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EARLY STAGES IN THE DEVELOPMENT OF STRESS CORROSION CRACKS

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Early Stages in the Development of Stress Corrosion Cracks

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ABSTRACT

Processes involved in the growth of short cracks and stage I of long stress corrosion cracks have been identified and evaluated. There is evidence that electrochemical effects can cause short stress corrosion cracks to grow at rates which are either faster or slower than long cracks. Short cracks can grow at faster rates than long cracks for a salt film dissolution growth mechanism or from reduced oxygen inhibition of hydrolytic acidification. An increasing crack growth rate with increasing crack length could result from a process of increasing crack tip concentration of a critical anion, such as Cl^- , with increasing crack length in a system where the crack velocity is dependent on the Cl^- or some other anion concentration. An increasing potential drop between crack tip and mouth would result in an increased anion concentration at the crack tip and hence an increasing crack velocity. Stage I behavior of long cracks is another early development stage in the life of a stress corrosion crack which is poorly understood. This stage can be described by $da/dt = AK^m$ where da/dt is crack velocity, A is a constant, K is stress intensity and m ranges from 2 to 24 for a variety of materials and environments. Only the salt film dissolution model was found to quantitatively describe this stage; however, the model was only tested on one material and its general applicability is unknown.

1. Introduction

Environment-induced cracks generally progress from initiation to component failure through a sequence of stages such as 1) initiation, 2) short crack growth, 3) stage I growth, 4) stage II growth, 5) crack coalescence and 6) failure. While our level of understanding of each of these stages is

variable, perhaps most is known about stage II growth and failure processes. Little is known about the fundamental processes controlling stages 1,2,3, or 5.

Crack initiation and short crack growth stages are often inseparable because it is difficult to measure the behavior of cracks less than $10\text{ }\mu\text{m}$ in length and because the transition between initiation and growth may be continuous and not distinct. Therefore, information on the behavior of short cracks is more available than details about initiation. Short crack growth behavior may differ from long crack growth behavior because of chemical gradients in the crack and because cracks are short relative to microstructural features plastic zone size and stress intensity (K) fields. There is evidence that short stress corrosion cracks (SCC) can either 1) crack slower than long cracks (1,2) or 2) crack faster than long cracks (3) or 3) crack at no applied K (4). These three cases have been demonstrated for 1) intergranular stress corrosion cracking (IGSCC) of Type 304 SS (1), 2) aqueous hydrogen-induced crack growth of high-strength steels (3) or 3) IGSCC in Ni (4).

Stage I SCC usually exhibits a very strong K dependence with Paris Law exponents (m) of up to 24. Existing SCC models do not adequately describe the stage I regime, but a recently developed salt film dissolution model was found to give an excellent correlation with experimentally determined stage I behavior for Ni. Model calculations indicate that the crack velocity in this regime is controlled by electromigration through a resistive salt film and that the K dependence results from crack opening-controlled salt film dissolution. In other materials there is some suggestion that the value of m is a function of the local crack tip chemistry and crack opening displacement or crack angle.

2. Growth of Short Stress Corrosion Cracks

Observations that short and long cracks behave differently were first made for fatigue crack growth (5-12). These cracks were characterized as mechanically small comparable to the extent of crack tip plasticity, microstructurally small relative to the scale of the microstructure, physically small (typically less than 1 mm) and chemically small. A similar size characterization can be

made for short SCC's; however, the length of chemically short SCC and CF cracks may differ because of the added effect of convection from mechanical unloading on the crack tip chemistry during cyclic loading. Short fatigue cracks were found to propagate at rates considerably higher than long cracks for what was thought to be similar crack driving forces described by K . Ritchie and Lankford (12) recommended the use of an equivalent elastic-plastic field parameter to normalize the driving force between short and long fatigue cracks. Essentially the addition of a ΔK_{eff} to the mechanically short cracks was sufficient to produce a flaw-size-independent crack driving force. While the use of a ΔK_{eff} solved the problem for mechanically short crack behavior, the effect of chemically short crack behavior remained.

