Near-threshold fatigue-crack propagation in steels

by R. O. Ritchie

The characteristics of fatigue-crack propagation in metals and alloys have been the subject of several extensive reviews in recent years, but in very few instances have the details of ultralow growth rate, near-threshold fatigue-crack propagation been similarly discussed. In this review the effects are examined of various mechanical, microstructural, and environmental factors which influence fatigue-crack propagation in steels at growth rates less than 10^{-6} mm/cycle, where the alternating stress intensity ΔK approaches the so-called threshold stress intensity ΔK_0, below which crack growth cannot be experimentally detected. The marked influences of load ratio, material strength, and microstructure on such near-threshold growth are analysed in detail and rationalized in terms of possible environmental contributions and crack-closure concepts. These effects are contrasted with crack-propagation behaviour in other engineering materials and at higher growth rates.

LIST OF SYMBOLS

a = crack length

a_0 = constant characteristic of material/material condition in expression for ΔK, equation (3)

da/dN = fatigue-crack propagation rate per cycle

A = constant in Paris power law for fatigue-crack growth, equation (1)

B, B' = constants dependent upon material and temperature in expression for ΔK_0, equation (13)

C = constant in expression for ΔK_0 in absence of environment, equation (10)

C_H = local concentration of hydrogen at point of maximum dilatation

C_0 = equilibrium concentration of hydrogen in unstressed lattice

ΔCOD = cyclic crack-tip opening displacement

E = elastic modulus

K_c, K_{Ic} = stress intensities at final failure (fracture toughness)

K_{max}, K_{min} = maximum and minimum stress intensities during cycle

ΔK = alternating stress intensity ($K_{max} - K_{min}$)

ΔK_{eff} = effective alternating stress intensity taking into account crack closure

ΔK_0 = threshold stress intensity for no crack growth

m = exponent in Paris power law, equation (1)

N = number of cycles

R = load or stress ratio (K_{min}/K_{max})

R_0 = gas constant

T = absolute temperature

$\bar{v}$ = partial molar volume of hydrogen in iron (2 cm^3 mol^{-1})

α = constant relating reduction in cohesive strength to local hydrogen concentration C_H

ρ' = Neuber's effective crack-tip radius

ρ^* = limiting microstructural dimension at threshold

$\bar{\sigma}$ = hydrostatic stress

σ_F = critical local fracture stress at threshold

σ_u = ultimate tensile strength (UTS)

σ_y = monotonic yield strength

σ_y' = cyclic yield strength

$\Delta\sigma$ = alternating stress

$\Delta\sigma_e$ = fatigue limit or endurance strength

$\Delta\sigma_H$ = reduction in cohesive strength due to hydrogen

$\Delta\sigma_{TH}$ = alternating threshold stress for no crack growth

R. O. Ritchie, MA, PhD, MIM, CEng, is in the Department of Mechanical Engineering, Massachusetts Institute of Technology, Cambridge, Mass., USA.

In many safety-critical engineering applications, where design against fatigue is based on the so-called defect-tolerant approach, the prime consideration in dictating the lifetime of a component is taken as the time or number of cycles to propagate a subcritical crack from an assumed initial defect size (often taken as the largest undetected flaw) to critical size where final failure occurs either catastrophically or at the limit load. The widespread adoption of this approach over the last decade for design against cyclic loading has spurred a multitude of investigators to characterize the rate of growth of fatigue cracks in the majority of engineering materials as a function of such variables as microstructure, mean stress, environment, frequency, and so forth. The majority of these data, however, have been generated for growth rates typically in excess of 10^{-6} mm/cycle. Although this information is vital for many structural engineering applications (i.e. determining safe non-destructive inspection intervals in aircraft), in recent years there has been a rapidly increasing need for fatigue-crack propagation data pertaining to extremely low growth rates (less than 10^{-6} mm/cycle) where the alternating stress intensity ΔK approaches a so-called threshold value ΔK_0, below which cracks remain dormant or grow at undectable rates. The reasons for this are several. First, there are fundamental aspects in that this topic represents a comparatively unexplored area of fatigue research compared to the vast amount of information on fatigue-crack propagation at higher growth rates. Essentially, little is known from a mechanics or metallurgical point of view about micromechanisms of crack propagation at near-threshold growth rates, and furthermore, there is still a substantial lack of reliable engineering data. Secondly, there is the practical concern of structural and materials engineers to design components which can withstand extremely high frequency, low-amplitude loadings for lifetimes in the range 10^{10}–10^{12} cycles. A high-speed rotor operating at 3000 rev min^{-1} in a steam turbine, for example, may be expected to see 10^{10} cycles of stress over a typical lifetime of 20 years. Should fatigue-crack propagation be occurring at the seemingly insignificant near-threshold growth rate of 3×10^{-9} mm/cycle, this would still represent a total growth of 30 mm during the life of the rotor, which could clearly result in catastrophic failure. In fact, a knowledge of low growth rate, fatigue-crack propagation data and, in particular, information regarding the existence of a threshold stress intensity, has been shown to be essential in the analysis of such problems as cracking in turbine blades,[1] turbine shafts,[2,3] and alternator rotors,[3] and acoustic fatigue of welds in gas circuitry in nuclear-reactor systems.[2–4]

There now exist results in the literature on near-threshold fatigue-crack growth for a wide range of materials, including ferritic,[5–20] martensitic,[5,19–29] and austenitic,[19,30–32] steels, titanium[5,33,34] and titanium alloys,[34–41] and several aluminium,[5,19,22,41–43] copper,[5,44] and nickel,[1,5] alloys. It is apparent from such data that near-threshold crack-growth rates and the value of the threshold ΔK_0 are particularly sensitive to several mechanical and microstructural variables, namely, mean stress or load ratio,[5,28,31–44] prior stress history,[9,19] crack size,[45] cyclic frequency,[22,32,42] monotonic[11,16,18,24] and cyclic[26,27,43,44] strength, grain size,[16,18,26,27,29,33,36] and grain-boundary composition.[25] However, the influence of environmental factors on near-threshold crack growth has remained somewhat of a controversy. Although it is well documented that fatigue-crack propagation at rates exceeding 10^{-5} mm/cycle is generally accelerated in the presence of an environment compared to inert conditions, initial studies by Paris and co-workers, in low-strength steels[6,7] and titanium alloys[35] in air, water, hydrogen, and dry argon, did not reveal any such environmentally-sensitive propagation at ultralow growth rates. Furthermore, by extrapolation of higher (midrange) growth-rate data to near-threshold rates (a procedure, incidentally, which should be regarded with extreme caution), results of certain authors[46,47] suggest that near-threshold crack propagation may indeed be *decelerated* in the presence of an 'aggressive' environment (e.g. salt water) compared to seemingly more inert environments such as air. Other data,[21,22,31,39–41] however, for various materials tested *in vacuo* suggest the contrary behaviour of an environmental contribution to cracking at near-threshold rates. Conclusive information to resolve these issues is still lacking.

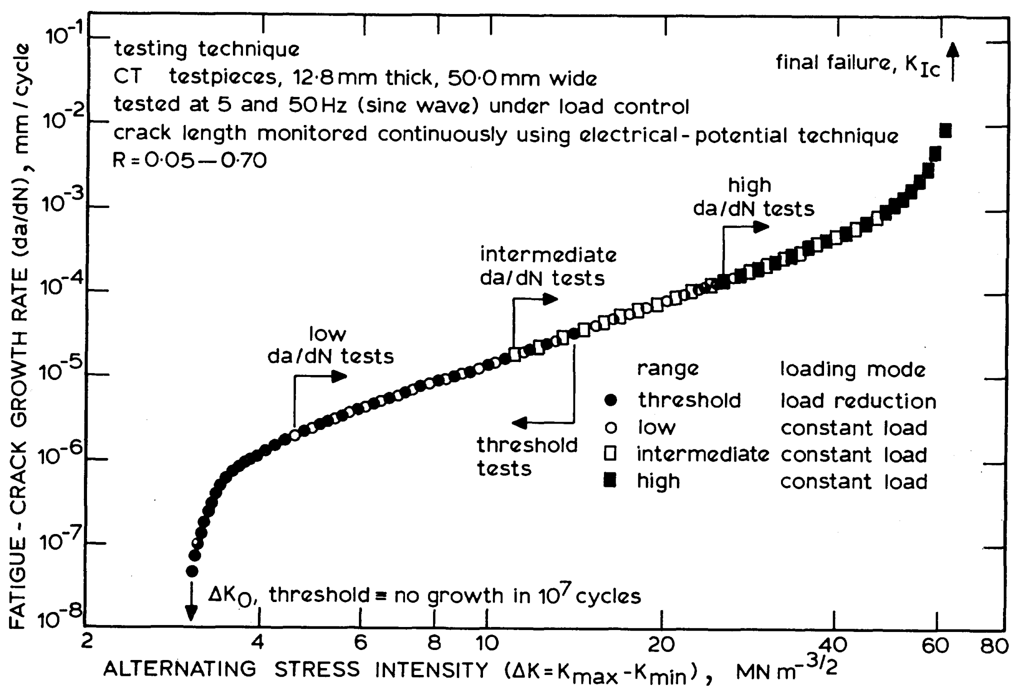
In design, the concept of a fatigue-crack propagation threshold is still infrequently utilized, except where there is an obvious source of very high frequency loading,[1–4] since it is often considered to be overly conservative. However, there has been a recent growing awareness in the nuclear industry that, for design and continued safe operation of reactor plants, such near-threshold data are required for both pressure-vessel and reactor-coolant piping steels (e.g. A533B, A508 low-alloy steels, and 304 and 316 stainless steels, respectively) tested in simulated reactor environments. Such data, measured in air, pressurized-water, and boiling-water reactor environments, are now appearing in the literature.[48,49] Similar data on candidate pressure-vessel steels for coal-gasifier and coal-liquefaction use, e.g. 2.25Cr–1Mo steels, in simulated environments, however, are not available, although near-threshold fatigue-crack growth is a possibility in the pressure-vessel wall as a result of high-cycle low-amplitude stresses generated from small pressure and temperature fluctuations of the system. Such 'operational transients' could arise from real variations in coal feedrates, 'clusters' of pulverized coal, combustion instabilities, and pressure fluctuations in feed and discharge lines for gases. Furthermore, typical atmospheres in coal-conversion pressure vessels, namely, mixtures of water vapour, hydrogen, hydrogen sulphide, carbon monoxide, carbon dioxide, ammonia, methane, and other hydrocarbons at pressures exceeding 10 MPa with metal-wall temperatures around 350°C (Ref. 50), may well accelerate this slow subcritical growth.

It is the purpose in the present paper to provide a critical review of existing information on near-threshold fatigue-crack propagation in steels, and, in particular, to discuss these data in the light of possible environmental interactions and crack-closure concepts. Where possible, effects are contrasted with crack-propagation behaviour at higher growth rates (exceeding 10^{-6} mm/cycle) and in other engineering materials.

EXPERIMENTAL MEASUREMENT OF NEAR-THRESHOLD GROWTH

It is pertinent at this stage to examine how such low growth rates are measured experimentally and, in particular, how the value of the threshold stress intensity ΔK_0 can be defined. Whereas fatigue-crack propagation at conventional growth rates (i.e. greater than around 10^{-5} mm/cycle) is typically measured under constant load (increasing stress intensity K) conditions, it is generally more realistic to measure near-threshold growth rates, and the value of the threshold ΔK_0 under decreasing K conditions, in order to minimize transient residual-stress effects.

In strict terms, the threshold stress intensity ΔK_0 should represent the alternating stress intensity where the growth rate is infinitesimal. However, for the purposes of practical measurement it is more useful to adopt an operational definition for ΔK_0. This is best achieved in terms of a maximum growth rate, calculated from the accuracy of the crack monitoring technique and the number of cycles elapsed.[21] In the system used by the present author,[24–27] crack length is continuously monitored using the dc electrical-potential technique,[51–53] and the threshold ΔK_0 is computed from the highest stress intensity at which no growth can be detected within 10^7 cycles. In this particular case, the crack monitoring technique is at least accurate to 0.1 mm on absolute crack length such that the threshold can be defined in terms of a maximum growth rate of 10^{-8} mm/cycle. Threshold levels are approached using a load-shedding technique involving a procedure of successive load reduction followed by crack growth. Measurements of crack growth rate are taken at each load level, over increments of 1–1.5 mm increase in crack length, after which the load is reduced by not more than 10%, and the same procedure followed. Larger reductions in load are liable to give premature crack arrest from retardation effects owing to residual plastic deformation. The increments, over which measurements of growth rate are taken, should represent distances at least four times larger than the maximum plastic zone size generated at the previous (higher) load level to minimize these retardation effects caused by change in load. Furthermore, frequency must be maintained constant during this procedure since significant environmentally-induced transient crack-growth rate effects can result from variations in cyclic frequency.[54] Following ΔK_0 measurement, the load may be increased in increments and a similar procedure adopted to measure growth rates. In this way, near-threshold data can be monitored under both decreasing K and increasing K conditions in the same specimen. Provided care is taken to minimize transient effects caused



1 **Typical test procedures for obtaining fatigue-crack propagation data spanning entire range of growth rates from threshold levels to final failure**

by changing the load, growth-rate data should be identical when measured by these two procedures.

A typical test procedure for obtaining fatigue-crack propagation-rate data spanning the entire range of growth rates from threshold levels to final failure is shown schematically in Fig. 1.

Other techniques have been used to measure the threshold. These include determining an *S/N* curve for cracked specimens with lifetime plotted against the initial value of ΔK, rather than stress,[5,17,20] and a decreasing K technique achieved by cycling under constant deflection control.[42,55] Although relatively simple to instrument, for conventional compact and centre-cracked tension specimens, the latter technique suffers from the fact that the decrease in K is not rapid enough for efficient near-threshold crack-growth measurement.[55]

The load-shedding (decreasing K) technique can be easily automated using a suitable crack monitoring technique and computer-controlled closed-loop testing machines. Such systems, utilizing a programmed constant decrease of the normalized K-gradient, i.e. $|\Delta K^{-1} . \mathrm{d}\Delta K/\mathrm{d}a|$ or $|K_{max}^{-1} . \mathrm{d}K_{max}/\mathrm{d}a|$, have been developed using several crack measurement techniques, namely, crack-opening displacement monitoring,[56] electrical-potential methods,[30] eddy-current crack following,[28] and elastic-compliance techniques.[57]

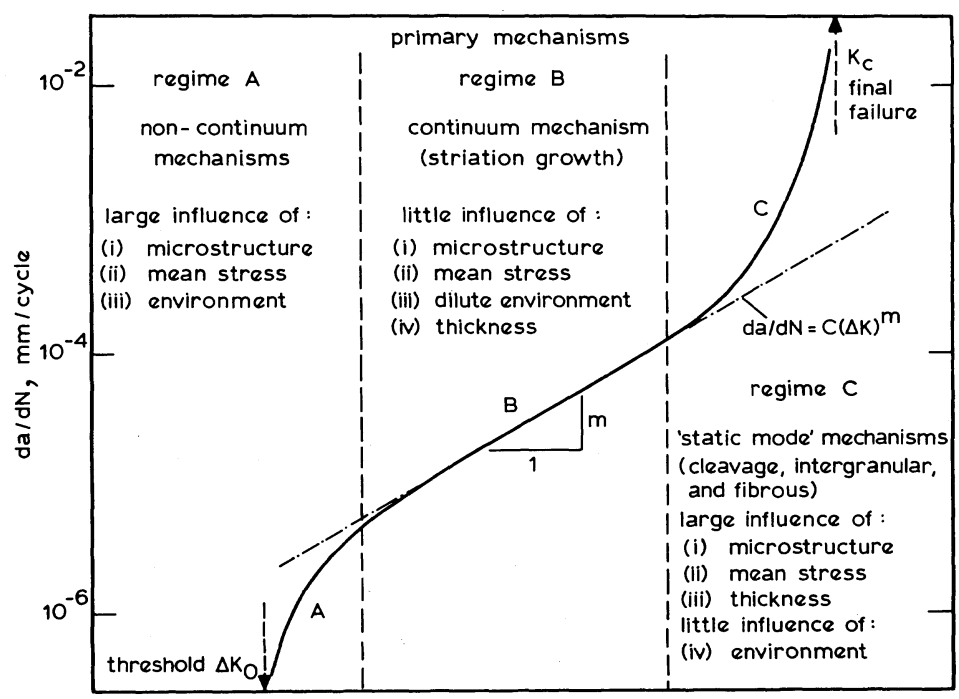
To provide some consistency in the measurement of near-threshold crack-growth rates and ΔK_0 values, the American Society for Testing and Materials E. 24. 04 subcommittee is presently developing a standard for proposed test methods similar to the procedures described above. Such guidelines for the establishment of fatigue-crack growth rates below 10^{-5} mm/cycle are likely to be incorporated as modifications to the recently proposed ASTM Standard E647–78T for measurement of constant-load-amplitude growth rates above 10^{-5} mm/cycle.

