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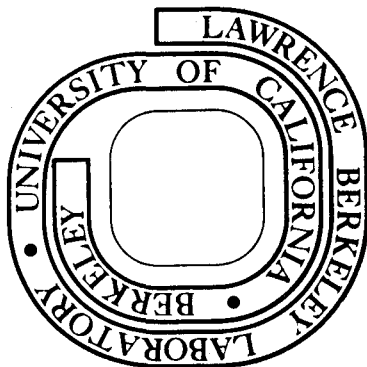
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THE MECHANICS AND MECHANISMS OF
FATIGUE CRACK GROWTH IN METALS,

A Review

*T.C. Lindley, *C.E. Richards and †R.O. Ritchie

ABSTRACT

The influence of alternating and mean stress intensity on fatigue crack propagation in metals has been studied in relation to the different microstructure, mean (or maximum) stress intensity and specimen thickness, particularly in the range of medium and high growth rates.

The simple growth law

$$\frac{da}{dN} = C \Delta K^m$$

was found to be obeyed with little variation in C and m for the striation mechanism of crack growth over a wide range of testing conditions. Modification of this law is necessary for very low and very high rates of propagation where non-striation (cleavage, intergranular, and void coalescence) mechanisms are involved. The results are briefly discussed in relation to fatigue crack growth in practice.

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I. INTRODUCTION

Very broadly, there are two important types of cyclic loading that give rise to concern in service. The first involves very large numbers of stress reversals, probably small in magnitude, in components such as rapidly rotating machinery and structures subjected to vibrations or pressure fluctuations. Design is then based on the number of cycles required to fracture small laboratory specimens at given stress amplitudes (the S-N curve). At the other extreme, components may suffer damage as a result of small numbers of large load changes incurred for example by filling and emptying a pressure vessel or stopping and starting a turbine. Such large machines and pressure vessels, especially those with welds, may contain defects or fabrication cracks. The S-N curve provides no information on the remaining life of a defective component since N is the number of cycles, both to initiate and grow a crack to complete failure. Under these circumstances, there was frequently no choice but to remove the defect.

The subject of fracture mechanics has developed [simply introduced by Barnby (1) and Pook (2)], which can describe fast fracture or fatigue crack growth from pre-existing defects. The stress intensity factor characterizes the stresses ahead of a sharp crack in an infinite elastic body and is related to the remotely applied stress, σ , and crack length, $2a$, by the equation:

$$K = \sigma\sqrt{\pi a} \quad \dots (1)$$

The fracture condition, due to Irwin (3), occurs when the stress intensity reaches some critical value, K_c , given by:

$$K_c = (E G_c)^{1/2} \quad (\text{for plane stress}) \quad \dots (2)$$

where G_c is the fracture toughness of the material and E is Young's Modulus. The commercial success of fracture mechanics lies in the fact that K can be calculated for finite sized specimens and for cracked components.

Paris and Erdogan (4) originally proposed that the rate of fatigue crack propagation, $\frac{da}{dN}$, is related to the alternating stress intensity, ΔK , by the equation:

$$\frac{da}{dN} = C\Delta K^m \quad \dots (3)$$

For a given material, equation (3) has been shown to provide a reasonable fit to the data. However, collections of several sets of data (4,5) indicate that, for a given cyclic and mean stress intensity condition, there can be a variation or 'scatter' in rate of propagation of more than an order of magnitude. Similarly, wide variations in the quantities C and m in equation (3) have been noted (4, 6-13). The exponent 'm', for example, is typically found to lie in the range 2-3, but values as high as 10 have been measured (14). Equation (3) does not describe fully the relationship between $\frac{da}{dN}$ and ΔK , being valid only for the intermediate range of growth rates (typically 10^{-3} - 10^0 $\mu\text{m}/\text{cycle}$). The variation of $\frac{da}{dN}$ with ΔK is a actually sigmoidal in form, being bounded at extremes by the values of K_c (characterizing the material's fracture toughness) and ΔK_0 (characterizing a threshold for crack growth).

