic embrittlement' to describe the intergranular brittle fracture in (boron and hafnium-doped) Ni_3Al tensile specimens tested in air at 600°C. (Taub et al. [8] had previously observed a loss of

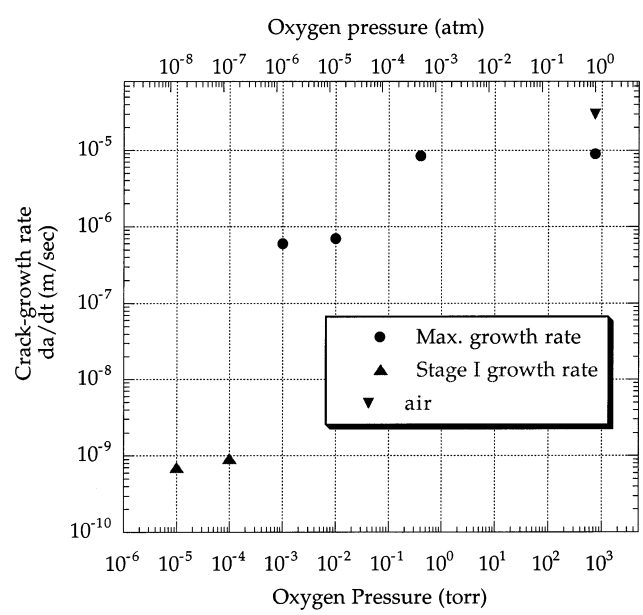

Fig. 3. Effect of oxygen pressure on the crack-growth rate in IN718 in a pre-cracked four-point-bend specimen loaded to a fixed displacement (Ref. [2]).

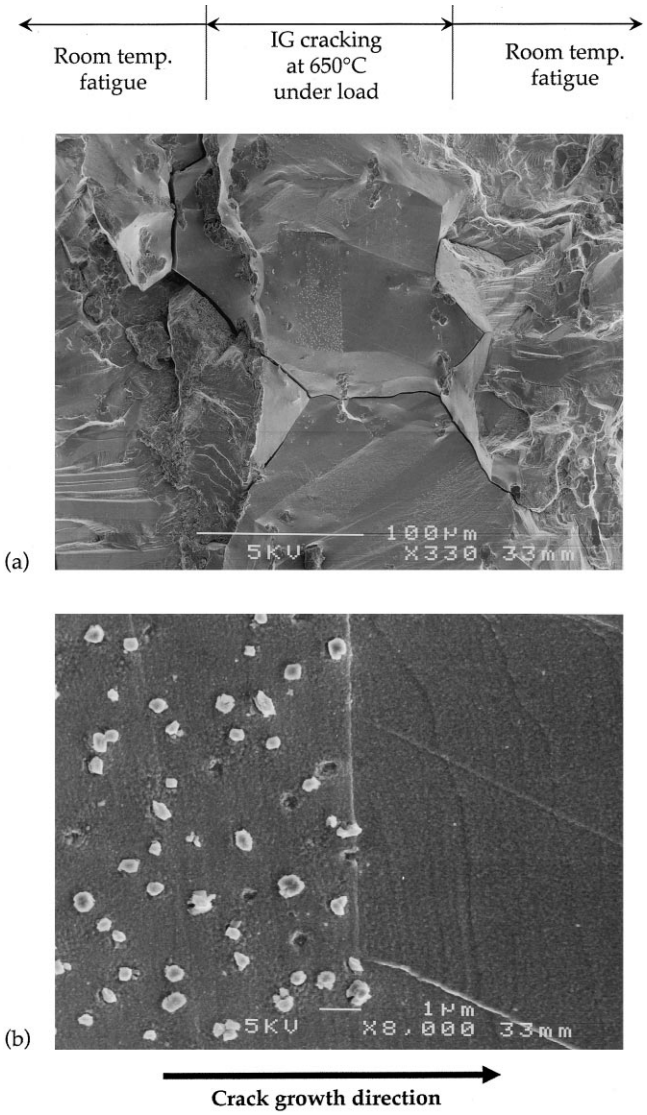

Fig. 4. (a) Intergranular cracking that occurred during two minutes of loading of IN718 in 1 atm oxygen in a pre-cracked four-point-bend specimen loaded to a fixed displacement; this occurred along about 15% of the pre-crack front. (b) Detail of a portion of (a) at higher magnification, showing oxidized carbides on part of the surface, as well as striations that appear to indicate that the cracking occurred in a step-wise manner (Ref. [2]).

ductility above 400°C in melt-spun Ni_3Al ribbons tested in argon.) Specimens tested by Liu and White in vacuum exhibited transgranular rupture after about 50% elongation, compared with the 4% elongation of the air-tested specimens. They proposed that adsorption of oxygen at crack tips produced the embrittlement. Subsequently, Hippsley and DeVan [6] measured the crack-growth rates in a similar alloy in air at 500–760°C in a constant-displacement load-relaxation test using the DC potential-drop method. They showed that the intergranular brittle fracture propagated at rates in the range 0.1–100 μm s^{-1}, depending on temperature. They estimated the distance of oxygen penetration