GENERAL NATURE OF FATIGUE-CRACK PROPAGATION IN STEELS

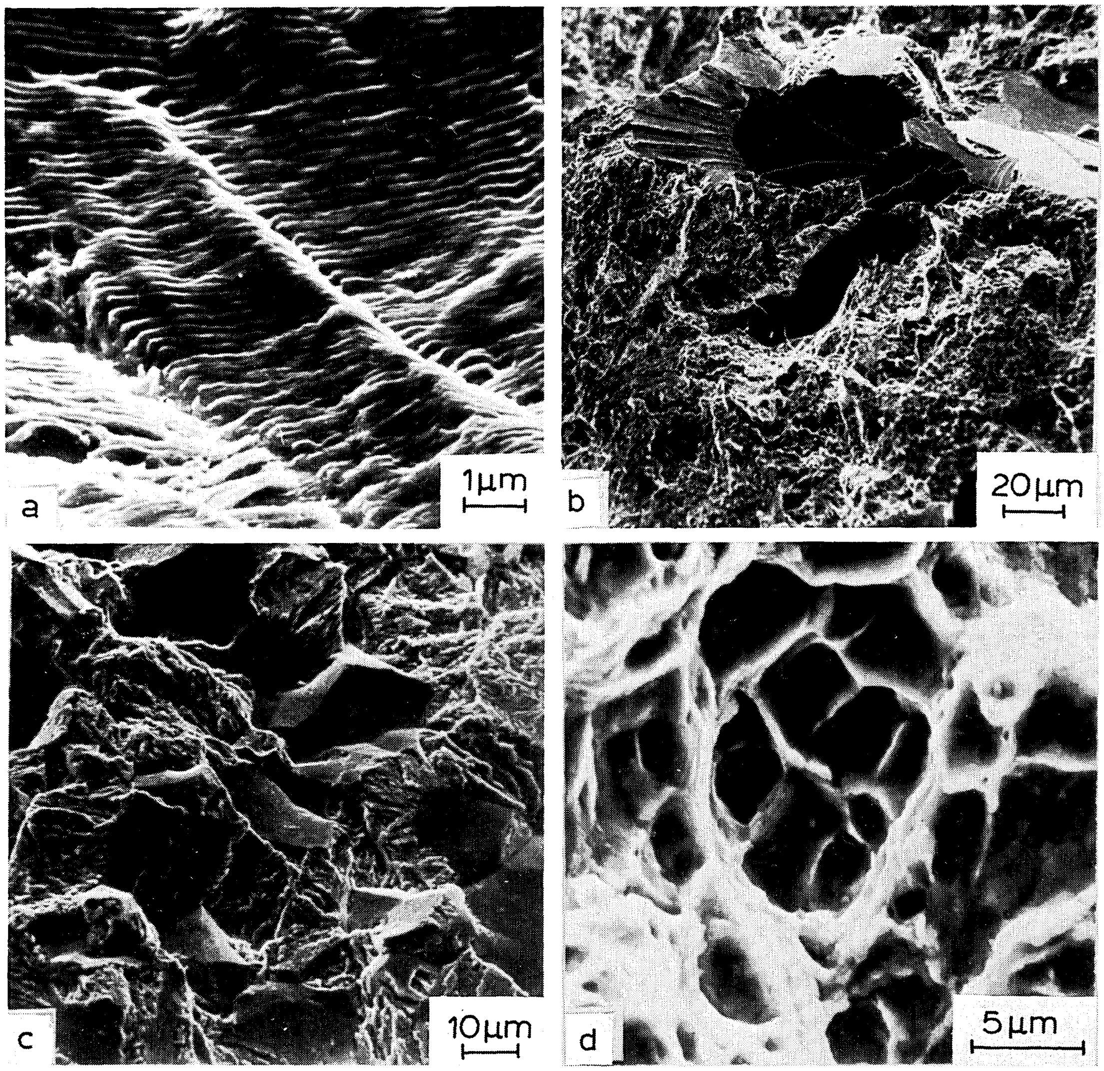
The application of linear-elastic fracture mechanics and related small-scale crack-tip plasticity has provided an empirical basis for describing the phenomenon of fatigue-crack propagation.[58] Most studies have confirmed that the crack-growth increment per cycle ($\mathrm{d}a/\mathrm{d}N$) is principally a function of the alternating stress intensity ΔK through a power-law expression of the form[59]:

$$\mathrm{d}a/\mathrm{d}N = A(\Delta K)^m \quad . \quad . \quad . \quad . \quad . \quad . \quad (1)$$

where A and m are experimentally-determined scaling constants, and ΔK is given by the difference between the maximum and minimum stress intensities for each cycle, i.e. $\Delta K = K_{max} - K_{min}$. This expression provides an adequate engineering description of behaviour at the midrange of growth rates, typically 10^{-5}–10^{-3} mm/cycle. At higher growth rates, however, when K_{max} approaches the fracture toughness (K_{Ic}, K_c) or limit-load failure, equation (1) often underestimates the propagation



2 Schematic variation of fatigue-crack growth rate d*a*/d*N* with alternating stress intensity Δ*K* in steels, showing regimes of primary crack-growth mechanisms



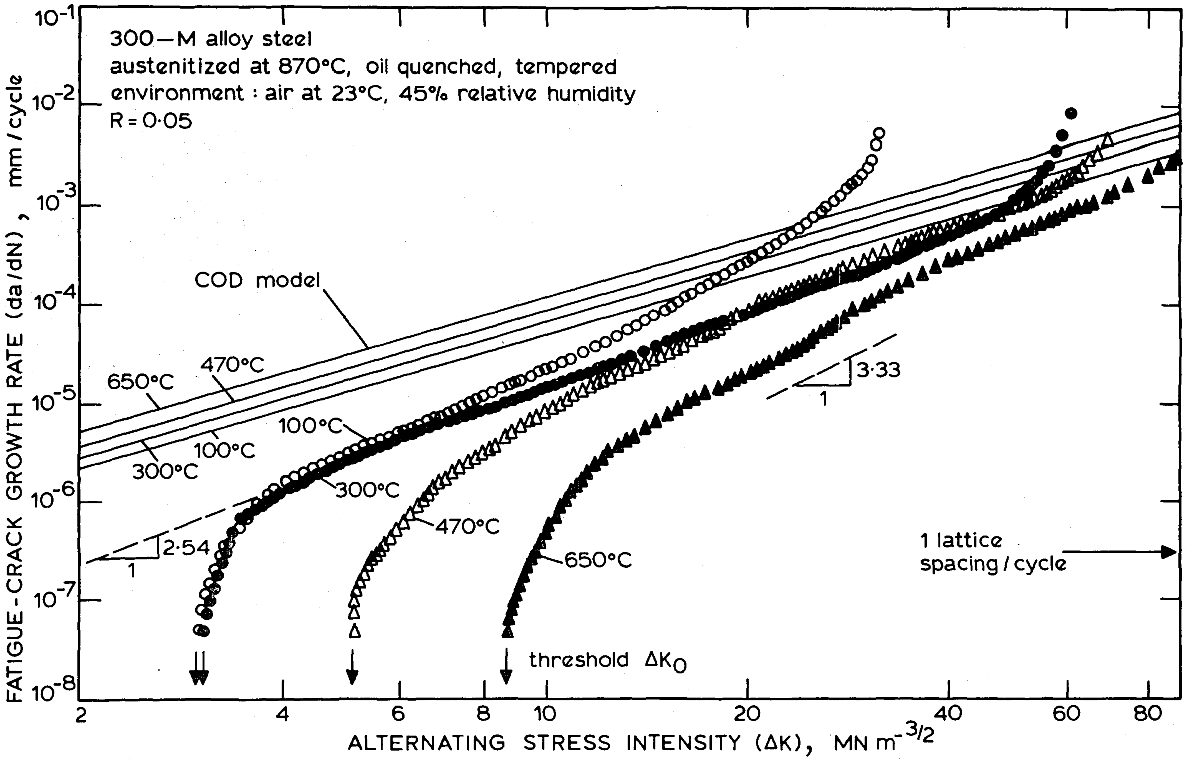
a ductile striations in 9Ni–4Co steel at $\Delta K = 30$ MN m$^{-3/2}$; *b* additional cleavage fracture in mild steel at $\Delta K = 40$ MN m$^{-3/2}$; *c* additional intergranular fracture in 4Ni–1.5Cr steel at $\Delta K = 40$ MN m$^{-3/2}$; *d* microvoid coalescence in 9Ni–4Co steel at $\Delta K = 70$ MN m$^{-3/2}$

3 Fractography of fatigue-crack propagation at intermediate (regime B) and high (regime C) growth rates in steels tested in moist air at $R = 0.1$

rate, whereas at lower (near-threshold) growth rates it is generally conservative as ΔK approaches the threshold stress intensity ΔK_0 (Fig. 2).

In steels, this sigmoidal variation of growth rates with ΔK has been characterized in terms of different primary mechanisms of fracture (Fig. 2). At the midrange of growth rates (regime B, where equation (1) applies), fatigue failure generally is observed to occur by a transgranular ductile striation mechanism[60,61] as shown in Fig. 3*a*, and there is often little experimentally observed variation of growth rates with microstructure and mean stress.[62–64] At higher growth rates (regime C), when K_{max} approaches K_{Ic}, static fracture modes, such as cleavage, intergranular and fibrous fracture (Fig. 3*b*–*d*), occur in addition to striation growth, resulting in a marked sensitivity of propagation rates to both microstructure and mean stress.[62–65] At low (near-threshold) growth rates (regime A), there is similarly a strong influence of microstructure and mean stress, although it is uncertain whether this can be *directly* related to a change in fracture mode. However, at such near-threshold stress intensities, the scale of plasticity approaches the order of the microstructural size scales, and measured propagation rates become less than an interatomic spacing per cycle, indicating that crack growth is not occurring uniformly over the entire crack front.

An indication of this deviation from continuum crack-growth mechanisms at near-threshold levels can be seen by comparing measured crack-propagation data with predictions based on crack-tip opening displacements as shown in Fig. 4. The latter model, originally proposed by McClintock,[66] considers striation growth to occur by alternating



4 Variation of fatigue-crack propagation in moist air at $R = 0.05$ with ΔK for ultrahigh-strength 300—M martensitic steel, quenched and tempered between 100° and 650°C to vary tensile strength from 2300 to 1190 MN m^{-2}, respectively; solid lines indicate predictions based on COD model for crack growth (equation (2))

shear[61] such that the crack-growth increment per cycle should equal the cyclic crack-tip opening displacement ΔCOD, i.e.

$$\frac{da}{dN} = \Delta COD \approx 0.49 \frac{\Delta K^2}{2\sigma_y' E} \quad . \quad . \quad . \quad . \quad (2)$$

where σ_y' is the cyclic flow stress and E the elastic modulus. It is clear from Fig. 4 that whereas such continuum models provide a reasonable description of crack-propagation behaviour at the midrange of growth rates, marked deviations are apparent at near-threshold levels.

There is no unifying picture of how the environment may influence such fatigue-crack propagation. At medium to high growth rates, several mechanisms have been proposed which rely on an increased chemical-reaction rate at the crack surface. In one class of mechanisms, the accelerating effect of the environment has been ascribed to increased mechanical failure caused by hydrogen (produced by an enhanced cathodic-reaction rate or by adsorption from a hydrogen-containing gas such as H_2 or H_2S) entering the lattice. This is the basis of 'hydrogen embrittlement' theories, which suggest that the hydrogen atoms are transported by diffusion or dislocation motion,[67] ahead of the crack tip, where they may induce hardening or softening,[38] or accumulate at interfaces (grain boundaries, internal voids, and cracks) and lead to decohesion,[68,69] or, in some circumstances, internal gas pressures.[38] In certain materials, such as titanium and zirconium alloys, the presence of hydrogen may lead further to the precipitation of brittle hydrides.[38,70,71] Other mechanisms interpret environmentally-enhanced fatigue-crack growth in terms of an increased anodic dissolution rate at the crack tip (active-path corrosion theories). Whereas it is now generally accepted that environmental effects in *high-strength* steels are principally hydrogen embrittlement,[72] in many materials, e.g. aluminium alloys, it is possible that both anodic and cathodic processes act in concert for moist or aqueous environments. Both processes result in increased growth rates from the rupture of protective oxide, and, in film-forming solutions, the environmental contribution to cracking can be considered as a function of the interaction between the rate of oxide rupture from emerging dislocations at the crack tip, the rate of passivation, and the rate of metal dissolution or hydrogen production at the bared surface.[46]

Lower strength steels, i.e. with yield strengths below 750 MN m^{-2}, post a particularly interesting problem in this regard since they are largely insensitive to environmentally-assisted cracking under sustained loading in low-pressure hydrogen-containing and dilute aqueous environments. However, on cyclic loading such steels, which range from AISI 1020 mild steel to ASTM A533B and A542 nuclear and coal-gasifier pressure-vessel steels, suffer significant environmental-assisted fatigue-crack propagation rates in hydrogen[73–75] and hydrogen sulphide gas,[49,73,74] distilled and salt water,[47,76] simulated nuclear-reactor environments,[48,49] and *low-pressure* gas mixtures typical of gasifier atmospheres.[73,74] Furthermore, measured crack-propagation rates appear to be frequency dependent at the midrange of growth rates (i.e. regime B in Fig. 2), and frequency independent at somewhat lower propagation rates.[76] However, little information has been obtained for the environmental contribution to cyclic cracking at near-threshold growth rates below 10^{-6} mm/cycle in these steels, and, moreover, no mechanistic basis for this effect has been developed, although it is widely considered to be hydrogen related in origin.

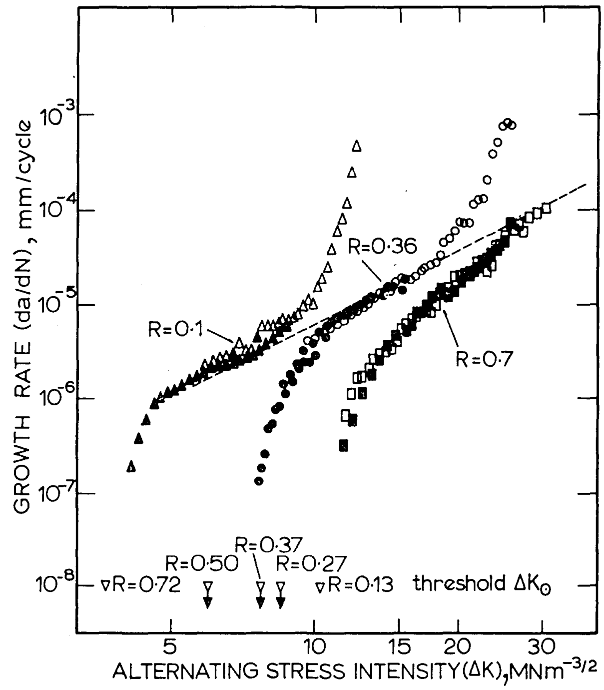
Mechanical and environmental behaviour associated with fatigue-crack propagation has been well documented, being the subject of many extensive reviews (*see*, for example, Refs. 65 and 77). However, in most cases, behaviour at very low, near-threshold rates has been overlooked. To rectify this, the effects of various mechanical, metallurgical, and environmental factors on near-threshold fatigue are examined in detail below.

NEAR-THRESHOLD FATIGUE-CRACK PROPAGATION

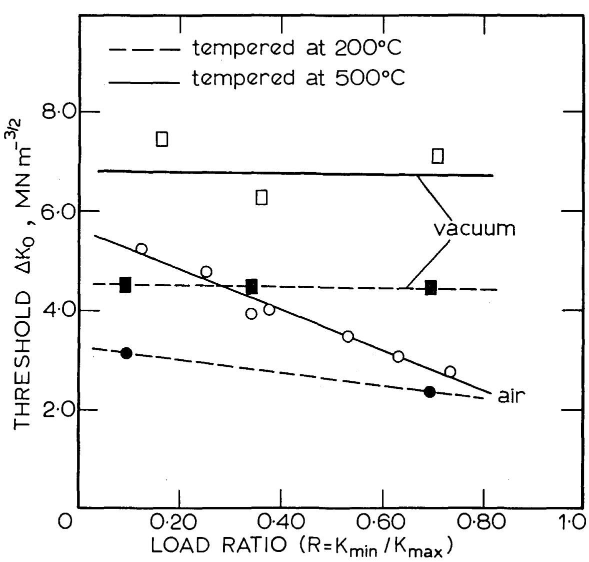
Mechanical factors

Mean stress (or load ratio)

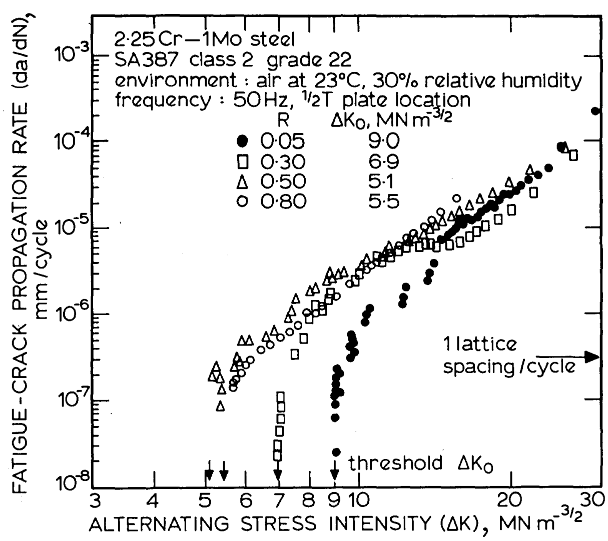
In fatigue studies, the effect of mean stress is often expressed in terms of the stress or load ratio $R(= K_{min}/K_{max})$. Whereas little influence of R can be seen for the midrange of growth rates, near-threshold propagation is generally extremely sensitive to the load ratio. Studies in a wide range of steels and non-ferrous alloys, tested in ambient-temperature air,[5–28,31–44] indicate that the value of ΔK_0 is markedly decreased, and that propagation rates are increased, as the load ratio is raised within a range of R from 0 to 0.9. Typical data, for a normalized medium-carbon steel,[78] are shown in Fig. 5. The load-ratio dependence on near-threshold growth, however, is found to be reduced at negative R values,[79] with increasing temperature,[6] with increasing strength in tempered martensitic steels,[26] and in inert atmospheres.[21,37,39–41,80] For the last case, Beevers and co-workers[21,37] observed that, for tempered martensitic En 24 steel and Ti–6Al–4V tested *in vacuo*, near-threshold crack-propagation rates and the value of ΔK_0 were completely independent of load ratio (Fig. 6). This lack of an R-dependence for tests under inert conditions has been confirmed for low-alloy martensitic steels *in vacuo* by Irving and Kurzfeld[81] and for 316 stainless steel *in vacuo* and in helium by Priddle *et al.*[31] Other studies of martensitic stainless steels[80] and of Ti–6Al–4V (Refs. 39, 40), however, show that by testing *in vacuo*, the dependence of ΔK_0 on load ratio is markedly reduced yet not completely eliminated, although in all cases the value of the threshold was significantly higher than in air. Such results are a strong indication that marked effects of load