It is important to realize that the linear elastic fracture mechanics approach in equation (3) is meaningless from a mechanistic viewpoint since the specimen either fails during loading or returns to its original state on unloading, with no crack growth. In practice, there is a yielded

region ahead of the crack where the shear stress exceeds the yield stress of the material, σ_y . Assuming this region to be circular, the radius, r_p , is approximately given by:

$$r_p = \frac{1}{2\pi} \left(\frac{K}{\sigma_y} \right)^2 \quad (\text{in plane stress}) \quad \dots (4)$$

The term K also appears in the expressions for the displacement fields about a crack and by combining such expressions with equation (4), it can be shown that the crack tip opening displacement, δ , is given by the equation

$$\delta = \frac{4K^2}{\pi E \sigma_y} \quad (\text{in plane stress}) \quad \dots (5)$$

Having shown that K can be related to the irreversible creation of new surface at the crack tip and cyclic plastic damage of material ahead of the crack, the process of fatigue crack growth and the general success of equation (3) can be more readily understood.

The relationships between these cyclic displacements and the fatigue crack growth rate, and, therefore, the precise form of equation (3) is the subject of debate of many publications. It is not the purpose of this paper to review such work, but simply to describe the various mechanisms of growth that can occur in steels and other metals as a result of cyclic displacements. The metallurgical factors that give rise to these mechanisms are outlined and it is shown that the form and applicability of equation (3) for a particular material can be estimated without recourse to fatigue testing.

In the application of data to service cracking, it is especially important to understand the effect of residual or superimposed mean stress,

characterized by K_{MAX} , the maximum stress intensity of the cycle, on fatigue crack growth rate. Fig. 1 shows the two most common types of laboratory tests. Fig. 1(a) is a constant load test where ΔK , K_{MAX} , and K_{MIN} (the minimum stress intensity of the cycle) gradually increase with increasing crack length. The rate of crack growth is obtained from the gradient of a curve of 'a' against N. Fig. 1(b) is a test where the stress intensity conditions remain constant and 'a' increases linearly with N. This is usually achieved either by testing specially shaped test-pieces or by continually adjusting the load levels.

Finally, it is also important to know the effect of thickness on growth rate in order to relate laboratory data to large components. For example, the size and shape of plastic zones and the stresses that are able to develop within these zones is different in plane stress (thin specimens - through thickness relaxation) than in plane strain (thick specimens - deformation only in the plane of the specimen).

Fatigue cracks often grow on a plane normal to both the applied stress and the surface of the specimen, but as crack length (and degree of plane stress deformation) increases the crack may rotate to a plane at 45° to the specimen surface but still normal to the stress axis. These thickness and mean stress aspects will receive special attention.

2. BASIC FATIGUE CRACK PROPAGATION DATA

Fig. 2 shows the relationship between crack propagation rate and the alternating stress intensity for a large variety of ferrous alloys of widely differing microstructures, over varying conditions of mean stress (or K_{MAX}), specimen thickness and specimen geometry (15). All

the tests were tension/tension loaded at room temperature for conditions where the maximum stress was less than about 0.8 of the general yield stress. The experimental techniques for these and other fatigue tests described in this paper are given in references 16 and 17.

It is clear from this diagram that the mechanism of fatigue crack propagation is not solely that of the classical transgranular striation growth, the symbols referring to fracture surfaces exhibiting 20% or more of other modes of crack propagation. It is evident that rates of growth by striation formation fall into a comparatively narrow scatter band, and that excessive deviations only result from other mechanisms of failure. These additional modes, namely cleavage cracking, intergranular separation, and microvoid coalescence, which we may term "static" or monotonic fracture modes, thus give rise to larger values of the exponent 'm' in the Paris relationship (equation 3). Further, since these additional modes are strongly sensitive to the maximum tensile stress at the crack tip, such as cleavage and intergranular cracking, or to the hydrostatic component, such as void coalescence, we may expect that where such mechanisms occur, fatigue crack growth will be increasingly dependent upon the level of mean stress (or K_{MAX}) and the specimen thickness (i.e., the achievement of high triaxial stresses characteristic of plane strain deformation).

