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Subcritical Crack Growth: Fatigue, Creep and Stress Corrosion Cracking [and Discussion]

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Subcritical crack growth: fatigue, creep and stress corrosion cracking

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Subcritical crack growth can occur under steady or varying loads. In the former it is precipitated by specific environmental conditions that encourage the operation of time-dependent processes controlling crack advance. These include aggressive environments leading to stress corrosion cracking, or elevated temperature conditions leading to creep cavitation. The result is a time-dependent maintenance of a sharp crack profile during crack extension. Under varying loads such a sharp profile is readily achieved by plastic deformation on load reduction. Net crack advance in fatigue therefore occurs in each load cycle by this blunting–resharpening process, and empirical crack growth laws reflect this physical basis. Parameters such as K and J, which define crack tip deformation, are useful for correlating fatigue crack growth. In that they define crack tip stress–strain fields under load, they also partly describe crack advance for steady load creep and stress corrosion cracking. In particular they can define a threshold state for crack extension by all three processes. Under varying loads, if fatigue conditions are combined with an aggressive or high-temperature environment the description of crack growth can be complex. These areas of corrosion fatigue and creep fatigue are of considerable current practical interest.

Introduction

Engineers are increasingly willing to accept that the structures and components that they design may contain defects or cracks when fabricated. In addition, during service the operation of various time-dependent processes (e.g. fatigue, creep, corrosion and wear) can both create and develop such defects. It is desirable from both the safety and reliability viewpoints to guarantee that cracks do not develop to such a degree that the structure fails. The definition of the failure condition in terms of a critical crack size, a_f, is both a function of the material and its structural context. The life of the structure or component depends on the rate at which the crack develops from its initial size, a_0, to the critical size, a_f. To this stable 'crack propagation life', t_p, may be added on 'initiation life', t_i, for quality components which are free from initial cracks. Total life, t_f, is the sum of t_i and t_p and this must exceed the design life of the component. Estimation of t_i is extremely difficult, but t_p can be deduced from a knowledge of crack growth rate ($\mathrm{d}a/\mathrm{d}t$) during service. This paper examines the factors that determine $\mathrm{d}a/\mathrm{d}t$ for various time-dependent cracking processes.

Three such processes are experienced by materials: fatigue, creep and stress corrosion. Fatigue, which is caused by variability of applied loading, is by far the most widely encountered and is a contributory factor to most service failures. Creep crack growth occurs under steady loading at elevated temperature as a result of the degradation of material ahead of the main crack by time-dependent cavity or micro-crack formation. Stress corrosion crack growth also occurs under a steady load but the rate of crack advance is controlled by a continuing material–environment interaction in the vicinity of the crack tip. In addition to these three basic processes, the growth of a fatigue crack at elevated temperature or in an aggressive environment can produce the synergistic processes of creep-fatigue and corrosion-fatigue.

A common factor in all these stable crack growth processes is the maintenance of a relatively sharp crack tip. This enables the high stresses and strains necessary for local material failure to be sustained at or near the crack tip. Severe blunting of a crack dramatically reduces its effectiveness in propagating a failure. In this regard the tendency to assume that all fabrication defects are crack-like can lead to gross pessimism in life assessment. This paper examines in broad terms the mechanisms and laws controlling crack advance by fatigue, creep and stress corrosion, and also comments on creep-fatigue and corrosion-fatigue crack growth. First, it is necessary to establish the behaviour of sharp cracks in ductile metals under load.

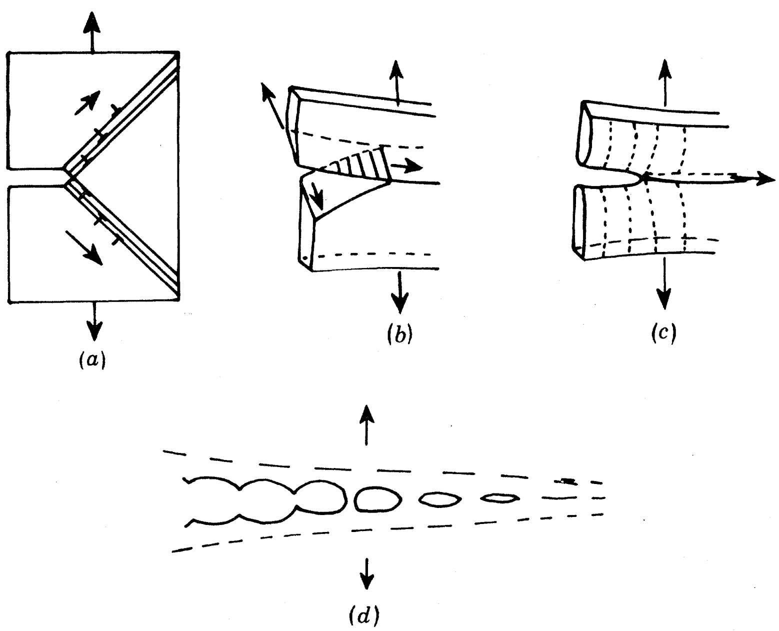

FIGURE 1. Continuous (*a*, *b*, *c*) and discontinuous (*d*) modes of advance of a crack under tensile loading. Mode *a* gives stable crack advance below general yield whereas *b*, *c* and *d* can give unstable advance. After Cottrell (1965).