2.1 Experimental Evidence of Short Stress Corrosion Crack Behavior

A schematic showing the possible da/dt - K response of short SCC's is shown in Figure 1 based on observations for short fatigue cracks (5-12) and conceptualization of short crack (SCC) behavior (SCB). The conceptual short crack curves in this schematic, Figure 1, presume a constant load test such that both the crack length and K increase simultaneously. Some aspects of the short crack curves in Figure 1 could perhaps be rationalized by mechanical crack length effects and adjusted by use of a K_{eff} but there is also considerable evidence that chemical and electrochemical processes induce a distinct crack length effect superimposed on the mechanical length effect. It is also conceivable that there is a K independent regime for SCB-1 at very short crack lengths such that the short crack curve exhibits K independence up to some crack length.

Intergranular corrosion (IGC) behavior can be used to indicate the stress independent aspects of electrochemical crack length effects. Examples where the IGC rate decreased with increasing length were reported by Jones et al. (4) for Ni and by Lee et al. (13) for Fe. Jones et al. observed that IGC, of Ni with S segregated to the grain boundaries, was arrested at about 0.1 mm when tested in 1N H_2SO_4 at +900 mV. This arrest was explained by the retention of S on the sidewalls of the IGC penetration inducing active corrosion of the sidewalls and saturation of the crevice. The saturation led to arrest of the IGC. Lee et al. (13) reported a stage of increasing

penetration rate followed by a retardation of the penetration rate for IGC in Fe with segregated P tested in 55% $\text{Ca}(\text{NO}_3)_2$ at 60°C. The time required for these stages was reduced by P segregation to grain boundaries. Lee et al. explained the retardation as a result of the repassivation rate at the crack tip increasing faster than the dissolution rate. A critical penetration depth of about 1 mm was noted for the transition from increasing to decreasing penetration rate which was independent of grain boundary chemistry; however, an electrochemical explanation for the critical depth was not given. Retardation of the IGC penetration does not always occur as noted by Jones et al (4) for Ni with P segregated to the grain boundaries and tested in 1N H_2SO_4 at +1200mV. In this case, IGC continued at an average rate of 7×10^{-6} mm/s without evidence of retardation.

The four steps in crack initiation of a pipeline steel proposed by Parkins (14) are presented as a function of crack velocity versus time in Figure 2. Presumably, crack length is increasing with time such that stage 4 consists of conventional long crack-subcritical crack growth behavior. The reduction in crack velocity in stage 2 was suggested by Parkins as being due to decreasing crack tip strain rate resulting from multiple crack initiation. Parkins schematic is quite similar to SCB-1 shown in Figure 1 where Parkins stage 3 is an extended minimum followed by long crack behavior. While Parkins explains stage 2 by multiple crack initiation, it is also quite possible that individual cracks will exhibit a decreasing velocity with increasing crack length because of electrochemical effects.

Eickemeyer (1) tested the effects of the initial crack depth in a low alloy steel on the da/dt vs K response in solutions of saturated CaSO_4 at pH 7 and $\text{Ca}(\text{OH})_2 + \text{CaSO}_4$ at pH 12. Eickemeyer found that the stage II crack velocity increased with increasing initial crack length for samples tested in the pH 12 solution, but no crack length dependence was found for the pH 7 solution. These results follow the SCB-2 type of behavior shown in Figure 1. There was no mention as to whether the cracks were intergranular or transgranular. Only the longest cracks in the pH 12 solution exhibited a crack velocity equal to those observed in the pH 7 solution. Eickemeyer explains the crack length dependence as resulting from an increasing anion concentration at the crack tip with increasing crack length. It was suggested that an increased anion

concentration results from the increased potential drop down the crack with increasing crack length. A greater degree of depassivation with increasing anion concentration is proposed as the mechanism for the increased crack velocity.