Until fairly recently, the vast majority of fatigue crack propagation data collected has not been accompanied by parallel examination of fatigue crack propagation mechanisms. We now show that the excessive crack propagation rates that have been observed (in the absence of strong

environmental effects) are invariably associated with departures from the conventionally accepted mechanism of fatigue crack growth, namely that of striation formation.

3. GROWTH BY THE STRIATION MECHANISM

Zappfe and Worden (18) first observed a pattern of ripples or striations on a fatigue fracture surface, and subsequently this has been characterized by many authors as the accepted mechanism of fatigue failure. Programmed loading has shown that each striation is produced by one cycle of stress (19, 20), although every cycle does not necessarily produce a striation. Considerable fractography has been carried out on striation growth mechanisms, particularly in aluminium alloys (reviewed in ref. 21) where the extremely clear definition of the striations has much facilitated morphological investigations.

The strength level, or more correctly, the work hardening characteristics dictate the clarity of the markings, since in low strength steels where the work hardening exponent is high, striations are very clearly formed (Fig. 3a). In the higher strength martensitic structures, however, the lower work hardening rate does not facilitate "ear" formation, and the striations are correspondingly much less visible (Fig. 3b). The morphology of the striations differs widely (21), and depends on the symmetry and number of available slip systems (22, 23). In aluminium alloys, two sorts of fatigue striations have been distinguished (24-27), namely ductile striations (Type A) which form on facets substantially parallel to the general fracture surface, and "brittle" striations (Type B)

which form on crystallographic planes. The latter markings were thought to be due to a cleavage mechanism on {100} planes, but it has been recently shown that their formation is in principle a flow process similar to that of ductile striations (28) with the influence of the environment restricting their occurrence to certain crystallographic planes ["stress-sorption" (29)].

The precise mechanism of striation formation has yet to be established, although it is generally accepted that it involves alternate blunting and sharpening of the crack tip, a mechanism first popularized by Laird and Smith (30,31) and Tomkins (32). Two such models due to Laird (31) and Pelloux (33) are illustrated in Fig. 4. In both cases the area of the crack tip is increased during the loading part of the cycle and the crack tip is sharpened during the unloading portion. Under such conditions growth is controlled by the local crack tip alternating plastic strain, and thus when applied specifically to propagation under linear elastic conditions, we would expect the growth rate to be dependent upon ΔK , the alternating stress intensity.

The effect of K_{MAX} (or mean stress) on rates of crack propagation by the striation mechanism is typified by Fig. 5, which shows intermediate growth rate data for an unembrittled low alloy steel (En 30A), tested at various mean stresses (17). Here the same initial ΔK was maintained, but the stress ratio $R = K_{MIN}/K_{MAX}$ was increased from 0.05 to 0.6. Throughout growth, the mechanism of failure was by the striation mode (Fig. 3b) and it is clear that the influence of mean stress is negligible, growth being dependent primarily on the alternating stress intensity ΔK .

Similar observations of the insensitivity of striation growth to K_{MAX} have been observed in a 3% Cr Mo Steel (15), spheroidized 1% carbon steel (34), a fine-grained mild steel (35) and in several high toughness commercial aluminium alloys (36,37). There is evidence, however, that at low growth rates and small ΔK values (typically $\Delta K < 15\text{MNm}^{-3/2}$) growth becomes increasingly sensitive to K_{MAX} (38,29). Under these conditions, the influence of the environment becomes more marked and non-continuum mechanisms of growth become operative (40). This aspect of fatigue crack propagation, i.e., that of very slow growth rates at low ΔK s has been discussed by Beevers and co-workers (40,41). The striation mechanism of growth has provided the basis for most, if not all, quantitative models of fatigue crack propagation. Such continuum models have been based on damage accumulation arguments (for example, due to Liu and co-workers (42-44), shear decohesion on planes of maximum shear strain gradient at the crack tip [due to Tomkins (32)] and alternating shear rupture at the crack tip [due to Pelloux (33)], and generally predict an exponent (m) of 2 in the Paris power law relationship (equation 3). This value of m, however, is merely at the minimum end of the range found experimentally, and it is the authors' contention that large departures from this value are due to a change in mechanism of fatigue crack growth, i.e., the occurrence of additional "static" modes of fracture during, or replacing, striation growth (see following sections).

