Applied Stress Affecting the Environmentally Assisted Cracking

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Stress corrosion cracking (SCC) is affected by the mode of applied stress, *i.e.*, tension, compression, or torsion. The cracking is measured in terms of initiation time to nucleate a crack or time to failure. In a simple uniaxial loading under tension or compression, it is observed that the initiation time can vary in orders of magnitude depending on the alloy and the environment. Fracture can be intergranular or transgranular or mixed mode. Factors that affect SCC are solubility of the metal into surrounding chemical solution, and diffusion rate (like hydrogen into a tensile region) of an aggressive element into the metal and liquid metallic elements in the grain boundaries. Strain hardening exponent that affects the local internal stresses and their gradients can affect the diffusion kinetics. We examine two environments (Ga and 3.5 pct NaCl) for the same alloy 7075-T651, under constant uniaxial tension and compression load. These two cases provide us application to two different governing mechanisms namely liquid metal embrittlement (7075-Ga) and hydrogen-assisted cracking (7075-NaCl). We note that, in spite of the differences in their mechanisms, both systems show similar behavior in the applied K vs crack initiation time plots. One common theme among them is the transport mechanism of a solute element to a tensile-stress region to initiate fracture.

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I. INTRODUCTION

IT is recognized that a tensile stress is necessary for SCC to occur.^[1] In the absence of any external or internal stress (like residual stress), metals can undergo general corrosion. For a given alloy/microstructure, environmentally assisted cracking (EAC) behavior under an applied tensile stress is commonly discussed in general terms as due to

- (a) Active path dissolution (APD or AD) involving dissolution of metallic elements, sometimes preferentially at slip bands or at grain boundaries (GBs) or of specific precipitates at the boundaries^[2];
- (b) Liquid metal embrittlement (LME) involving a liquid metal penetrating into GB or dissolving the host metal to induce embrittlement^[3];
- (c) Hydrogen-assisted cracking (HAC) involving hydrogen-enhanced decohesion, hydrogen-enhanced localized plasticity (HELP), adsorption-induced dislocation emission or reduction in cohesive energy, particularly, as in Fe-, Ti-, Ni-, and Al-based alloys^[4];

In the literature, there has been greater emphasis on the kinetics of crack growth than on the initiation thresholds, even though damage involves both the initiation of a crack and its growth. Not much experimental data are available on the initiation/incubation times for SCC. Incubation time appears to depend on the tensile or compressive stress at the crack tip. In

particular, under uniaxial tension or compression, the initiation time varies significantly with alloy–environment system. It is observed that SCC initiation times under tensile stress can be one to two orders of magnitude shorter compared to that under compression load. For a crack to initiate under compression, there has to be a local tensile stress at some point away from the notch. Even if this crack forms, it will not grow until the crack-tip driving force becomes tensile. The current article examines the role of applied stress on the EAC crack initiation time in terms of applied stress intensity factor, K_{app} , and the factors controlling the crack initiation time.

II. EARLY EXPERIMENTAL OBSERVATIONS

Early investigators^[5] had observed that SCC in stainless steels could occur also under compression load, using the horseshoe-type specimens. They observed that corrosion occurs on the compressive side of a bent sample, albeit at a lower rate than on the tensile side. Since then, there have been several qualitative experimental observations, using bent sheet samples, showing that SCC can occur under compressive stress in 304 stainless steels. Similar results were observed in other steels. For example, SCC occurs under compression in boiling 42 pct $MgCl_2$ solution^[6] in 1015 mild steel, and boiling nitrate solution in 1017 mild steels.^[7] It was observed that the time to nucleate a SCC crack under far field compression was about one to two orders of magnitude longer than that under tension. The results were attributed to slower dissolution under compression than in tension.^[6,7] Dissolution occurs on freshly created surface, when the passive films are broken by local plastic deformation. The process can be continuous on a

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macroscopic scale. The dissolution of a fresh surface is slower under compression than in tension. Alternatively, slower dissolution under compression can be independent of the slip deformation process. On the other hand, if the incubation time and the related stress effects are due to HAC, then two possibilities can exist: (1) HAC crack initiation occurring close to the crack/notch tip; and (2) crack initiation occurring ahead of the crack/notch tip. HAC can occur at or ahead of the crack tip depending on where the peak tensile hydrostatic stress is located. For the case in which HAC occurs ahead of the crack tip, hydrogen diffusion via GB or matrix is needed; in this case, H-diffusion kinetics can be a controlling factor. On the other hand, for HAC occurring at the crack/notch tip, no H-diffusion into metal is required. HAC at the crack tip can be reaction-rate controlled if H-generation is slow, or surface diffusion is controlled along the crack/notch surface if H-generation is fast. The relative kinetics of crack-tip reaction and the associated surface diffusion *vs* the diffusion of hydrogen into the metal become the governing factor in determining the incubation time for crack formation as well as for crack growth kinetics^[8]. In either case, the relative roles of applied tensile or compressive stresses need to be understood and characterized.

III. MATERIALS AND EXPERIMENTS

A set of data was published by Chu *et al.*^[9,10] under uniaxial tension and compression loads for 7075-T651. The tests on the 7075 alloy were performed in two different environments: (1) liquid gallium, and (2) 3.5 pct NaCl solution at pH = 3.5. The fractographic analyses of SCC failures under tension and compression were also reported for each of the alloys.^[9,10] The mechanical properties for 7075-T651 were yield strength (YS) = 510 MPa, and strain hardening coefficient $n = 0.113$. 7075-T6 alloy deforms by mixed planar and wavy slip.

In the following sections, for convenience, we label the two main alloy–environment systems as: 7075-Ga and 7075-NaCl. In all the tests, the modified wedge opening load (WOL) samples^[10] under plane stress (2.5 mm thick with notch radius $\rho = 0.1$ mm) were used without precracking for the case of 7075-NaCl. 7075-Ga test was under plane strain (10-mm-thick samples with $\rho = 0.15$ mm). The crack initiation time was monitored using clip gage and optical methods with accuracy ranging from 10 to 30 μ m in crack length measurement. See original papers^[9,10] for details. Testing in un-precracked condition has some advantages. For notches of finite radius, the peak stress $\sigma_{\max}$, remains the same, except for a change in sign, for both tension and compression loadings. In all the tests reported, the notch tip radius remained finite. Owing to lack of specific data, we assume that, for the purpose of our analysis, the local chemistry at the notch tip is not different from the bulk chemistry. In essence, the homogenization of the aqueous solution in the notch region is faster than the reaction or diffusion kinetics.