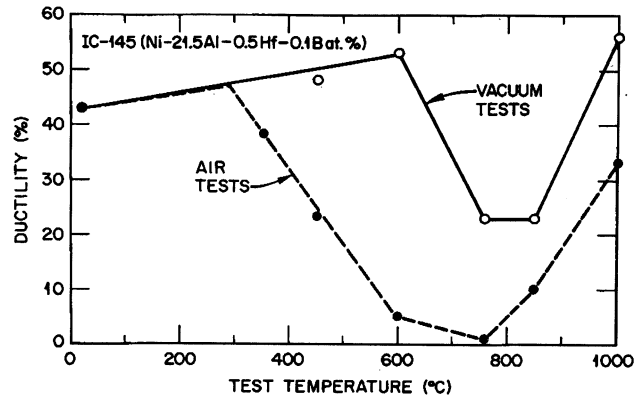

Fig. 5. Effect of temperature on the tensile ductility of a Ni-21.5 at.%Al-0.5 at%Hf 0.1 at%B alloy tested in air and at 10^{-3} Pa vacuum (Ref. [9])

along the grain boundaries in terms of normal diffusion, without taking account of the effect of stress. Takeyama and Liu [9] showed the large ductility minimum that occurs in tensile tests in air in an alloy similar to those used by Liu and White [7] (Fig. 5), as well as the dependence of the ductility on the pressure of air, down to the UHV range (Fig. 6). Pre-oxidation of fine-grained specimens had no effect on the ductility of subsequent tests in vacuum, but pre-oxidation did degrade the ductility in samples with coarser grain sizes. They found that a surface layer of aluminum-rich oxide formed on the fine-grained samples, presumably acting as a barrier to oxygen penetration, but that the oxides formed on the coarser-grained samples was nickel-rich and presumably not protective. This points up the potential difficulty in separating the stress-induced penetration of a surface element from the penetration that comes from a concentration gradient alone. A ductility minimum at 600°C has been demonstrated in Ni_3Si [10], and Takasugi et al. [11] have reported that the addition

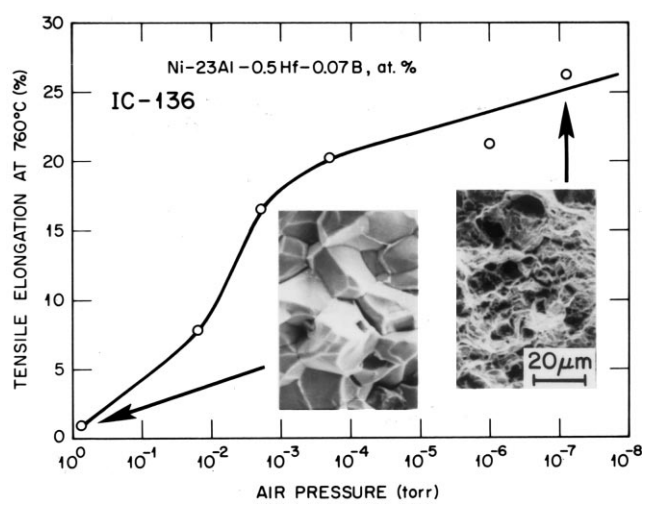

Fig. 6. Effect of air pressure on the tensile ductility of a Ni-23 at.%Al-0.5 at.%Hf 0.07 at.%B alloy tested at 760°C (Ref. [9]).

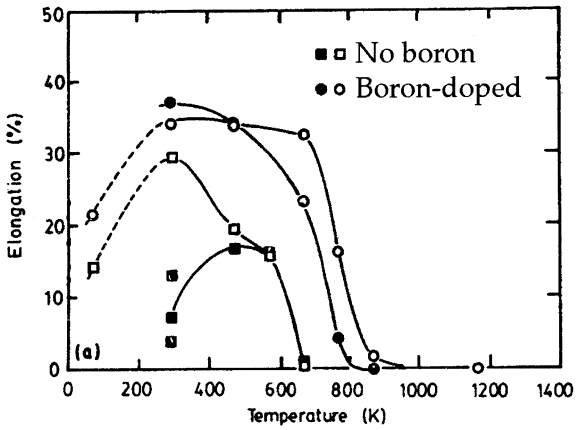

Fig. 7. Effect of temperature on the tensile ductility of $Ni_3(Si, Ti)$ alloys with and without boron tested in air and in vacuum (Ref. [11]).

of boron to $Ni_3(Si, Ti)$ moves the ductility minimum to higher temperatures, as shown in Fig. 7. This suggests that boron, which segregates to grain boundaries, interferes with the diffusion of oxygen along the boundaries.

3. Copper-base alloys and steels

An alloy of Cu-0.25%Be in the precipitation-hardened condition has been used as a simple model material for oxygen-induced cracking [12]. The dramatic effect of the applied stress is shown in Fig. 8 for constant-load tests on notched tensile bars in air at 200°C. At a net-section stress of about 50% of the yield stress, the failure occurred in less than 1 h. The fractures occurred by intergranular decohesion, as shown by Fig. 9. A specimen loaded almost to the yield stress in a vacuum of 2×10^{-6} torr did not fail in 148 h, but sectioning of the specimen revealed a crack about 400 μm long. This would correspond to a crack growth rate of about 45 nm s^{-1}. The cracking was highly temperature dependent, as shown by the load-relaxation curves in Fig. 10, and the temperature dependence of the maximum growth rate was found to match that of the estimated rate of intergranular diffusion of oxygen in copper (R.C. Muthiah, J.A. Pfaendtner, C.J. McMahon Jr., 1997, unpublished data). This would indicate that the zone of penetration of the oxygen into the copper is extensive enough that the rate is controlled by the diffusivity along the grain boundary. This is consistent with the observation in Fig. 4 that cracking in the IN718 occurs faster along some grain boundaries than along others and with similar observations made with regard to sulfur-induced cracking in steels [5]. It is notable that in the commercial Cu-0.25%Be alloy an addition of 400 ppm zirconium has been found to get rid of the problem of hot shortness (i.e. cracking during hot working) in this material, and it was also found to suppress completely the cracking illustrated in Figs. 8 and 9 [13]. The zirconium is presumably segregated at the grain boundaries and interferes somehow with the oxygen penetration.

