

# Elastic-Plastic Fracture Mechanics Analysis of SCC in a Low Strength Steel

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## ABSTRACT

Stress corrosion crack growth during constant displacement rate testing was investigated using elastic-plastic fracture mechanics. Compact tension specimens of a low strength steel were tested in 2 N ammonium carbonate solution at 75°C under freely corroding conditions. The crack growth rates were correlated with the time derivative of the J-integral,  $dJ/dt$ , and the modified J-integral known as  $C^*$  or  $J^*$ . Based on the energy approach to SCC, the chemical energy contribution was evaluated and compared with that of the mechanical energy factor.

## KEYWORDS

Stress corrosion crack growth,  $C^*$ -integral, energy balance approach.

## INTRODUCTION

There is considerable evidence that the slip-dissolution model, also called the film-rupture model, is applicable to the stress corrosion cracking of a wide range of alloys in passivating solutions (Bursle and Pugh, 1977). Failure caused through this mechanism is characterized mainly by ductile fracture and discontinuous crack propagation. As an example, Fig. 1. shows the fractograph taken from a low strength ferritic steel in 2N  $(NH_4)_2 CO_3$  solution (Sarioğlu, 1985). Apart from the voids formed during crack propagation, the fracture surface consists of dimples, which in turn shows the ductile nature of fracture. Crack propagation is highly discontinuous partly due to the interaction between the major crack and the small cracks initiated on planes parallel to that of the main crack (Fig. 2).

According to the slip-dissolution mechanism proposed for SCC a protective surface film is ruptured by localized plastic deformation at the crack tip, where the exposed metal becomes a very small anodic region with respect to the non-deforming crack walls which are protected by the film acting as a

cathode. Therefore for the stress corrosion crack to grow the crack tip must be in the film-free condition. This can be achieved if the protective films are not allowed to grow over the crack tip or if this occurs, the film is repeatedly broken under the applied stress. The maintenance of film-free conditions is therefore both dependent on the strain rate and the repassivation rate determined by electrochemical conditions.

This study aims to explain the role of the crack tip deformation rate on SCC in an alloy-environment system where the slip-dissolution model would be applicable. For this purpose, the crack growth kinetics is investigated using elastic-plastic fracture mechanics.

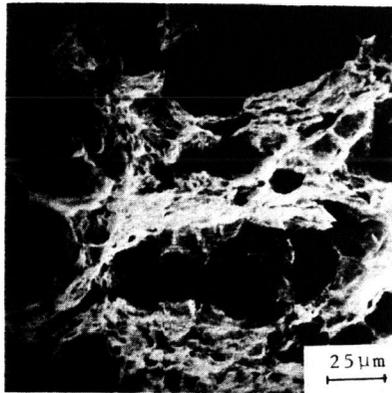


Fig. 1. Stress corrosion fracture surface of a low strength ferritic steel in an ammonium carbonate solution.

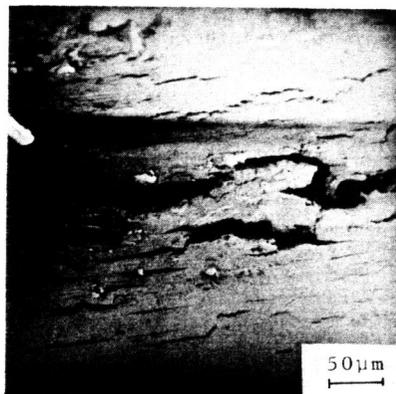


Fig. 2. The stress corrosion crack profiles on the surface of side groove of the same specimen in Fig. 1.

## EXPERIMENTAL

The material used was a hot-rolled sheet of a steel having a composition (wt%) of 0.22%C, 0.70%Mn, 0.05%Si, 0.02%P, 0.03%S and 0.2%Cu. The mechanical properties of the material are given in Table 1.

Table 1. Tensile properties.