IV. GENERAL TRENDS IN 7075-GA AND 7075-NACL SYSTEMS

Initiation times in LME can be affected by the nature of the applied stress as shown in Figure 1(a). In some systems, LME is so fast that incubation times cannot be measured easily. In 7075-Ga system, the oxide scale on the base metal can prevent liquid metal contact providing a fictitious incubation times. In those cases, a simple scratching of the oxide scale can hasten the LME process.^[11] Most importantly, the trend in Figures 1(a) and (b) indicate that the behavior of LME is not different from that in the aqueous NaCl system, implying that any mechanistic explanation of the behavior should be generic and applicable to both.

Figures 1(a) and (b) indicate that there is a power-law relation between the initiation time and applied K (K_{app}). In the current analysis, notch with large depth was considered as a crack in the computation of K_{app} and is computed using standard specimen compliance relation. Slope “ m ” for the linear segments in each plot (Figures 1(a) and (b)) was obtained by applying regression analysis, and the equation is expressed as

$$K_{\text{app}} = A (t_{\text{init}})^{-m} \quad [1]$$

where t_{init} = crack initiation time at a given K_{app} . As noted, this relation (Eq. [1]) is valid for both the systems. The above power law is valid in the intermediate range of K , with K_{Ic} and K_{Isc} forming the upper and lower limiting values, respectively, indicating that the same mechanism of cracking remains throughout the range. As applied K is reduced, initiation time increases and finally reaches a plateau value that corresponds to the SCC threshold K (K_{Isc}). In both cases, we note that crack initiation occurs faster under tension than in compression. Some similarities and differences in the exponent “ m ” are noteworthy. The exponents “ m ” differ for SCC in 7075-NaCl but are same for LME in 7075-Ga system. Most importantly, the magnitude of “ m ” is nearly the same for both tension and compression although the kinetics is slower under compression.

Since K_{Ic} (upper bound) and K_{Isc} (lower bound) are independent of time, the slope in the two sets of data in Figure 1 can be represented in terms of

$$K_{\text{app}} = C (t_{\text{init}})^{-m}, \quad \text{where } m > 0 \quad [2]$$

where $t \rightarrow 0$, and $C = K_{\text{Ic}}$. Then, Eq. [2] reduces to $K_{\text{app}} = K_{\text{Ic}} (t_{\text{init}})^{-m}$, suggesting that fracture stress K_{Ic} is lowered with increasing time. When $t \rightarrow t_{\infty}$, $K_{\text{appl}} \rightarrow K_{\text{Isc}}$, as indicated by the beginning of the lower plateau in Figures 1(a) and 1(b). This lower limit indicates a saturation limit for the embrittlement mechanism, since K cannot be reduced any further than the K_{Isc} limit under static load. The slope “ m ” should depend on the governing embrittlement mechanism, such as APD, HAC, or LME, *etc.* What particular m -values should be for APD, HAC, or LME is not known, even though these variables are also stress dependent. Thus, it is not known how to attach a value to the slope “ m ” that corresponds to a particular mechanism.

Fig. 1—Applied stress intensity K vs time to initiate a crack in notched samples: (a) 7075-T651 alloy with liquid gallium; (b) 7075-T651 alloy in 3.5 pct NaCl solution.

The relationship of K_{app} with time can be rewritten, taking into consideration that the upper limiting value persists up to some finite t_0 value, as

$$\{K_{app}[t_{init}/t_0]^{-m}\} = K_{Ic} \text{ or } t_{init} = t_0 [K_{app}/K_{Ic}]^{1/m} \quad [3]$$

where $t_{init} = t_0$, and $K_{app} = K_{Ic}$; then, t_0 can either be zero or some finite value (like in Eq. [1]). In essence, linear segment is only between the lower and the upper plateaus.

Upper plateau is fracture toughness K_{Ic} , which is a material property in an inert environment, and it depends on the alloy microstructure, its yield stress, and slip mode. The lower plateau is K_{Isc} is the stress corrosion threshold that depends on the environment and material. The threshold K_{Isc} is proportional to the minimum force required to move a preexisting crack for a material/environment condition. The initiation time “ t_{init} ” is then the time required to build the necessary chemical force to help a crack move against the material resistance at an applied stress intensity $K < K_{Ic}$. This can be viewed as either the environment reduces the mechanical resistance by the amount represented by $K_{env} = K_{Ic} - K_{Isc}$ or helps the applied stress with the required amount to overcome the intrinsic material resistance. This representation involving the linear addition or subtraction of K is valid within the limits of linear elastic fracture mechanics. In essence, K_{env} , represents a mechanical equivalent of chemical force (or work) involved in the embrittlement process. In the following sections, we will consider this reduction in the crack-tip driving force due to environment for each of the specific system shown in Figures 1(a) and (b). Thus, the two most important results that need to be accounted are (a) crack initiation occurs even under remote compression, and (b) the slopes “ m ” in Figures 1(a) and (b) are the same under both tension

and compression. Under both modes of uniaxial load, we need at least a local tensile stress to initiate a crack.

V. BENDING-INDUCED TENSILE STRESS

At the notch tip, magnitude of stresses for tension and compression loads and their gradients are the same except for the sign. If deviatoric stress is considered, they are of the same magnitude with opposite signs, since the shear stresses under tension and compression are in the opposite directions. However, if some bending is involved, then it can induce tensile and shear stresses that increase with increasing distance away from neutral axis. Chu *et al.*^[9,10] used a WOL sample that is screw driven to apply tension and compression loads, in their experiments. This can induce bending stresses giving rise to some local tensile stresses. Such stresses are induced near the notch tip for the tensile load and at some distance away under compression load. An incipient crack can form in the tensile region (Figure 2(a)) and probably grow toward the notch until it is arrested at the notch tip because of the dominating compressive stresses at the notch tip (Figure 2(b)). These bending tensile stresses were estimated using FEM for the WOL sample geometry (10 mm thick) used by Chu *et al.*^[9,10] in 7075-Ga system. FEM analysis showed that at ~4 mm from the notch tip, the compression stress transitioned to tension.^[12] FEM analysis was conducted using tri-linear hardening model with data points: $E = 70$ GPa, $\nu = 0.3$, and stress/strain values at yield and ultimate stresses.