5 Influence of load ratio from $R = 0.10$ to 0.72 on fatigue-crack growth in a normalized 0.55C–2.23Mn pearlitic–martensitic steel of $\sigma_y = 743$ MN m^{-2} (after Ref. 78)



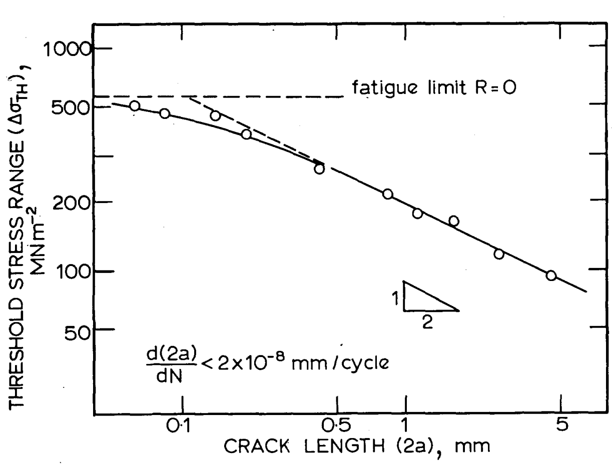
6. Variation of threshold ΔK_0 with load ratio R for martensitic 1.4Ni–1Cr–0.3Mo high-strength steel (En 24), tempered at 200°C ($\sigma_y = 1570$ MN m^{-2}) and at 500°C ($\sigma_y =$ 1274 MN m^{-2}); tests in laboratory air (40% relative humidity) and vacuum (10^{-3} Pa) (after Refs. 21 and 81)



7 Influence of load ratio from $R = 0.05$ to 0.80 on near-threshold fatigue-crack growth in air for normalized 2.25Cr–1Mo pressure-vessel steel (SA387) of $\sigma_y = 290$ MN m^{-2} (after Ref. 83)

ratio on near-threshold fatigue-crack propagation can be attributed, at least partially, to some environment interaction. The lack of a load-ratio effect on higher (midrange) growth rates is consistent with this argument, since at such faster propagation rates, the pertinent environmental reactions may not be able to keep pace with the crack velocity. Other explanations[6,7,19,35,42,79] of the load-ratio effect have centred around the phenomenon of crack closure, first identified by Elber[82] in aluminium alloys at much higher stress intensities. The relative merits of the environmental and closure explanations are discussed below.

In low-strength ferritic steels, both Cooke and Beevers,[12] and Masounave and Baïlon[15] found for a particular microstructure and strength



8 Effect of crack size on fatigue-threshold conditions in Japanese HT80 steel (after Ref. 45)

level that the effect of load ratio was consistent with the threshold occurring at a constant value of K_{max} or $\Delta K_0/1-R$. Such behaviour is without explanation, and has not been observed in higher strength steels or other materials.

The influence of load ratio on the fatigue-crack propagation threshold ΔK_0 in normalized 2.25Cr–1Mo steel (SA387 Class 2, Grade 22), a potential steel for coal-gasifier pressure-vessel construction, is shown in Fig. 7 for tests in ambient-temperature moist air at a strength level of 290 MN m^{-2} (Ref. 83). It is apparent that the values of ΔK_0 decreases with increasing load ratio up to $R = 0.5$, consistent with a constant value of K_{max} at the threshold. Above $R = 0.5$, ΔK_0 remains constant, similar to behaviour reported for aluminium alloys.[42] As noted above, explanations for this effect are uncertain.

Testpiece geometry and crack size

Provided conditions of linear elasticity are reasonably valid, fatigue-crack growth can be generally considered to be geometry independent, apart from variations in growth rate owing to changes in specimen thickness, e.g. departure from plane-strain conditions. In such instances, small differences in the growth rate between specimens of different thickness can arise because of nominal yielding, changes in fatigue-fracture mechanism near final failure, and perhaps closure effects.[84] However, for near-threshold fatigue-crack growth, effects of thickness and specimen geometry should be minimal because the low stress intensities involved invariably impose a condition of predominately plane strain. This has been borne out by experiments in low-strength steels[55] and aluminium alloys[32] tested over a range of specimen geometries and thicknesses.

The influence of crack size, on the other hand, has been shown to be particularly important. Classic experiments by Kitagawa and Takahashi[45] in low-strength steel have shown that, whereas the condition for the non-propagation of a surface flaw could be related to a constant threshold stress intensity ΔK_0 for crack lengths in excess of around 1 mm, below this crack size, a transition occurred in which the fatigue-limit stress became the threshold condition for growth (Fig. 8). Topper and co-workers[85,86] have attempted to rationalize these data by defining the alternating stress intensity ΔK at elastic stresses $\Delta\sigma$ in terms of

$$\Delta K = \Delta\sigma \sqrt{\pi(a + a_0)} \qquad (3)$$

where a is the crack length and a_0 a constant characteristic of a given material and material condition. Thus, by defining the threshold ΔK_0 in the usual way as the minimum value of ΔK for crack growth, the threshold stress range $\Delta\sigma_{TH}$ is given by

$$\Delta\sigma_{TH} = \frac{\Delta K_0}{\sqrt{\pi(a + a_0)}} \qquad (4)$$

such that as crack sizes become small, of the order of a_0

$$\Delta\sigma_{TH} = \frac{\Delta K_0}{\sqrt{\pi a_0}} = \Delta\sigma_e \quad . \quad . \quad . \quad . \quad . \quad (5)$$

where $\Delta\sigma_e$ is the fatigue limit corrected for the appropriate R value. Equations (4) and (5) reproduce exactly the data trend shown in Fig. 8, but at present there is no physical interpretation of the constant a_0.

There is also limited evidence that near-threshold crack-growth rates are accelerated, and the value of the threshold ΔK_0 decreased, at shorter crack lengths, i.e. for microcracks,[85-87] which, if substantiated, would provide a very feasible explanation to the existence of non-propagating cracks. A more complete review of the differences and similarities between the near-threshold fatigue-crack propagation behaviour of micro- and macrocracks can be found in Ref. 88.

Frequency and wave shape

For fatigue-crack propagation in the midrange of growth rates, the general effect of decreasing the cyclic frequency is to increase the crack-growth increment per cycle, due to enhanced environmental effects.[77] At near-threshold growth rates, however, very limited data exist, simply because of the limitations imposed by the high-frequency response of conventional testing machines, and the time factor involved in measuring low growth rates at very low frequencies. Results[55] on 2219–T851 aluminium alloy indicated no effect on near-threshold growth over a frequency range of 25–150 Hz. Tests in 2024–T3 aluminium alloy at frequencies between 342 and 832 Hz (Ref. 42), conversely, showed an increase in crack propagation rate and a twofold decrease in ΔK_0 with increasing frequency, which was tentatively attributed to creep effects due to crack-tip heating. In D6ac steel, however, increasing the frequency from 100 to 375 Hz resulted in lower near-threshold growth rates for tests in dry argon, and no effect in room air.[22]

For low-strength ferritic steels, such as HSLA pipeline steels X–65 (Ref. 47) and A533–B and A508 nuclear pressure-vessel steels,[49] tested in water environments at frequencies between 0.01 and 10 Hz, there is evidence showing marked frequency-dependent crack propagation behaviour for growth rates in excess of 10^{-5} mm/cycle, which becomes frequency independent at lower growth rates as the threshold is approached. Similar results have been seen in higher strengh HY 130 marine steels tested in salt solution.[76] Companion tests in hydrogen sulphide gas[49] resulted in similar growth rates without any effect of frequency, even above 10^{-5} mm/cycle. The environmentally-enhanced growth rates in water and hydrogen sulphide, compared to those measured in air, were accounted for in terms of hydrogen embrittlement, although the mechanism for the embrittlement was not defined and no attempt was made to explain the transition from frequency-independent to frequency-dependent propagation rates in water above 10^{-5} mm/cycle.

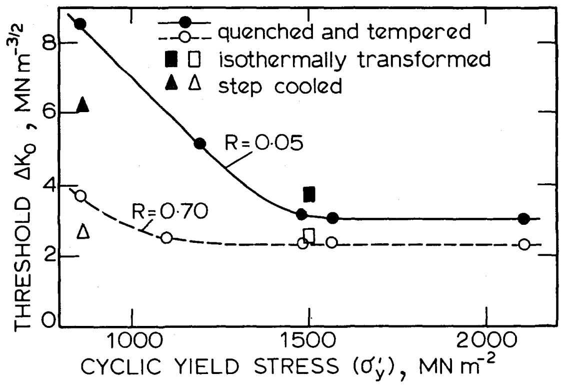
The effect of wave form (i.e. sinusoidal, square, and triangular) on near-threshold growth has been examined in aluminium alloys and stainless steels in room air.[32] For both materials no change in ΔK_0 was observed for the various wave forms. In the same experiments, decreasing the frequency from 130 to 0.5 Hz for all wave shapes led to a decrease in ΔK_0, perhaps reflecting the influence of the environment at low growth rates.

Microstructural factors

Material strength

Fatigue-crack propagation in metals has been generally found to be largely unaffected by yield strength.[89] In fact, for steels, raising the strength by nearly an order of magnitude does not change crack-propagation rates over the midrange of growth rates by much more than a factor of two or three.[65] However, at near-threshold levels below 10^{-6} mm/cycle, a surprisingly large dependence of material strength has been observed on the value of the threshold ΔK_0 and on subsequent growth rates. In low-strength ferritic–pearlitic steels with yield strengths less than 500 MN m^{-2}, for example, values of the threshold have been observed to decrease significantly with increasing strength.[11,16,18] An even larger effect has been reported for high- and ultrahigh-strength martensitic steels (yield strengths between 1000 and 2000 MN m^{-2}) with the exception that the controlling measure of strength was the cyclic, rather than the monotonic, yield stress.[26,27] Specifically, increasing the cyclic strength led to marked increases in near-threshold propagation rates (Fig. 4), and a significant reduction in ΔK_0 (Fig. 9). Cyclic softening can thus be regarded as extremely beneficial to improving near-threshold fatigue-crack growth resistance in steels.[26,27] In low-alloy pressure-vessel steel, variations in monotonic yield strength between 300 and 400 MN m^{-2} resulting from differences in cooling rate experienced through the thickness of a 184 mm thick plate, did not give rise to significant changes in near-threshold fatigue-crack growth resistance in a 2.25Cr–1Mo steel, SA387 (Ref. 83). However, in 0.5Cr–0.5Mo–0.25V steel,[90] coarse-grained precipitation-hardened ferritic microstructures showed significantly lower growth rates near ΔK_0 than higher strength bainitic or martensitic structures.

The effect of strength on near-threshold crack-propagation behaviour is significantly less at high R values,[26,27,80,91] and not so evident in ferritic–pearlitic steels.[91] The latter probably results from the fact that data on the lower strength steels cover a relatively narrow range of strengths and are subject to considerably more scatter. Furthermore, varying the strength in such steels generally involves variations in the



9 Variation of threshold ΔK_0 with cyclic yield strength at $R = 0.05$ and 0.70 for 0.4C–1.76Ni–0.76Cr–1.6Si ultrahigh-strength steel (300–M) tested in moist laboratory air

ferrite grain size which is also known to have a marked influence on near-threshold behaviour.[16,18,29]

Results in non-ferrous alloys, however, reveal somewhat different behaviour. Small reductions in the threshold ΔK_0 have been observed in aluminium bronze[44] and Al–Zn–Mg alloys[43] as the cyclic strength is decreased. This is precisely the opposite of behaviour observed in steels, but the magnitude of the effect is very much smaller, for example, increasing the cyclic flow stress by a factor of nearly two in the aluminium alloy only results in an increase in ΔK_0 from 2.5 to 3.7 MN m$^{-3/2}$ (Ref. 43), whereas a similar increase in strength in a tempered martensitic 300–M steel leads to a decrease in ΔK_0 from 8.5 to 3.0 MN m$^{-3/2}$ (Ref. 26).

Explanations for these effects are again uncertain. The small increase in ΔK_0 with increase in flow stress for the non-ferrous materials have been attributed to a decrease in the alternating plastic crack-tip opening displacement.[43,44] However, reference to Fig. 4 clearly shows that such explanations are not applicable to steels. In this figure, growth-rate predictions based on the COD model (equation (2)) are compared with experimental data for an ultrahigh-strength martensitic steel 300–M, tempered over a range of temperatures from 100° to 650°C to vary the tensile strength from 2340 to 1190 MN m^{-2}, respectively. It is apparent that not only does the COD model predict a far too small dependence of growth rates on strength but also predictions are in the wrong direction, i.e. higher strength results in a marked increase in near-threshold growth rates whereas COD predictions show a small decrease.

The dependence of near-threshold growth-rate behaviour on material strength in steels has been rationalized, however, in terms of environmental arguments[24,27] and notch-sensitivity effects,[92] as described below and in the 'Discussion' section of this paper. A comparison of growth rates in a single material, with varying strength, *in vacuo* is required to help resolve this issue.

An intriguing aspect of the strength effect in steels is the fact that whereas the fatigue-crack propagation threshold ΔK_0 is *decreased* with increasing strength, the well known fatigue limit $\Delta\sigma_e$, or endurance strength, is *increased*.[26,88,90] Both parameters represent limits for fatigue damage,[88] but the threshold ΔK_0 must be regarded as the minimum stress intensity below which long macrocracks do not grow, whereas the fatigue limit is generally the minimum stress below which short macrocracks do not *initiate* (i.e. by coalescence of microcracks). Following the work of Topper and co-workers[85,86] and McEvily,[92] one can rationalize this apparent inconsistency in the effect of strength by equating the characteristic length constant a_0 in equations (3)–(5) to Neuber's effective crack-tip radius ρ', such that

$$\Delta K_0 = \Delta\sigma_{TH}\sqrt{\pi(a+\rho')} = \lim_{a\to\rho'} \Delta\sigma_e\sqrt{\pi\rho'} \qquad (6)$$

where $\Delta\sigma_{TH}$ is the threshold stress and a the crack length. Since the fatigue limit for steels (at $R = -1$) is generally found to be half the ultimate strength ($\sigma_u/2$), correcting for $R = 0$ using the Goodman relationship yields

$$\Delta\sigma_e = 2\sigma_u/3 \qquad (7)$$

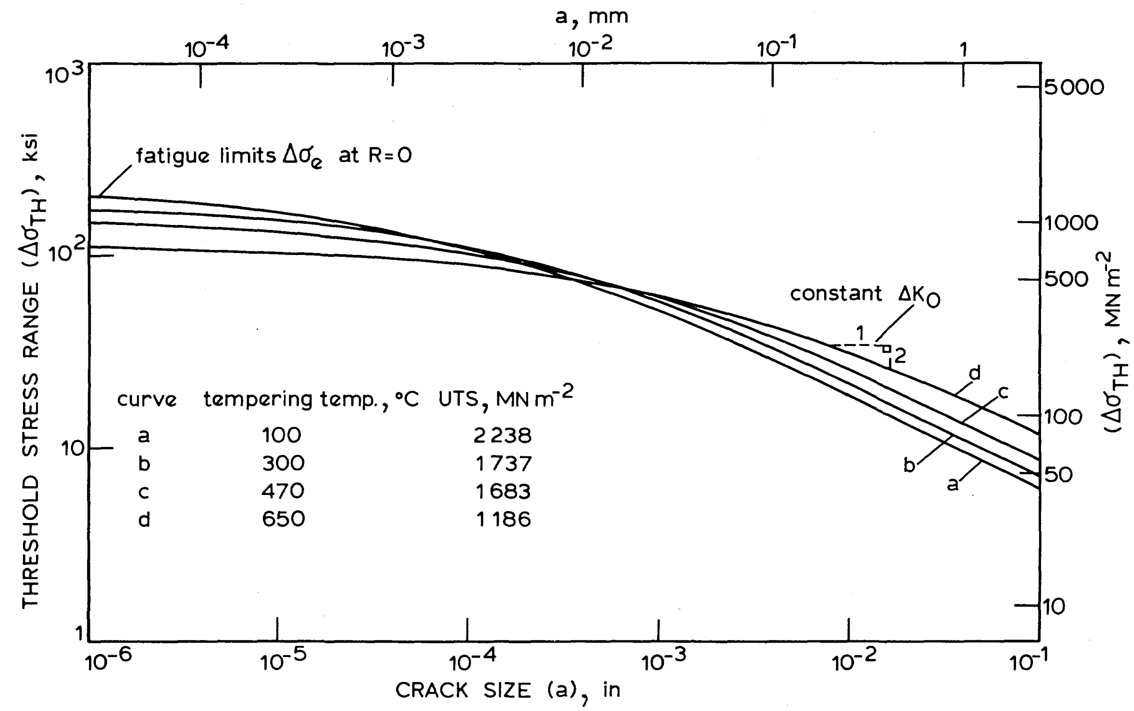
consistent with the experimentally observed *increase* in fatigue limit with increasing strength. However, from the studies of Kuhn and Hardrath,[93] the Neuber constant ρ' appears inversely proportional to the ultimate tensile strength (raised to the 4th power), and the following empirical relationship is found for steels

$$\sqrt{\rho'} = \frac{530}{\sigma_u^2} \qquad (8)$$

where ρ' is in inches and σ_u in ksi. Combining equations (6)–(8) gives

$$\Delta K_0 = \Delta\sigma_e\sqrt{\pi\rho'} \approx \left(\frac{2\sigma_u}{3}\right)\left(\frac{530\sqrt{\pi}}{\sigma_u^2}\right) \propto \frac{1}{\sigma_u} \qquad (9)$$

consistent with the experimentally observed *decrease* in threshold ΔK_0 with increasing strength. By comparison with Fig. 8, this indicates that at short crack lengths ($a \approx \rho'$), the threshold stress $\Delta\sigma_{TH}$ will be *increased* as the strength level is raised, whereas at long crack lengths ($a \gg \rho'$), $\Delta\sigma_{TH}$ will be *decreased*. These predictions are plotted in Fig. 10 using data from 300–M high-strength steel tempered between 100° and 650°C (Refs. 26, 27, 94) and equations (6)–(9). As discussed by Fine and Ritchie,[88] this result is