0.2% Proof Strength 295 MPa	Tensile Strength 450 MPa	Reduction in Area 60%	Young's Modulus $1.76 \times 10^5$ MPa
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13 mm thick and 60 mm wide compact tension specimens were prepared such that the crack growth direction was parallel to the rolling direction. Side grooves were machined to have a ratio of net to gross thickness of 0.76. The specimen surfaces except the grooves were chromium plated to satisfy the condition of inactive crack surface. After fatigue precracking in air, SCC tests were conducted under constant load-line displacement rates using a closed-loop hydraulically activated servo-controlled testing machine. The machine was set in the horizontal position to facilitate complete immersion of the test piece in the corrosive solution. The test solution was 2N  $(\text{NH}_4)_2\text{CO}_3$  at 75°C. A clip-on-gage was used to record the load-line displacement values and crack lengths were measured optically to an accuracy of 0.01 mm. Displacement rates ranged from  $7 \times 10^{-6}$  mm/sec to  $8 \times 10^{-5}$  mm/sec.

## EXPERIMENTAL RESULTS

The crack growth-time data were fitted to a second order polynomial from which the crack growth rates could be determined at any test time.

The instantaneous J-integral values were calculated as (Clarke *et al.*, 1979):

$$J = (2A/Bb) \left[ (1 + \alpha) / (1 + \alpha^2) \right] \quad (1)$$

where

$$\alpha = 2 \left[ (a/b)^2 + (a/b) + 1/2 \right]^{1/2} - (a/b + 1/2)$$

A is the area under the load-displacement record in energy units, B is the thickness of the specimen, b is the remaining uncracked ligament length of the specimen, and a is the crack length.

The modified J-integral  $C^*$  also known as  $J^*$ , was originally developed for crack growth under steady state creep conditions by Landes and Begley (1976). The graphical technique to determine  $C^*$  described by Landes and Begley is quite cumbersome. Analytical solutions to  $C^*$  have been estimated by considering the strain and strain rate as power functions of stress (Koterazawa and Mori, 1977, Harper and Ellison, 1977, Saxena, 1980). One of the solutions for the compact tension geometry in plane strain condition is the one given by Koterazawa and Mori (1977):

$$C^* = (2P/Bb) dv/dt \quad (2)$$

where P=load,  $dv/dt$ =load-line displacement rate, b=remaining uncracked ligament and B=thickness of the test piece. Results of the crack growth rate

analysis by the parameters which represent the changing rate of the stress-strain fields i.e.,  $dJ/dt$  and  $C^*$  are given in Figs. 3 and 4.

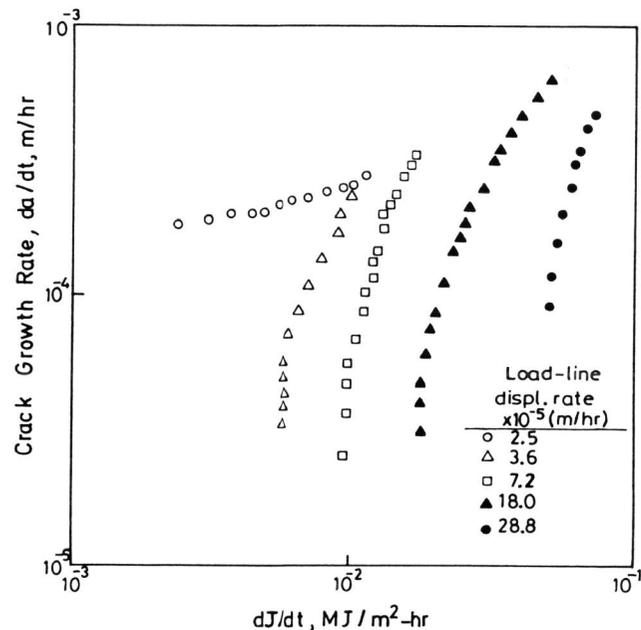


Fig. 3. Crack velocity as a function of the time derivative of the J-integral for different load-line displacement rates.

#### DISCUSSION

Crack growth rate versus  $dJ/dt$  plots shown in Fig. 3. resembles the well known crack growth rate-stress intensity,  $K$  relation obtained in high strength materials which give a threshold value known as  $K_{ISCC}$ . It is observed from Fig. 3. that thresholds shift to lower values as displacement rates decrease. This can be explained as approaching to SCC conditions with decreasing displacement rates.