Andresen (2) also found an increasing crack velocity with increasing crack length for a sensitized Type 304 SS tested in deaerated water at 288°C, as shown in Figure 3. Andresen's tests were performed with cyclic loads at an R of 0.6 and a frequency of 0.01 Hz. He found that the short crack regime extended to about 40 μm after which long crack growth velocities were observed. Andresen's results, like Eickmeyers, follow the SCB-2 type of behavior shown in Figure 1. These results were obtained for cracks which initiated from a machined notch of a cyclically loaded compact tension sample while those of Eickmeyer (1) were obtained for surface generated cracks in a statically loaded bend sample. Clearly, the applied stress intensity in the tests used by Eickmeyer and Andresen are substantially different so direct comparison is difficult.

Accelerated short crack velocities were observed by Gangloff (3) for fatigue crack growth behavior of 4130 steel in 3% NaCl at 24°C, as shown in Figure 4. Crack velocities for 0.1 to 1.0 mm deep elliptical surface cracks were about a factor of 10 faster than 30 to 44 mm long cracks in compact tension samples. The observed acceleration is not a mechanical crack length effect as the crack velocities were independent of crack length, from 0.1 to 40 mm, for vacuum or moist air. Therefore, the faster crack velocities shown in Figure 4 are primarily an electrochemical crack length effect. This behavior is opposite of that observed by both Eickmeyer and Andresen in that the crack velocity decreased with increasing crack length. These results are consistent with the SCB-3 type of behavior shown in Figure 1. Gangloff explained the accelerated crack velocities as resulting from enhanced hydrogen production in the crack resulting from increased hydrolytic acidification and reduced oxygen inhibition within the crack for short relative to long fatigue cracks. Because crack advance is caused by H uptake, the pH within the crack is a critical parameter in determining the crack velocity.

Experimental evidence suggests that there are at least two and perhaps more processes controlling crack length effects in SCC. Processes which could explain the three crack length behaviors depicted in Figure 1 are : 1) salt film dissolution controlled cracking, 2) critical crack tip concentration processes and 3) crack tip hydrogen concentration controlled by anion transport down the crack. The distinction between these curves is that SCB-1 results from corrosion processes that occur at high rates for short cracks and low K 's but that these processes are impeded as the crack increases in length in spite of the increasing K . Once long crack behavior is established the crack growth behavior matches the long crack behavior curve. The corrosion processes that produce SCB-2 require some critical depth to create a critical crack tip concentration to develop; therefore, it is expected that these cracks will require a higher mechanical driving force K , than long cracks so that their crack growth rates will be slower than long crack growth rates until steady state is achieved. The distinction for SCB-3 is that the environment in short cracks is electrochemically more aggressive long cracks as observed by Gangloff (3).

2.2. Predictions of Short Crack Behavior Based on the Salt Film Dissolution Model

Short crack propagation as suggested by SCB-1 can be limited by the lack of crack opening necessary to allow transport of corrosion products from the crack tip to the bulk solution. For tight cracks in anodically polarized nickel, the concentration of species from wall and tip corrosion processes can be sufficient to exceed the solubility limit of corrosion products and salt precipitation at the crack tip is expected. The presence of the precipitate at the crack tip subsequently inhibits crack tip dissolution and hence crack growth. An analysis of stage I crack growth kinetics for anodically polarized nickel in 1N H_2SO_4 indicated that precipitation of $NiSO_4$ was controlling crack growth. The assumed resistance of the salt film caused a reduction in the estimated anodic polarization and a reduction in the estimated corrosion rate. A transport model was developed to calculate the effect of crack geometry on Ni^{++} concentration in a crack tip and to estimate the thickness of a $NiSO_4$ film at the crack tip.(15)

In the present analysis, this model is applied to the prediction of short crack behavior. To accomplish this, mechanics estimates for crack geometry are evaluated as a function of crack length and related to calculated crack-tip chemistry for anodically polarized nickel. Two difficulties are encountered in the mechanics estimates for short stress corrosion cracks: (1) continuum far-field mechanics calculations are not valid when cracks are of the size of microstructural features such as the grain size, and (2) stress corrosion cracks are moving cracks, i.e., the crack moves through its own crack tip plastic zone. The conventional stationary crack calculations have to be modified to account for wake effects ahead and behind the moving crack.