The maximum tensile stresses induced due to bending component for the 7075-Ga system were estimated using such a FEM model^[12] for both tension and compression samples, and the results are shown in Figure 3. The tensile stresses due to bending component occurred at the notch tip (shown as “X” in Figure 2(a)) for the

Fig. 2—FEM calculations for tension and compression loads at $K = 19 \text{ MPa(m)}^{1/2}$ showing the maximum tensile stress locations. The optical micrographs show that the FEM calculations are in agreement. (X) marks the location of transition to tensile stress due to bending.

tensile-loaded sample, whereas for compression, it occurred at a distance of $\sim 5 \text{ mm}$ away from the notch tip (shown as “X” in Figure 2(b)). These results are estimated for the case $K_{\text{app}} = 19 \text{ MPa(m)}^{1/2}$. The magnitude of the tensile stress due to bending decreases with the decreasing values of the applied K for both tension and compression loads. Figure 3 summarizes the magnitudes of the maximum tensile stresses that arise near the notch tip under remote tension and compression stresses due to bending for the various applied K . The results show that to obtain the same tensile stress higher applied K is needed under compression. The optical metallographic figures taken from Chu *et al.*^[9] show a crack at $\sim 5 \text{ mm}$ from the notch tip in compression, which is in fair agreement with the FEM results. FEM results are intentionally matched to the same magnification as the optical micrographs for clarity. As an example, for an applied $K = 30 \text{ MPa(m)}^{1/2}$ under compression, the corresponding tensile stress due to bending is about 350 MPa. To arrive at the same tensile stress of 350 MPa, K required in tension is about $13 \text{ MPa(m)}^{1/2}$. Thus, in essence, to reach the same $350 \text{ MPa(m)}^{1/2}$ ahead of the notch, K of $13 \text{ MPa(m)}^{1/2}$

Fig. 3—Maximum tensile stress variation with applied K for 7075-Ga system. For compression tests, the maximum tensile stress is due to bending. If the applied $K = 30 \text{ MPa(m)}^{1/2}$, the corresponding tensile stress is $\sim 350 \text{ MPa}$. To compensate this stress, the corresponding compression K is reduced to $13 \text{ MPa(m)}^{1/2}$.

Fig. 4—Comparison of compression data from Fig. 1(a) scaled to the estimated maximum tensile stress due to bending. Insert shows the IG fracture.

in tension or $30 \text{ MPa(m)}^{1/2}$ in compression is required. Based on this analysis, all the compression data can be scaled to the tension results as shown in Figure 4. The insert shows the IG fracture of a 7075-Ga alloy taken from Benson and Hoagland.^[11] Chu *et al.*^[9] also observed IG fracture under both tension and compression loads; however, their fractographs were not clear. Based on these results, LME in 7075-Ga system occurs because of the diffusion Ga atoms to the tensile regions ahead of the notch tip in compression and at the notch tip for tension. Compression tests thus provide clear discrimination to analyze the above process. The need to reproduce such results by careful experimentation is warranted.

Similarly, transition to tensile stresses due to bending loads was also estimated using FEM analysis^[12] for the case of 7075-NaCl system. It occurred at a distance of ~20 mm from the notch tip under compression, at the same $K_{app} = 19 \text{ MPa(m)}^{1/2}$. Hence, in 7075-NaCl samples, the bending stresses were negligible. Because of the differences in sample geometry, there is tensile stress with one order of magnitude lower in 7075-NaCl samples compared to that used for 7075-Ga. In spite of this, cracking occurred in compression, and the results in Figure 1(b) are not nearly parallel for tension and compression. Further study is required to ascertain these results and the reason for the disparities.

VI. ANALYSIS OF LIQUID METAL EMBRITTLEMENT: 7075-GA SYSTEM

For LME to have continuous crack growth to occur, the embrittling element has to be moving along the crack front, making contact with the base metal as the crack tip moves. If the cracking is ahead of the main crack, then diffusion of the element into the metal is required. If tensile stresses are required to open the crack tip, then liquid metal element has to diffuse to the region where tensile stresses exist. GB can be the preferential path for

diffusion, thus establishing a trajectory for crack growth. GB diffusion coefficient^[11,13,14] of Ga in an unstressed AL-bicrystal is $D(\text{Ga-GB}) = 4 \times 10^{-10} \text{ m}^2/\text{s}$, and is orders of magnitude higher than that for bulk with $D(\text{Ga-bulk}) = 2.4 \times 10^{-26} \text{ m}^2/\text{s}$. Therefore, Ga diffuses mainly along the GBs without dispersing into the matrix. The exponent, “ m ,” in Figure 1(a) relating the stress intensity to initiation time can be explained as follows, using this diffusion kinetics. For diffusion to occur, there should be a chemical potential gradient for diffusion. In the case of cracking, potential gradient is set up by having concentration gradient as well as stress gradient, if solubility is a function of stress. We can arrive at the stress gradient (qualitatively) using basic simple fracture mechanics relation between K and crack length “ a ” as

$$K_{app} = \sigma(\pi a)^{1/2} \quad [4]$$

where “ a ” is the crack length, and “ σ ” is the remote applied stress. The local tensile stress field at position distance “ x ” ahead of the crack tip is given by $\sigma(x) = K_{app}/(2\pi x)^{1/2}$. From the diffusion equation, we can write the diffusion distance “ X ” as a function of time “ t .” This is given as $X = (2Dt)^{1/2}$ or $X^2 = 2Dt$, where “ D ” is the diffusion coefficient, and “ t ” is the time for diffusion. Substituting for “ X ” for “ x ” we can write

$$\begin{aligned} \sigma(X) &= K_{app}/(2\pi[2Dt]^{1/2})^{1/2} \\ &= K_{app}/[(2\pi)^{1/2}(2D)^{1/4}t^{1/4}] = \alpha(t)^{-(1/4)} \end{aligned} \quad [5]$$

Since the diffusion distance is given by “ X ,” the diffusion time “ t ” is related to the square root of distance. In the linear approximation, the stress field ahead of a crack tip varies inversely with square root of distance. The implication is in terms of external applied load, or K_{app} is related to the diffusion kinetics that is proportional to $t^{-1/4}$. The exponent “ m ” relating K_{app} and initiation time “ t ” will then be close to 0.25, if the process is related to the diffusion of an embrittling element ahead of the crack tip. This conclusion is just based on stress-assisted diffusion. Assumption in Eq. [5] is that the diffusion distance is equated to the crack length.