The model due to Pelloux (33) predicts that the growth increment per cycle is some proportion of the cyclic crack opening displacements δ_c such that

$$\frac{da}{dN} \propto \delta_c = \beta \frac{\Delta K^2}{4E\sigma_y} \dots (6)$$

where β is the efficiency of blunting (or degree of irreversibility) and can be taken as constant (45). The correlation of the growth rate with the elastic modulus E has been well established (46,47), but the inverse dependence on yield stress is rarely observed. For example, the large variation of yield stress (x 7) of the materials in Fig. 2, yields only a variation of crack growth rate (by the striation mode) of around 2. However, since σ_y in equation 6 represents a flow stress characteristic of the cyclic plastic zone, it is possible at high strains that this approaches a common value. Further, the microstructural factors which affect σ_y may also alter the proportionality constant β , or the ΔK dependence in opposing fashion. With regard to the latter fact, there is increasing evidence that, for striation growth, although high strength steels generally show a ΔK dependence of 2 (see ref. 17), the exponent for low strength steels, such as mild steel (16, 35) and stainless steels (48) is closer to 3 or 4. Since this is not a result of a change in fatigue fracture mechanism, it is conceivable that the small variation in 'm' of between 2 to 4 for striation growth is related to the difference in work hardening characteristics of low and high strength steels, a phenomenon more adequately described by the Tomkins' model (32).

The influence of specimen thickness on fatigue crack propagation by the striation mechanism is small, and critically dependent on the specimen geometry and on how close the stresses are to general yield. There is much conflicting evidence for the thickness effect in the literature (reviewed in refs. 49 and 50), but in very few cases has there been a parallel examination of the mechanisms of crack growth. Fig. 6 shows the effect of a wide variation of specimen thickness (1.5 - 75 mm) in Ducol W30B beneath ~ 0.7 of the general yield stress, where in all cases the mechanism of fatigue crack growth was striation formation. It is clear that there is a very small influence of thickness, although for a given range of stress intensity there can be a considerable variation in the degree of through thickness constraint as indicated by the angle of the fracture surface. However, small differences in crack growth rates between thin and thick specimens can arise from variations in test-piece geometry and the accompanying displacement restraints. For example, growth rates have been shown to be faster in thick specimens when tested in bend, whereas the reverse has been found when the same materials are tested in tension (50-52). In this case, the displacement restraints in the through thickness direction were quite different. The fracture face of the thin bend specimen remained at 90° to the face of the specimen whereas, for the same stress intensity conditions, the fracture surface of the thin tension specimens were slanted at 45° to the specimen face. A corollary of these observations is that the angle of the fatigue surface to the loading direction is a poor indication of whether the conditions are plane stress or plane strain.

Rates of fatigue crack propagation by striation formation in thin specimens may also exceed those in thicker specimens due to the earlier onset of general yielding in the thinner test pieces. This has been shown by constant load tests conducted in a low alloy-weld metal where the stress on the uncracked portion of the specimen was initially $\sim 70\%$ of that to cause general yielding (49). Here the propagation rates in the thin specimens were observed to be greater than in the thick as crack length increased and general yield was approached. This thickness effect is due to the higher general yield stress in thick specimens than in thin, which can be very considerable, especially for deep notches (53). Stress ranges exist, therefore, where plastic strains in the thin specimens can be large, while in the thick specimens they are restricted by unyielded material. Thus, in general, the difference in crack propagation rates between thin and thick specimens will depend on the type of loading, the difference in thickness, and the depth of crack since these are the parameters that govern the general yield stress. In all the cases reported previously where cracks propagated faster in thin specimens (10, 13,54), the stress conditions were certainly approaching or had exceeded the general yield stress. In other instances, slightly faster growth rates are to be expected in thick specimens where the plane strain conditions at the crack tip create a greater concentration of strain (50,51) and restrict the crack closure stresses (55) which may limit the "effective" stress intensity experienced at the crack tip (56).

