

Stress Rate Effects in Stress Corrosion Cracking

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ABSTRACT

This paper presents a conceptual model for the stress corrosion cracking of quenched and tempered low alloy steels in high temperature, high purity water. The model is based on the cracking mechanism of anodic dissolution. It is based on the postulate that the crack tip strain rate is the controlling parameter, with the crack velocity increasing with an increase of crack tip strain rate. Furthermore, it is assumed that the relationship between crack velocity and crack tip strain rate is valid up to very high crack velocities due to strain assisted dissolution. Following Parkins we propose that K_{ISCC} is defined by a critical crack tip strain rate, $\dot{\epsilon}_{cr}$. That is, the proposed model of SCC explains K_{ISCC} as being the stress intensity at which significant dislocation motion starts to occur to give significant crack tip creep.

KEYWORDS

Stress corrosion cracking; stress rate; crack tip strain rate; model.

INTRODUCTION

This paper presents a conceptual model for the stress corrosion cracking of quenched and tempered low alloy steels in high temperature, high purity water. The model is based on the cracking mechanism of anodic dissolution and the experimental observations (Rieck 1988, Rieck et al 1988a,b) that crack tip plasticity is an integral part of crack propagation, with there being considerable experimental evidence (Rieck et al 1988b) contrary to crack advance by a mechanism of brittle mechanical fracture assisted by corrosion processes.

Much study (Atrens et al 1989) has gone into the system of interest and several attempts have been made to propose crack growth models. None to-date have been able to explain the observations of Speidel (1984) who showed that the crack velocity increases exponentially with increasing strength, increasing from a crack velocity of 1.0×10^{-11} m/s at a yield strength of 700 MPa to a crack velocity of 1.0×10^{-4} m/s at a yield strength of 1700 MPa.

It is the purpose of this paper to present a conceptual model for this system, which can provide a qualitative explanation of the experimental observations. It indicates where there are large gaps in our knowledge and thereby provides a focus for research. The model is conceptual and not mathematical because of these gaps in knowledge.

TABLE 1, Effect of Prior Creep on SCC

specimen	metal	creep time	environment	cracking
1	3.5NiCrMoV	0	H ₂ O	yes, $v = 1 \times 10^{-7}$ m/s
2	3.5NiCrMoV	72 h	dried air	no
3	4340	0	H ₂ O	yes, $v = 8 \times 10^{-5}$ m/s
5	4340	48 h	dried air	no

CRACKING MECHANISM

For the system: low alloy steels undergoing stress corrosion cracking in high purity water, recent work by Magdowski and Speidel (1988) and Magdowski (1987) has indicated that the cracking mechanism is different at high temperatures and at low temperatures; they suggested anodic dissolution (AD) at high temperatures and hydrogen embrittlement at low temperatures. A review of the literature by Atrens et al (1989)

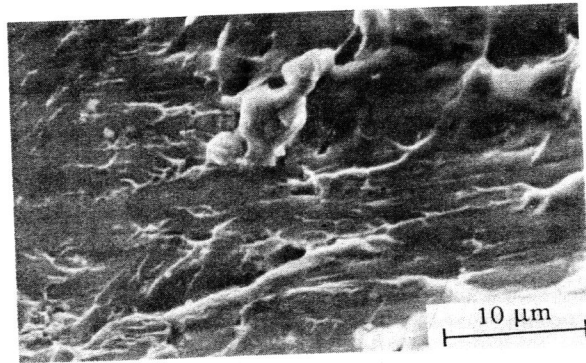


Fig. 1. High magnification views for 3.5NiCrMoV of the interface between SCC and subsequent overload fracture which is by dimple rupture. The surface markings on the intergranular fracture facets can be seen as an array of ledges oriented approximately in parallel directions.