In contrast to the environmental effects discussed so far, steels can suffer the same kind of cracking from the effects of dissolved sulfur that segregates to free surfaces, either internal or external. This has been studied extensively in the context of stress-relief cracking of steels [5], in which the sulfur is supplied by sulfides that dissolve when the steel is austenitized at a high temperature and then cooled somewhat rapidly. The sulfur segregates to austenite grain boundaries and is trapped there, either as solute or as fine precipitates, during cooling. When the steel is then re-heated and the relaxation of stress, either internal or external, is allowed to occur by creep, intergranular cracks can occur in regions of maximum tensile stress, and these cracks can propagate in the same manner as indicated above. In this case, the role of the embrittling element is taken on by the sulfur that segregates from the alloy itself.

After the mechanism of stress-relief cracking of steels was deduced [14], it was postulated [15] that this was but one example from a generic set and that one should be able to observe similar behavior with any structural alloy having a surface-adsorbed, low-melting-point element when the alloy is loaded in tension at an appropriate temperature. This idea was tested with an alloy of Cu-8%Sn, which is well known to be hot short. Tin, a low-melting element, is known to be surface active in such alloys [16], and it is known to cause temper embrittlement in steels. It was found [17] that the Cu-Sn at around 250°C behaved in a manner identical to the steels at around 550°C, so this was used as a simple model material for the steel.

This model material has recently been used to study the effect of intergranular diffusivity on the cracking rate (R.C. Muthiah and C.J. McMahon, Jr., 1997, unpublished data). Bicrystals of Cu-7%Sn having a $\Sigma = 5$ (031) [001] symmetrical tilt boundary were tested as pre-cracked specimens at constant displacement in bending at 265°C in vacuum. In one set of experiments, the pre-crack was oriented so that the cracking direction was parallel to the tilt axis, that is, along the 'diffusion pipes' (Fig. 11), and in the other the cracking direction was perpendicular to the tilt axis. Previous diffusion studies in bicrystals have shown that the diffusion coefficient in the parallel direction is generally significantly greater than in the perpendicular direction [18,19]. Cracking in the parallel direction was found to occur at a very low stress intensity (Fig. 12b) and at a maximum rate of about 2 μm s^{-1} (Fig. 12a). It was found to be impossible to cause cracking in the perpendicular direction; the grain boundary cavitated under load, but it would not crack. This is compelling evidence that the stress-induced tin penetration involves a significant amount of grain-boundary diffusion.

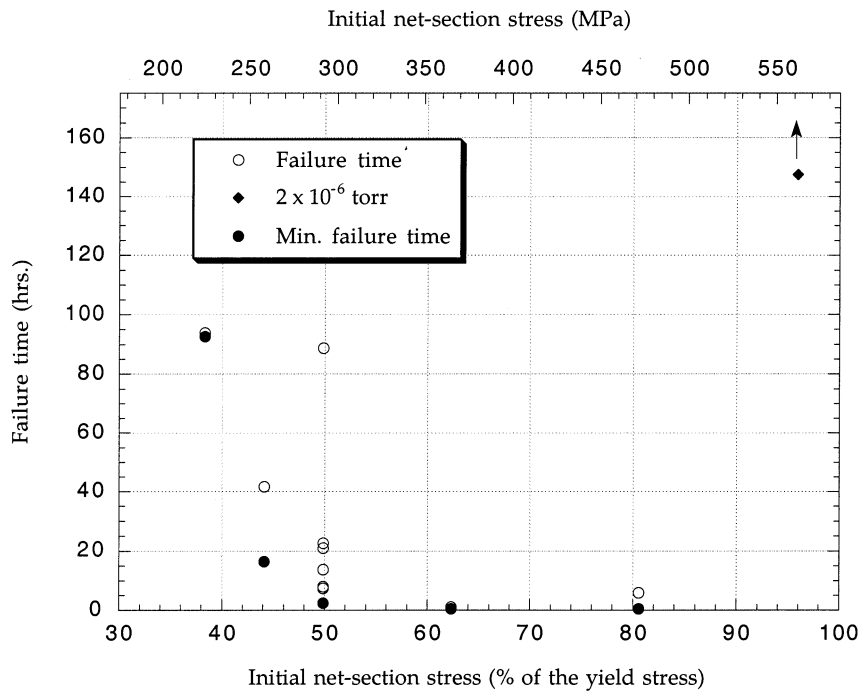

Fig. 8. Effect of applied stress on the time to failure of notched tensile bars of a Cu 0.25% Be alloy tested at constant load at 200°C in air and at 2×10^{-6} torr (Ref. [12]).