On the other hand, if the diffusion distance is linear with time, $X = Dt$, then “ m ” will be close to 0.5, and if $X = (Dt)^{1/3}$, then $n = 0.17$.^[15] Thus, the value of “ m ” can vary depending on a given type of diffusion process and its boundary conditions. In this analysis, the stress field singularity with $\sqrt{a}$ is maintained.

Thus, the slope “ m ” can be related to Ga diffusion in 7075-GB using Eqs. [2] and [5]. Figure 1(a) shows the m -values for tension (0.36) and compression (0.34) which are higher than 0.25 predicted by Eq. [5]. These slopes are a bit higher than those observed by others.^[13,14] In Figure 5, the slope $m = 0.26$ observed by Watanabe *et al.*^[13] is close to that of the Eq. [5] in pure AL-Ga bicrystal (failure time) compared to a higher slope $m = 0.35$ in Figure 1(a) (or Figure 4) for the polycrystal (initiation time) 7075 alloy. In this analysis, we assume that the stress affects the diffusion distance but not the diffusion kinetics.

Watanabe *et al.*,^[13] using a 90 deg notched single-edge sample ($40 \times 30 \times 7$ mm), measured the failure time as a function of applied stress in a bicrystal (misorientation angle of 28 deg) AL-Ga system. Their results, shown in Figure 5, indicate the slope $m = 0.26$, complying with Eq. [5], suggesting that Ga diffusion is the rate-controlling process for LME failure in AL-Ga bicrystals. The crack initiation times was less than 10 seconds implying that decohesion occurs right at the crack tip for this initiation, but further propagation is assisted by diffusion of Ga along the open crack, and that Ga diffusion occurs primarily along the GB.

Ga diffusion is affected by the misorientation angle in both bicrystals and polycrystals. This is reflected in the variation in the slope “ m ” of applied stress *vs* failure time data from Watanabe.^[13,14] Figure 6(a) gives the slope “ m ” as a function of GB misorientation angle for AL-Ga bicrystals. This is due to Ga penetration rate varying with the misorientation angle of the GB.^[14]

Fig. 5—Fracture stress *vs* failure times in AL-bicrystals with liquid Ga. The misorientation angle was about 28 deg.

Slope “ m ” increases linearly to a peak value of about 0.5 at 38 deg and then decreases with further increase in the misorientation angle. The data at 55 deg had limited data points. The large variation of diffusion coefficient increasing with misorientation shown in Figure 6(b) supports the change in the “ m ” value observed.

The embrittlement in bicrystals is comparatively easy to understand, since the diffusion process appears to depend on GB orientation. In polycrystalline alloys, the embrittlement by Ga appears somewhat complicated. For example, $(t)^{-1/2}$ dependence was noted even in the absence of stress.^[13,16,17] Incubation times varied from few seconds to several hundred hours depending on the type of GB boundary encountered. In addition, no compression data are available for comparison with tension.

Chu *et al.*^[9] interpreted their data in Figure 1(a) that Ga adsorption induces local plastic deformation and a microcrack ahead of the crack formed when the plasticity reaches a critical value under the applied tensile load. Their interpretation implies that (a) plasticity is a prerequisite for crack growth by LME, and (b) plasticity overrides the direct de-cohesion or embrittling process. On the other hand, if micro-crack formation is ahead of the notch under compression, then Ga atoms need to diffuse through the GB. Then diffusion kinetics is involved. Their interpretation falls short in accounting whether plasticity is a helping or is the controlling factor in forming a crack ahead of the notch. The intergranular cracking and crack formation way ahead of the notch tip, particularly under compression, rules out the proposition that local enhanced plasticity contributing to embrittlement. The chemically enhanced plasticity can help in building up the needed tensile hydrostatic stresses ahead of the notch tip, but cannot be the rate determining factor. Authors do state that the Ga-surface-diffusion occurs and helps in forming a crack ahead of the notch. Under compression cracks can only form because of some tensile stress; but will arrest and not grow.

Fig. 6—Misorientation angle in AL-bicrystals affecting (a) the slope “ m ” in $\log \sigma_f$ – $\log t$ plots; (b) diffusion rates. Arrows indicate the two critical “ m ” values.

Fig. 7—Schematic illustration of the Ga migration to the tensile field on a GB region under tension and compression loads. Stress gradients for the two modes of loading are shown for comparison. Schematic indicates in compression the diffusion path distance is longer than in tension.

Under tensile load, the decohesion can be accomplished by stretching the bonds slightly in the presence of Ga. Crack nucleation ahead of the main crack is not required along with crack-tip plasticity. This process can continue as the crack moves forward. In this case, no diffusion is required to form a nano-size crack. For a small crack extension of this crack to be measurable, we need some diffusion distance. This may require some plasticity. Then the liquid metal should keep moving as the crack tip moves because of capillary forces. In the case of compression, this process cannot happen since the local bonds are not stretched at the notch tip. Therefore Ga has to migrate to the tensile place where there is a stretching of bonds; thus affecting the Ga diffusion under tension versus compression resulting in time differences to reach the tensile region to crack nucleation.

As shown earlier, it is the decrease of the tensile stress due to bending in compression that gives the time delay in Ga penetration (Figure 2), thus resulting in the same slopes for tension and compression (Figure 1(a)). The role of oxide, GB sliding (to break the oxide film prior to Ga penetration), and Ga reaction with GB precipitates are all minor effects to explain the overall trend observed in Figure 1(a). Anisotropy in diffusion rates due to GB misorientation and applied stress seem to be the two main contributions to the time difference between tension and compression and the slope “*m*.”

We can summarize LME due to Ga in 7075 alloy in the following manner: (1) stress does not affect Ga diffusion kinetics since the slope “*m*” is the same under tension and compression which implies that sign of stress is not affecting the diffusion kinetics; (2) stress is shifting the position of the tensile peak stress in compression (Figures 2 and 3), such that the maximum position shifts from the notch tip position due to plasticity arising from bending moments. It is not the diffusion kinetics per se affecting the delay times but the diffusion distance is shifted to where cracking can occur. This qualitative description is illustrated schematically in Figure 7.

One can query as to why diffusion is needed for embrittlement if Ga can embrittle right at the notch tip by debonding the atomic bonds. For this to occur, large amount of energy must be supplied to debond one atom at a time by diffusion. For this to happen, critical number of Ga atoms has to populate on a GB plane, which can allow a lower stress to debond. This minimum concentration of Ga (discussed in a later section) is required for the debonding process. Hence the role of stress is only in shifting the diffusion distance “*X*” without affecting diffusion kinetics. Tensile stress region is required for cracking under both tension and compression loads.