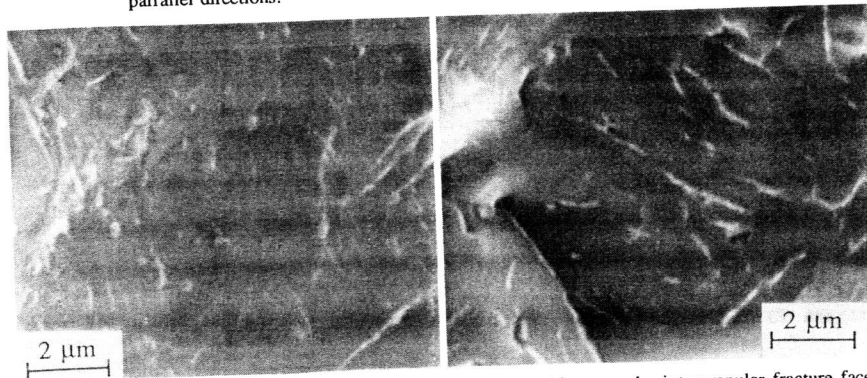


Fig. 2. High magnification views of the surface markings on the intergranular fracture facets on (a) 4340 and (b) 3.5NiCrMoV.

supports this view. The critical data are as follows.

In high temperature water, the threshold stress intensity for stress corrosion cracking, K_{ISCC} is independent of the yield strength of the steel, σ_y , Magdowski and Speidel (1988), Magdowski (1987) and Speidel (1984), whereas the threshold stress intensity for cracking in hydrogen, K_{IH} increases as σ_y decreases (Nelson and Williams 1977). These different trends of K_{ISCC} and K_{IH} with strength indicate that hydrogen embrittlement is not the cracking mechanism. This is reinforced by the fact that the crack velocity in region II under mode I loading, $v_{mode I}$, is the same as under mode III loading (Rieck 1988), $v_{mode III}$.

In contrast, in low temperature water, the threshold stress intensity for stress corrosion cracking, K_{ISCC} increases as the yield strength of the steel, σ_y decreases (Magdowski and Speidel 1988 and Magdowski 1987) and the threshold stress intensity for cracking in hydrogen, K_{IH} increases as σ_y decreases (Nelson and Williams 1977). These observations are in agreement with hydrogen embrittlement being the cracking mechanism. This is reinforced by the fact that the crack velocity in region II under mode I loading, $v_{mode I}$, is greater than under mode III loading (Chu et al 1982), $v_{mode III}$.

IMPORTANCE OF STRAIN RATE

Creep experiments, Table 1, (Rieck 1988 and Rieck et al 1988a) have demonstrated that crack tip plasticity is an integral part of crack propagation; if primary creep is exhausted in a benign environment, then no stress corrosion cracking occurs in a cracking environment despite an applied stress intensity four times K_{ISCC} . Moreover, the detailed examination of the fracture surfaces (Rieck et al 1988b) indicates traces of crack tip plasticity on the intergranular fracture surface with there being considerable experimental evidence (Rieck et al 1988b) contrary to crack advance by a mechanism of brittle mechanical fracture assisted by corrosion processes, Fig. 1 & 2.

These show similar features on both steels with a spacing of approximately 1 µm. This similarity in fractography is to be contrasted to very different crack velocities: 1×10^{-4} for 4340 compared with 5×10^{-8} . In addition Ramamurthy et al (1988) have measured the crack velocity as a function of applied stress rate using the newly developed LIST test (linearly increasing stress test) (Atrens et al 1989). Typical data are given in Fig. 3 (Ramamurthy et al 1988) which shows crack velocity as a function of applied stress rate at crack initiation. This test uses a smooth un-notched tensile specimen. Crack initiation is in the elastic region and consequently it is valid to discuss crack velocity also as a function of applied strain rate at crack initiation.

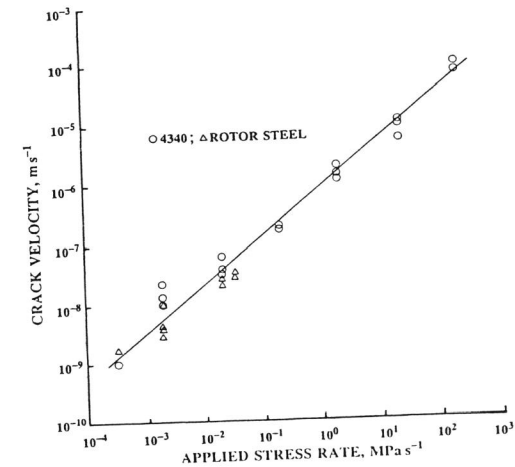


Fig. 3. Crack velocities measured with the LIST test (Ramamurthy et al 1988) as a function of applied stress rates for both as-quenched 4340 and rotor steel in aerated distilled water at 90 C.