CHARACTERISTICS OF A CRACK UNDER LOAD

In his Bakerian and Tewksbury lectures, Cottrell (1963, 1965) described in some detail the factors controlling stable and unstable crack advance in nominally ductile metals. Figure 1 summarizes the mechanisms of cracking which he examined for simple elastic–plastic materials. Figure 1*a* shows a crack opened by applied tensile load under plane strain conditions. Crack advance can be thought of as being produced by the injection of dislocations into slip lines at $\pm 45°$ to the crack plane at the tip. New crack surface is produced by 'sliding off' or 'shear decohesion' at the tip and the crack advance step, Δa, is equal to $\frac{1}{2}\delta$, where δ is the crack tip opening. Crack growth by this process is stable and leads to a high local ductility failure with the fracture stress, σ_f, not less than the yield stress, σ_y. If the same cracked material is loaded under plane stress conditions, however, crack advance can be achieved at applied stresses below σ_y. Crack-tip deformation is confined to a local region in the plane of the crack, and crack advance occurs by the continuous operation of a simple array of screw (figure 1*b*) or edge (figure 1*c*) 'dislocations'. Cottrell termed these fractures 'cumulative' because the contribution of a single dislocation to fracture increases continuously with distance travelled by the dislocation. In contrast, the stable crack advance of figure 1*a* he termed 'non-cumulative', as each new amount of growth requires the operation of new dislocations. The plane stress fractures of figure 1*b*, *c* are unstable because the applied stress needed for crack growth decreases with crack advance. Finally, Cottrell noted the observation of crack advance situations where some material cracking occurs ahead of the main crack tip (figure 1*d*). This is observed in plane strain and can precipitate discontinuous unstable fracture at stresses below σ_y. The modes of crack growth shown in figure 1*a*, *d* will be shown later to be important in several time-dependent stable crack growth processes.

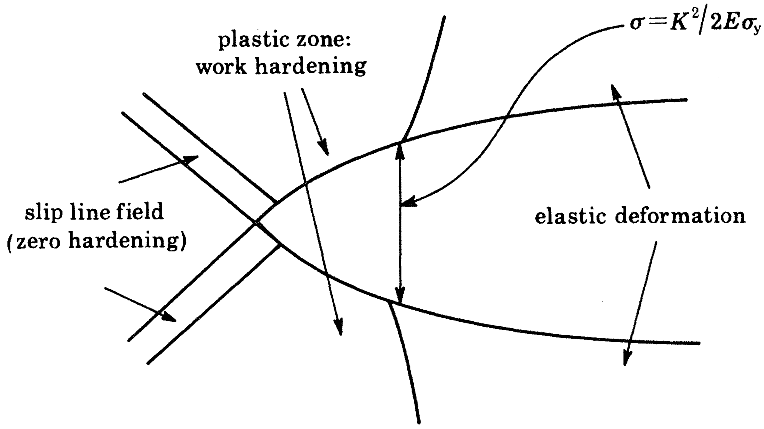

FIGURE 2. Crack tip profile for a realistic material.

The characterization of stresses and strains ahead of a crack and the resulting changes in crack tip profile were a major advance in fracture studies. For a crack in linear elastic material of size W, the stress intensity factor $K(= Y(a/W)\sigma(\pi a)^{\frac{1}{2}})$ emerged as a parameter that controls the strength of the crack tip singularity. More recently, the J contour integral has been shown to characterize crack tip stresses and displacements in nonlinear materials.

For material that obeys the stress–strain law

$$\epsilon/\epsilon_0 = \alpha(\sigma/\sigma_0)^n, \tag{1}$$

stresses and displacements are given by

$$\sigma_{ij} \propto (J/r)^{1/(1+n)}\,\tilde{\sigma}_{ij}(\theta)$$

and

$$u_i \propto (J/r)^{n/(1+n)}\,r\tilde{u}_i(\theta), \tag{2}$$

where

$$J = \alpha\sigma_0\epsilon_0(\sigma/\sigma_0)^{1+n}a\tilde{J}(a/W, n), \tag{3}$$

and r and θ are the polar coordinates related to the crack tip. A typical metal shows behaviour that contains both elastic and power law hardening components before achieving some limiting flow stress value $(\sigma_u) > \sigma_y$. Figure 2 shows schematically a strained crack tip in such a material under plane strain conditions. The simple elastic–plastic crack tip opening occurs at the elastic–work hardening boundary and is given by

$$\delta \approx K^2/2E\sigma_y \tag{4}$$

for the special conditions of $n = \infty$; this crack opening is that created by operation of the limiting slip line field.

Having briefly discussed crack advance modes and stress–strain characterization, we are in a position to examine stable, time-dependent crack growth.