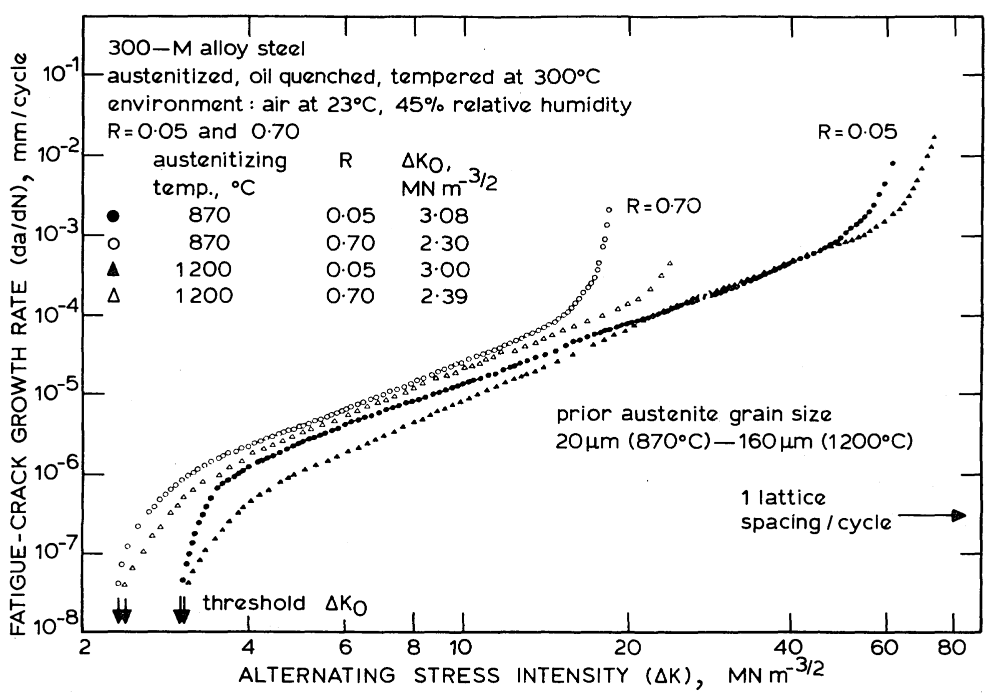


10 **Predicted variation of threshold stress $\Delta\sigma_{TH}$ at $R = 0$ with crack size a from equations (6)–(9) based on data for 300–M ultrahigh-strength steel tempered between 100° and 650°C to vary tensile strength**

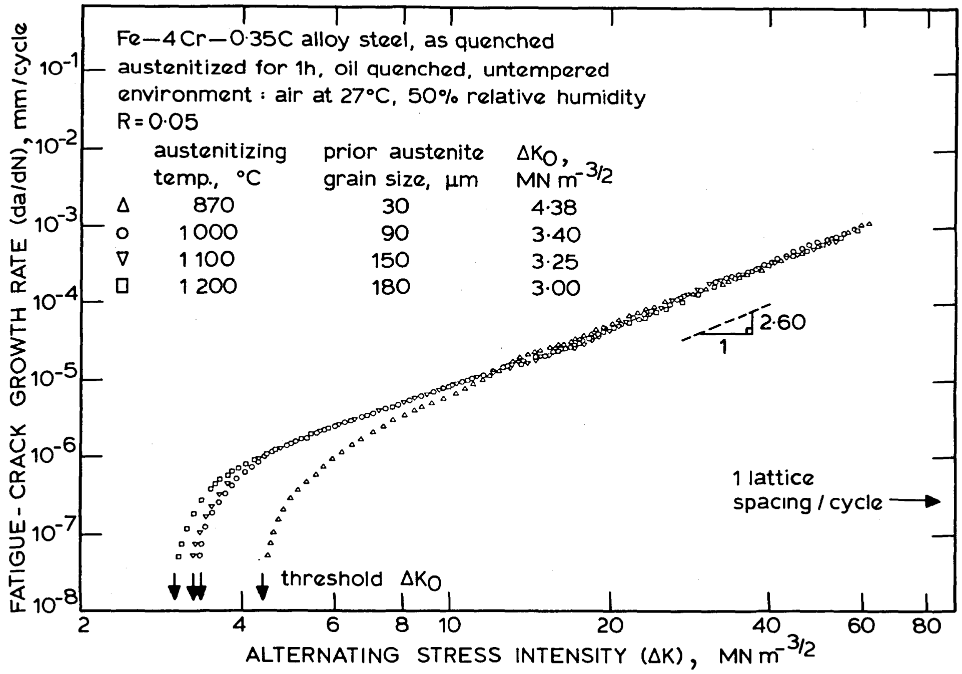
particularly significant for the alloy design of steels with improved resistance to very high cycle fatigue damage, since the optimum microstructures desired will depend upon whether structural design is to be based on initiation or propagation of a 'fatal' flaw.

Grain size

Whereas refining grain size can be beneficial in raising the fatigue limit or endurance strength of (planar slip) materials,[95] the effect of grain size on fatigue-crack propagation has been observed to

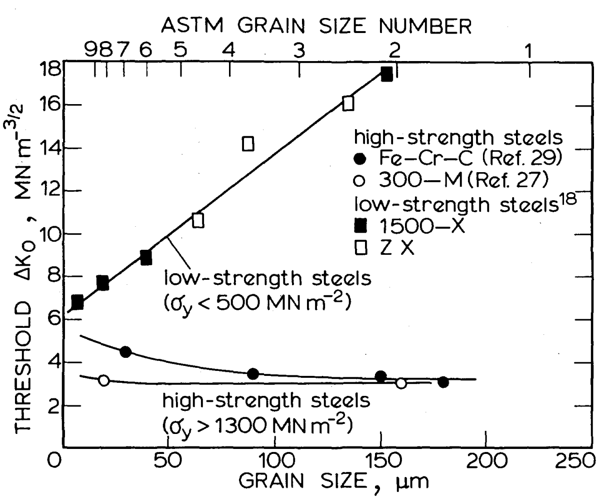


11 **Effect of prior austenite grain size (20–160 µm) on fatigue-crack growth in cyclic softening 300–M ultrahigh-strength steel (σ_y = 1700 MN m^{-2}) at R = 0.05 and 0.70 in moist laboratory air**



12 Effect of prior austenite grain size (30—180 μm) on fatigue-crack growth in cyclic-hardening as-quenched Fe—Cr—C high-strength steel ($\sigma_y = 1300$ MN m^{-2}) at $R = 0.05$ in moist laboratory air

be negligible in most studies at intermediate growth rates.[29,95–97] At low growth rates in room air, however, several workers[16,18,24,27,33,36,98] have observed improved resistance to near-threshold crack propagation with coarser grain sizes. Robinson and Beevers[33] report an order of magnitude decrease in near-threshold growth rates in α-titanium after coarsening the grain size from 20 to 200 μm. Similar effects have been seen in Ti—6Al—4V (Refs. 36, 98) and other basic $\alpha + \beta$



13 Variation of threshold ΔK_0 with grain size for steels at $R = 0.05$: for ferritic—pearlitic low-strength steels grain size refers to ferritic grain size, whereas for martensitic high-strength steels, grain size refers to prior austenite grain size

titanium-alloy systems.[98] Furthermore, a marked increase in threshold ΔK_0 values has been observed in a range of low-strength steels by increasing the ferrite grain size.[16,18] In all of these studies, however, no attempt was made to control strength, and the effect of coarsening the grain size may well have been masked by a concurrent decrease in strength, which is known to increase markedly the threshold in steels.[27] Comparisons at constant yield strength have been made in two high-strength steels, 300—M (Refs. 24—27) and Fe—Cr—C (Ref. 29), where it was found that, in the cyclic softening 300—M, coarsening the prior austenite grain size by almost an order of magnitude decreased near-threshold growth rates yet left ΔK_0 unchanged (Fig. 11), whereas, in cyclic hardening Fe—Cr—C, similar coarsening of the structure increased near-threshold rates and reduced the threshold (Fig. 12). The variation of threshold with grain size for low- and high-strength steels is shown in Fig. 13, and clearly shows the contrasting behaviour between the two classes of steels.[29]

Several explanations have been proposed to explain the grain-size effect, none of which is entirely satisfactory. It has been suggested, for example, that since near-threshold growth is 'microstructurally-sensitive', it may be confined to specific crystallographic planes so that in coarser structures, greater deviations of the crack path may occur from the plane of maximum tensile stress.[16] Other authors[24,99] have reasoned for high-strength steels that, since plastic zones are often confined within a single grain during near-threshold growth, the probability that hydrogen atoms, generated by chemical reactions with the environment at the crack tip, can be swept into the grain boundaries by dislocation motion,[67] is much smaller if the grain size is very large. Effectively, this hypothesis states, that by coarsening the grain size, the environmental contribution to near-threshold growth is reduced, and this is consistent with an observed decrease in hydrogen-embrittlement susceptibility of high-strength steels as the grain size is increased.[100] However, the conflicting results observed in Fe—Cr—C steel[29] are not consistent with this explanation. A study of the influence of grain size (at constant strength) in inert environments is required before this effect can be resolved.

Structure

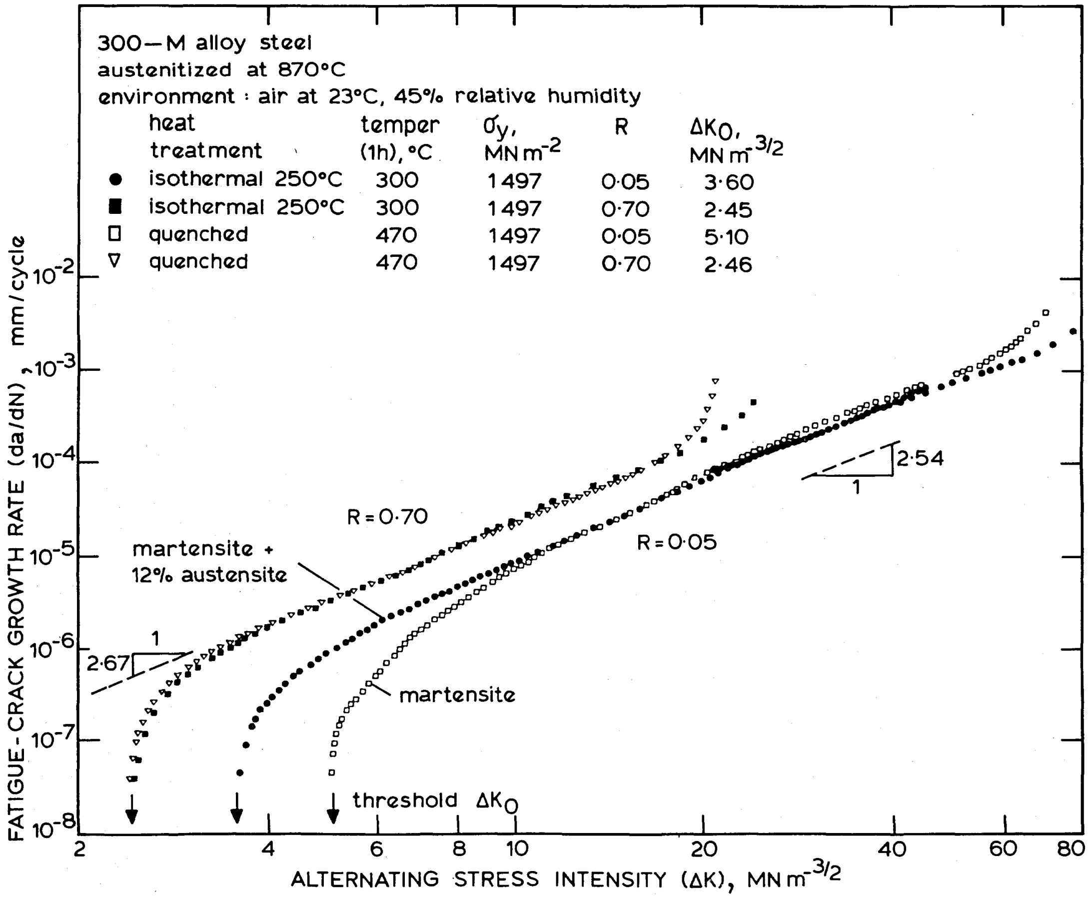
Very little information is available at present on the relative resistance to near-threshold fatigue-crack propagation of particular microstructures in metals (i.e. pearlite *v.* tempered martensite *v.* bainite, etc.), aside from data where particular structures have been compared at different strength levels. For example, in aluminium alloys,[43] underaged structures appear to have fractionally better resistance than peak and overaged structures whereas in tempered martensitic steels, spheroidized structures offer far the best resistance.[27] Comparisons at constant strength have been made in 300—M ultrahigh-strength steel[26,27] between isothermally-transformed structures containing an interlath network of retained austenite (volume fraction ~12%) within a lower bainite—tempered-martensite matrix, and quenched and tempered fully martensitic structures containing no austenite. Here, if the structures are compared at equivalent monotonic yield strength, quenched and tempered microstructures offer greater resistance owing to their lower cyclic yield strength* (Fig. 14). However, when compared at equivalent cyclic strength, isothermally-transformed structures offer marginally superior resistance, in the form of higher thresholds (Fig. 9), which is indicative[101] of their superior resistance to environmentally-induced cracking under monotonic loads. A similar small beneficial effect of interlath retained austenite on near-threshold crack-propagation resistance in tempered martensitic steels has been observed for a 9Ni—4Co—0.2C high-strength aerospace steel HP 9—4—20 tested in moist air.[102] It was reasoned in this case that the role of the austenite effectively was to reduce the environmental contribution to cracking (i.e. from hydrogen embrittlement) by slowing down the rate of diffusion of adsorbed hydrogen atoms ahead of the crack tip. Alloy additions of silicon, similarly, may lead to a reduction in diffusivity of hydrogen in iron.[103–105] Accordingly, silicon-modified 4340 (300—M), with its lower susceptibility to sustained-load environmentally-assisted cracking,[101,106,107] displays superior near-threshold crack-propagation resistance in moist air compared with unmodified 4340 at the same strength level, as shown in Fig. 15 (Ref. 108). In general, it appears that microstructures with superior resistance to hydrogen-assisted or stress corrosion cracking under monotonic loads will have a similar superior resistance to near-threshold fatigue-crack growth.[108]

*Isothermally-transformed structures do not cyclically soften due to deformation-induced transformation of retained austenite to martensite.[26]

A particularly striking effect of microstructure has been observed in AISI 1018 mild steel, where the production of duplex ferrite—martensite structures can lead to increased near-threshold crack-propagation resistance and increased strength compared to conventionally heat-treated steel.[109] By suitable heat-treatment procedures, duplex microstructures where the martensitic phase α' is continuous and totally encapsulates the ferritic phase α were found to have increased strength ($\sigma_y = 452$ MN m^{-2}) and a significantly higher threshold ($\Delta K_0 \approx 20$ MN m$^{-3/2}$), than similar structures where the ferritic phase surrounds the martensite ($\sigma_y = 293$ MN m^{-2}, $\Delta K_0 \approx$ 10 MN m$^{-3/2}$), as shown in Fig. 16. The result is contrary to what one might expect for the relationship between strength and near-threshold behaviour and is without explanation, yet it does introduce a very promising way of improving the fatigue-crack propagation resistance of low-carbon steels without compromising other mechanical properties.

For non-ferrous metals, studies in Ti—6Al—4V at roughly constant strength (monotonic yield strengths between 835 and 1000 MN m^{-2}) have revealed a strong effect of structure near the threshold.[36] Ranked in order of greatest resistance to near-threshold growth, β-annealed structures were superior to transformed β, followed by as-received and martensitic structures. These microstructural differences were more pronounced at low R values, as has been similarly observed in ultrahigh-strength steels,[26] and far less pronounced for tests *in vacuo.*[37] This latter fact once again reinforces the argument that microstructural influences on near-threshold fatigue-crack propagation are often primarily a consequence of environmental effects, rather than inherent mechanical effects.