Estimates of crack opening for selected specimen geometry and for stationary cracks have been reported by Kumar et al.(16) For stationary cracks, the crack is assumed to be introduced prior to the application of a load. Movement of the crack to its current position is not considered. The influence of load and crack length on angle are known for the compact specimen geometry and a single-edge cracked plate. The application of a tensile load on a specimen induces opening of the notch and crack walls. The angle of opening is calculated from estimates of displacements at the crack tip compared to displacements at the load line. The method does not provide estimates of the microscopic opening at the crack tip but nonetheless indicates an estimate of change in the notch and crack wall angle caused by the application of a load.

As a first approximation for a stationary crack, the opening angle dependence on stress intensity in a compact tension specimen is shown in Figure 5. The opening angle is defined as the angle between the crack walls and the crack plane. The effect of elastic and fully plastic assumptions are compared. The opening angles which allow for plasticity are much greater than for the elastic-only assumption. The elastic-plastic case is not significantly different from the fully plastic case. The crack opening angle for stage I cracking in nickel was measured to be less than 0.1 degrees. The stationary crack assumption, with plasticity included, significantly over-estimates the measured angle which suggests that the stationary method of estimating opening angle is not correct.

The stationary crack analysis assumes that the specimen is deformed after the short stress corrosion crack forms in advance of the notch. In contrast during stress corrosion crack growth, the specimen is first placed under stress which caused deformation and then the stress corrosion crack is introduced. The sequence of stress application and crack formation is important in that the response of the crack walls depends on the prior deformation in the region of the crack tip.

An estimate of opening angle caused by an incremental increase in crack length following the application of a stress was calculated. The estimated incremental increase in angle caused by an assumed short-crack growth of 100 μm is shown in Figure 6 as a function of stress intensity. The estimated opening angle of the short-crack is consistent with the measured angle of less than 0.1 degrees. The present method of comparing incremental positions of stationary cracks has the advantage of allowing estimates for crack lengths and stress intensity factors which approach zero as is required for evaluation of initiation of short cracks.

Examination of crack length and angle limits for salt formation during anodic polarization of nickel indicates geometric conditions required for formation of a salt film at a crack tip. Propagation of a crack requires a length and opening angle that ensures no salt formation, i.e., the crack angle must be opened sufficiently with increase in crack length to allow transport of Ni^{++} out of the crack volume. Using the above analysis of crack opening angle, it is possible to estimate the magnitude of opening with increasing crack length to determine if a crack will propagate without being arrested by salt formation. The critical rate of opening is 0.146 degrees/mm of crack extension and is shown by the dashed line in Figure 7.

There is a stress intensity above which crack growth is predicted to occur without salt formation. The effect of stress intensity factor on the calculated rate of crack opening angle is shown in Figure 7 by the solid line. Based on the intersection of the solid and dashed lines in Figure 7, the critical stress intensity for propagating salt-free cracks is estimated to be 6.5 $\text{MPa}\sqrt{\text{m}}$.

The predicted crack growth rate as a function of crack length is shown in Figure 8 for stress intensities near the transition of $6.5 \text{ MPa}\sqrt{\text{m}}$. This response is an example of SCB-I behavior. Cracks are expected to propagate freely until they achieve a critical length for salt formation. At $6.4 \text{ MPa}\sqrt{\text{m}}$, the crack growth is arrested but at $6.6 \text{ MPa}\sqrt{\text{m}}$ the crack growth is uninhibited. In the transition, cracks are significantly retarded.

3. Stage I Stress Corrosion Behavior

3.1 Examples of Stage I Behavior

Experimental crack velocity-stress intensity data for SCC is generally presented in the following form:

$$da/dt = AK^m \quad (1)$$

Values of m in Equation 1, describing the slope of the stage I regime for a variety of materials [17-30], range from 7 to 24; for Ni it is 11; for an aluminum alloy it is 19; for iron tested in $\text{Ca}(\text{NO}_3)_2$ it is 4, for Fe tested in Na_2SO_4 it ranges from 11 to 20, and for X-750 tested in high temperature water it is 5. For Fe and Ni tested at cathodic potentials it is 12 and 7, respectively and for C-275 charged with H it is 12 and 5; depending on crack growth mode.