4. Model for dynamic embrittlement

The idea of stress-induced penetration along grain boundaries by a surface-adsorbed embrittling element [14] was later [15] elaborated by a model in which the relevant diffusion equation could be used, in principle, to calculate a crack-growth rate. This is a modification of Fick's second law to include the effect of a tensile stress σ:

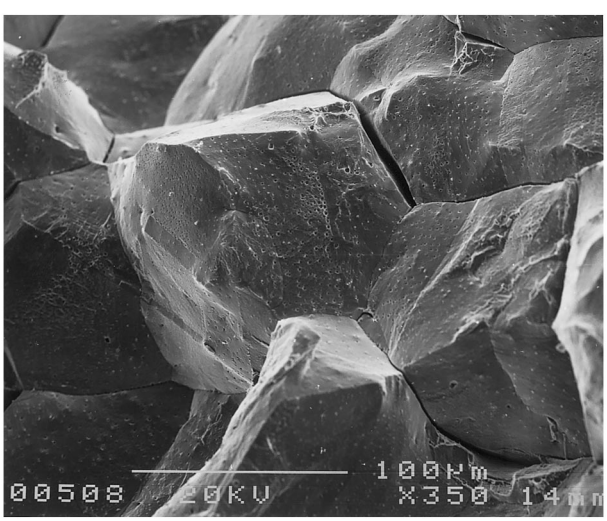

Fig. 9. Fracture surface of the Cu-0.25%Be alloy that failed under constant load at 63% of the yield stress at 200°C in air (Ref. [12]).

$$\frac{\partial C}{\partial t} = D_b \frac{\partial^2 C}{\partial x^2} - \frac{D_b \Omega}{kT} \frac{\partial}{\partial x}\left[C \frac{\partial \sigma}{\partial x} \right]$$

where D_b is the intergranular diffusion coefficient, and C and Ω are the concentration and the atomic volume of the diffusing element. To calculate a crack-growth rate, one would have to know the intergranular diffusion coefficient of the embrittling element, the stress profile ahead of the crack (presumably time-dependent), and a fracture criterion, meaning the dependence of the cohesive strength on the concentration of the embrittling element. It was assumed that the cracking would be intermittent: The embrittling element would diffuse in, and the diffusion zone would crack open when the concentration reached a critical value, which would depend on the stress level. Calculations were made using assumed stress profiles that gave reasonable cracking rates [4], but the weakness of the calculations was in these assumptions. More recently, Xu [20] has assumed a steady-state type of cracking and has employed a cohesive-zone model for the stress profile. Her equation for the concentration ahead of the crack as a function of time has two terms in the exponent.

$$C(x) \approx C_0 \exp[(-\dot{a}/D_b)x + \Omega/kT(\sigma(x) - \sigma_0)]$$

When the appropriate values of crack-growth rate $\dot{a}$ in the Cu–Sn bicrystals, the diffusion coefficient, and other constants are put into the equation, the second term in the exponent would be of order unity, since the stress term would be a few tens of MPa. For the first term to be of roughly the same magnitude, the diffusion

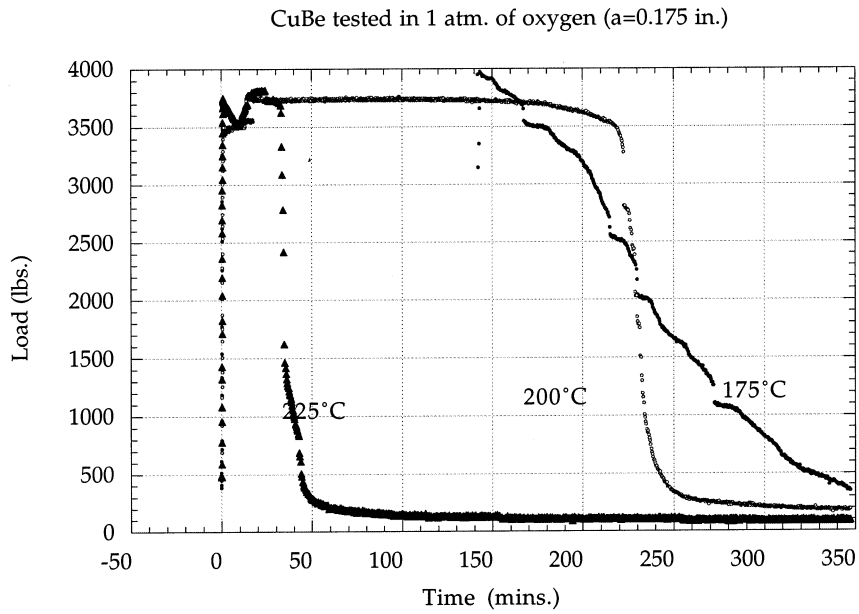

Fig. 10. Load relaxation due to intergranular cracking in a Cu-0.25%Be alloy tested as pre-cracked four-point-bend specimens loaded to a fixed displacement in air at three different temperatures (R.C. Muthiah, J.A. Pfaendtner, C.J. McMahon Jr., 1997, unpublished data).

distance x would have to be on the nanometer scale. In other words, the zone of penetration would be on the scale of the core of a sharp crack, as would have to be the case for steady-state crack growth. Thus, steady-state cracking could not be intermittent in a freely cracking bicrystal. This model appears to be consistent with the observations on cracking of the Cu–Sn bicrystals.