The effect of non-metallic inclusion content on near-threshold fatigue behaviour has been examined in a medium-strength pearlitic rail steel.[110] Here it was found that, whereas decreasing the volume fraction of inclusions (sulphide stringers and oxide-type) led to marked increases in the fatigue or endurance limit, no systematic effect was observed on the value of the threshold ΔK_0. However, other studies[111] on the influence of steelmaking practice on somewhat higher fatigue-crack propagation rates in nuclear pressure-vessel steel A533B have shown that the absence of elongated Type 2 MnS inclusions and galaxies of alumina inclusions by calcium treating or



14 Comparison in martensitic 300—M high-strength steel at constant monotonic strength of isothermally-transformed structure, containing 12% retained austenite, with quenched and tempered structure containing no austenite, at $R = 0.05$ and 0.70 in moist laboratory air; quenched and tempered structure undergoes cyclic softening ($\sigma_y{}' = 1200$ MN m^{-2}) whereas isothermally-transformed structure remains cyclically stable ($\sigma_y{}' = 1500$ MN m^{-2})

electroslag remelting results in a significant improvement in the isotropy of fatigue-crack growth behaviour and lower overall growth rates.

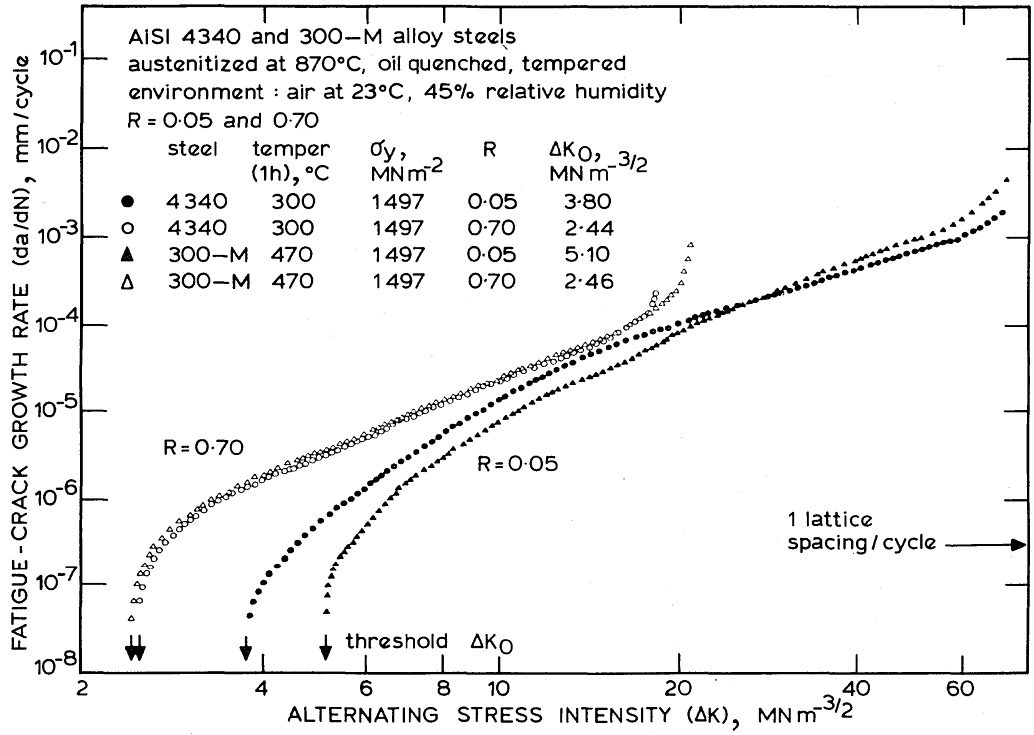
Impurity-induced grain-boundary segregation

It is well known that the toughness of low-alloy steels can be severely impaired by heat-treatment procedures which result in the segregation of residual-impurity elements, such as P, S, Sb, and so forth, to grain boundaries (temper embrittlement). Recently, it has become clear that impurity segregation can also degrade creep, stress corrosion, and hydrogen-induced cracking resistance.[25] There has been one study[25] on the effect of prior impurity segregation on near-threshold fatigue-crack propagation, conducted in an ultrahigh-strength martensitic steel (300—M), tested in moist room air in the unembrittled and temper embrittled* conditions (at the same yield strength and prior austenite grain size). Here it was found that, although no difference in crack propagation rates was observed between unembrittled and embrittled structures at the midrange of growth rates, at near-threshold levels, impurity-induced embrittlement gave rise to vastly accelerated growth rates and a reduction in ΔK_0 by almost 30% at both low and high load ratios (Fig. 17). This was accompanied by a significant increase in the proportion of intergranular fracture in the embrittled steel close to the threshold. This large effect at very low growth rates, compared to little or no effect at higher intermediate rates, is again indicative of an enhanced environmental contribution to cracking in the impurity-embrittled structure, perhaps caused by an interaction between residual-impurity elements and hydrogen atoms (generated by chemical reactions with the moist air atmosphere at the crack tip) in prior austenite grain boundaries (*see* 'Discussion' section below).

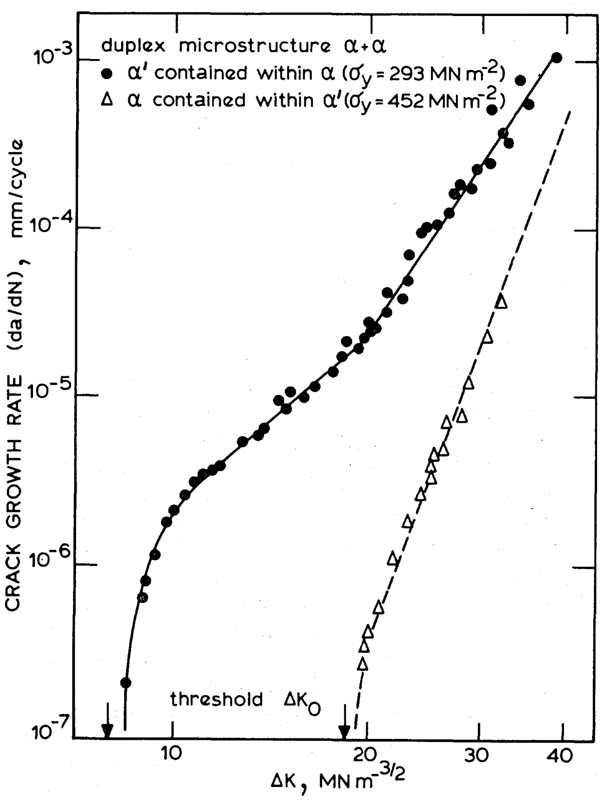
*Auger studies indicated that the embrittlement resulted from a build-up of P, Si, Mn, and Ni in prior austenite grain boundaries.[25]

Weld microstructure

Little information exists on the influence of welds and weld microstructure on near-threshold fatigue-



15 Comparison, at constant yield strength ($\sigma_y = 1497$ MN m^{-2}), of fatigue-crack growth behaviour of quenched and tempered 4340 and 300—M (4340 modified with 1.3%Si) ultrahigh-strength steels in moist laboratory air at $R = 0.05$ and 0.70 (after Ref. 108)



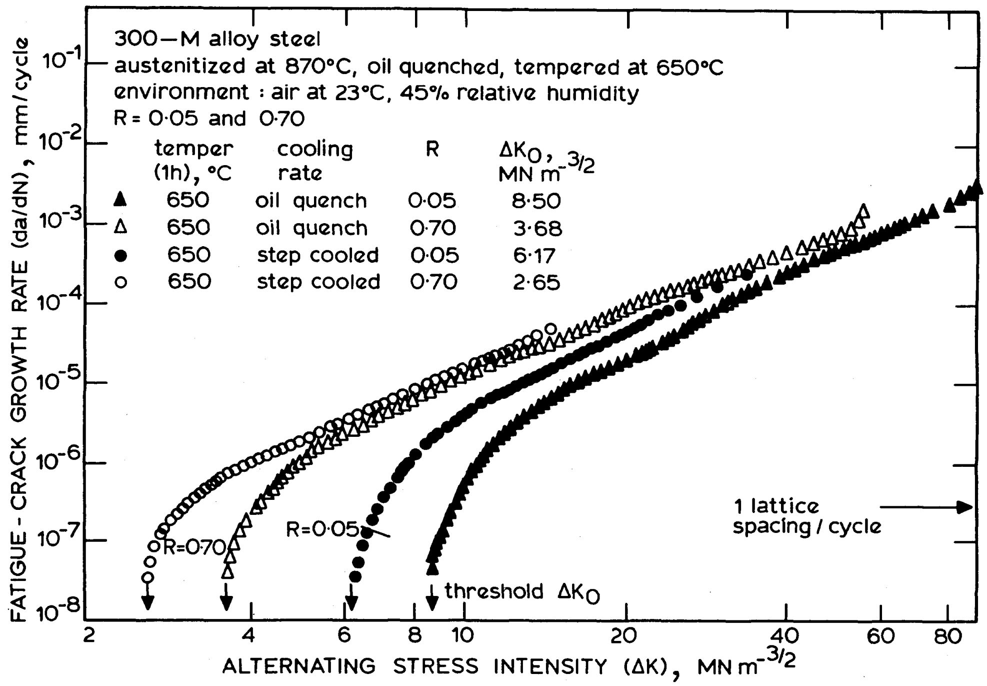
16 Variation of fatigue-crack growth rate with ΔK for duplex microstructures in AISI 1018 mild steel; duplex structures consist of ferrite α and martensite α' (after Ref. 109)

crack propagation, except for a few studies on austenitic stainless steels.[30,112] In general, stainless steel weldments show little or no deterioration in fatigue-crack propagation resistance when compared to parent metal at intermediate growth rates.[113] At low growth rates, however, Type 304 and 316 weldments show faster crack propagation rates than in parent metal, both at ambient[30] and higher[112] temperatures in air. No explanations exist for this behaviour, apart from suggestions regarding the possible deleterious effect of the variable and coarser grained structure within the weld.[30] Limited results for weldments of mild steel indicate no change in threshold ΔK_0 values between parent metal, heat-affected zone areas, and weld metal.[17]

Environmental factors

Temperature

Effects of temperature on near-threshold fatigue-crack growth behaviour have been studied primarily in steels.[6,17,31] Pook and Greenham[17] observed no change in ΔK_0 between ambient temperature and 300°C for tests on mild steel in air at low R values, whereas thresholds were higher at 300°C for tests at high R values. Conversely, Paris *et al.*,[6] in A533B and A508 nuclear pressure-vessel steels, found the largest sensitivity of the threshold to temperatures at low R values (i.e. $R = 0.1$), and observed no change in ΔK_0 over the temperature range, ambient to 350°C, at high R values (i.e. $R = 0.7$). Furthermore, although fatigue-crack growth (at $R = 0.1$) was



17 Effect of prior impurity-induced embrittlement on fatigue in 300—M high-strength steel at $R = 0.05$ and 0.70 in moist laboratory air; step-cooled structure is temper embrittled, oil-quenched structure is unembrittled at same strength level ($\sigma_y = 1070$ MN m^{-2})

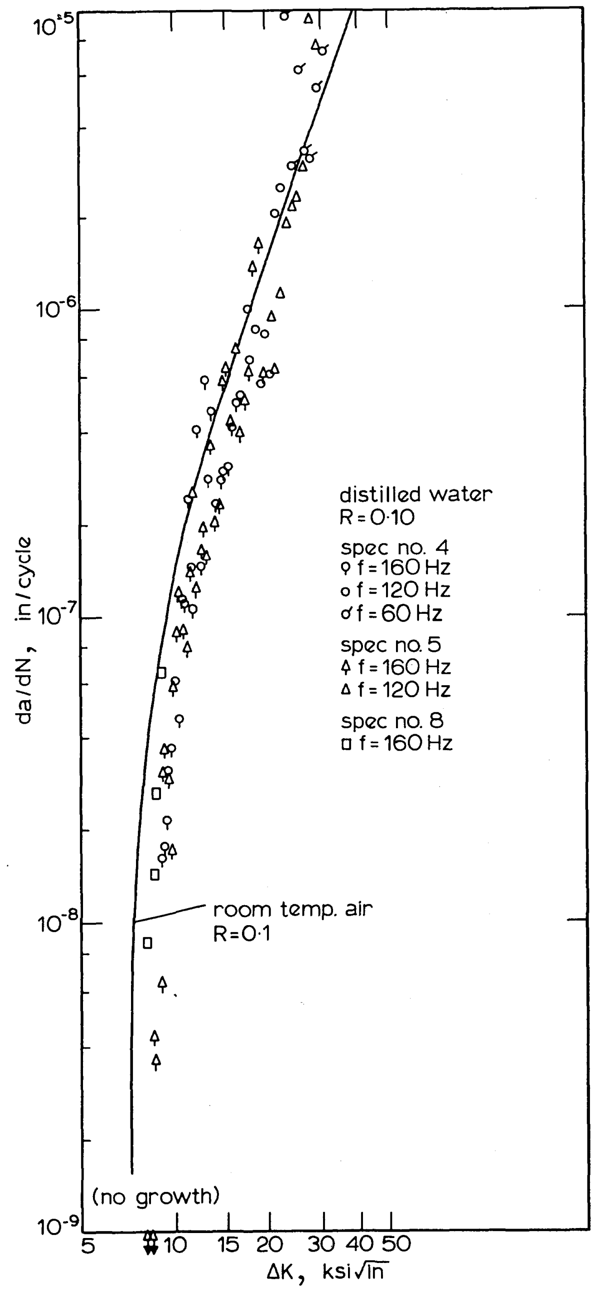
most sensitive to temperature close to the threshold, there was no systematic increase in ΔK_0 with increasing temperature. Threshold values were found to be highest at ambient temperature and at 350°C, and lowest at 180°C. No explanations have been proposed for this behaviour. Threshold values for 316 stainless steel, however, were observed to increase with temperature over the range 20°—700°C for tests in air, and to remain constant over the same range of temperature for tests *in vacuo* and in helium.[31] In the latter case, the evidence implies that the influence of temperature on near-threshold fatigue behaviour in stainless steel at least is primarily a function of environmental factors.

Pressure

For high-strength martensitic steels tested in low-pressure (10—100 kPa) hydrogen gas, rates of crack growth under sustained loading have been found to be sensitive to the pressure of the surrounding gas to a degree dependent upon temperature (*see* e.g. Refs. 58, 114—117). At temperatures below about 0°C, steady-state growth rates appear to vary with the square root of the hydrogen pressure, between 0° and 40°C growth rates appear proportional to the first power of pressure, whereas between 40° and 80°C the dependence is to the 1.5-power of pressure.[115] Limited studies under cyclic loading, where the environmental contribution to cracking can be considered to be hydrogen embrittlement, tend to support these pressure dependences. However, there are no reported studies on the effect of pressure, in any environment, on near-threshold crack-propagation rates, either for low pressure (10—100 kPa) atmospheres where hydrogen embrittlement predominates, or for high-temperature (350°C), high-pressure (10 MPa) atmospheres where hydrogen attack[118] predominates. Since the latter environments are particularly important with regard to pressure-vessel steels for coal gasification or liquifaction processes and pipeline steels for potential hydrogen transport, this is an area where there is a great need for future study.