There is an apparent trend in the Type 304 SS data where m is less than 10 for tests in pure water, 8-14 in Na_2SO_4 , 11-17 in $\text{Na}_2\text{S}_2\text{O}_3$ and 24 in 22% NaCl. This trend suggests an increasing value of m with increasing effectiveness and concentration of critical species. Another trend is revealed by the data in Figure 9 where the value of m is correlated with K_{ISCC}^2 . This correlation suggests that the value of m is a function of K such that higher values of m occur at higher values of K_{ISCC} . This suggests a dependence on the crack tip opening displacement (CTOD), $K^2/\sigma_y E$, and the concept that the crack velocity in stage I is uniquely dependent on the CTOD.

3.2 Comparison of SCC Models and Stage I Behavior

Existing SCC models were not developed to specifically describe stage I behavior and in fact many SCC models do not contain stress or K as a variable. Stress corrosion models that do contain K as a variable are summarized below and in more detail in ref. 31.

Andresen [32] has recently expanded on Ford's [33] model for stress corrosion crack growth rate expressed in terms of strain rate. This analysis resulted in a model where the crack growth rate is proportional to K to the 1.9 to 3.4 power. These values of m are substantially less than those observed experimentally and are more consistent with the K dependence observed in the stage II regime. However, the rate-controlling step in the stage II regime is thought to be either ion transport in the crack electrolyte or the environment-crack tip reaction rate and not the passive film rupture rate as assumed by the Ford/Andresen analysis.

Newman [34] presented a film rupture stress corrosion model for steel in sodium hydroxide in which he assumed that crack extension was limited to the thickness of the corrosion product film, L_c , which fractured in a time period, t_c , determined by K and hence the crack tip strain rate. Numerical solution of Newman's model resulted in curves for the crack velocity versus K for $f(K)$ proportional to K and K^2 . The two functions for K result in similar-shaped curves with K^2 dependence giving a slightly higher threshold value. The predicted slope, m , of the stage I regime was 25 for both cases. Newman concluded that the stress corrosion model was consistent with experimental data, but no comparison was made between calculated and experimental crack velocity-stress intensity relationships. The predicted slope of the stage I regime exceeds the upper end of the range of 4-20 observed for Fe and mild steel. Attempts to calculate the stage I behavior of Fe tested in $\text{Ca}(\text{NO}_3)_2$ resulted in a poor correlation with experimental data for both the linear and square K dependence, as shown in Figure 10.

Vermilyea [35] considered the effect of the stress intensity and the solution chemistry by proposing that the relationship between crack velocity and K has the following form:

$$da/dt = k' C K^m \quad (2)$$

where k' is a constant and C is the concentration of some species in the crack tip solution. Then, assuming that the difference between the crack tip and mouth concentration is a function of crack velocity, he obtained a relationship where m in Equation 2 is equal to R/η , where η is a small constant in the crack tip concentration/crack velocity relationship for which Vermilyea assumed a value of about 0.1. He suggested that small changes in η and hence in the crack tip chemistry could have large effects on the K dependence of the crack velocity. A systematic study of the effects of crack tip chemistry on the parameter m has not been carried out, but data for Type 304SS suggests that there is a trend as mentioned previously. Increasing values of m suggest decreasing values of η in Vermilyea's model; this behavior results from increasing concentrations of a critical species at the crack tip for a given crack velocity.

A recently published crack-tip chemistry-based model by Macdonald and Urquidi-Macdonald [36] for intergranular stress corrosion cracking (IGSCC) of Type 304 SS suggests that the K dependence results from a combination of the passive film rupture rate and crack opening. In essence, this model relates the crack tip potential to K through the balance of the positive current produced by the crack with the reduction of oxygen at the external crack surfaces. A plot of the Urquidi-Macdonald model is given in Figure 11 where the results fit a form of $\log da/dt = A + BK^2$ with a threshold less than $1 \text{ MPa}\sqrt{\text{m}}$. This model predicts a decreasing value of m with increasing values of K . Experimental values for K_{ISCC} for Type 304 SS are approximately $20\text{-}30 \text{ MPa}\sqrt{\text{m}}$, so this model clearly misrepresents the process controlling K_{ISCC} . This model also suggests a continuous transition from stage I to stage II with an m value ranging from large to small depending on the value of K , which is inconsistent with experimental results.