5. Environmental effects at ambient temperatures

The cracking phenomena discussed so far have involved embrittlement by mobile elements at temperatures well above the ambient. The most mobile embrittling element of all is hydrogen, and hydrogen-induced cracking is usually observed mainly at ambient temperatures, for example in steels, which have received an extraordinary amount of attention for many decades. More recently, it has been found that the L_{12} ordered intermetallics that exhibit brittle intergranular fracture and low tensile ductility at ambient temperatures are not intrinsically brittle as had been previously believed [21,22]. Liu and George et al. [23–25] first reported that the grain boundaries in Ni_3Al can be quite strong and that the brittle intergranular fracture is caused mainly by moisture-induced hydrogen embrittlement. The embrittlement process is thought to involve the reaction of moisture in the air with a reactive element in intermetallic alloys, such as A1 in Ni_3A1 alloys, resulting in the generation of hydrogen. Environmental embrittlement has been observed in almost every strongly ordered intermetallic alloy containing reactive elements, such as Al, Si, Ti, V, etc. [26–28]. The severity of moisture or hydrogen-induced embrittlement is dependent on the kinetics of molecular dissociation and hydrogen diffusion at crack tips and on the strength of grain-boundary cohesion. In the case of Ni_3Si, both low cohesion along grain boundaries and environmental embrittlement are important. On the other hand, the environment is apparently the sole source of intergranular embrittlement in Co_3Ti and $(Co, Fe)_3V$. Because of the time dependence of the reaction, the environmental embrittlement is sensitive to strain rate and test temperature, as well as to the gaseous atmospheres employed in tests.

Liu et al. [23,24] discovered the moisture-induced environmental embrittlement in binary Ni_3Si and Ni_3Al alloys tested at room temperature in various environments. George et al. [25,29] reported an even larger environmental effect in polycrystalline Ni-24at%Al prepared by recrystallization of cold-worked monocrystalline material. The tensile ductility of this material increased from 3% in air to 23% in an ultra-high vacuum of 10^{-8} Pa, as indicated in Fig. 13.

A recent XPS study of the chemical reaction of H_2O with Ni_3Al alloys [30] reported that H_2O absorbs on the surface and then dissociates into hydroxyl ions and atomic hydrogen. The hydroxyl ions were found to associate preferentially with Al atoms, whereas atomic H was associated preferentially with Ni atoms. This reaction can be written as:

$$Al + H_2O \rightarrow Al(OH) + H$$

This generates atomic hydrogen that could penetrate into crack tips and cause hydrogen-induced embrittlement. Consistent with this interpretation, a high ductil-

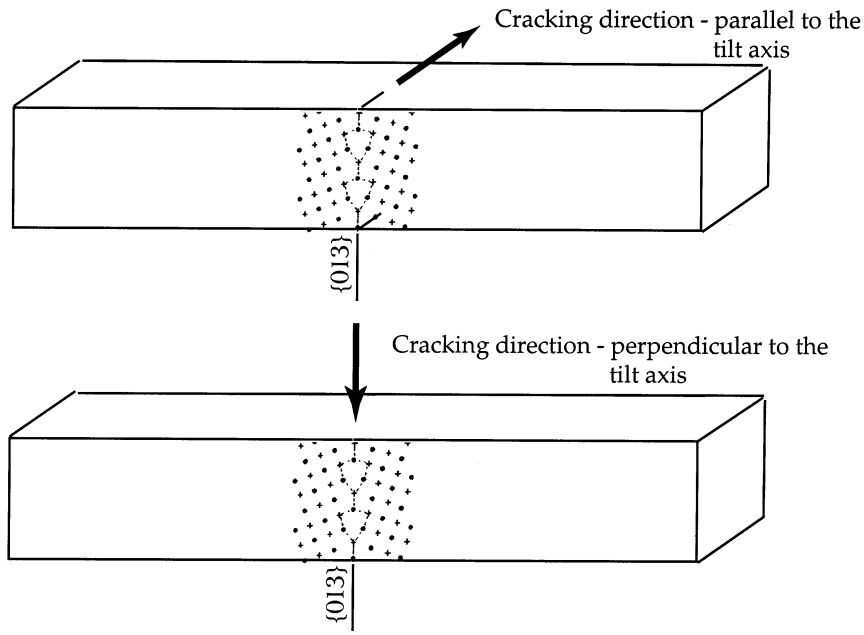

Fig. 11. Schematic illustration of the two directions of crack growth in the Ω: = 5 (031)[001] bicrystals of Cu-7%Sn (R.C. Muthiah and C.J. McMahon, Jr., 1997, unpublished data).

ity is obtained in a vacuum back-filled with dry oxygen, which would form A1 oxides without hydrogen as a byproduct [26–28].

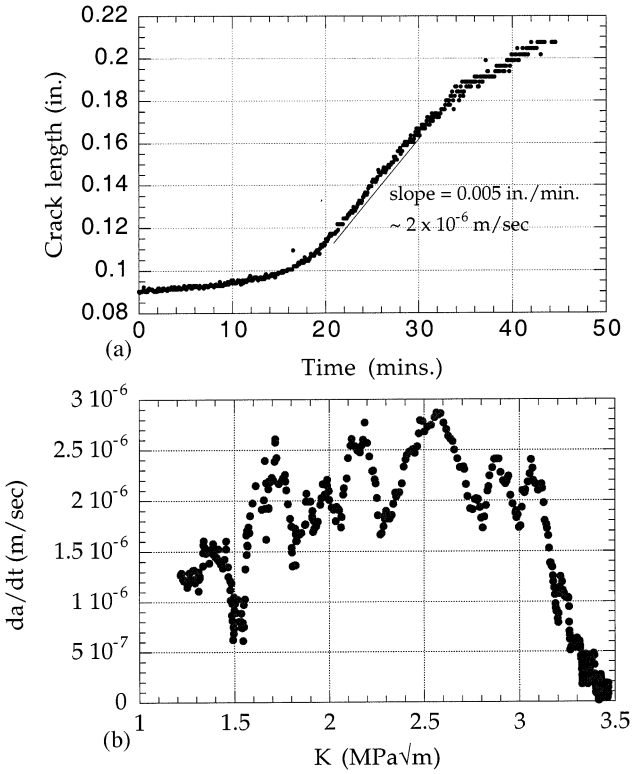

Fig. 12. (a) Crack length versus time in a Cu-7%Sn bicrystal cracked parallel to the tilt axis. (b) Crack velocity versus stress intensity in the same specimen (R.C. Muthiah and C.J. McMahon, Jr., 1997, unpublished data).