VII. EMBRITTLEMENT IN AQUEOUS SOLUTIONS: 7075-NaCl SYSTEM

Crack initiation results under the uniaxial tension and compression [$K_{app} - t_{init}$] for 7075-T651 in 3.5 pct NaCl solution at pH = 3.5 is given in Figure 1(b). The result indicates that the slope of the two curves is similar on a log-log scale with about 10 pct difference in magnitude; indicating the kinetics of embrittlement is similar under both tension and compression. The minimum value of K_{Isc} for tension is about $8.3 \text{ MPa(m)}^{1/2}$ and that for compression is $27.6 \text{ MPa(m)}^{1/2}$. Tension K_{Isc} is comparable to the data obtained by others in the literature^[18] for the same NaCl concentration.

Chu *et al.*^[10] observed metallographically that under compression a micro crack nucleated slightly ahead of the notch at about 0.2 mm, as in Figure 8(a). The fracture surface in tension show predominantly intergranular (Figure 8(b)) while under compression fracture show striation-like like features (Figure 8(c)) which was described as mix-mode. The authors interpreted their data in terms of HAC under tensile load and slip dissolution under compression. Matching fracture surface was not done to show that Figure 8(c) is a dissolution process. On the other hand, a microcrack was observed under compression, at about 0.2 mm

Fig. 8—(a) Cracking under compression load showing a microcrack formed ahead of the notch; (b) IG fracture surface under tension; and (c) mix mode fracture under compression.

ahead of the notch; when the sample was intentionally unloaded to induce a tensile residual stress. Both stress modes show some IG features. If author's interpretation is correct, then slopes under tension and compression are expected to be vastly different. Even though the slope under tension is about 10 pct lower than that under compression; the difference may be due to lack of sufficient data resulting in scatter. If H is causing the embrittlement, then its diffusion to a tensile region ahead of the crack tip can be involved. This difference is discussed in the later section.

7075-NaCl system involves chemical reaction at the notch in contact with the NaCl solution. This results in anodic dissolution that produces H at the notch to diffuse into AL. Assuming NaCl is a catalyst, AL surface will react with H₂O to produce nascent H following the reaction:

In the 7075-NaCl system, the aqueous NaCl solution reacts with the notch tip region of the 7075 alloy to produce nascent hydrogen which can then diffuse to the region of tensile stress at the notch tip or at the hydrostatic location at some distance ahead of the notch tip to nucleate a microcrack.^[4] Anodic dissolution in 7075 alloy is a precursor to promote the necessary reaction to create the high fugacity at the notch to produce hydrogen. Such type of mechanism can result in a time dependence on applied $K [K_{\text{app}} - t_{\text{init}}]$ plots.

The differences in the tension and compression results in 7075-NaCl system (Figure 1(b)) can be due to

- (a) anodic dissolution affected by applied stress;
- (b) If H is involved in the embrittling process, H generation and its diffusion distances are affected by the applied stress,
 - Stress can affect the H solubility and its diffusion kinetics.
 - Stress and chemical gradients affecting the SCC behavior in Figure 1(b).

VIII. ANODIC DISSOLUTION

There is only a limited quantitative experimental result^[19] to support that metal dissolution is stress dependent in the elastic region. Figure 9 provides experimental data on un-notched flat cantilevered bent-samples of 2024-T3 and pure Ni showing that the dissolution rate is dependent strongly under tension load and weakly under compression. Dissolution rate, $\text{DR} = \alpha \sigma_{\text{app}}$ linearly, where α is the slope that gives the dissolution rate per unit stress. DR is estimated by normalizing the dissolution depth with time. DR under compression is about ten times smaller than in tension. However, the reciprocal ratio of the slope α is proportional to time. In stress vs α^{-1} plot (not shown), compression slope is about 11 pct higher than under

Fig. 9—Dissolution rate variation with applied stress is shown for 2024-T351 and pure Ni, with similar behavior. Inserts show the preferential dissolution in the two materials, pitting in 2024 and GB grooving in Ni using AFM.

that in tension. This is comparable to that in Figure 1(b) where the compression–tension difference in the slope is about 10 pct. Inserts in Figure 9 show the type of dissolution on the polished surface exposed to aqueous solution of each material under tensile load. The authors point out that the electrochemical reactivity on the cantilevered bent sample, is higher on the tensile side compared with the compression side. The implication of the figure is that preferential dissolution is very sluggish under compression. Translating this observation to a notch sample suggests that material away from the notch tip can dissolve faster (and on the flat side surface of the sample) than close to the notch that can cause some blunting effect. While comparing the two materials in Figure 9, 2024 alloy shows pitting in the matrix, whereas Ni showed GB dissolution; but the overall DR trends in these two systems are similar. The article mentions (without showing data) that the dissolution rate increases drastically beyond the yield stress due to exposure of fresh surface metal to the environment. Their dissolution results are summarized schematically in Figure 10. The data only show that the dissolution rate is decreasing slightly under compression compared to that under zero stress. Several other studies^[20,21] on localized aqueous corrosion indicate that tensile stresses accentuates corrosion rate, but the results are inconclusive.

Since K vs time plots in Figure 1(b) have 10 pct difference in the slopes under tension and compression, dissolution can be in part the controlling mechanism under both modes of stress. Such stress-assisted dissolution study in 7075 alloy is not available. However, in AL alloys, dissolution is a necessary step to promote nascent H at the notch. If the dissolution rates are slower in compression than in tension, then the notch-

Fig. 10—Schematic illustration of Fig. 9 showing regions of dissolution with the mode of stress. Tension slope is higher than in compression.

tip H concentrations can be different. This may give rise to the observed 10 pct differences in the slopes as shown in Figure 1(b).

IX. H-SOLUBILITY AND DIFFUSION

There are few studies on alloys with high H solubility that show solubility is affected by the applied stress. One such experimental study is by Wriedt and Oriani^[22] in 75Pd-25Ag system. Their results show that in hydrogen-saturated environment [10^2 torr H_2 pressure at 348 K (75 °C)], the solubility of hydrogen is lower with stress in uniaxial compression stress region compared with tension, Figure 11. Overall decrease in solubility is

Fig. 11—Hydrogen solubility in Pd-Ag alloy is affected by stress. In tension, the solubility is increased, while in compression, it is decreased. Overall solubility change is about a 1 pct.

about 1 pct due to tension and compression stresses even though Pd-Ag alloys have large H solubility. Similar H solubility results are also observed in Ti-based alloys.^[23] However, H solubility in the lattice of 7075 alloy (and diffusivity) is considerably lower than in Pd, Steels, and Ti, to account for the large stress effect on solubility. Thus, H solubility cannot account for the large slope differences in Figure 1(b). These observations are more related to the solubility in matrix than that in the GB.