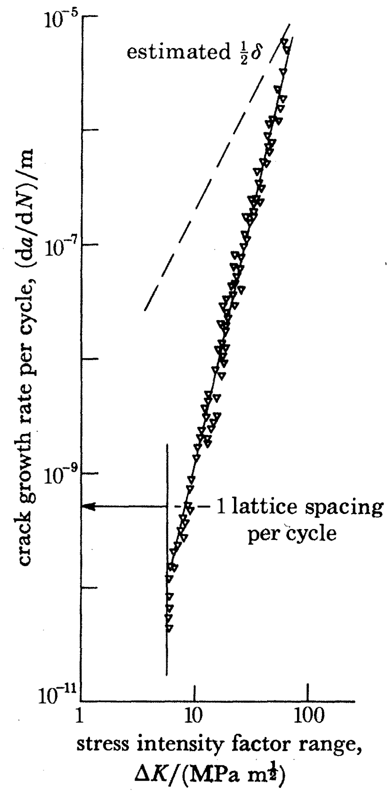

Figure 3. Variation of fatigue crack growth rate with stress intensity factor range for type 316 stainless steel at 22 °C. Data from Lloyd (1976).

Fatigue crack growth

The basic means of crack advance during fatigue, which is predominantly a plane strain phenomenon, is the repeated operation, during each loading cycle, of the process shown in figure 1*a*. The operation of crack tip slip lines results in the well known striation patterns observed on fatigue fracture surfaces for intermediate crack opening displacements (typically 0.1–10 μm). For smaller crack openings only one slip system operates, giving crack growth up one or other of the slip planes. In oxidizing environments, the process is almost irreversible, a net amount of crack advance occurring on each reloading cycle. Figure 3 shows crack growth data in terms of $\mathrm{d}a/\mathrm{d}N$ against ΔK for type 316 stainless steel at 22 °C loaded by zero-tension loading (Lloyd 1976). The first point to note is that crack growth occurs over a very wide range (10^{-10} to 10^{-5} m per cycle). The lower limit to growth seems to be related approximately to an average crack advance of one lattice spacing per cycle, and a definite limiting ΔK for growth exists. This has been termed the threshold (ΔK_{th}). When a tensile mean stress is introduced in the cycle this threshold is reduced. Just above the threshold, crack growth tends to be crystallographic, merging later into the single slip line shear cracking mentioned previously. This has been termed stage I crack growth by Forsyth (1961), although in fracture mechanics terminology it is predominantly by mode II crack opening. For growth rates above *ca.* 10^{-8} m per cycle, both slip lines operate, and crack advance is by the mechanism shown in figure 1*a*. Striations are observed and this growth mode has been termed stage II, although it is essentially mode I crack opening.

Finally, for growth rates above 10^{-5} m per cycle, microfracture occurs ahead of the crack around second phase particles and the potential for operation of the discontinuous fracture mode of figure 1*d* is present. This third stage (III) of fatigue crack growth marks the onset of the approach to final unstable fracture as ΔK tends to the limiting value of K_{Ic}. At this point $a = a_f$. At least up to the onset of stage III, one might expect a good correlation between da/dN and $\frac{1}{2}\delta$ because crack advance is by simple shear decohesion. However, in alloys with significant work hardening capacity, da/dN falls increasingly short of $\frac{1}{2}\delta$ as ΔK decreases. Only at higher values of ΔK does $da/dN = \frac{1}{2}\delta$. An estimate of the $\frac{1}{2}\delta$ line from (4) is shown in figure 3. In fact, the crack growth relation above the threshold follows a power law,

$$da/dN = C\Delta K^m, \tag{5}$$

where the exponent $m \geqslant 2$. This law was first noted by Paris & Erdogan (1963). The discrepancy between da/dN and $\frac{1}{2}\delta$ is probably a result of the contribution of crack flank deformation to δ as defined by (4) and as shown in figure 2. As n tends to infinity one would expect a closer correlation. This is observed in higher strength materials and also as temperature increase leads to an increase in n for a given alloy (Tomkins & Wareing 1977). However, $\frac{1}{2}\delta$ does represent an upper limit to the amount of crack advance by the shear decohesion process.

This mechanism of fatigue crack growth is not confined to long cracks at low stresses where formulations in terms of K are valid. Short fatigue cracks grow under post-yield stresses and applied plastic strain (ϵ_p) in exactly the same way, but the external elastic field is replaced by a general work hardening field. This effective reduction in modulus greatly increases the crack tip opening, which can then be approximated by

$$\delta \approx \frac{K^2}{2E\sigma_u} + \frac{\pi n \sigma \epsilon_p a}{(1+n)\sigma_u} \tag{6}$$

(Tomkins 1975). The plastic component of δ dominates crack opening at high strain levels and its integration with cycles forms the basis of the well known Coffin–Manson law of high strain fatigue. Now as a parameter to characterize cracks in nonlinear stress–strain fields, J can be used instead of K to relate fatigue crack growth in linear elastic and elastic–plastic conditions:

$$J = \beta\delta\sigma_u, \tag{7}$$

where the constant $\beta \approx 2$, so that a plot can be made of ΔJ against da/dN, which will reflect the link between da/dN and $\frac{1}{2}\delta$. Dowling (1976) has used this approach to correlate a wide range of fatigue crack growth data in pressure vessel steels, but the limitations of ΔJ as a unifying parameter are the limitations of δ.