The severity of the embrittlement is thought to be affected by two kinetic steps: dissociation at surfaces and hydrogen diffusion from surface to interior. The chemical reaction is expected to occur at the places where 'fresh' Al atoms are available because of plastic deformation or cracking. Both slip steps on surfaces and surface cracks produced during the early stages of plastic deformation are possible sites for this reaction. In consideration of the reaction kinetics, moisture-induced hydrogen embrittlement would be expected to decrease at high strain rates. This has indeed been observed experimentally, as shown in Fig. 14 [27,29–31].

With increasing temperature, the rates of dissociation of adsorbed water and of hydrogen penetration should decrease; the sticking probability of the H_2O (or H_2) molecules on intermetallic surfaces should also decrease, and the rate of oxide formation should increase. Both nickel and iron aluminides show a ductility minimum at around room temperature [29]. As shown in Fig. 15, Fe-36.5wt%Al goes through a minimum at around 70°C; in dry oxygen, on the other hand, its ductility remains relatively high over the entire temperature range. Li and Liu [32] considered several thermodynamic and kinetic factors and produced a quantitative model of the effect of temperature on the ductility of FeAl that successfully reproduced the shape of the experimental curve in Fig. 15. They concluded that the minimum in ductility due to environmental embrittlement was satisfactorily explained by a kinetic effect at low temperatures and an equilibrium effect at high temperatures.

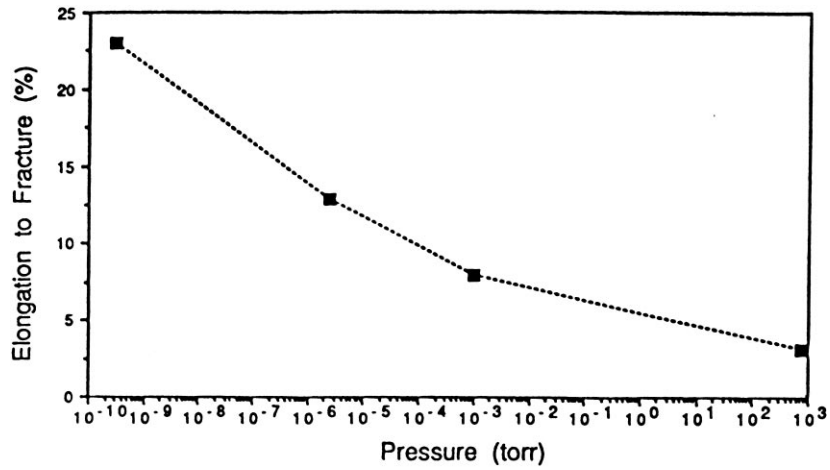

Fig. 13. Room temperature tensile ductility of polycrystalline Ni-23.4 at.%Al as a function of vacuum level [25,29].

Moisture-induced embrittlement of aluminides at ambient temperatures is considered to be a dynamic process involving a number of sequential steps: (i) formation of 'fresh' surfaces containing unreacted Al atoms during the initial stage of deformation; (ii) chemical reaction of Al atoms with moisture and generation of atomic hydrogen; (iii) hydrogen penetration into the stress field ahead of crack tips; and (iv) inward growth of surface cracks along grain boundaries in Ni_3Al or cleavage planes in FeAl and Fe_3Al. No detailed studies have yet been conducted on the micromechanics of crack growth in the time-dependent intergranular fracture of these intermetallic alloys.

6. Effects of boron additions

It has been found that environmental embrittlement in Ni_3Al and other intermetallic alloys at room temperature can be alleviated or eliminated by alloying additions [33,34]. Boron appears to be most effective in eliminating brittle grain-boundary fracture and increasing the tensile ductility of Ni_3Al with hypo-stoichiometric compositions (< 25at.%Al) [20]. Unalloyed Ni_3Al containing 24% A1 exhibited grain-boundary fracture with little tensile ductility when tested in air at room temperature, but the ductility increased sharply with boron doping and reached more than 40% at > 200 wt ppm B, as shown in Fig. 16. The increase in ductility was accompanied by a change in fracture mode from brittle intergranular to ductile transgranular. When > 200 wt ppm B is added to Ni_3Al (24% Al), the moisture-induced embrittlement was completely suppressed.

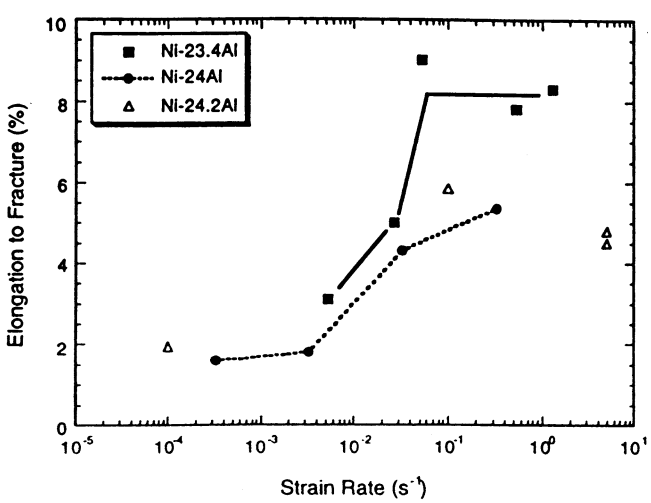

Fig. 14. Effect of strain rate on the room-temperature tensile ductility of three hypo-stoichiometric Ni_3A1 alloys in air [27].