3.3 Modeling Stage I Behavior in Ni + P

Stage I SCC behavior has been modeled using the salt film dissolution model proposed by Simonen et al, and a description of this analysis has been given elsewhere.(37) The predicted crack growth rate is compared to the measured

growth rate in Figure 12. Assumed parameters of resistivity equal to 5600 ohm-cm, tip width equal to 1 nm, and the mechanics beta parameter equal to 4.7 result in excellent agreement between prediction and measurement. This resistivity corresponds to a film porosity of 6×10^{-4} . It is assumed that this porosity is nonhomogeneous because such a small volume fraction would not provide a continuous path if homogeneously distributed. The stress intensity was assumed to be given by the equations of Rice, Drügan, and Sham [37] and the measured size of the crack tip plastic zone. The critical crack opening angle was 0.06 degrees, which agreed with measured angles at the stage I-II transition equal to values less than 0.1 degrees.

4. Summary

Early stages in the development of stress corrosion cracks have been evaluated and modeled. These stages include the growth of short stress corrosion cracks and the stage I region of long crack growth. There are possibly three or more processes which control short crack growth behavior. These include: 1) a salt film formation process which results in fast crack velocities at very short lengths followed by a decreasing velocity for K values less than some critical value, 2) a critical crack tip anion concentration effect which results in an increasing velocity with increasing length up to the limiting long crack velocity and 3) an increase in pH because of increased oxygen inhibition of hydrolytic acidification resulting in a decreasing crack velocity with increasing length. Stage I is considered an early stage in the development of a stress corrosion crack because it precedes the steady state stage II region in which the majority of the crack extension occurs. This stage can be described by a Paris law relationship and experimental data indicates that materials exhibit exponents ranging from 2 to 24. Most SCC models do not treat stress or K effects and therefore do not describe this very distinct stage. However, of those SCC models which do consider stress or K, only the film dissolution model correctly predicts the stage I characteristics. This comparison has only been made for one material however and further work is needed.

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6. References

1. J. Eickemeyer, Corrosion, Vol 43 (1987) p. 693.
2. P. L. Andresen, I. P. Vasatis and F. P. Ford, Corrosion/90, Paper 495, NACE, Houston, TX. x date?
3. R. P. Gangloff, Metall. Trans A, Vol. 16A (1985) p. 953.
4. R. H. Jones, M. J. Danielson and D. R. Baer, J Materials for Energy Systems, Vol 8 (1986) p. 185.
5. S. J. Hudak, J. Eng Mater. Technol., 103 (1981) p. 26.
6. J. Schijve, in Proceedings 1st Int. Conf. on Fatigue Thresholds, J. Backlund, A. F. Blom and C. J. Beevers, eds., Stockholm, June 1-3, 1981, Vol. 7, Engineering Materials Advisory Services, Warley, 1982, p. 881.
7. K. J. Miller, Fatigue Eng. Mater. Struct. 5 (1982) p. 223.
8. S. Suresh and R. O. Ritchie, Int. Metall. Rev., 29 (1984) p. 445.
9. B. N. Leis, M. F. Kanninen, A. T. Hopper, J. Ahmad and D. Broek, Tech. Rep. AFWAL-TR-83-4019, January 1983 (U.S. Air Force Wright Aeronautical Laboratories, Wright-Patterson Airforce Base, OH 45433).
10. R. P. Gangloff and R. O. Ritchie, in Fundamentals of Deformation and Fracture, B. A. Bilby, K. J. Miller and J. R. Willis, eds. Cambridge University Press, Cambridge, Cambridgeshire, 1985, p. 529.
11. J. Lankford, Fatigue Fract. Eng. Mater. Struct., 8 (1985) p. 161.
12. R. O. Ritchie and J. Lankford, Materials Science and Engineering, 84 (1986) p. 11.
13. S. Lee, K. Kim and S. Pyun, Scripta Met., Vol 22 (1988), p. 31.
14. R. N. Parkins, "Localized Corrosion and Crack Initiation" in Proceedings of Mechanics and Physics of Crack Growth: Application to Life Prediction, R. B. Thompson, R. O. Ritchie, J. L. Bassani and R. H. Jones, eds., Keystone, CO, Aug 4-7, 1987, Materials Science and Engr, Vol A103, No.1, p. 143.