Creep crack growth

When a static load is applied to a cracked section in a ductile metal under plane strain conditions at temperatures below approximately one-third of the melting point T_m, some small amount of crack advance, of order $\frac{1}{2}\delta$, will occur by the mechanism shown in figure 1*a*. In

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addition, if the load is high enough, some additional amount of discontinuous advance (figure 1*d*) may occur by ductile tearing. However, at temperatures above 0.3 T_m, crack advance may occur by discontinuous growth below this tearing limit. This has been termed creep crack growth and the fracture damage ahead of the crack tip is occasioned by time-dependent creep processes in the material. It most often takes the form of discontinuous grain boundary cavitation or cracking, giving a localized failure path for crack advance. As with lower temperature tearing, the ratio of crack advance, Δa, to crack opening increment, $\Delta\delta$, varies considerably with material, temperature and applied stress level. In very creep ductile materials, $\Delta a \approx \frac{1}{2}\Delta\delta$, while for creep brittle materials ratios of $\Delta a/\Delta\delta$ of up to 14 have been observed (Gooch *et al.* 1978). While a constant finite ratio $\Delta a/\Delta\delta$ is maintained, crack advance is stable.

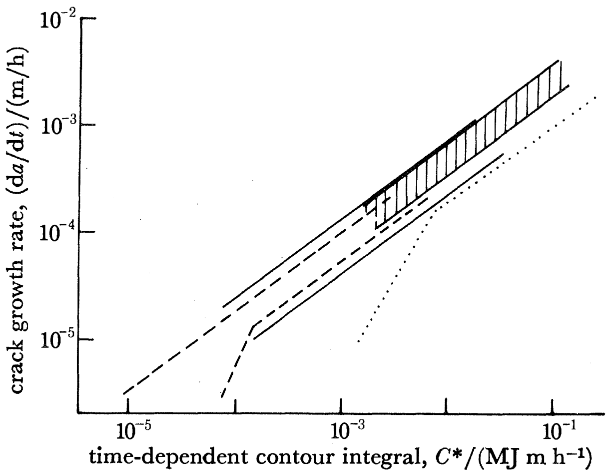

FIGURE 4. Summary of results of creep crack growth rate against C^*. Solid lines, compact tension; broken lines, single edge notch bend for 1 Cr-Mo-V steel at 565 °C; hatching, compact tension; dotted line, centre crack panel for Discaloy at 650 °C. After Harper and Ellison (1977).

Creep crack growth can occur under a steady applied load P or increasing displacement Δ applied to the cracked section. For material undergoing power law creep, strain, ϵ, is replaced by strain rate, $\dot{\epsilon}$, in (1) and the stress–strain field ahead of a crack is given by the time-dependent version of (2). A time-dependent J integral, C^*, has the form shown in (3). Goldman & Hutchinson (1975) have shown that the crack and overall displacements are then proportional to C^*/P, so for a constant load test, one might expect the crack increment rate, $\mathrm{d}a/\mathrm{d}t$, to be proportional to C^*. This assumes that even though unloading of material occurs on crack advance, thus invalidating stress–strain descriptions in terms of C^* just at the tip, C^* will still be characterizing the cavitation conditions further ahead. A plot of $\mathrm{d}a/\mathrm{d}t$ against C^* is shown in figure 4 for data on 1 CrMoV steel and Discaloy collated by Harper & Ellison (1977). C^* seems to correlate crack growth data from various specimen geometries, but more work is needed to demonstrate that C^* is a general correlating parameter for creep crack growth. The change in slope at lower growth rates is probably due to stress redistribution during early parts of the test, but a threshold for creep cracking analogous to that for fatigue must exist. Such a true threshold would be related to either the development of creep fracture damage ahead of the crack or the achievement of a large enough crack tip opening increment to achieve linkage. In either case, it is effectively a threshold determined by creep deformation. After early attempts to use K

or net section stress to correlate creep crack growth, it was clear that, for creep ductile materials, if a simple reference stress, σ_d, ahead of the crack could be estimated, it could be used to correlate crack growth data (Haigh & Richards 1973). Creep crack growth should then relate to σ_d^n and this was indeed found to rationalize data that earlier had shown some limited correlation with K or net section stress, σ_n, raised to the power n (Neate & Siverns 1973). More recently, Freeman & Neate (1978) have suggested that if a critical value of reference stress, σ_{cr}, operating in the region ahead of the crack tip, where cavity formation and linkage is occurring, can be estimated, it can relate failure by creep crack growth to a simple stress rupture plot. This approach suggests that for engineering applications crack growth itself can be ignored and structural failure determined by the effect of the crack in providing a local higher creep stress condition. Certainly the estimation of C^* for structures is difficult, so perhaps the reference stress approach offers the most simple method for engineers to approach the effect of cracks in creep. Elevated temperature fracture mechanics has recently been reviewed by Ashby & Tomkins (1979).

Finally, it should be noted that creep crack growth involves a considerable increase in overall displacement, Δ, for ductile materials. In structures, large displacements are difficult to accommodate and creep crack growth should be considered in the same way as ductile tearing from a structural point of view.