MODEL

General Principles

The proposed model of SCC is based on the following postulates:

1. Crack tip strain rate is the controlling parameter up to the maximum velocity observed experimentally for this system.
2. There is a relationship between crack velocity and crack tip strain rate which is valid up to the maximum velocity observed experimentally for this system.
3. The relationship between crack tip strain rate and the applied stress intensity is as shown in Fig. 4.

There is good experimental evidence for postulate 1 (Rieck 1988, Rieck et al 1988a,b). Postulate 2 is a plausible generalization of postulate 1. It is central to this conceptual model. If postulate 2 is correct, then there is good evidence for postulate 3.

Importance of Crack Tip Strain Rate

The experimental evidence (Rieck 1988, Rieck et al 1988a,b) in support of crack tip strain rate was summarized previously.

Crack Tip Strain Rate -v- K_{ISCC}

For loading such that there is zero stress or zero stress intensity, the crack tip strain rate must be zero. For reasons given below, there is a critical stress intensity, termed K_p after Speidel (1971), above which the strain rate is independent of the applied stress intensity. Fig. 4 reflects the above joining the two extremes with a region wherein the crack tip strain rate increases very quickly, even exponentially, with stress intensity.

Above K_p

Above K_p , the crack tip has a well developed plastic zone with the maximum stress in this plastic zone being $3\sigma_y$ (1973). Higher stress intensities increase the size of the plastic zone at the crack tip but do not affect the stress within the zone. Consequently, it is a plausible inference that, above K_p , the creep rate is independent of stress intensity because the deviatoric stress which drives creep is independent of stress intensity.

It is important to emphasise that we are dealing with crack tip strain rate or creep rate at the crack tip under conditions of full plasticity in precracked specimens with the additional factor of constraint. As highlighted

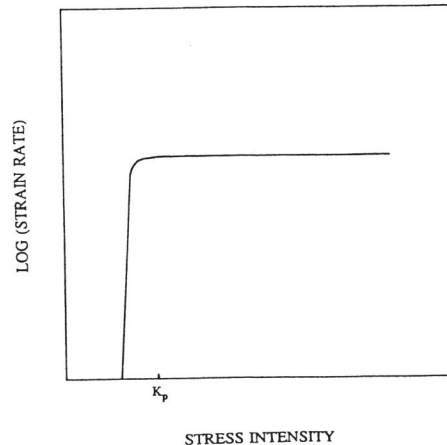


Fig. 4. Expected relationship between crack tip strain rate and applied stress intensity.

previously, this produces a higher stress at the crack tip for a higher strength steel (Knott 1973). High strength steels also have a much smaller zone of plasticity, so that dislocation motion will be more concentrated at the crack tip. These two factors combine to produce a higher strain rate at the crack tip in high strength steels under conditions of full plasticity.

On the assumption that there is a relationship between stress corrosion crack velocity and crack tip strain rate, this gives a region of crack velocity independent of applied stress intensity but dependent on the yield strength σ_y of the material as illustrated in Fig. 5 and in agreement of the experimental results of Speidel and co-workers (Speidel 1984, Magdowski 1987, Magdowski and Speidel 1988) who showed that the crack velocity increases exponentially with increasing strength increasing from a crack velocity of 1.0×10^{-11} m/s at a yield strength of 700 MPa to a crack velocity of 1.0×10^{-4} m/s at a yield strength of 1700 MPa.

Creep data as described by conventional creep experiments refer to steady state creep, whereas it has been shown that stress corrosion cracking is controlled by primary creep (Rieck et al 1988a, Parkins 1979). Primary creep in newly stressed metal occurs at a faster rate than secondary creep. An experimental program on the primary creep of low alloy steels would be required to test this proposal, as experimental results are not presently available.

Threshold Stress Intensity

Following Parkins (1979) we propose that K_{ISCC} is defined by a critical crack tip strain rate, $\dot{\epsilon}_{cr}$. That is, the proposed model of SCC explains K_{ISCC} as being the stress intensity at which significant dislocation motion starts to occur to give significant crack tip creep. The process which limits plastic deformation such that a significant strain rate is only found above K_{ISCC} has not been identified. The observation that K_{ISCC} is independent of σ_y and temperature implies that these parameters have no effect on the processes limiting strain at the crack tip (Speidel 1984, Magdowski 1987, Magdowski and Speidel 1988).