15. R. H. Jones and E. P. Simonen, in Proceedings of the Parkins Symposium on Fundamental Aspects of Stress Corrosion Cracking, S. M. Bruemmer, E. I. Meletis, R. H. Jones, W. W. Gerberich, F. P. Ford and R. W. Staehle, Eds., TMS, Warrendale, PA, 1992, p. 69.
16. V. Kumar, M. D. German and C. F. Shih, EPRI NP-1931, Electric Power Research Institute, Palo Alto, CA, 1981.
17. R. H. Jones, M. J. Danielson, and D. R. Baer, in Corrosion of Nickel-Base Alloys, Journal of Materials for Energy Systems, Vol. 18, 1986, pp. 185.
18. F. P. Ford, in Embrittlement by the Localized Crack Environment, R. P. Gangloff, Ed., TMS-AIME, Warrendale, Pa., 1984, pp. 117-148.
19. M. O. Speidel, EPRI NP-2531, Electric Power Research Institute, Palo Alto, Ca, 1982, p. 1-10.
20. R. C. Newman, K. Sieradzki, and H. S. Isaacs, Metallurgical Transactions, Vol. 13A, 1982, pp. 2015.
21. P. Chung, A. Yoshitake, G. Cragolino, and D. D. Macdonald, Paper No. 166, Corrosion/84, New Orleans, La., 1984.
22. M. O. Speidel, in Predictive Capabilities in Environmentally Assisted Cracking, R. Rungta, Ed., 1985, American Society of Mechanical Engineers, New York, pp. 443.
23. D. Singbeil, and D. Tromans, Metallurgical Transactions, Vol. 13A, 1982, pp. 1091.
24. L. D. Lee and Tromans, D., in Environment-Sensitive Fracture of Engineering Materials, Z. A. Foroulis, Ed., TMS-AIME, Warrendale, Pa., 1979, pp. 232-240.
25. R. H. Jones, M. A. Friesel and W. W. Gerberich, Metall. Trans. A, Vol. 20A, (1989), p. 637.
26. R. H. Jones, B. W. Arey, D. R. Baer and M. A. Friesel, Corrosion, Vol. 45, (1989) pp. 494.
27. R. H. Jones, B. W. Arey, C. F. Windisch and D. R. Baer, Corrosion, Vol. 47 (1991) p. 542.
28. T. Kawakubo, and B. Rosborg, Stress Corrosion Cracking Threshold Testing-Scc Growth Behavior of Alloy X-750 and Type 304 Stainless Steel Under Slow Rising Load, Proc. Second Int. Atomic Energy Agency Specialists Meeting on Subcritical Crack Growth, Vol. 1, NUREG/CP-0067, U.S. Nuclear Regulatory Commission, 1986, pp. 477-485.
29. R. H. Jones, Acta Metall. Mater. Vol 38 (1990) p. 1703.

30. N. Sridhar, J. A. Kargol and N. F. Fiore, Hydrogen-induced Crack Growth in a Ni-base Superalloy, Scripta Met., V. 14, p. 225, 1980.
31. R. H. Jones and E. P. Simonen, Mater. Sci. and Engr. A160 (1993) p. 127.
32. P. L. Andresen and F. P. Ford, Matl. Sci. and Eng., A10J (1988) P. 167.
33. F. P. Ford, EPRI NP-2589, Electric Power Research Institute, Palo Alto, CA, 1982.
34. J. F. Newman, Corrosion Science, Vol. 21, 1981, p. 487.
35. D. A. Vermilyea, in Stress Corrosion Cracking and Hydrogen Embrittlement of Iron Base Alloys, R. W. Staehle, J. Hochmann, R. D. McCright and J. E. Slater, Eds., 1973, National Association of Corrosion Engineers, Houston, TX, pp. 208-217.
36. D. D. Macdonald and M. Urquidi-Macdonald, Corrosion Science, Vol. 32 (1991) p. 51.
37. E. P. Simonen, R. H. Jones and M. J. Danielson, Corrosion Science, in press.
38. J. R. Rice, W. J. Drugan, and T-L. Sham, "Elastic-Plastic Analysis of Growing Cracks," Fracture Mechanics: Twelfth Conference, ASTM STP 700, American Society for Testing Materials, 1980, pp. 189-221.