Stress corrosion crack growth

Even at temperatures well below 0.3 T_m, cracks can grow under a static load if the local environment is sufficiently aggressive. In aqueous environments, electrochemical activity at the crack tip can maintain a sharp crack profile while producing an increment of crack advance in conjunction with an applied tensile stress. Electrochemically, the important thing is to maintain a critical balance between active and passive behaviour. Under tensile load, new material surface is exposed at the crack tip by the decohesion process of figure 1*a* and in an active environment this region becomes anodic relative to the adjacent crack flanks, which act cathodically. During crack advance, as crack tip material becomes crack flank material it can change from an active to a passive state. The required critical rate of this active–passive transition can be determined from potentiodynamic polarization experiments, in which the variation in anodic current density, i_a, with potential, E, is observed for different potential sweep rates, dE/dt (Sutcliffe *et al.* 1972). This electrochemical condition for stress corrosion cracking (s.c.c.) requires a rapid variation in i_a as the potential sweep rate varies. Adjacent to the s.c.c. region, if i_a remains high, even though there is little variation with dE/dt, local pitting or general dissolution rather than crack growth would occur, i.e. the crack flanks would remain active. It now appears that s.c.c. can occur over a wide range of material–environment conditions if the critical electrochemical balance is achieved.

The second necessary condition for crack growth to be maintained is an adequate renewal rate of fresh surface at the crack tip. This can be thought of as an adequate crack tip strain rate, $\dot{\epsilon}_t$, although the actual surface exposed must be localized. This can be either a specific slip step or an electrochemically weak path such as a grain boundary, e.g. in sensitized stainless steel where chromium depletion has occurred. The large slip steps generated in low stacking fault energy materials make them particularly susceptible to slip-dissolution controlled transgranular s.c.c. In both selective corrosion and slip-dissolution, it is likely that slip activity is needed to break or prevent the formation of oxide films, which form under passivation. A final factor for sustained s.c.c. is the provision of adequate renewal of the crack tip solution. This type of stress corrosion crack growth is continuous in the sense that no cracking occurs ahead of the main crack tip, but it is probably discontinuous in terms of da/dt.

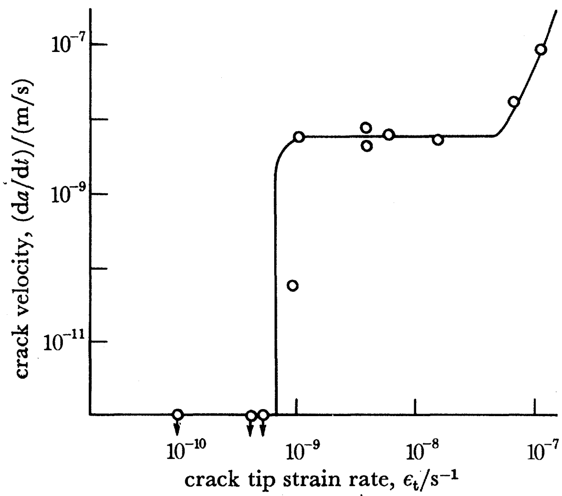

FIGURE 5. Intergranular crack velocities for various applied crack tip strain rates in C–Mn steel immersed in CO_3–HCO_3 solution at 75 °C and -650 mV s.c.e. After Parkins (1979).

If the crack tip strain rate is raised by varying the applied load, P, or strain rate, $\dot{\epsilon}$, a characteristic relation between da/dt and $\dot{\epsilon}_t$ can be observed (figure 5). A threshold for crack growth exists at low $\dot{\epsilon}_t$ where fresh surface is not exposed quickly enough at the crack tip. However, above this value, the crack velocity rises rapidly to a plateau value, $(da/dt)_{II}$ (the II designates a second stage of crack growth), which is independent of $\dot{\epsilon}_t$. Finally, at a large enough value of $\dot{\epsilon}_t$ the crack growth rate rises rapidly, indicating an instability condition. For steady loading, the threshold and final condition can be defined in terms of an applied K level, designated $K_{Is.c.c.}$ and K_{Ic}. The plateau velocity obviously reflects corrosion controlled growth and can be expressed in terms of Faraday's second law,

$$(da/dt)_{II} = i_a\, M/zF\rho, \tag{7}$$

where M is atomic mass, z is valency, ρ is density and F is Faraday's constant; i_a is the effective current density which may depend on the passivation rate and oxide rupture rate (Ford 1979). This influence on i_a can lead to a loss of strain rate independence and also give a variation in velocity with water oxygen content.