It is concluded that fatigue crack propagation rates by striation formation in thin specimens can exceed those in thick when conditions approach or exceed that of general yield or when gross out-of-plane sliding is allowed by the loading system.

4. GROWTH BY NON-STRIATION MECHANISMS

Recent studies by the authors have shown that, in low toughness materials particularly, the mechanism of fatigue crack growth can include contribution from fracture modes other than striation formation. These additional "static" fracture modes may include cleavage and intergranular cracking and microvoid coalescence.

Fatigue crack propagation by combined striation formation and cleavage cracking has been observed by Richards (22) in a coarse grained silicon iron along {110} planes. The cleavage cracks propagated only a short distance and their uniform distribution produced a macroscopically steady growth rate. Similar observations in silicon-iron have been made by Wright and Argon (59), and by Pearson in a 12% chromium rotor steel (60). Further, Ritchie and Knott (35) have observed that rates of crack propagation in a high-nitrogen mild steel are substantially greater at temperatures beneath the ductile-brittle transition temperature due to the formation of cleavage cracks, nucleated at brittle grain boundaries carbides, during striation growth (Fig. 7). The same material tested above this temperature showed no excessive rates of crack propagation; the mechanism of growth being merely striation formation. Microcleavage may also occur

during fatigue crack propagation in materials containing brittle second phase particles. Fig. 8 shows areas of fractured cementite in a coarse pearlitic steel (34). Several other examples of microcleavage have been observed in high strength aluminium alloys (61-63).

Fatigue crack propagation involving cleavage and microcleavage of particles clearly results in enhanced rates of crack propagation when compared with a purely striation mechanism. The amount of increase in rate has been shown to be critically dependent on microstructure in ferrite/pearlite steels(34,35). A comparison (34) between spheroidized, mixed spheroidized/pearlite and coarse pearlite structures in a 1% carbon steel showed that rates of crack propagation were least in the spheroidized condition where there was no microcleavage (Fig. 9). As the proportion and coarseness of pearlite was increased by raising the reheat-treatment temperature, the amount of microcleavage and rate of crack propagation increased.

Heald et al (34) have proposed the following equation to describe the enhanced growth rates associated with a non-striation mechanism:

$$\frac{da}{dN} = A \left(\frac{\Delta K^4}{\sigma_1^2 (K_C^2 - K_{MAX}^2)} \right)^n$$

where A is a material constant; σ_1 a strength parameter; K_{MAX} the maximum stress intensity in the fatigue cycle; K_C the fracture toughness of the material.

Examination of the literature revealed considerably greater reference to void or dimple formation during fatigue crack growth in steels (21,58, 54,64-68) aluminium alloys (37,69-72) and titanium alloys (72) than cleavage, because structural alloys are usually heat-treated to produce fine microstructures and high toughness. Most observations of voids during fatigue crack growth in ferrous materials have been made on medium to high strength steels (e.g., 17,58). However, void coalescence accompanying striation growth has also been observed in a low strength stainless steel weld metal (Fig. 10) (48).

A mechanism for fatigue crack propagation by void coalescence has been proposed by Forsyth and Ryder (73), in which voids form ahead of the main crack and eventually link up by thinning of the bridging material during subsequent cyclic loading. Tearing at the peak load of the cycle, interrupted by the unloading and loading parts of the cycle, is unlikely to occur except possibly very near the point of failure. For instance, Griffiths et al, (58) have demonstrated a strong influence of ΔK on both the rate and mechanism of crack propagation by void coalescence for the same K_{MAX} conditions. An interrupted tearing mechanism should not be dependent on ΔK . Furthermore, the rates of crack propagation by the void coalescence mechanism were independent of frequency, and hold times of several hours at peak loads produced no measurable growth.