DISCUSSION

Strain Assisted Dissolution

A crack velocity of 10^{-4} m/s requires a crack tip anodic current density of 300 A/cm² for cracking by AD. Such high current densities have not been observed in scratching electrode tests. Lacombe and Parkins (1977) report a maximum bare surface current density of 3 A/cm², correlating with a crack velocity of 10^{-6} m/s. The problem for AD models of SCC, is accounting for crack velocities higher than 10^{-6} m/s. This requires a mechanism for accelerating the dissolution rate occurring at the crack tip.

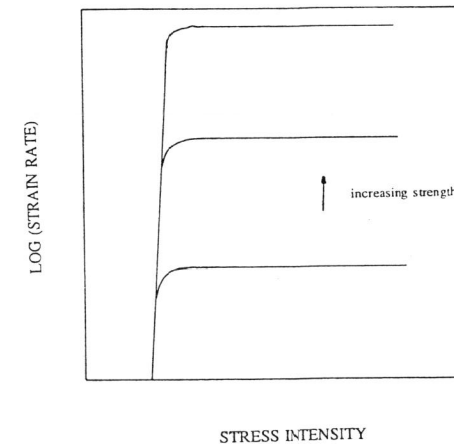


Fig. 5. Expected effect of metal yield strength on the relationship between crack tip strain rate and stress intensity.

This acceleration could be provided by strain assisted dissolution at a bare crack tip. The large stress concentration at the crack tip ensures that dislocation motion is concentrated at the crack tip itself. Each dislocation which impinges upon the crack tip produces a ledge of bare metal in contact with the cracking solution. The atoms on the ledge so produced can dissolve very quickly due to a number of effects. Clean metal (ie without any films) is in contact with the solution. Atoms on the ledge are not strongly bound to the bulk metal due to the missing bonds on the two free faces of the ledge. These features cause the metal on the ledge to dissolve much more quickly than would occur on a film free surface. Moreover, the dissolution rate is strain rate controlled through the rate of production of ledges by dislocation motion.

Moreover, there is also a small anodic area (active crack tip) in contact with a large cathodic area (crack sides), due to the geometry of the crack (Vermilyea and Diegle 1973), which increases the rate of reaction. The likelihood of such an effect being important cannot be assessed directly due to the unknown nature of the crack tip electrochemistry. However an area effect cannot account for the effect of strength on v_{II} , since the analysis of Doig and Flewitt (1984), which included strength defining the area of dissolution, could not account for the exponential relationship between σ_y and v_{II} . The independence of cracking kinetics on metal composition also makes it unlikely that area effects are important to SCC, as the composition controls the corrosion potential and areas of anodic and cathodic reaction. Variations in these parameters should change v_{II} and the lack of such change in v_{II} makes it unlikely that area effects are important to SCC. This indicates that strain assisted dissolution may be the only feasible method of accounting for the high crack velocities found in the high strength steels.

Prior Creep

Rieck et al (1988a) have shown that prior creep in a non-cracking environment can inhibit cracking when a specimen is subsequently placed in a cracking environment even if the applied stress intensity is well above K_{ISCC} .

On loading above K_p , into region II of SCC, the crack tip strain rate will be that shown in Fig. 4, unless subcritical cracking does not occur, in which case this metal will undergo primary creep. Primary creep results in work hardening of the metal at the crack tip and a decreasing creep rate with time to such an extent that it is less than that required for the initiation of SCC, that is the crack tip strain rate falls below $\dot{\epsilon}_{cr}$.

The value of $K_{ISCC} = 8 \text{ MPa}\sqrt{\text{m}}$ measured by Speidel (1984) for low alloy steels cracking in pure water is only applicable to metal freshly stressed in the cracking solution, or metal which has cracked from a higher stress intensity down to K_{ISCC} with the stress intensity decreasing as the crack extends. This ensures that the strain hardened metal at the crack tip is continuously removed by dissolution.

Intergranular Cracking

The intergranular nature of SCC, has normally been attributed to segregation effects with phosphorus being the major species (Bandyopadhyat and Briant 1982, 1983, Burnstein and Woodward 1983). Recent work by Magdowski and Speidel (1988) has shown that for low alloy steels in high temperature water, segregation has no effect on SCC. Therefore, a new explanation is needed for the crack path.