In addition to the anodic dissolution reaction, hydrogen is released by the cathodic reaction and is readily adsorbed in atomic form on surfaces in the tip region, followed by absorption and diffusion into material ahead of the crack tip. This hydrogen can play a variety of roles in crack advance (Bernstein & Thompson 1977), one of which is in crack formation ahead of the tip following the clustering of H atoms to a critical size (Raj & Varadan 1977), followed by ductile ligament rupture between the hydrogen-induced crack and the tip. This process gives a discontinuous crack advance mechanism (cf. figure 1*d*). Hydrogen cracking of this type obviously requires a critical amount of hydrogen and a high enough local stress in the crack tip region. High strength alloys at ambient temperatures are most susceptible. Hydrogen can often lead to intergranular cracking as grain boundaries provide a clear diffusion path and the

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presence of hydrogen can lower the effective surface energy (Howard 1979). In the presence of temper embrittling trace elements, hydrogen can produce a synergistic effect (Banerji *et al.* 1978). Even if hydrogen is responsible for the crack advance step Δa, the time for this advance, Δt, can be controlled by the initial anodic reaction or oxide film rupture. Thus hydrogen-produced and dissolution-produced s.c.c. both tend to show a $\mathrm{d}a/\mathrm{d}t$–$\dot{\epsilon}_t$ relation, or K curve, of the form given in figure 5.

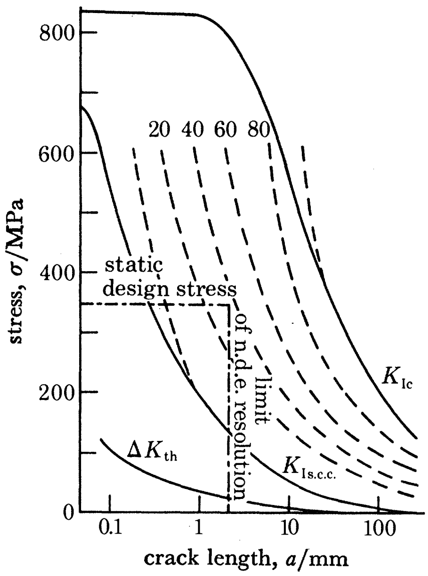

FIGURE 6. Stress–crack length diagram for steam turbine disk/rotor NiCrMoV and CrMo steels in water at 110 °C. After Ford (1975).

It is worth noting that the threshold K value for s.c.c. and threshold ΔK for fatigue can be very low for engineering alloys. Ford (1975) has shown how design stress levels and tolerable defect sizes relate to the value of K for s.c.c. in steam turbine disk/rotor steels in water at 110 °C. The essential features of his plot are shown in figure 6 along with estimates of fatigue threshold values. In the latter case a low value of 2 MPa $\mathrm{m}^{\frac{1}{2}}$ is taken for ΔK_{th}, consistent with a high mean stress situation. This plot illustrates the dilemma of the engineer in estimating subcritical crack growth in his components. If pre-existing defects are considered as cracks, current n.d.e. techniques often cannot guarantee defect sizes below threshold at design stress levels. Two points should be made. First, many pre-existing defects are not cracks and the life of many components and structures is bound up in the time or cycles taken to initiate cracks from defects. Secondly, estimates of acceptable amounts of crack growth linked to in-service inspection can now be made by using fracture mechanics information on subcritical crack growth rate.

Corrosion fatigue crack growth

As discussed earlier, the crack advance process by fatigue involves the repeated creation of a small amount of new crack surface each cycle. If this process operates in a corrosive environment at applied strain rates above the stress corrosion threshold, some cyclic s.c.c. can occur. On this basis, Wei & Landes (1969) argued a simple superposition of fatigue and s.c.c. for the prediction of corrosion fatigue crack growth rates. However, more recent work, particularly in lower strength structural steels not normally susceptible to s.c.c., has shown that a dramatic enhancement of fatigue crack growth rates can occur due to corrosion effects. Two factors are important: (*a*) the cycling of crack tip material can effectively increase its strain rate (cyclic creep), thus suppressing $K_{\mathrm{Is.c.c.}}$ (Parkins & Greenwell 1977), and (*b*) by mechanically renewing crack tip surface each cycle, the film rupture step is automatically achieved and a small amount of dissolution controlled growth can occur. In this latter case, the small amount of growth can be much larger than the Δa due to fatigue. As the whole of the crack tip region defined by δ is

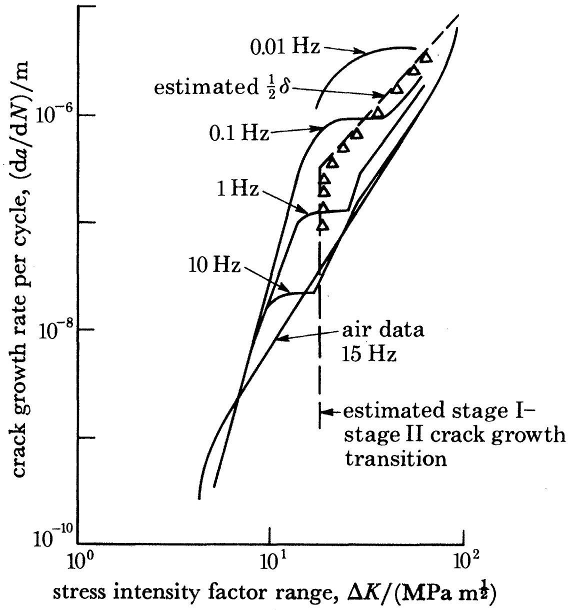

FIGURE 7. Fatigue crack growth data for A533–B steel (△) in KOH solution, pH = 9.5, $R = 0$ (after Tomkins 1977); and curves for a C–Mn pipeline steel tested in a 3.5% NaCl solution at applied potential of –1.04 V, $R = 0.2$ (after Vosikovsky 1975).