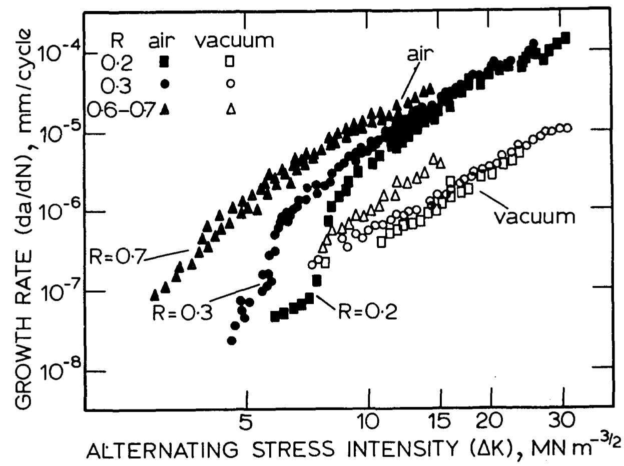
Environmental species

Whereas it is generally accepted that the nature of the environment plays a prominent role in influencing intermediate growth-rate fatigue-crack propagation behaviour, opinion is still somewhat divided with respect to the influence of the environment at near-threshold rates. On the one hand, there are the initial results of Paris and co-workers[6,7,35] on low-strength steels and titanium alloys which showed no variation in threshold behaviour with change in environment, for tests at high frequency (160 Hz). Specifically, no change in near-threshold growth rates and the value of ΔK_0 at ambient temperature were observed for: (a) A533—B nuclear pressure-vessel steel (Fig. 18) tested in air and distilled water[6]; (b) low-strength T—1 steel tested in air, distilled water, and hydrogen gas[7]; and (c) Ti—6Al—4V tested in air and dry argon of unspecified purity.[35] Somewhat similar results have been seen in high-strength D6ac steel and a 7050—T73651 aluminium alloy,[22] where threshold ΔK_0 values were un-



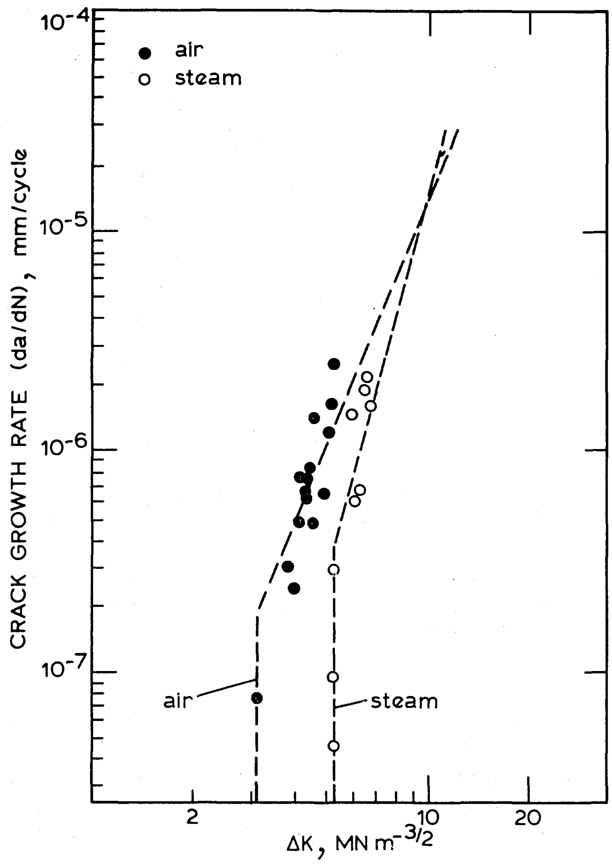
18 Near-threshold fatigue-crack propagation in ASTM A533B nuclear pressure-vessel steel (0.2C–1.5Mn–0.7Ni–0.5Mo), tested between 60 and 160 Hz at $R = 0.1$, showing a comparison of data for distilled water with 'best-fit' data line for room-temperature air (after Ref. 6)

changed in air and dry and wet argon atmospheres at 100–374 Hz, although near-threshold growth rates were marginally lower in dry argon compared to air in the steel and significantly lower in the aluminium alloy. On the other hand, there are results where significant decreases in near-threshold growth rates were observed by comparing crack propagation *in vacuo* to laboratory

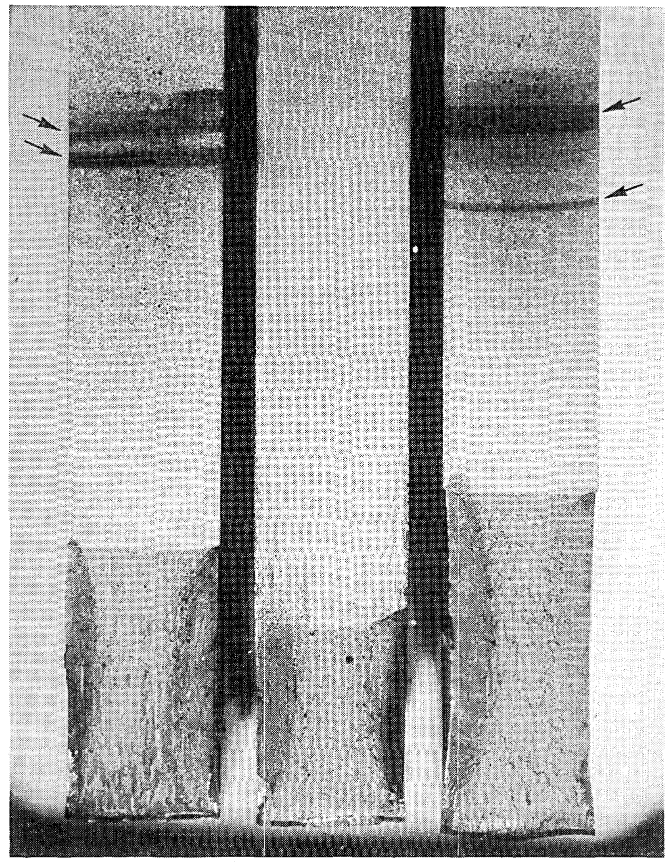


19 Near-threshold fatigue-crack propagation in high-strength martensitic En 24 steel (tempered at 500°C), showing influence of load ratio from $R = 0.17$ to 0.73 in moist laboratory air and in vacuum (10^{-3} Pa) environments (after Ref. 21)

air.[21,31,37–40,80,81] Cooke *et al.*[21] found an increase in ΔK_0 from 5 to 7 MN $m^{-3/2}$ at $R = 0.1$ and from 3 to 7 MN $m^{-3/2}$ at $R = 0.7$ for En 24 steel, tempered at 500°C, tested *in vacuo* (10^{-3} Pa) compared to air (Figs. 6 and 19). Similar studies on both high-purity and commercial-purity heats of the same steel (UK equivalent of 4340) tempered at 200°C, confirmed the marked reductions in near-threshold growth rates and the increase in ΔK_0 for tests *in vacuo* compared to air[81] (Fig. 6). In both cases, growth rates *in vacuo* were independent of load ratio over the range $R = 0.1$–0.7 (Refs. 21, 81). More recent data by Lindley and Richards[80] for 13%Cr martensitic stainless steels (En 56 and FV520B) also reveal appreciable increases in ΔK_0 *in vacuo* (10^{-3} Pa) compared with air at 70 Hz, although for these steels a slight R-ratio dependence on ΔK_0 was still apparent in the inert environment. Similar behaviour has been observed in non-ferrous alloys. In Ti–6Al–4V, for example, several independent studies[37,39,40] have shown increases in ΔK_0, for tests *in vacuo* compared with air, of the order of 3–8 MN $m^{-3/2}$ at $R = 0.35$. Furthermore, Bathias *et al.*[32] measured an increase in ΔK_0 from 3.5 to 6.1 MN $m^{-3/2}$ at $R = 0.01$ in T651 aluminium alloy by testing in a vacuum of 10^{-3} Pa compared with air. One is led to believe here that the difference between these two opposing sets of results lies principally in the nature of the 'inert' environment, i.e. high vacuum (better than 10^{-3} Pa) *v.* dry argon. First, how pure must the argon atmosphere be to produce no environmental action at the particular test frequency, and secondly, are rewelding effects occurring for cracking *in vacuo*? Typically impurity contents of argon atmospheres are generally quoted as being in the region of 20–50 ppm water vapour, plus smaller quantities of oxygen, nitrogen, and hydrogen, which correspond to an



20 Near-threshold fatigue-crack propagation in Ni—Cr—Mo—V A471 rotor steel (σ_y = 880 MN m^{-2}) tested at R = 0.35, 100 Hz in air and steam at 100°C, showing reduction in growth rates owing to environmental effects (after Ref. 120)



a b c

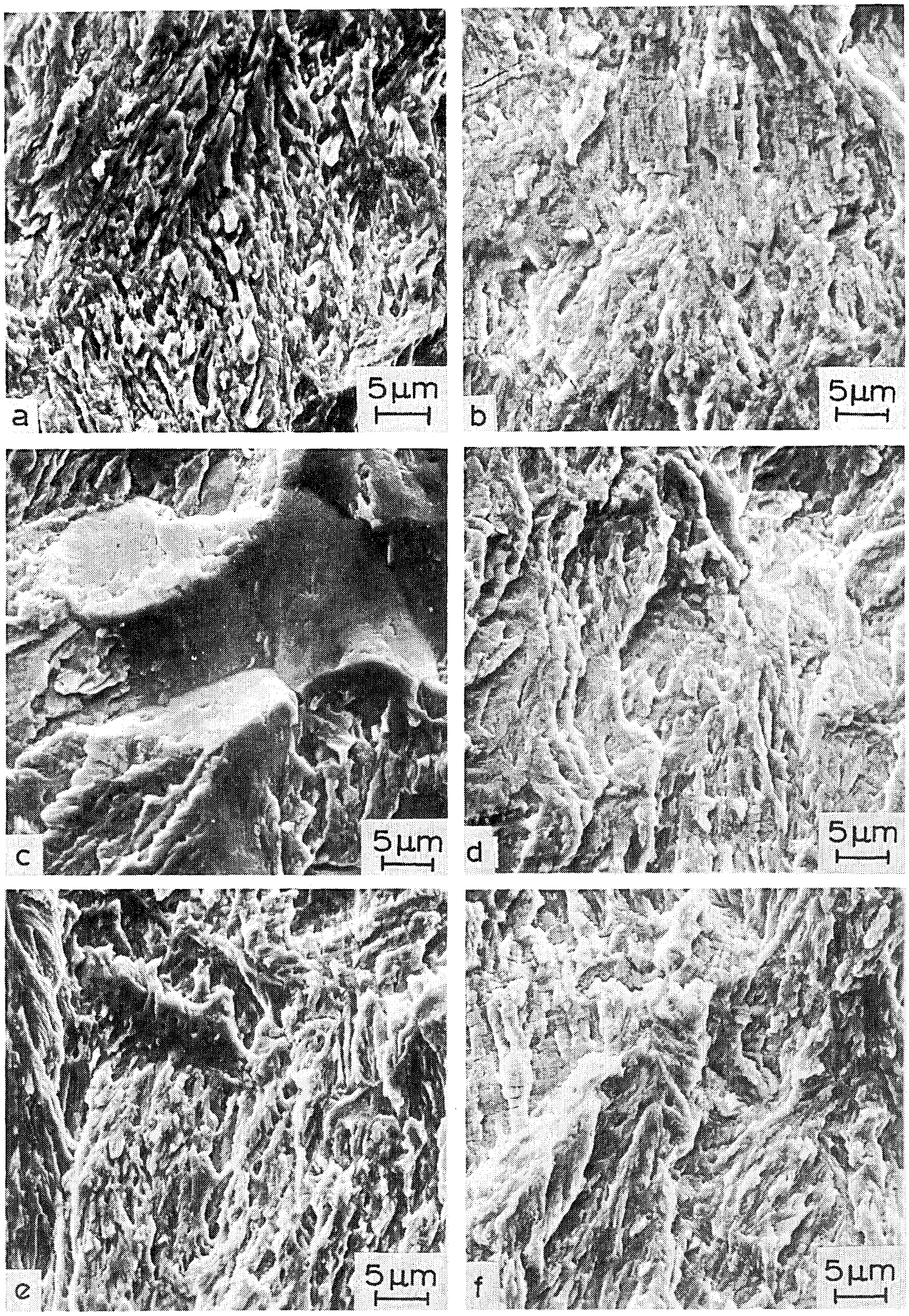
a and *c* near-threshold growth rates; *b* growth rates >10^{-6} mm/cycle

21 Bands of corrosion products on fatigue test-pieces of 9Ni—4Co high-strength steel (HP 9—4—20)

effective partial pressure of water vapour in the region of 2 to 10^{-2} Pa, constituting a poor vacuum.[37] Comparison of fatigue behaviour in Ti—6Al—4V in rigorously purified argon and in a 10^{-5} Pa vacuum revealed that propagation rates were almost identical at high growth rates, but differed by a factor of two in the region 10^{-5}—10^{-4} mm/cycle (Ref. 119). Differences at near-threshold growth rates would be expected to be even more significant. Accordingly, it is felt that the absence of environmental effects in the former set of results[6,7,35] can be rationalized in terms of the fact that comparisons were not made with a truly inert reference environment.[37] Further, the extremely high frequencies utilized in these tests would reduce significantly any environmental influence. Precise resolution of this issue, however, must await near-threshold fatigue-crack propagation data, for a single material, *in vacuo,* purified argon, and more aggressive environments, such as air, H_2O, H_2, H_2S, etc., over a range of frequencies. Additional evidence for an environmental effect on near-threshold behaviour can be found in the work of Pook and co-workers.[17,20] They observed that, for zero-tension loading in mild steel at ambient temperature, the value of ΔK_0 was reduced from 7.3 MN m$^{-3/2}$ in SAE 30 oil to 6.0 MN m$^{-3/2}$ in air, and significantly reduced to 2.0 MN m$^{-3/2}$ in brine.

There is also an added complication in aggressive or wet corrosive environments that near-threshold growth rates may be *reduced* compared with more inert atmospheres. This may arise from 'oxide-wedging' effects, where the presence of thick corrosion products in the crack may prevent further access for environmental species (akin to passivation), or from certain environments which promote crack branching or meandering. An example of these effects can be seen from the work of Tu and Seth[120] on A470 and A471 Ni—Cr—Mo—V rotor steels where near-threshold growth rates were reduced and the threshold increased when tests were conducted in steam as opposed to air at 100°C (Fig. 20).

Thus, although the extent of data is still somewhat limited, there is good direct evidence in the literature to support the hypothesis that near-threshold fatigue-crack propagation is environmentally sensitive. However, as with any corrosion-



a and *b* transgranular mode close to threshold at $\Delta K = 5.5$ MN m$^{-3/2}$; *c* and *d* transgranular and intergranular modes at $\Delta K = 6.5$ MN m$^{-3/2}$; *e* and *f* total absence of intergranular mode at higher growth rates at $\Delta K = 11$ MN m$^{-3/2}$

22 Fractography of near-threshold fatigue-crack growth in HP 9–4–20 high-strength aerospace steel ($\sigma_y = 1300$ MN m^{-2}), tested in moist laboratory air at $R = 0.1$

related process, there is unlikely to be a single unifying mechanism for the extent of the environmentally-assisted contribution to cracking in all materials, and clearly much research is still needed to characterize these interactions.

FRACTOGRAPHY OF NEAR-THRESHOLD GROWTH

Although the precise mechanisms of fatigue-crack propagation at low stress intensities are unknown, the morphology of near-threshold fracture surfaces has been well characterized. Macroscopically, a band of corrosion product is generally seen on fracture surfaces at the exact location where growth rates have been less than 10^{-6}–10^{-7} mm/cycle (Refs. 13, 21, 23, 102). An example of this is shown in Fig. 21 for ambient-temperature air tests on a high-strength 9Ni–4Co aerospace steel (HP 9–4–20); no such corrosion product is apparent at higher growth rates.[102] This is clearly indicative of significant environmental action at near-threshold levels, even in air.

Microscopically, near-threshold growth has been termed 'microstructurally-sensitive'[13,21,33,34,36,37,41] owing to the presence of isolated planar transgranular or intergranular facets within a flat, ductile transgranular mode. Examples of such microstructurally-sensitive growth are shown in Fig. 22 for ambient-temperature air tests on HP 9–4–20 steel, and should be compared with growth mechanisms at much higher growth rates (i.e. regimes B and C in Fig. 2) shown in Fig. 3*d* for the same steel. Note the presence of intergranular facets within a planar transgranular mode at stress intensities just above the threshold (Fig. 22*c*). Very close to the threshold, the proportion of facets is generally small (~1%), increasing to a maximum of anywhere from 10 to 80% as ΔK is increased, and then gradually diminishing at higher K values in the midgrowth rate regime. The maximum proportion of facets appears to occur when the *cyclic* plastic-zone size approaches the grain size,[13,21,33,34,36,37,41,81,99] whereas the disappearance of facets seems to occur when the *maximum* plastic-zone size exceeds the grain size.[99] Such facets are intergranular in ferritic steels,[10,13,21,23–27,41,63,78,80,81,99,102] and transgranular in austenitic stainless steels[30–32,41,121] and alloys of titanium,[33–37,41,70] aluminium,[41] copper,[41,44] and nickel.[41] A survey of such fractures presented by Beevers,[41] identifies the transgranular facets, termed 'cyclic cleavage', with {111} planes in stainless steels,[121] nickel superalloys, and Al 2219–T6, {001} planes in Al–Zn–Mg, and $\{10\bar{1}7\}$ in Ti–6Al–4V (Ref. 70), and so forth. It has been demonstrated, by spiking the loading sequence, that such cyclic-cleavage facets in titanium alloys[70] and 316 stainless steel,[121] form as a result of continued load cycling, and thus are not cleavage cracking in the true sense.

In high-strength martensitic steels, the presence of intergranular facets at near-threshold growth rates appears to be additionally related to the possibility of impurity-induced grain-boundary segregation.[25,27,81,122,123] This can be appreciated by varying the tempering temperature in 4340-type martensitic steels, where microstructures subject to impurity segregation during austenitizing,[122,123] tempered martensite embrittlement,[122,123] or temper embrittlement[25] show significantly more evidence of intergranular separation at near-threshold growth rates. Similar results have been observed by comparing high-purity heats with commercial (impure) heats of En 24 steel,[81] although, somewhat surprisingly, in many cases the presence or absence of intergranular fracture did not appear to have much direct effect on near-threshold crack velocities.

A more important and clearly interrelated aspect of such near-threshold facets is that they appear to be primarily environmentally-induced. First, facets are not observed at intermediate growth rates[25–27] and are largely suppressed for tests *in vacuo*, dry air, and helium atmospheres.[21,31,37,80,81] Furthermore, intergranular cracking in high-strength steels is commonly associated with hydrogen embrittlement,[68,69] and cleavage on $\{10\bar{1}7\}$ in Ti–Al alloys coincides with the habit plane of the hydrides.[71]

Aside from the presence of the facets, few mechanistic conclusions have been made from fractographic studies of near-threshold fatigue surfaces, principally because the fine-scale details (i.e. striation spacings should they exist) are below the resolution obtained from scanning microscopy or replica work. Fractures appear microscopically very planar and transgranular, and, in general, resemble low-magnification images of fractures at intermediate growth rates. Certain authors[14] have claimed that near-threshold propagation occurs locally by a Mode II mechanism involving regions of intense shear, although such features have not been observed in most fractographic studies on steels. For a more complete survey of near-threshold fracture-surface morphology the reader is referred to Refs. 41 and 81.