Considerable effort has been focused on understanding the beneficial effects of boron additions. Both experimental and theoretical studies indicate that boron may have three beneficial effects:

1. Segregation to grain boundaries and enhancement of grain-boundary cohesion,
2. Inhibition of hydrogen diffusion in grain boundaries, and
3. A decrease in the surface reactivity with H_2O.

Early atomistic calculations indicated that boron atoms may occupy atomic size defects at Ni_3Al grain boundaries and enhance the grain-boundary cohesion [35,36]. Bond et al. [37] first provided experimental evidence, based on in situ TEM studies, that boron enhances grain-boundary cohesion.

Wan et al. [38] reported that boron doping reduced hydrogen diffusivity along Ni_3Al grain boundaries. By cathodically charging with hydrogen and measuring the depth of the intergranular fracture region, they estimated that doping with boron lowers the hydrogen diffusivity by more than an order of magnitude.

Cohron et al. [39] recently investigated the effects of boron doping on the room-temperature ductility and

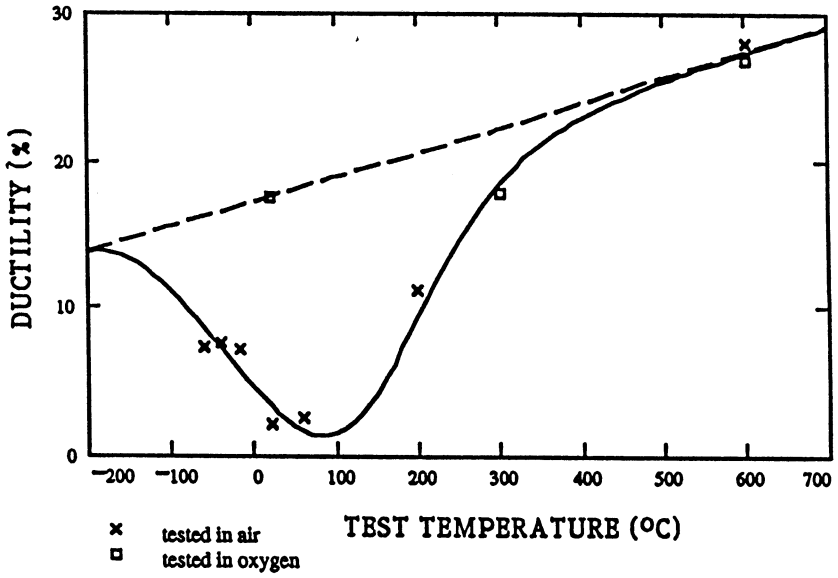

Fig. 15. Tensile elongation of B2 FeA1 (36.5 wt% A1) tested in moist air and oxygen [32].

fracture behavior of Ni_3Al (24% Al) in dry H_2 gases. They found that H_2 was not embrittling in B-free Ni_3Al as long as the filament of the ion gage used for measuring the degree of vacuum was turned off [39,40]. When the H_2 gas was dissociated by the hot filament, the ductility decreased dramatically, and the fracture mode switched from transgranular to intergranular. They concluded that pure H_2 does not dissociate easily on a clean surface of B-free Ni_3Al.

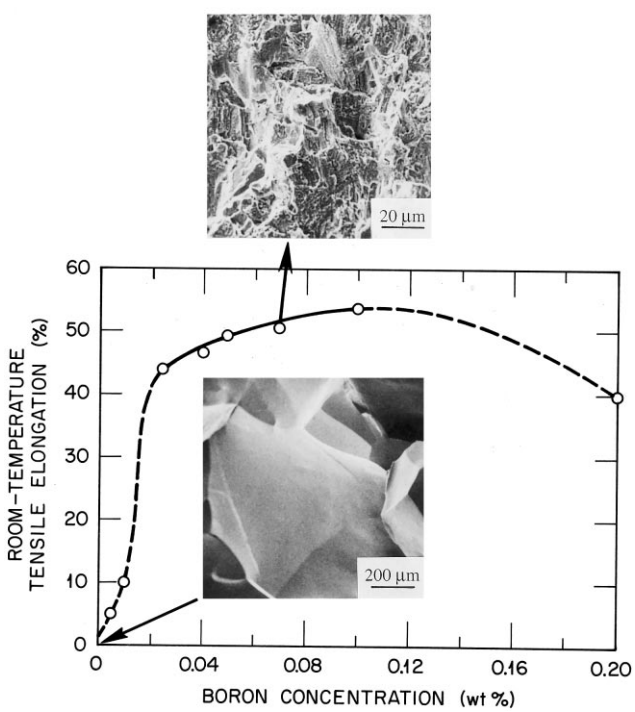

Fig. 16. Effect of boron additions on tensile elongation and fracture behavior of Ni-24at.%Al) tested at room temperature [22].

When they tested a boron-doped alloy in H_2, they found little change in ductility with increasing H_2 pressure at first, but when the pressure reached a critical value, the ductility dropped precipitously (Fig. 17). At all H_2 pressures employed, the higher the boron concentration the higher the elongation to fracture. It did not matter whether the filament of the ion gage was on or off for the specimens that contained boron. At high H_2 pressures they found that the boron-doped alloy showed intergranular fracture and significantly lower ductility, compared with the boron-free alloy. They suggested that boron promotes the dissociation of H_2 on the surface of Ni_3Al. At high pressures this appears to outweigh any effect of boron on the diffusion of hydrogen in grain boundaries.