electrochemically active, in the absence of a weak dissolution path, the local crack tip geometry can be modified by corrosion to give an effective growth rate up to $\frac{1}{2}\delta$. Figure 7 shows corrosion fatigue data for a pressure vessel steel in distilled water doped with KOH. For stage II fatigue crack opening, the crack growth rate achieves a level of $\frac{1}{2}\delta$. At lower cyclic frequencies (i.e. lower $\dot{\epsilon}_t$), excessive dissolution can lead to crack blunting above $\frac{1}{2}\delta$ with consequent reduction in $\mathrm{d}a/\mathrm{d}N$ (Atkinson & Lindley 1977). At the stage I–stage II fatigue crack growth transition, solution access becomes restricted and effectively produces a corrosion fatigue threshold for dissolution controlled growth. High mean stresses, reflected by the ratio $R = K_{\min}/K_{\max}$, suppress this transition in terms of ΔK (Tomkins 1977). Under these conditions, fatigue crack growth is controlled by $\Delta\delta$ but the exposure of a much larger δ to the environment can give corrosion fatigue effects on crack growth of more than two orders of magnitude for high R cycling (Bamford 1977).

Under conditions where hydrogen evolution is encouraged at the crack tip (low pH, H_2S, cathodic polarization) local fatigue crack growth enhancement due to hydrogen embrittlement has been observed in low strength materials not susceptible to such embrittlement under static loads (Vosikovsky 1975; Johnson *et al.* 1978). This discontinuous corrosion fatigue crack growth can give dramatic increases in crack growth rate that do not saturate readily with cyclic frequency. Some data are included in figure 7 for a pipeline steel of similar strength to A533-B tested in NaCl solution under cathodic polarization of -1.04 V. As the true fatigue threshold ΔK_{th} is related to the limit for exposure of new crack surface at the tip it also seems to represent a threshold for corrosion fatigue crack growth even when hydrogen cracking is involved.

From an engineering viewpoint, corrosion fatigue presents a problem in that it erodes traditional methods of fatigue data collection because it is most damaging at low cyclic frequencies. For dissolution-enhanced crack growth, the crack tip opening represents an upper limit below $K_{Is.c.c.}$, but for hydrogen-assisted growth it is difficult to define the limit to growth. Also for hydrogen, Wei (1979) has shown that long dwell periods between cycles can be very damaging on account of the incubation time that they provide for cracking ahead of the tip. Finally, however, because crack tip profiles can be easily modified by corrosion, below $K_{Is.c.c.}$ corrosion fatigue crack growth is often difficult to maintain under load variations.

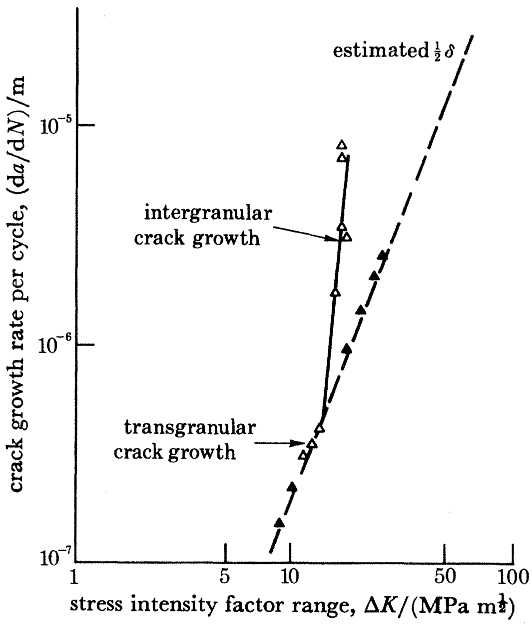

FIGURE 8. Effect of creep on fatigue crack growth in type 316 stainless steel at 625 °C, $R = 0.6$. ▲, sawtooth cycle, frequency 10^{-2} Hz; △, tensile dwell cycle, frequency 10^{-4} Hz. After Lloyd & Wareing (1979).

Creep fatigue crack growth

In considering creep crack growth, it was seen that crack advance is a discontinuous process caused by cavity or crack formation ahead of the main crack tip. Under slow cycling conditions, particularly in low creep ductility materials, increments of creep crack growth can occur cycle by cycle. This is most likely where a simple tearing condition for crack advance exists such that half the crack opening displacement is of the order of the cavity spacing or crack ligament $(\lambda - p)$, where p and λ are the cavity size and spacing respectively (Tomkins 1975). Thus if

$$\tfrac{1}{2}\delta \geqslant (\lambda - p), \tag{8}$$

crack advance can occur. Large δ are most often achieved in high strain fatigue, which is where cyclic creep crack growth was first observed (Solomon & Coffin 1972). More recently it has been observed in low stress crack growth (Ohmura *et al.* 1973).