DISCUSSION

Environmental effects

It is apparent from the above that, unlike fatigue-crack propagation at intermediate growth rates, near-threshold behaviour is particularly sensitive to microstructural factors such as cyclic strength, grain size, grain-boundary composition, structure, and so forth, in addition to being markedly dependent upon the mean load or R ratio. This microstructural dependence is particularly encouraging for the metallurgist since it implies that there is a potential for the alloy design of materials with improved resistance to very high cycle, low-amplitude fatigue-crack propagation, a procedure which is extremely limited for the microstructurally-insensitive intermediate growth rates short of increasing the elastic modulus. Furthermore, on the basis of somewhat restricted and often conflicting results, it appears that such near-threshold growth is dependent on the nature of the environment. The question which arises is whether such microstructural effects can be traced to this environmental interaction or to some other phenomena such as crack closure.[82] In view of the limited data available and the complexities of varying environmental mechanisms between different materials, a conclusive answer to this question is not possible at present. However, by examining behaviour in high-strength steels, where the primary mechanism of environmental attack during fatigue-crack growth in moist air, water, and low-pressure hydrogen environments is hydrogen embrittlement,[72] some consistent, yet necessarily simplistic, explanation can be developed.[27]

In the absence of any environment, there are two physical models[124,125] for the existence of a fatigue-crack propagation threshold which give an expression for ΔK_0 in the form

$$\Delta K_0 = C\sqrt{\pi\rho^*}\ \sigma_F \quad . \quad . \quad . \quad . \quad . \quad . \qquad (10)$$

In the model of Weiss and Lal,[124] ρ^* is taken as the limiting microstructural dimension over which a certain local critical fracture stress σ_F must be attained ahead of the crack tip for propagation to occur, the constant C being numerically equal to $\sqrt{3/2}$. The model of Sadananda and Shaninian,[125] on the other hand, equates the threshold with the minimum stress σ_F to nucleate a dislocation at the crack tip, where ρ^* is the minimum distance that the dislocation can exist away from the tip (taken as the Burgers vector). In this form, equation (10) does not predict any variation of ΔK_0 with mechanical and microstructural factors (e.g. load ratio, strength, and so forth) under totally inert conditions, consistent with the very limited experimental data available.[21,32,37,81]

However, in the presence of 'hydrogen-producing' environments, cyclic stressing will lead to chemical reactive surface at the crack tip where atomic hydrogen, evolved by cathodic reactions or adsorbed from the gas phase, can enter the lattice and diffuse under the driving force of the stress gradient into the region ahead of the crack tip. In the proposed hydrogen-embrittlement mechanisms for high-strength steels,[38,68,69] hydrogen is considered to diffuse to the point of maximum hydrostatic tension (i.e. maximum dilatation), and lead to a reduction in cohesive strength. The enrichment of hydrogen ahead of the crack tip is a function of the magnitude of the hydrostatic tension $\bar{\sigma}$, and is given thermodynamically by

$$\frac{C_H}{C_0} = \exp\left(\frac{\bar{\sigma}\bar{v}}{R_0 T}\right) \quad . \quad . \quad . \quad . \quad . \quad . \qquad (11)$$

where C_H is the local hydrogen concentration at the point of maximum dilatation, C_0 the equilibrium hydrogen concentration in the unstressed lattice, $\bar{v}$ the partial molar volume of hydrogen in iron, R_0 the gas constant, and T the absolute temperature.[68] Thus, if the reduction in cohesive strength due to hydrogen $\Delta\sigma_H$ is assumed to be proportional to C_H, i.e. $\Delta\sigma_H = \alpha C_H$, then in the presence of a hydrogen-producing environment, the Weiss and Lal expression for the threshold becomes:

$$\Delta K_0 = \sqrt{3\pi\rho^*/2}\ (\sigma_F - \Delta\sigma_H) \quad . \quad . \quad . \quad . \qquad (12)$$

By combining equations (11) and (12), and incorporating an expression for the maximum hydrostatic tension, it can be shown[27] that

$$\Delta K_0 = \sqrt{3\pi\rho^*/2}\left(\frac{\sigma_F - \alpha C_0 \exp(B\sigma_y)}{1 + \dfrac{\alpha C_0 \rho^{*1/2}}{(1-R)} B' \exp(B\sigma_y)}\right) \quad . \quad . \quad . \qquad (13)$$

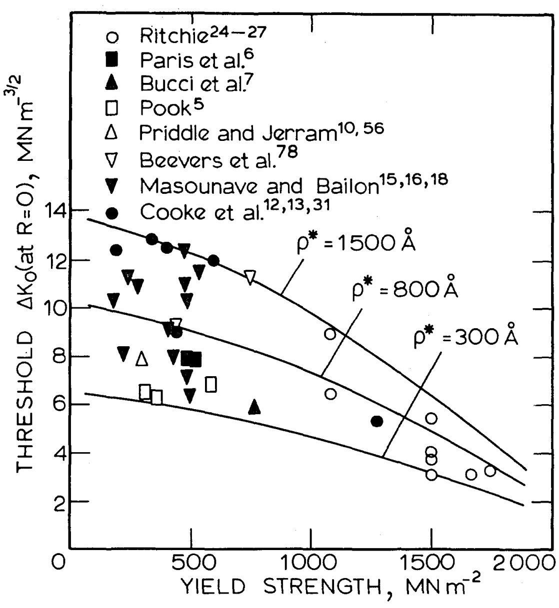
where σ_y is the yield strength, R the load ratio, and B and B' are constants dependent upon temperature. At ambient temperatures in steels $B = 8 \times 10^{-4}$ (MN m^{-2})$^{-1}$ and $B' = 5 \times 10^{-2}$ (MN m$^{-3/2}$)$^{-1}$.

Because of uncertainty in the magnitude of various parameters in the model[68,69] for hydrogen embrittlement (i.e. C_0 and α), it is perhaps premature at this stage to utilize the above relationships in a predictive capacity. However, such equations do provide a useful basis for rationalizing the observed near-threshold fatigue-crack propagation behaviour in high-strength steels. First, the occurrence of intergranular fracture in high-strength steels during near-threshold crack growth in moist air (Fig. 22) or hydrogen-containing gas atmospheres, and the absence of such fracture for tests under inert conditions[21,31,80,81] lends strong support to a hydrogen-embrittlement contribution to crack growth. Secondly, on the basis of equation (12), values of ΔK_0 should be larger in inert environments, consistent with all reported experimental results (e.g. Figs. 6 and 19). Thirdly, the load-ratio and yield-strength effects on near-threshold growth (e.g. Figs. 5–7 and 19, and 4 and 9, respectively) are capable of interpretation since an increase in either parameter will result in a larger hydrostatic stress state $\bar{\sigma}$, which, in turn, raises the local concentration of hydrogen ahead of the crack tip (equation (11)) leading to a reduction in ΔK_0 (equations (12) and (13)). Moreover, since the influence of load ratio arises from such environmental interactions, it follows that in inert atmospheres, near-threshold growth rates and threshold values should be less affected by load ratio, consistent with results[21,80,81] for austenitic and martensitic low-alloy and stainless steels tested in air and vacuum (e.g. Figs. 6 and 19). Thus, the observed effects of load ratio and strength on fatigue-threshold behaviour in high-strength steels can be thought of in terms of an enhanced environmental contribution to crack growth arising from an increase in hydrostatic tension. Additionally, increasing the load ratio will raise K_{max}, which, in turn, leads to a larger plastic stress gradient ahead of the crack tip, thus providing a greater driving force for the transport of hydrogen into the region of maximum triaxiality.

Results,[26,102] showing improved resistance to near-threshold fatigue-crack propagation in tempered martensitic structures containing almost continuous networks of interlath retained austenite (e.g. Fig. 14) are also consistent with this environmental model, since hydrogen diffusivity in the fcc austenite is three to four orders of magnitude lower than in the bcc phase.[101,126] Furthermore, the lower growth rates observed[27] in coarser grained structures (e.g. Fig. 11) can be rationalized in terms of a lower probability of hydrogen atoms reaching a grain boundary,[99] as discussed above, although more recent data showing a decrease in ΔK_0 with increasing prior austenite grain size (Fig. 12) cast some doubt on this explanation.[29] However, heat treatments used to vary prior austenite grain size in steels (e.g. by increasing austenitizing temperatures) are liable to result in other microstructural changes,[29] such as varying the distribution of grain-boundary embrittling impurity elements.[127] Finally, the effect of such impurity segregation in reducing the threshold (Fig. 17) is readily explained in terms of this analysis, since the presence of residual impurities in grain boundaries will also lead to reduced cohesion.[25] There is also a possibility of synergistic effects between hydrogen and impurity atoms[25] arising from the fact that such impurities may act as recombination poisons for atomic hydrogen.[128]

It is apparent, therefore, that by considering the environmental contribution of hydrogen to crack growth, near-threshold fatigue behaviour in high-strength steels can be usefully rationalized in terms of microstructure and load-ratio effects. As described in Ref. 27, the model can be utilized semiquantitatively by assigning reasonable values to equation (13), such that the experimentally observed trend of decreasing ΔK_0 with increasing strength in steels can be correctly reproduced (Fig. 23). Further verification, however, must await more extensive data showing the influence of mechanical and microstructural factors on near-threshold fatigue-crack propagation in high-strength steels under environmental and, particularly, inert conditions.

It is important to realize that the model described above is only strictly valid for higher strength steels where the environmental contribution to cracking in the presence of hydrogen-producing environments can be modelled in terms of a 'decohesion' mechanism of hydrogen embrittlement.[38,68,69] Lower strength steels, such as mild steel and certain pressure-vessel steels (e.g. A387) where yield strengths are below 500 MN m^{-2}, similarly show marked effects of microstructure and mean stress at near-threshold growth rates in moist air, and yet the possible enrichment of hydrogen ahead of the crack tip, predicted from equation (11), would be very small owing to the lower magnitude of the hydrostatic tension. Thus, any explanation of the characteristics of near-threshold fatigue-crack growth behaviour for such low-strength steels in terms of a decohesion hydrogen-embrittlement model must be regarded as unlikely. However, as discussed above, such steels do show marked environmentally-enhanced fatigue-crack propagation above 10^{-6} mm/cycle in H_2, H_2S, and water environments,[47–49,73–75] and there are indications that similar hydrogen-assisted crack



23 Variation of published data on threshold ΔK_0 at $R = 0$ with strength for steels, showing predictions of environmental model (equation (13))

propagation occurs at much slower near-threshold growth rates.[83] Mechanisms for this hydrogen-assisted growth, which certain authors[47,49] have termed hydrogen embrittlement, are presently unknown, yet are unlikely to be similar to classic hydrogen-embrittlement mechanisms (i.e. decohesion) operative in higher strength steels. Thus, until this hydrogen-induced contribution to cracking is understood, particularly close to ΔK_0, rationalization of near-threshold fatigue-crack propagation behaviour in low-strength steels is clearly not possible.

Crack-closure effects

To justify the existence of a fatigue-crack propagation threshold and to explain various characteristics of subsequent near-threshold crack growth, many authors have used arguments based on the phenomenon of crack closure.[6,7,14,19,28,35,42,79,129,130] However, in the light of existing information, the role of closure on ultralow crack-propagation behaviour is somewhat open to question, as discussed below.

The concept of crack closure, first described by Elber[82] for crack growth at high stress intensities, relies on the fact that, as a result of plastic deformation left in the wake of a growing fatigue crack, it is possible that some closure of the crack surfaces may occur at positive loads during the loading cycle. Since the crack is unable to propagate while it remains closed, the net effect of closure is to reduce the applied ΔK value (computed from applied load and crack-length measurements) to some lower effective value ΔK_{eff} actually experienced at the crack tip. It follows from this argument that a threshold for crack growth will be reached when the crack remains closed throughout the entire loading cycle. Certain Japanese workers have claimed to verify this by reporting that at the threshold the range of effective stress intensity ΔK_{eff}, based on surface compliance measurements, has no finite value.[129,130] Subsequent more extensive studies on a wider range of steels, however, failed to substantiate this claim.[19] Other authors[6,7,28,35,42] have utilized the closure concept to explain the load-ratio effect on the assumption that as the mean load is raised, the crack will remain open for a larger portion of the cycle, thereby increasing ΔK_{eff} and hence the growth rate. Such arguments have been applied to near-threshold behaviour in steels,[6,7,28] titanium,[35] and aluminium[42] alloys, but with little or no experimental verification. There are, however, several basic problems with this explanation for near-threshold behaviour. First, it is difficult using closure arguments to account for the fact that the influence of load ratio becomes minimal at intermediate growth rates (regime B in Fig. 2) where closure is equally likely to occur. To counter this, it has been suggested[14,131] that the closure effect may be enhanced in the near-threshold region because of the presence of a shear mode of fracture although, as discussed above, the notion that near-threshold failure mechanisms are essentially Mode II in nature does not appear to be universal and, further, it is by no means certain why this in itself should increase closure loads. Secondly, certain workers have observed that the level of closure in inert environments is greater than[132–134] (or at least equal to [135]) the level in air. If the origin of the load-ratio effect were simply crack closure, this would imply that near-threshold growth rates would be more sensitive to load ratio in inert atmospheres, which is contrary to all experimental observations.[21,31,37,39,40,80,81] Thirdly, there is now a large body of evidence[131,136,137] to suggest that crack closure is essentially a surface (plane stress) effect, having a minimal consequence on crack growth under plane-strain conditions. Since near-threshold growth is invariably measured under plane-strain conditions, it seems unlikely that crack closure could be primarily responsible for the marked dependence on load ratio. The issue, however, must remain somewhat unresolved since there is still a certain degree of uncertainty in the experimental measurement of closure; the electrical-potential and compliance techniques used often yield inconsistent results.[138,139]

Undoubtedly, crack closure does occur during fatigue-crack growth, but in the light of recent evidence by Wei *et al.*,[140] which showed no sensible correlation between crack-propagation kinetics and ΔK_{eff}, models based on this concept to explain crack-growth behaviour patterns at near-threshold levels must be regarded as questionable.

CONCLUDING REMARKS

In this paper, an attempt has been made critically to review existing information on the characteristics of near-threshold fatigue-crack propagation in steels with particular emphasis on the possible role of environmental contributions to crack growth. However, despite a rapidly increasing interest in this field over the past five years, there is still virtually no mechanistic basis for near-threshold behaviour and still a comparative lack of reliable engineering data. Further, it is clearly apparent that many fundamental questions remain unanswered, such as the precise role of the environment especially in low-strength steels and the relevance of crack closure. Additionally, the usefulness of threshold data to the problem of short cracks, which are most often found in service, is largely unsolved, particularly in the light of the relationship between threshold data, representing the fracture-mechanics approach to growth (or lack of growth) from long cracks, and the *S–N* and fatigue-limit data, representing principally initiation from short cracks. At first glance, it would seem that both parameters are closely related since they both describe limits for fatigue damage, and yet, as described, by raising the strength of a steel the fatigue limit is generally increased whereas the threshold is decreased. Thus, for alloy-design purposes, procedures designed to optimize high-cycle crack-propagation resistance will not necessarily guarantee similar optimum resistance to crack initiation. Hence, before recommendations can be made for the selection of a suitable material to withstand high-frequency low-amplitude cyclic loading, the relative importance of crack initiation and crack propagation must be clearly established.

The relevance of near-threshold crack-growth data, however, cannot be underestimated. Defects introduced during fabrication or developed in service become potential sources for catastrophic failure from subcritical cracking at seemingly immeasurable growth rates where components are subject to high-frequency low-amplitude cyclic loading. This is especially important where such fluctuations are superimposed on a mean operating stress which results in extremely high load ratios. Examples of the use of threshold data for design and post-failure analysis are now becoming more numerous, particularly in the electrical-supply industry for high-speed rotating equipment and components subject to acoustic fatigue. However, in other applications, such as large-scale nuclear and coal-conversion pressure vessels where small operational transient stresses may result in significant near-threshold crack growth over long periods of time, relevant low growth-rate data for typical environments are still lacking. Although widely recognized in certain countries (*see* e.g. Refs. 2–4, 80, and 141), relatively few studies are in progress in the USA and, despite the achievements of the last five years, there still remains an important need for both fundamental and applied research on near-threshold fatigue behaviour.

ACKNOWLEDGMENT

The work was supported by the Fossil Energy Research Division of the US Department of Energy under contract No. EX—76—A—01—2295.