The energy balance approach to mechanical fracture was modified for the environmental sensitive fracture by summing the strain energy release and the electrochemical energy release to provide the total surface energy change and the plastic work done near the crack tip (Parkins and Craig, 1977). The surface energy component is usually negligible in comparison with the plastic work component in SCC of ductile materials and, hence:

$$\gamma_p = \text{strain energy release} + (zF\rho/M)\delta\eta \quad (3)$$

The second expression on the right side of eq. (3) gives the electrochemical

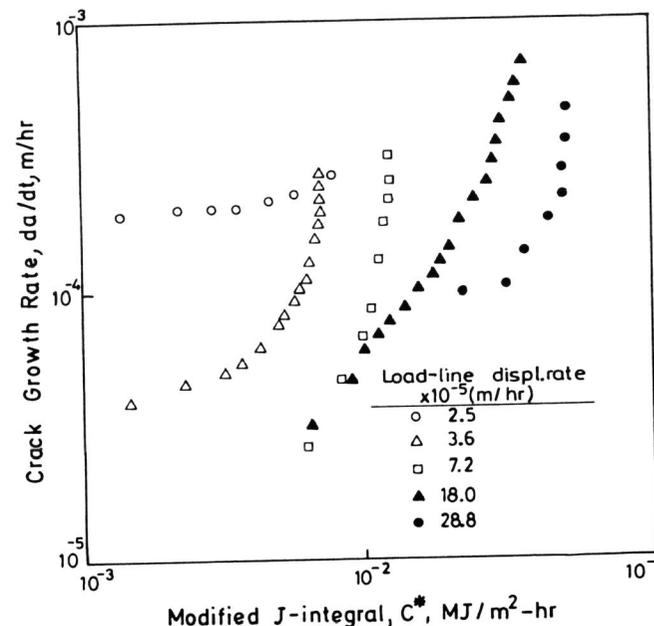


Fig. 4. Crack growth rate as a function of the modified J-integral for different load-line displacement rates.

energy where  $z$ =the valency of the solvated ions,  $F$ =Faraday's constant,  $\rho$ =density,  $M$ =molecular mass of the metal,  $\delta$ =the height of the advancing crack front which can be approximated to the crack tip opening displacement and  $\eta$ =anodic overpotential.

Considering the rate of change of the energy terms:

$$d\gamma_p/dt = \text{Strain energy release rate} + (zF\rho/M)\eta (d\delta/dt) \quad (4)$$

For a stationary crack the crack tip displacement,  $\delta_t$ , can be estimated by using the J-integral as derived from the Hutchinson-Rice-Rosengren singularity (Shih, 1981):

$$\delta_t = (d_n/\sigma_0) J \quad (5)$$

where,  $\sigma_0$ =yield stress, and  $d_n$ =a function of strain hardening exponent and the yield strain. Values of  $d_n$  for a range of materials are given by Shih (1981). Considering the effect of strain rate on  $\sigma_0$  and  $d_n$  to be negligibly small, crack tip deformation rate can be obtained by time derivative of eq. (5) so that the crack tip deformation rate is characterized by  $dJ/dt$ .

$$d\delta_t/dt = (d_n/\sigma_0) dJ/dt \quad (6)$$

The electrochemical energy term will also depend on  $\eta$  which is a function of electrochemical conditions within the crack, i.e., of pH, anion activity, metal composition and electrode potential. Direct measurement of the

potential drop along a crack is rather difficult. Turnbull and Thomas (1980) have predicted that the potential drop with distance from the tip of a stationary crack in BS4360 steel in 3.5% NaCl varies in the order of 10 mV. By comparing the potential ranges for cracking of pre-cracked and plain specimens, Parkins (1977) has found this value to be 20 mV for a C-Mn steel in a CO<sub>3</sub>-HCO<sub>3</sub> solution.

Taking the overpotential term,  $\eta$  as 20 mV for our case, the strain energy rate represented by  $dJ/dt$  and  $C^*$  and electrochemical energy rates are calculated using eqs. (6) and (4) for all applied displacement rates (Table 2).

It is found from these calculations that the electrochemical contribution to the driving force for crack growth is the same for all crack lengths and applied displacement rates. This contribution for our case is found to be 35% when calculated from  $dJ/dt$  and between 36-42% from  $C^*$ . The evidence from the results is that the driving force for SCC is dictated by the electrochemical conditions in the crack.