Practical high temperature situations are more likely to involve load–strain-time histories with higher rate cycles interspersed with load or strain dwell periods. The latter can induce creep cavitation in the bulk or crack tip plastic zone. If the condition given by (8) is then achieved during an intermittent tensile cyclic load, rapid discontinuous crack advance can occur (Tomkins & Wareing 1977) up to the order of the plastic zone size (Tomkins & Ashby 1979). Figure 8 shows the onset of discontinuous growth in type 316 stainless steel tested at 625 °C. Transgranular crack growth is observed for both the sawtooth and dwell cycle at a rate of $\frac{1}{2}\delta$ (this rate is approached at elevated temperature in fatigue because $n \rightarrow \infty$) at low levels of ΔK. At a transition point given by (8), crack growth accelerates and takes an intergranular crack path.

Concluding remarks

Subcritical crack growth can occur in several ways, producing a wide range of growth rates in terms of cycles or time. In simple fatigue, $\frac{1}{2}\delta$ represents a clearly defined upper limit for growth by the shear decohesion process, although this limit is only achieved in real materials whose work hardening capacity is very small. In dissolution-enhanced corrosion fatigue this limit can be achieved by the modification of the local crack tip geometry by corrosion. Subcritical discontinuous crack growth can occur under static loading conditions in the creep range and for hydrogen-controlled stress corrosion cracking. In the creep range, the achievement of a critical crack opening condition for crack advance related to ligament rupture is required and must be incrementally sustained. This is probably also true for hydrogen-assisted growth in many circumstances. The same ligament rupture condition also determines the critical rate change in creep-fatigue crack growth where a transition from continuous to discontinuous crack growth is achieved. For dissolution-controlled stress corrosion cracking, a critical electrochemical condition must be achieved after which the crack growth rate is under strict electrochemical control. Even though in terms of time the growth is discontinuous, it does not involve material failure ahead of the crack tip.

The stability of subcritical cracks is maintained by the fact that they are essentially displacement controlled at the crack tip. Local discontinuous crack advance can occur but only within a strict limit. As the microcracking ahead of the tip is an essential feature of final unstable, critical crack growth, the achievement of the critical condition is simply related to a loss of limitation on the crack tip displacement increment. This is clearly reflected in the J–R curve.

For the engineer, subcritical crack growth can occupy much of the life of his component. In such cases, whether aware of it or not, he is 'living with defects'. In this respect the presence of discontinuous subcritical crack growth in creep, stress corrosion, corrosion fatigue and creep fatigue is disturbing, and knowledge of its onset in terms of material and applied stress–strain parameters may prove to be as important as knowing the final critical crack condition.

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Discussion

H. K. Grover (*ERA Technology Ltd, Leatherhead, U.K.*). It is generally accepted that the present state of development of fracture mechanics analyses has enabled more efficient use of materials in wrought condition. However, it is fabrication by welding that introduces defects, and all of us professionals, physicists, metallurgists, engineers, designers, etc., have to put up with their consequences, i.e. initiation of cracks from these defects which may be due to microstructural inhomogeneity or mechanical stress raisers.

Dr Tomkins has stated that designers do not always allow for the presence of defects in their calculations. Perhaps one of the reasons is a lack of reliable data on behaviour of defects under service conditions, e.g. creep.

Reproducibility of creep crack growth rates is not always good, and there can be significant differences in results from different laboratories on material of very similar composition. Added to this is the problem of applying the results to components in service. These difficulties have already led people to use equivalent stress or reference stress techniques and relate these to materials' known stress–rupture properties, the scatter in these properties being within known limits.

It is clear from one of the ship's failures discussed by Mr Smedley that cracks can initiate and propagate completely within the weld zone for considerable distances. Therefore, it is incumbent on metallurgists to improve the techniques used for measuring crack growth rates so that data produced on material in different metallurgical conditions can be confidently used in design calculations.

T. G. F. Gray (*Department of Mechanics of Materials, University of Strathclyde, Glasgow, U.K.*). Would Dr Tomkins comment on the use that we make of theoretical analyses that are strictly for stationary or non-propagating cracks, in describing the moving cracks found in fatigue or, more importantly, in creep and stress corrosion? To put the question another way, does Dr Tomkins think that the 'monotonic loading' type of analysis adequately describes situations where the crack tip stress distributions may be modified by 'screening' or 'wake' effects in quasi-static propagation.

B. Tomkins. With regard to the use of stationary crack analyses in crack propagation problems, a distinction should be made between continuous and discontinuous crack advance. In stage I and stage II fatigue, where continuous crack advance is a fraction of the crack tip opening, stationary analyses give a good estimate of the observed opening, although account should be taken of the effect of previously worked material along the crack flanks, which will modify the crack tip stress–strain field. The strong crack closure effect of a 'wake' of plastic deformation in plane stress is well known. For discontinuous crack growth (tearing as in stage III fatigue, creep crack growth and stress corrosion cracking by hydrogen embrittlement), however, the stationary crack field is of value in defining the crack advance condition but not necessarily the crack advance step. This means that the current use of static crack parameters such as J in defining crack advance Δa is at present empirical. That good correlations are found between J and Δa is probably a result of the fact that Δa is often small in comparison with a and hence the range of the J-dominated field. Unloading of a small portion of the field on crack advance does not seriously affect this field ahead of the tip as the crack advances.

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