REFERENCES

1. S.W. Hopkins, C.A. Rau, G.R. Leverant, and A. Yuen: 'Fatigue crack growth under spectrum loads', STP 595, 125; 1976, Philadelphia, Pa, American Society for Testing and Materials.
2. I. Gray, M.D. Heaton, and G. Oates: Proc. of British Steel Corporation Conf. on 'Mechanics and mechanisms of crack growth', Cambridge University, April 1974, 264.
3. M.D. Heaton: 'The mechanics and physics of fracture', 34; 1975, London, The Metals Society.
4. J.R. Griffiths and G. Oates: *Met. Sci.*, 1977, **11**, 285.
5. L.P. Pook: 'Stress analysis and growth of cracks', STP 513, 106; 1972, Philadelphia, Pa, American Society for Testing and Materials.
6. P.C. Paris, R.J. Bucci, E.T. Wessel, W.G Clark, and T.R. Mager: *ibid.*, 141.
7. R.J. Bucci, W.G. Clark, and P.C. Paris: *ibid.*, 177.
8. M. Klesnil and P. Lukáš: *Mater. Sci. Eng.*, 1972, **9**, 231.
9. M. Klesnil and P. Lukáš: *Eng. Fract. Mech.*, 1972, **4**, 77.
10. E.K. Priddle: 'Constant amplitude fatigue crack propagation in a mild steel at low stress intensity: the effect of mean stress on propagation rate', Technical Report RD/B/N2233, May 1972, Berkeley Nuclear Laboratories, Central Electricity Generating Board.
11. H. Kitagawa, H. Nishitani, and J. Matsumoto: 'Proc. 3rd int. congr. on fracture', Vol. 5, Paper V—444/A; 1973, Düsseldorf, Verein Deutscher Eisenhüttenleute.
12. R.J. Cooke and C.J. Beevers: *Eng. Fract. Mech.*, 1973, **5**, 1061.
13. R.J. Cooke and C.J. Beevers: *Mater. Sci. Eng.*, 1974, **13**, 201.
14. A. Otsuka, K. Mori, and T. Miyata: *Eng. Fract. Mech.*, 1975, **7**, 429.
15. J. Masounave and J.-P. Baïlon: *Scr. Metall.*, 1975, **9**, 723.
16. J. Masounave and J.-P. Baïlon: *ibid.*, 1976, **10**, 165.
17. L.P. Pook and A.F. Greenham: 'Proc. fatigue testing and design conf.', Vol. 2, 30.1; 1976, London, Society of Environmental Engineers.
18. J. Masounave and J.-P. Baïlon: 'Proc. 2nd int. conf. on mechanical behavior of materials', 636; 1976, Metals Park, Ohio, American Society for Metals.
19. M. Kikukawa, M. Jono, and K. Tanaka: *ibid.*, 716.
20. L.P. Pook: *Met. Sci.*, 1977, **11**, 382.
21. R.J. Cooke, P.E. Irving, G.S. Booth, and C.J. Beevers: *Eng. Fract. Mech.*, 1975, **7**, 69.
22. J. Mautz and V. Weiss: 'Cracks and fracture', STP 601, 154; 1976, Philadelphia, Pa, American Society for Testing and Materials.
23. E.K. Priddle: 'Fracture 1977', (ed. D.M.R. Taplin), Vol. 2, 1249; 1977, Waterloo, Ont., University of Waterloo Press.
24. R.O. Ritchie: *ibid.*, 1325.
25. R.O. Ritchie: *Metall. Trans.*, 1977, **8A**, **1131**.
26. R.O. Ritchie: *J. Eng. Mater. Technol. (Trans. ASME, H)*, 1977, **99**, 195.
27. R.O. Ritchie: *Met. Sci.*, 1977, **11**, 368.
28. A. Ohta and E. Sasaki: *Eng. Fract. Mech.*, 1977, 9, 307.
29. M.F. Carlson and R.O. Ritchie: *Scr. Metall.*, 1977, **11**, 1113.
30. A.C. Pickard, R.O. Ritchie, and J.F. Knott: *Met. Technol.*, 1975, **2**, 253.
31. E.K. Priddle, F. Walker, and C. Wiltshire: 'Proc. conf. on influence of environment on fatigue', 137; 1977, London, Institution of Mechanical Engineers.
32. C. Bathias, A. Pineau, J. Pluvinage, and P. Rabbe: 'Fracture 1977', (ed. D.M.R. Taplin), Vol. 2, 1283; 1977, Waterloo, Ont., University of Waterloo Press.
33. J.L. Robinson and C.J. Beevers: *Met. Sci. J.*, 1973, **7**, 153.
34. J.L. Robinson, P.E. Irving, and C.J. Beevers: 'Proc. 3rd int. congr. on fracture', Vol. 5, Paper V—343; 1973, Düsseldorf, Verein Deutscher Eisenhüttenleute.
35. R.J. Bucci, P.C. Paris, R.W. Hertzberg, R.A. Schmidt, and A.F. Anderson: 'Stress analysis and growth of cracks', STP 513, 125; 1972, Philadelphia, Pa, American Society for Testing and Materials.
36. P.E. Irving and C.J. Beevers: *Mater. Sci. Eng.*, 1974, **14**, 229.

P.E. Irving and C.J. Beevers: *Metall. Trans.*, 1974, **5**, 391.

38. A.W. Thompson and I.M. Bernstein: 'Advances in corrosion science and technology', (eds. M.G. Fontana and R.W. Staehle), Vol. 7, 53; 1979, New York, Plenum Press.
39. R. Ebara, K. Inoue, S. Crosby, J. Groeger, and A.J. McEvily: 'Proc. 2nd int. conf. on mechanical behavior of materials', 685; 1976, Metals Park, Ohio, American Society for Metals.
40. A.J. McEvily and J. Groeger: 'Fracture 1977', (ed. D.M.R. Taplin), Vol. 2, 1293; 1977, Waterloo, Ont., University of Waterloo Press.
41. C.J. Beevers: *Met. Sci.*, 1977, **11**, 362.
42. R.A. Schmidt and P.C. Paris: 'Progress in flaw growth and fracture toughness testing',

STP 536, 79; 1973, Philadelphia, Pa, American Society for Testing and Materials.
43. J. F. Knott and A. C. Pickard: *Met. Sci.*, 1977, **11**, 399.
44. A. C. Pickard, R. O. Ritchie, and J. F. Knott: 'Proc. 4th int. conf. on strength of metals and alloys', Aug. 1976, 473; Nancy, France, INPL.
45. H. Kitagawa and S. Takahashi: 'Proc 2nd int. conf. on mechanical behavior of materials', 627; 1976, Metals Park, Ohio, American Society for Metals.
46. F. P. Ford and T. P. Hoar: 'The microstructure and design of alloys', Vol. 1, 467; 1974, London, The Metals Society.
47. O. Vosikovsky: *J. Eng. Mater. Technol. (Trans. ASME, H)*, 1975, **97**, 298.
48. T. R. Mager, D. M. Moon, and J. D. Landes: *J. Pressure Vessel Technol. (Trans. ASME, J)*, 1977, 99, 238.
49. W. H. Bamford and D. M. Moon: 'Corrosion 79', Proc. NACE Meeting, March 1979; Houston, Texas, NACE.
50. Proc. of Workshop on Materials Problems and Research Opportunities in Coal Conversion, April 1974, Vols. I and II; Washington, DC, ERDA.
51. R. O. Ritchie, G. G. Garrett, and J. F. Knott: *Int. J. Fract. Mech.*, 1971, **7**, 462.
52. R. O. Ritchie and K. J. Bathe: *Int. J. Fract.*, 1979, **15**, 47.
53. G. H. Aronson and R. O. Ritchie: *J. Test. Eval.*, 1979, **7**, 208.
54. P. S. Pao, W. Wei, and R. P. Wei: 'Environment sensitive fracture of engineering materials', (ed. Z. A. Foroulis), 565; 1979, New York, AIME.
55. S. J. Hudak, A. Saxena, R. J. Bucci, and R. C. Malcolm: Third Semi-Annual Report to AFML, Westinghouse Research and Development Laboratories, Pittsburgh, March 1977.
56. K. Jerram and E. K. Priddle: *J. Mech. Eng. Sci.*, 1973, **15**, 271.
57. A. Saxena, S. J. Hudak, J. K. Donald, and D. W. Schmidt: Scientific Paper 77—IE7—FANWL—P1, Westinghouse Research and Development Laboratories, Pittsburgh, May 1977.
58. H. H. Johnson and P. C. Paris: *Eng. Fract. Mech.*, 1968, **1**, (1), 3.
59. P. C. Paris and F. Erdogan: *J. Basic Eng. (Trans. ASME. D)*, 1963, **85**, 528.
60. C. Laird and G. C. Smith: *Philos. Mag.*, 1962, **7**, 847.
61. R. M. N. Pelloux: *Eng. Fract. Mech.*, 1970, **1**, (4), 697.
62. C. E. Richards and T. C. Lindley: *ibid.*, 1972, **4**, 951.
63. R. O. Ritchie and J. F. Knott: *Acta. Metall.*, 1973, **21**, 639.
64. R. O. Ritchie and J. F. Knott: *Mater. Sci. Eng.*, 1974, **14**, 7.
65. T. C. Lindley, C. E. Richards, and R. O. Ritchie: *Metall. Met. Form.*, 1976, **43**, 268.
66. F. A. McClintock: 'Fatigue crack propagation', STP 415, 170; 1967, Philadelphia, Pa, American Society for Testing and Materials.
67. J. K. Tien: 'Effect of hydrogen on behavior of materials', (eds. A. W. Thompson and I. M. Bernstein), 301; 1976, New York, AIME.
68. R. A. Oriani and P. H. Josephic: *Acta Metall.*, 1974, **22**, 1065.
69. W. W. Gerberich and Y. T. Chen: *Metall. Trans.*, 1975, **6A**, 271.
70. N. E. Paton, J. C. Williams, J. C. Chesnutt, and A. W. Thompson: Proc. AGARD Conf. on 'Alloy design for fatigue and fracture resistance', No. 185, April 1975, 4—1; Neuilly sur Seine, France, NATO.
71. N. E. Paton and R. A. Spurling: *Metall. Trans.*, 1976, **7A**, 1769.
72. R. P. Wei and J. D. Landes: *Mater. Res. Standards*, 1969, **9**, 25.
73. H. G. Nelson: 'Proc. 2nd int. conf. on mechanical behavior of materials', 690; 1976, Metals Park, Ohio, American Society for Metals.
74. H. G. Nelson: 'Effects of hydrogen on behavior of materials', (eds. A. W. Thompson and I. M. Bernstein), 602; 1976, New York, AIME.
75. R. P. Wei and G. W. Simmons: 'Fracture mechanics and surface chemistry studies of steels for coal gasification systems', Lehigh University Report No. IFSM—77—87, Dec. 1977.
76. O. Vosikovsky: *J. Test. Eval.*, 1978, **6**, 175.
77. W. J. Plumbridge: *J. Mater. Sci.*, 1972, **7**, 939.
78. C. J. Beevers, R. J. Cooke, J. F. Knott, and R. O. Ritchie: *Met. Sci.*, 1975, **9**, 119.
79. A. Ohta and E. Sasaki: *Eng. Fract. Mech.*, 1977, **9**, 655.
80. T. C. Lindley and C. E. Richards: 'Fatigue crack growth at low stresses in steels', Central Electricity Generating Board Laboratory Note No. RD/L/N 135/78, Aug. 1978, Central Electricity Research Laboratories, Leatherhead, Surrey.
81. P. E. Irving and A. Kurzfeld: *Met. Sci.*, 1978, **12**, 495.
82. W. Elber: 'Damage tolerance in aircraft structures', STP 486, 230; 1971, Philadelphia, Pa, American Society for Testing and Materials.
83. S. Suresh, C. M. Moss, and R. O. Ritchie: 'Proc. 2nd Japan Institute of Metals int. symp. on hydrogen in metals', Sendai, Japan, Nov. 1979, Paper 27B23.
84. R. O. Ritchie, R. F. Smith, and J. F. Knott: *Met. Sci.*, 1975, **9**, 485.
85. M. H. El Haddad, T. H. Topper, and K. N. Smith: *J. Eng. Mater. Technol. (Trans. ASME, H)*, 1979, **101**, 42.
86. M. H. El Haddad, T. H. Topper, and K. N. Smith: *Eng. Fract. Mech.*, 1979, **11**, 573.
87. N. E. Dowling: 'Cyclic stress—strain and plastic deformation aspects of fatigue crack growth', STP 637, 97; 1977, Philadelphia, Pa, American Society for Testing and Materials.
88. M. E. Fine and R. O. Ritchie: 'Fatigue and microstructure', (ed. M. Meshii), 245; 1979,

Metals Park, Ohio, American Society for Metals.
89. J. Barsom, E. J. Imhof, and S. T. Rolfe: *Eng. Fract. Mech.*, 1970, **2**, 301.
90. J. P. Benson and D. V. Edmonds: *Met. Sci.*, 1978, **12**, 223.
91. O. Vosikovsky: *Eng. Fract. Mech.*, 1979, **11**, 595.
92. A. J. McEvily: 'Recent research on mechanical behavior of solids'; 1979, Tokyo, University of Tokyo Press.
93. P. Kuhn and H. Hardrath: NACE Technical Report No. NACE TN 2805, 1952.
94. R. O. Ritchie: 'Proc. int. conf. on fundamentals of tribology'; 1979, Cambridge, Mass., MIT Press.
95. R. M. Pelloux: 'Ultrafine-grain metals', 231; 1970, Syracuse, NY, Syracuse University Press.
96. A. W. Thompson and R. J. Bucci: *Metall. Trans.*, 1973, **4**, 1173.
97. A. W. Thompson: *Eng. Fract. Mech.*, 1975, **7**, 61.
98. G. R. Yoder, L. A. Cooley, and T. W. Crooker: *J. Eng. Mater. Technol. (Trans. ASME, H)*, 1979, **101**, 86.
99. J. D. Frandsen and H. L. Marcus: *Scr. Metall.*, 1975, **9**, 1089.
100. J. F. Lesser and W. W. Gerberich: *Metall. Trans.*, 1976, **7A**, 953.
101. R. O. Ritchie, M. H. Castro Cedeño, V. F. Zackay, and E. R. Parker: *ibid.*, 1978, **9A**, 35.
102. R. O. Ritchie, V. A. Chang, and N. E. Paton: *J. Fatigue Eng. Mater. Structures*, 1979, 1.
103. A. S. Tetelman: 'Fundamental aspects of stress corrosion cracking', (ed. R. W. Staehle), 446; 1969, Houston, Texas, NACE.
104. E. E. Fletcher, W. E. Berry, and G. A. Elsen: *ibid.*, 457.
105. W. W. Gerberich: 'Hydrogen in metals', (eds. I. M. Bernstein and A. W. Thompson), 115; 1974, Metals Park, Ohio, American Society for Metals.
106. C. B. Gilpin and N. A. Tiner: *Corrosion*, 1966, **22**, 271.
107. C. S. Carter: *ibid.*, 1969, **25**, 423.
108. R. O. Ritchie: 'Environment sensitive fracture of engineering materials', (ed. Z. A. Foroulis), 538; 1979, New York, AIME.
109. H. Suzuki and A. J. McEvily: *Metall. Trans.*, 1979, **10A**, 475.
110. G. J. Fowler: *Mater. Sci. Eng.*, 1979, **39**, 121.
111. A. D. Wilson: *J. Pressure Vessel Technol. (Trans. ASME, J)*, 1977, **99**, 459.
112. P. Shaninian, H. H. Smith, and J. R. Hawthorne: *Weld. J. Res. Suppl.*, 1972, **51**, 527.
113. L. A. James: *At. Energy Rev.*, 1976, **14**, 37.
114. N. G. Nelson and D. P. Williams: *Metall. Trans.*, 1970, **1**, 63.
115. G. W. Simmons, P. S. Pao, and R. P. Wei: *ibid.*, 1978, **9A**, 1147.
116. S. J. Hudak and R. P. Wei: *ibid.*, 1976, **7A**, 235.
117. R. P. Gangloff and R. P. Wei: *ibid.*, 1977, **8A**, 1043.
118. P. G. Shewmon: *ibid.*, 1976, **7A**, 279.
119. R. P. Wei and D. L. Ritter: *J. Mater.*, 1972, **7**, 240.
120. L. K. L. Tu and B. B. Seth: *J. Test. Eval.*, 1978, **6**, 66.
121. E. K. Priddle and F. E. Walker: *J. Mater. Sci.*, 1976, **11**, 386.
122. R. O. Ritchie: *Eng. Fract. Mech.*, 1975, **7**, 187.
123. R. O. Ritchie: PhD thesis, University of Cambridge, 1973.
124. V. Weiss and D. N. Lal: *Metall. Trans.*, 1974, **5**, 1946.
125. K. Sadananda and P. Shaninian: *Int. J. Fract.*, 1977, **13**, 585.
126. J. H. Shively, R. F. Hehemann, and A. R. Troiano: *Corrosion*, 1966, **22**, 253.
127. G. Clark, R. O. Ritchie, and J. F. Knott: *Nature Phys. Sci.*, 1972, **239**, 104.
128. R. M. Latanision and H. Opperhauser: *Metall. Trans.*, 1974, **5**, 483.
129. H. Nisitani and K. Takeo: *Eng. Fract. Mech.*, 1974, **6**, 253.
130. A. Ohta and E. Sasaki: *Int. J. Fract.*, 1975, **11**, 1049.
131. A. J. McEvily: *Met. Sci.*, 1977, **11**, 274.
132. O. Buck, J. D. Frandsen, and H. L. Marcus: *Eng. Fract. Mech.*, 1975, **7**, 167.
133. P. E. Irving, J. L. Robinson, and C. J. Beevers: *ibid.*, 619.
134. P. E. Irving, J. L. Robinson, and C. J. Beevers: *Int. J. Fract.*, 1973, **9**, 105.
135. V. Bachmann and D. Munz: *ibid.*, 1975. **11**. 713.
136. T. C. Lindley and C. E. Richards: *Mater. Sci. Eng.*, 1974, **14**, 281.
137. F. J. Pitoniak, A. F. Grandt, L. T. Montulli, and P. F. Packman: *Eng. Fract. Mech.*, 1974, **6**, 663.
138. C. K. Clarke and G. C. Cassatt: *ibid.*, 1977, **9**, 675.
139. V. Bachmann and D. Munz: *ibid.*, 1979, **11**, 61.
140. K. D. Unangst, T. T. Shih, and R. P. Wei: *ibid.*, 1977, **9**, 725.
141. 'Crack arrest threshold ΔK_{th}'; French Round Robin Program on Threshold Testing, Groupe de la Commission de Fatigue, May 1978. (Translated by ASTM for E.24.04 Subcommittee on Subcritical Crack Growth.)