One suggestion is that H diffusion can be affected by an applied stress and its mode. However, there are no data in 7075 alloy suggesting H diffusion is affected by an applied stress. There are a few experimental results in other materials. In pure Ni, Grimes^[24] observed that H-diffusion rate and the activation energy were not affected by an applied tensile stress, but concluded that H solubility must be affected by stress. The positive stress gradient in tension and a negative gradient in compression, in relation to a stress-free state, can enhance or inhibit the H-diffusion rates. Diffusion affected by stress mode allows for entry of H into the lattice or GB. While higher/lower diffusion rates can be related to smaller/longer incubation times, it does not account for the cracking under compression by diffusion. Since the H diffusivity is very low in 7075, H may not go farther into the hydrostatic stress field ahead of the notch, but can enter short distance at the notch tip, under tension and compression loads. One can calculate the distance that H can diffuse into 7075 lattice for time $t = 1$ second by considering $D = 4 \times 10^{-12} \text{ m}^2/\text{s}$,^[25,26] with $X = (2Dt)^{1/2} = (2 \times 4 \times 10^{-12} \times 1)^{1/2} = 2.8 \times 10^{-6} \text{ m} \sim 3 \text{ } \mu\text{m}$, practically at the notch root. The GB diffusion rates (under tensile and compression load) of H in the pure AL are calculated using MD to be a difference of two orders^[27,28] of magnitude higher than that in the lattice. This would give $\sim 30\text{-}\mu\text{m}$ distance that H has diffused into the GB of AL which is substantial. These variations are affected by the type of GB (twist and tilt) for which the calculations are made. There is no experimental result for GB diffusion to validate MD

calculations for 7075 alloy. These observations lead us to say that stress does not affect H solubility to account for the trends in Figure 1(b). GB-diffusion rates (MD calculations) can be affected by the applied stress which can contribute to the time delay in compression load. However, mechanism for slope “ m ” is unclear.

X. STRESS VS CHEMICAL GRADIENTS

Equilibrium concentration of a chemical solute depends on several factors. One such factor may be the chemical affinity in terms of A–A, B–B bonds vs A–B bonds affecting the solubility of a solute in the lattice or GB for an Ideal solution. Other contributing factors depend on the atomic size factor, lattice site, electronic factors, *etc.* In general, the presence of a foreign atom distorts the parent lattice contributing to local strains and stresses, which in turn can affect the solubility.

Diffusion kinetics depends on the potential gradient of a solute. In the absence of a stress gradient, the diffusion kinetics is governed by the chemical potential gradient. Such diffusive currents normally cause the exponent “ m ” to be around 0.5. If stress gradients also affect the diffusion kinetics, then they can be in or out of phase with the chemical potential gradients. This can affect the kinetics. Consider now a stress-free crack or a notch. Then, even with a maximum chemical concentration at the notch/crack tip, crack cannot initiate or grow. When an external stress is applied, even though remote applied stress is uniform, a stress gradient is set up because of the presence of a notch and a crack can initiate and grow, particularly in the presence of an aggressive environment.

For an elastic case, the peak stress $\sigma_{yy, \text{max}}$ occurs at the notch/crack tip. The notch/crack-tip strain fields accommodate the presence of foreign chemical atoms, and as a result, the equilibrium concentration due to chemical and mechanical factors can be higher than when there is no stress. In the elastic case, the stress is maximum at the notch/crack tip itself and decreases with distance farther away from the tip. The chemical potential gradients cause the solute atoms (like H-atoms) to move into the material, while the stress gradient tries to concentrate it at the notch/crack tip itself. In addition, since the stress is also involved in opening the notch/crack, crack initiation most likely occurs at the notch tip. Since there is no perfect elastic notch, diffusion distances are slightly ahead of the notch tip under tensile load. The resulting slope “ m ” (in K vs t plot) is around 0.25. In essence, in the elastic case, larger notch stress gradient can reduce the incubation time. This is because for large K_{app} , one needs less help from chemistry to initiate a crack. The required concentration is reduced at high K_{app} to cause failure to occur at shorter time. At lower K_{app} , one needs help from chemistry to have higher concentration to afford longer initiation time. This is true only if the source for the solute and the peak stress are located at the same place, namely at the notch/crack tip with an external aggressive chemical solute source. In the case of uniformly distributed internal solute (like internal H), stress

gradients at the notch/crack tip can drive H toward the tip, while the chemical potential gradient will be an uphill diffusion. These points are summarized schematically for both elastic and elastoplastic cracks in Figure 12. We have added the reference solute concentration gradient as a down-hill gradient as an external source. Thus, stress gradients can have an effect on the diffusion kinetics. The stress has direct effect on the crack nucleation from the point of strain-energy release processes. Environment then affects the nucleation process via the surface energy term.

In Eq. [5], under 7075-Ga section, we briefly discussed the stresses ahead of a sharp crack. In the case of notches or blunted cracks, the stress fields are altered due to blunting. For the simple elastic case of a blunt crack, one can use Creager–Paris solution^[29] (Figure 13)

Fig. 12—Schematic illustration of stress (elastic and plastic) gradients under tension/compression can affect the diffusion of solute into the tensile region, for a given solute concentration gradient.