In low alloy steels, wavy slip ensures that slip cannot concentrate on any one plane, nor at any obstacle within the grains. In the early stages of deformation the most significant obstacle is the grain boundary so that dislocation motion will be concentrated in the grain boundary region. This causes cracking to be intergranular. In the quenched and tempered steels, the prior austenite grain boundaries remain high angle boundaries and as such are a more significant barrier to dislocation motion than the lower energy twin boundaries found in martensite. Thus cracking occurs along the prior austenite grain boundaries. Such a result also explains the importance of slip morphology to SCC.

CRITICAL TESTS FOR THE PROPOSED MODEL

The proposed model of SCC depends upon a number of assumptions which could be tested as to their validity, and by consequence, the validity of the model itself.

One critical assumption is that of the effect of stress intensity on crack tip creep. The fact that the creep rate at the crack tip is exponentially dependent on the stress at the crack tip is the most important assumption made. This is very difficult to test as it is creep at the crack tip rather than the region of general yielding which is important to SCC. Creep tests on DCBs measuring only the crack opening displacement are unable to define the creep rate at the crack tip as the COD is defined mainly by the size of the plastic zone

at the crack tip. It is this difficulty in measuring crack tip creep rate which makes the model difficult to test. Strain assisted dissolution was invoked to explain the very high crack velocities found in the high strength steels. It may be possible to test this assumption using electrochemical measurements on straining samples. However, to simulate the crack tip conditions of creep, it is necessary to localise the straining region while keeping the rest of the sample passive.

The easiest test of the proposed model is a stress analysis of a precracked sample loaded in mode III. The result of comparable crack velocities in modes I and III indicates similar crack tip creep rates in the two loading modes. A stress analysis of the mode III loading should allow a comparison of the deviatoric component of stress at the crack tip in both loading modes. The experimental results indicate that this stress should be similar in the two loading modes.

REFERENCES

- Arens, A, RM Rieck, S Ramamurthy, JD Gates, IO Smith (1989). *International Metal Review*, accepted for publication.
- Arens, A, C Brosnan, S Ramamurthy, IO Smith (1989). submitted to *J Phys E: Sci Inst*.
- Bandyopadhyat, N. and CL Briant, (1982). *Corrosion*, **38** 125.
- Bandyopadhyat, N. and CL Briant, (1983). *Met Trans*, **14A** 2005.
- Burstein, G. and J Woodward, (1983). *Met Sci*, **17** 111.
- Chu, WY, CM Hsiao, SY Ju, C Wang (1982). *Corrosion*, **38** 446.
- Doig, P. and PEJ Flewitt (1984). *Corrosion in Power Generating Equipment*, ed MO Speidel and A Arens, 133.
- Knott, JF (1973). *Fundamentals of Fracture Mechanics*.
- Lacombe, P, RN Parkins (1977). *NACE-5*, 521.
- Magdowski, RM and MO Speidel (1988). *Met Trans*, **19A** 1583.
- Magdowski, RM (1987). *Ergebnisse der Werkstoff-Forschung - Moderne Staehle*, ed PJ Uggowitzer 107.
- Nelson, HG, DP Williams, (1977). *NACE-5* 390.
- Parkins, RN (1979). *Br Corros J*, **14** 5.
- Ramamurthy, S, RM Rieck, A Arens, RM Rieck, JD Gates, IO Smith (1988). *ASIA CORROSION 88*, paper 21.
- Rieck, RM (1988). *The Mechanism of Stress Corrosion Cracking of Quenched and Tempered Low Alloy Steels in Water*, PhD Thesis, Univ Queensland.
- Rieck, RM, A Arens, IO Smith (1988a). *Met Trans*, accepted for publication.
- Rieck, RM, A Arens, IO Smith (1988b). submitted to *Materials Forum*.
- Speidel, MO (1984). *Corrosion in Power Generating Equipment*, ed MO Speidel and A Arens, Pleum, 85.
- Speidel, MO (1971). *The Theory of Stress Corrosion Cracking in Alloys*, JC Scully ed, 345.
- Vermilyea, DA, RB Diegle, (1973). *Corrosion*, **32** 26.