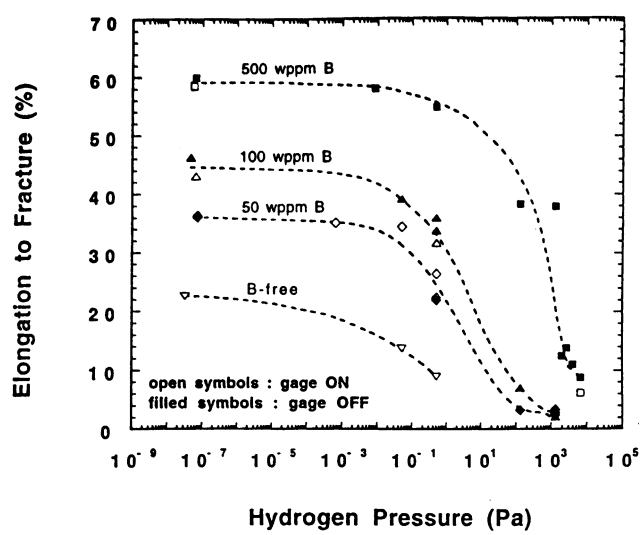

Fig. 17. Effect of H_2 pressure on the ductilities of B-free and B-doped Ni_3A1 alloys [40].

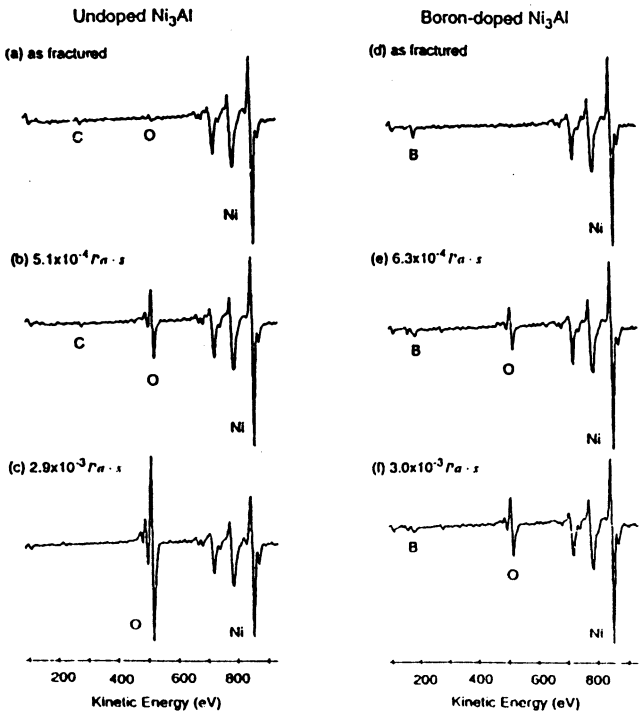

Fig. 18. Auger spectra from intergranular fracture surfaces of undoped (a); (b); (c) and boron-doped (d); (e); (f), Ni_3Al. (a) and (d) as fractured; (b) through (f) various pressures of water. Y-axis: $E\mathrm{d}N(E)/\mathrm{d}E$, arbitrary scale [42].

The concentration of boron at Ni_3Al grain boundaries increases with its bulk concentration, as observed by Choudhury et al. [41]. The increase in tensile ductility with bulk boron concentration in Fig. 17 is consistent with the report that segregated boron reduces hydrogen diffusivity in the grain boundaries and thereby hinders the uptake of hydrogen.

Recently, Lee et al. [42] studied the effect of boron segregation on the interaction of water vapor with Ni_3Al grain boundaries. They found that boron can suppress the moisture-induced embrittlement completely. In their study, Ni_3A1 specimens with and without boron (650 at. ppm) were fractured intergranularly in an Auger system in order to monitor the H_2O/Ni_3Al interaction on the fracture surfaces. The comparison of the oxygen peaks shown on the Auger spectra (Fig. 18) obtained from undoped and boron-doped fracture surfaces indicates that boron enrichment (at a level of 8at.%) at the grain-boundaries significantly reduces the kinetics of the surface reaction between water vapor and Ni_3Al.

In summary, there is good evidence that moisture-induced hydrogen embrittlement is the major cause of the intergranular brittle fracture of Ni_3Al alloys in air. The embrittlement appears to involve the reaction of Al with water in the air and the generation of atomic hydrogen, which penetrates into the stress field ahead of crack tips and causes brittle grain-boundary fracture. Similar embrittlement effects can be observed when L_{12} intermetallics are tested in H_2. Boron is found to be effective in suppressing moisture-induced hydrogen embrittlement in Ni_3Al alloys. Segregated boron appears to have four effects on environmental embrittlement in Ni_3Al alloys:

1. Enhancement of grain-boundary cohesion;
2. Slowing down hydrogen penetration along grain boundaries;
3. Decrease of surface reactivity with H_2O; and
4. Enhancement of surface reaction with H_2.The first three are beneficial effects and the last one is harmful.

It is believed that the cracks formed by environmental embrittlement propagate dynamically; however, there is a lack of understanding of the nature of the crack-tip phenomena. Further work is required to understand crack-tip processes in terms of the hydrogen-penetration depth, the stress-induced diffusion process, and the mode of crack propagation (stepwise or continuous). Of course, the modeling of the cracking process by micromechanics should be very useful for understanding the moisture-induced hydrogen embrittlement in intermetallic alloys.

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