Fig. 13—Hypothetical linear-elastic stress field near a sharp notch using Creager–Paris solution.

to find a relation between the stress gradient and notch radius “ ρ ” at distance $x = \rho$ ahead of the notch:

Stress components

$$\begin{aligned}\sigma_{xx}(x) &= \frac{K}{\sqrt{2\pi x}} \left(-\frac{\rho}{2x} + 1 \right) \\ \sigma_{yy}(x) &= \frac{K}{\sqrt{2\pi x}} \left(\frac{\rho}{2x} + 1 \right) \\ \tau_{xy}(x) &= 0\end{aligned}\quad [6]$$

Stress gradient

$$\frac{d\sigma_{yy}(x)}{dx} = -\frac{K}{2\sqrt{2\pi} \cdot x^{\frac{3}{2}}} \left(\frac{3}{2} \cdot \frac{\rho}{x} + 1 \right) \quad [7]$$

which is valid for $x \geq \frac{\rho}{2}$

The above Eq. [7] reduces to $[d\sigma_{yy}(x = \rho)/dx] = K/\rho^{1.5}$; the stress gradient is a direct function of K and an inverse function of notch radius ρ . Figure 1 shows that $K\alpha(1/t_{inc})^m$. Thus, the stress gradient decreases as ρ increases, and as ρ increases, the slope “ m ” (in Figure 1(b)) can decrease to a plateau around ~ 0.5 . ρ is related to $\sigma_{yy,max}$ and the gradient $(d\sigma_{yy}/dx)$. For a crack to nucleate and grow, a minimum required value in $\sigma_{yy,max}$ and $(d\sigma_{yy}/dx)$ has to be satisfied. The following discussion on stress and chemical effects is with respect to internal chemical source (like internal H) and external source (like external H).

For a uniform concentration of internal H, chemistry is not helping to segregate it toward the notch/crack tip. This is an uphill-driven process from the point of chemistry, since H has to segregate from a dispersed state. Hence, sustained load crack growth is stress driven, which therefore gives a higher thresholds. There is not only an elastic interaction (stress field of the crack with the defect), but also an additional interaction from the modulus due to elastic inhomogeneity. For the chemical solute source that is in the bulk of the alloy (as in internal H in Fe lattice), we need to separate the mechanical contribution from that of the chemical. The same stress gradient can still drive the solute toward the notch/crack tip. For a uniformly dispersed H at the notch/crack tip, the chemical diffusion kinetics can lead to $m > 0.5$. Thus, K vs t relationship will be different for external and internal H.

In the case of plasticity at the notch/crack with external H source, the location of maximum stress shifts away from the notch/crack tip while keeping the chemical concentration source at the tip. The stress gradient is now positive near the notch/crack tip with stress increasing to the peak stress, and decreasing away from the tip. Now, both chemical and mechanical gradients help in the diffusion process. However, beyond the peak stress, the stress gradient is similar to the elastic case, and therefore the stress gradient opposes further diffusion even though the chemical gradient may still be the driving factor. The latter factor helps H atoms to accumulate at the peak stress region. The same principles can be applied to the bulk H (internal source), where the strain energy drives H to move toward the peak stress and not away. In this case, the slope “ m ” can

be greater than 0.5 since both stress and chemistry effects are not helping. The above description of the coupled effect of stress and chemistry is schematically shown in Figure 12 for elastic and plastic cases along with solute concentration gradient. Stress gradient varies with the notch radius (or K_I).

Notch stress gradient (calculated for elastic case = $K/\rho^{1.5}$) varying with notch radius is compared for 4340 steel in vacuum ($\sim 10^{-8}$ torr)^[30] and in H_2SO_4 solution^[31,32] on log-log scale in Figure 14. The fracture toughness values for various notch radii in vacuum is measured to observe a K value corresponding to crack initiation, and this gives $K = K_q$ for each ρ . In H_2SO_4 solution^[31,32], applied $K = K_{Isc}$ is defined from the applied K vs incubation time at the lowest K value where the time is very long, for each ρ . When notch radius is very sharp like a crack, $K_q = K_{Ic}$ and the lowest

Fig. 14—Stress gradient variation on notch radius for 4340 Steel tested in vacuum and H_2SO_4 solution.

$K_{app} = K_{Isc}$. The mechanical stress gradient decreases linearly with ρ for both vacuum and H_2SO_4 in 4340 steel. The slopes (~ 1.3) are same for both environments. However, the H_2SO_4 environment is lower compared with vacuum. For the same notch radius ρ , the stress gradient in vacuum is higher than for H_2SO_4 , requiring higher gradient to nucleate the crack in vacuum. Here, we assume that the crack will initiate at a distance $x = \rho$. The H_2SO_4 results for 4340 steel^[32] in Figure 15(a) show the slope “ m ” decreasing with ρ , approaching a plateau around 0.5. The mechanical stress gradient decreases as the notch radius ρ is increased or as “ m ” increases, Figure 15(b), indicating that the chemical solute diffusion is slowed down or diffused away from the notch. Thus, in the limit of constant stress gradient for both tension and compression loading, the slope should be 0.5. Figure 15(b) is a re-plot of Figure 15(a), showing a complex relation of “ m ” varying nonlinearly with $(K_{Isc}/\rho^{1.5})$. These stress gradients are also affected by the material work hardening rate “ n .”

In the elastoplastic case, as stress increases with distance from the notch tip to the peak, the stress gradient is reversed to give $m > 0.5$. If the stress gradient is a second power dependence on K as in terms of strain energy density then you can have slope of 1.0 or 1.14 for diffusion. For 7075-T6 alloy, the slope of the HRR field, $(d\sigma_{yy}/dx) = -[(1/n) + 1] = -0.899$. Square root of this is ~ 0.95 and is proportional to the slope “ m ,” observed in Figure 1(b). This suggests that the work hardening characteristics of a material also plays a role in the embrittlement process like HELP. Work hardening is related to the generation of internal stresses in the notch plastic zone, which can affect the reaction rate of the environment with the notch. In the case of NaCl, this can result in changing the initial H concentration between tension and compression loads. Stress gradient and work hardening may be interrelated to affect the slope “ m .”

Fig. 15—(a) The slope “ m ” from applied K vs incubation time plots are shown to depend on (a) the notch radius, and (b) the stress gradient. “ m ” decreases with notch radius while it increases with the stress gradient.

Fig. 16—Single overload and compression under load yields longer and shorter Incubation times in 7075-T651 alloy in 1 pct NaCl solution.

XI. 7075-NACL VS 7075-GA

Notch $K_{Ic} = 43 \text{ MPa(m)}^{1/2}$ is the same for both 7075-Ga and 7075-NaCl systems. Comparing tension data in Figures 1(a) and (b) show that K_{Isc} is $8.3 \text{ MPa(m)}^{1/2}$ for NaCl vs 1.7 for Ga. These values show that there is a large difference in the embrittling efficiency of H from NaCl vs Ga. Implication is that there is a larger reduction in surface energy γ with Ga than with H. The reduction in γ can be related to (1) the number of Ga or H atoms occupying the GB sites, (2) reduction in surface energy γ per atom, and (3) reaction kinetics with AL. Ga has no reaction with 7075, while NaCl solution reacts strongly to produce nascent H. Reaction kinetics can then change the H concentration at the notch, not affecting the diffusion rates.