The evaluation of energy effects on the crack driving force in SCC was done by considering a stationary crack. For a growing crack, the analysis of Rice and Sorenson (1978) can be used, which is:

$$d\delta_t/dt = \alpha(dJ/dt)/\sigma_0 + \beta\sigma_0/E (da/dt) \ln (R/r) \quad (7)$$

where  $\alpha$  and  $\beta$  are constants,  $r$  is the distance from the crack tip, and  $R$  scales approximately with the maximum radius of the plastic zone, and  $da/dt$  is the crack velocity. Exact calculation of the second term on the right side of eq. (7) is ambiguous due to the arbitrary value of  $r$ . In this case, the contribution of the electrochemical energy to the total energy released as crack grows is expected to increase.

Table 2. Strain energy and chemical energy contributions to SCC for the applied displacement rates when  $a/W = 0.53$ .

Load-line disp.rate (m/hr)	Strain energy release rate (kJ/m <sup>2</sup> -hr)		Chemical energy release rate (kJ/m <sup>2</sup> -hr)	Crack tip def.rate (m/hr)	Chemical energy (%)	
	Calc'd from $dJ/dt$	$C^*$			Calc'd from $dJ/dt$	$C^*$
2.5x10 <sup>-5</sup>	3.6	3.5	2.0	4.0x10 <sup>-6</sup>	35.7	36.3
3.6x10 <sup>-5</sup>	7.8	6.7	4.3	8.0x10 <sup>-6</sup>	35.8	39.2
7.2x10 <sup>-5</sup>	14.5	12.3	8.0	1.5x10 <sup>-5</sup>	35.4	39.3
18.0x10 <sup>-5</sup>	42.6	33.7	23.9	4.4x10 <sup>-5</sup>	35.9	41.5
28.8x10 <sup>-5</sup>	68.0	50.7	37.5	6.9x10 <sup>-5</sup>	35.6	42.5

#### CONCLUSIONS

1. Under large scale yielding conditions, stress corrosion crack growth rates should be correlated with the parameters such as  $dJ/dt$  or  $C^*$  which estimate the crack tip deformation rate.
2. The threshold values of  $dJ/dt$  were found to decrease with decreasing deformation rate.

3. From the energy balance approach to fracture, the electrochemical energy contribution to the driving force for crack growth was calculated to be approximately 35% without being affected by the displacement rate. From this result it was concluded that the driving force for the stress corrosion crack to grow depends mostly on the electrochemical conditions at the crack tip which are dictated by the nature of the dissolving metal and the corrosive environment.

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# Fracture Toughness Testing in the Ductile-Brittle Transition Region

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## ABSTRACT

Fracture toughness testing in the ductile-brittle transition regime is often problematic. In this region both constraint effects as well as ductile tearing will affect the brittle fracture probability. This causes the scatter to grow very large in the transition region. Up till now there has not existed any kind of reliable means of how to validate such data. In this paper a new method for validation is presented.

Based on a theoretical cleavage fracture model, it is possible to evaluate the effect of ductile tearing on the cleavage fracture probability. When combining this model with the knowledge from finite element calculations of the plastic zone, it is not only possible to evaluate a minimum specimen size to obtain valid brittle fracture toughness results, but it is also possible to change the invalid results into valid results. It is also possible to describe the fracture probability of large specimens and structures from the results obtained with "invalid size" specimens.

## INTRODUCTION

The standardization of fracture toughness testing is presently undergoing rapid development. Many testing standards already exists and presently several new standards are being developed together with revision of the existing ones.

The most problematic toughness range seems to be the transition region where it is practically impossible to measure valid  $K_{IC}$  results and yet the final fracture is cleavage fracture. Presently only the British CTOD-test standard BS 5762 is applicable in this toughness region. The CTOD-standard does, however, in the present form contain weaknesses that deteriorates it's applicability in the transition region.

In this paper the effect of ligament size on cleavage fracture toughness in the elastic-plastic regime is examined. The large scale yielding effect of the ligament is combined with statistical cleavage fracture theory so that