7. Figure Captions

Figure 1. Schematic illustration of crack growth rate-stress intensity behavior of short and long cracks.

Figure 2. Schematic illustration of the variations in crack velocity with time for an operating pipeline (14).

Figure 3. Crack length versus time for compact tension specimen C33 (sensitized Type 304 SS) exposed to deaerated 288 C water (2).

Figure 4. Fatigue crack growth, da/dN vs K , for small elliptical surface and long compact tension cracks in 4130 steel exposed to aqueous 3% NaCl.

Figure 5. Opening angle as a function of stress intensity for a notched compact tension specimen using either fully plastic or elastic

assumptions. The measured opening angle at $18 \text{ MPa}\sqrt{\text{m}}$ was less than 0.1 degrees.

Figure 6. Opening angle as a function of stress intensity for incremental growth of quasi-stationary cracks and an assumed fully plastic compact tension specimen.

Figure 7. Effect of rate of crack opening with increasing length as a function of stress intensity for a crack growing in a notched compact tension specimen. The critical rate of opening to avoid salt formation is 0.146 degrees/mm. The estimates suggest that crack growth without salt formation should occur for stress intensities greater than $6.5 \text{ MPa}\sqrt{\text{m}}$. This value is independent of crack length.

Figure 8. Effect of crack length on crack growth rate for stress intensities near $6.5 \text{ MPa}\sqrt{\text{m}}$.

Figure 9. Stress intensity exponent, m , versus K_{2Isc} for Fe and Ni at anodic and cathodic potentials and for Type 304 SS (31).

Figure 10. Comparison of the experimental crack velocity versus stress intensity relationship for Fe tested in $\text{Ca}(\text{NO}_3)_2$ and the calculated relationship from Newman's model (34).

Figure 11. Calculated crack velocity versus stress intensity for Type 304 SS based on model by Macdonald (36).

Figure 12. Comparison of calculated and measured crack growth rates for variable salt resistivities and crack widths.

39304117.6

Log. Crack Velocity (mm/soc)

39304117.1

39304117.3

39304117.4

EPS Figures for ICFF May 1993 Paper

Figure 5 Opening angle as a function of stress intensity for a notched compact tension specimen using either fully plastic or elastic assumptions. The measure opening angle at about 18 MPa-m^{0.5} was less than 0.1 degrees.

Figure 6 Opening angle as a function of stress intensity for incremental growth of quasi-stationary cracks and an assumed fully plastic compact tension specimen.

Figure 7 Effect of rate of crack opening with increasing length as a function of stress intensity for a crack growing in a notched compact tension specimen. The critical rate of opening to avoid salt formation is 0.146 degrees/mm. The estimates suggest that crack growth without salt formation should occur for stress intensities greater than 6.5 MPa-m^{0.5}. This value is independent of crack length.

Figure 8
Effect of crack length on crack growth rate for stress intensities near
6.5 MPa-m^{0.5}.

39202091.1

39202091.6

Figure 5

Comparison of the experimental crack velocity versus stress intensity relationship for Fe tested in $Ca(NO_3)_2$ and the calculated relationship based on Equation 2-2.

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 Figure 8 Calculated crack velocity versus stress intensity for Type 304 SS based on Equation 2-4 with $[O_2] = 0.02$ ppm, $L = 1$ cm, $[NaCl] = 10^{-15}$ mol/cm², and $T = 573^\circ K$. Other parameters are as listed in Ref. 20.

Figure 12 Comparison of calculated crack growth rates with measured crack growth rates for indicated salt resistivities and crack tip widths.

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