One can calculate the number of Ga or H atoms that can occupy a given type of GB (for a $\Sigma 5(012)$) boundary by using the following data from Yamaguchi's first principle calculations^[33]: cohesive energy is related to surface and GB energy through the equation $\gamma_{CE} = 2\gamma_{SURF} - \gamma_{GB}$. For AL, the GB cohesive energy $\gamma_{AL} = 1.45 \text{ J/m}^2$, and the area of $\Sigma 5$ boundary = 37.3 AU^2 ; strain energy released by 1 Ga atom at GB is given by $\Delta E_{Ga} = 290 \text{ meV}$; that for H is $\Delta E_H = 560 \text{ meV}$; E (7075) = 70 GPa ; and $\nu = 0.33$. Here, pure AL data are assumed to be the same as those for 7075 alloy. ΔE_H is not much different for internal or external H^[33]. Using the equation for plane stress, we have $K_{Isc} = (\gamma_{CE} E')^{1/2}$, where $E' = E$ for plane stress for 7075-NaCl system, and $E' = E/(1 - \nu^2)$ plane strain for 7075-Ga system. The ratio of these two relations gives $\{\gamma_{Ga}/\gamma_H\} = 0.042$ where γ 's are cohesive energies.

Cohesive energies are related as $\gamma_{Ga} \propto (\gamma_{AL} - \Delta E_{Ga} N_{Ga})$ and $\gamma_H \propto (\gamma_{AL} - \Delta E_H N_H)$, where N is the number of atoms/unit area. Within the linear elastic approximations, we can re-write the above expressions as $\gamma_{Ga} = (\gamma_{AL} - \Delta E_{Ga} N_{Ga})$ and $\gamma_H = (\gamma_{AL} - \Delta E_H N_H)$. Substituting all the corresponding values listed above, we can estimate the number of atoms saturating

the $\Sigma 5$ GB for Ga ~ 12 atoms/unit cell and for H ~ 10 atoms/unit cell. The embrittling efficiency is then the reduction in cohesive energy or reduction in strain energy released per atom. For Ga, this is $\Delta E_{Ga} = -0.29 \text{ eV/atom}$ and that for H, it is $\Delta E_H = -0.56 \text{ eV/atom}$.

First principle calculations^[34,35] show that there are three layers of Ga penetrating into the GB. Two of the layers are bound with AL atoms on two sides of the GB plane, while the third mid-layer moves freely in between. This means that Ga is basically a liquid layer in the GB plane making it to be an efficient embrittler. *In situ* experiments on AL-Ga by Pereiro-Lopez *et al.*^[36] show that very small $K_I = 0.045 \text{ MPa(m)}^{1/2}$ can extend a Ga-penetrated GB crack over $500 \mu\text{m}$ in 147 seconds. They suggest that GB separation in AL-Ga alloy system can occur at very weak stress levels. In the case of H which is a proton, it is also free to move in the GB plane and may not be efficiently bound to AL atoms due to traps sites. These observations are also reflected in the $\Delta E_H > \Delta E_{Ga}$.

XII. CONCLUSIONS

Chu *et al.*^[6-9] observed that in compression crack initiation is delayed with respect to tension load by about an order of magnitude. Owing to lack of any confirmatory or contradictory data, we assume that Chu *et al.*'s results are correct. Jiang *et al.*^[37] observed the same trend of Chu's results in his 7075-T651 alloy (in 1 pct NaCl solution) by applying single overloads in tension and in compression at a constant $K = 10 \text{ MPa(m)}^{1/2}$, to measure incubation times. They observed the incubation times to be shorter by a factor of 100 under compression underload (due to tensile residual stress) vs tension (due to compressive residual stress) overload, at the same applied load of $\pm 100 \text{ MPa}$ (Figure 16). This validates indirectly Chu *et al.*'s observations as seen in Figure 1(b). Analysis suggests that tensile stress is necessary for a crack to initiate under both tension and compression loads. Hence, K_{Isc} is valid for only for the cases tension loads, similar to K_{Ic} . In Figures 1(a) and (b), K_{Isc} under compression, marked by the authors,^[9,10] is only an apparent value.

In the current article, we are using SCC under uniaxial tension vs compression loads to discriminate the governing mechanisms involved. We make the following conclusions:

1. In 7075-Ga system, GB fracture occurs via decohesion of AL atoms and is mainly due to Ga penetration into the GBs at the tensile stress region, and the rates are reduced in compression load, because of lack of tensile stress. GB diffusion of Ga depends on the misorientation angle. The sample geometry used for this system gives a transition in tensile bending stress at about 4 mm away from the notch. If the sample geometry was designed with no bending stress, then there should be no difference in Figure 1(a) between tension and compression.
2. The behavior of 7075-NaCl system can be ascribed to a mixed APD and HAC with H-diffusing into a tensile location ahead of the notch to give a GB fracture.

The exact mechanisms are not very clear. One can *speculate* that anodic dissolution and H solubility do not *directly* affect the results in Figure 1(b). Assuming HAC mechanism, the role of stress and stress gradients can be affecting H diffusion to the tensile stress region. In addition, strain hardening coefficient also affects the slope “*m*.” In this system, the stress affects the diffusion distance, while the diffusion kinetics affects the incubation time.

3. The K_{Isc} in 7075-Ga is very low (by a factor of 5) compared to that in 7075-NaCl system. This suggests that embrittling efficiency of Ga in 7075 is more potent compared with H. Based on first principle calculation, the differences lie in the higher occupancy number of Ga atoms compared with H on the GB, and their differences in cohesive energy.
4. Based on the above observations and analysis, we conclude that Ga transport is not stress dependent, but NaCl that produces H via chemical reaction is. And our theoretical analysis highlights two issues in terms of maximum stress and stress gradients that interact with the diffusion of the chemical species. While the computational techniques of getting at the stresses and its gradients are available, the systematic experimental data are missing.
5. The uniaxial tension and compression incubation time experiments at constant applied stress should be repeated in many alloy-environmental systems that are related to predominantly APD or HAC mechanisms. This should be supported by clear crack- or notch-tip metallographic observation of crack location at a tensile region along with mating fracture surfaces. Such experiments can offer better understanding about the failure mechanisms. We observe that the incubation time experiments at constant stress are key to the understanding the operating SCC mechanisms.

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