Fatigue crack propagation by an intergranular mode has been reported most frequently for quenched and tempered high strength steels (7,17,40, 50, 74, 75) where the fatigue crack tends to follow prior austenite grain

boundaries. Fig. 11 provides an example of fatigue crack propagation along prior austenite grain boundaries at small values of ΔK and K_{MAX} in quenched and tempered En30A steel (17). We do not intend to discuss the influence of environment on fatigue crack propagation, except to point out that intergranular fatigue crack propagation is encouraged by the presence of water vapour which is, of course, present in most tests in air (7,74). Furthermore, the intergranular mode of propagation is usually accompanied by higher growth rates (7,17,50,74,75). There is also evidence that this mechanism is encouraged by impurity elements (7, 17,50), which are known to reduce the strength of boundaries(76). These observations are probably related to those of Li et al., (77), who showed that the sensitivity of growth rate to moisture is dependent on toughness in ultra-high strength steel.

Several investigations have shown that intergranular fatigue crack propagation is less likely with increasing stress intensities (Fig. 2 of this article for EN24 and EN30B; (7,75). Seemingly, when crack propagation is slow, the environmentally assisted intergranular mode can predominate (40,41).

This intergranular fatigue mode is not confined to prior austenite grain boundaries. For example, it has been observed in single phase ferritic 3% silicon iron (15), in mild steel (35,78), stainless steel (79), and copper (79,80).

More recently, further modes of intergranular fatigue failure have been observed which predominate at higher stress intensities, in a manner similar to the occurrence of cleavage cracking. Ritchie and Knott (17,50)

have observed markedly accelerated growth rates in an En30A low alloy steel in a temper embrittled condition. In this instance, increasing amounts of brittle intergranular cracking (Fig. 12) were observed during striation formation, as the stress intensity was raised; no such cracking occurred in the unembrittled material which displayed correspondingly lower growth rates. Moreover, by testing a similar steel in an "overheated" condition (57) they showed that accelerated fatigue crack propagation rates could be obtained when the mechanism of growth involved a contribution from intergranular fibrous fracture (a process in which void coalescence occurred around a fine dispersion of α -MnS particles, precipitated during cooling from 1300°C in prior austenite grain boundaries (Fig. 13).

In summary, there appears to be several intergranular fatigue crack growth mechanisms: one favoured at small values of ΔK and is probably environmentally enhanced, one a cracking mode dependent on impurity weakened grain boundaries (as in temper embrittlement), and one involving microvoid coalescence along such boundaries (as in overheating).

5. EFFECT OF TESTING VARIABLES ON NON-STRIATION GROWTH

There are several important consequences of the occurrence of these additional "static" fracture modes during striation growth on fatigue fracture behaviour. Firstly, since all these mechanisms are critically dependent on the tensile stress at the crack tip (in the case of limited cleavage, microcleavage and intergranular cracking) or on the hydrostatic

component (in the case of fibrous fracture) we might expect a strong dependence of mean stress (or K_{MAX}) on the propagation rate where these modes occur. There is now an increasing body of evidence to show that, for growth rates above $\sim 10^{-3}$ $\mu\text{m}/\text{cycle}$, the sensitivity of the propagation rate to mean stress is almost entirely due to a change in mechanism from purely striation formation. For example, Fig. 14 shows markedly accelerated growth rates in a medium carbon steel as the stress ratio (K_{MIN}/K_{MAX}), R , is increased from 0.10 to 0.72, causing increasing amounts of cleavage cracking to occur during the striation growth (81). No such mean stress dependence on growth rate was observed in a similar steel where cleavage cracks did not form. Similar observations of a strong influence of K_{MAX} due to the occurrence of cleavage or microcleavage cracking have also been reported in a coarse pearlite 1% C steel (34), and a polycrystalline silicon iron (15). In all cases, by changing the microstructure to prevent formation of cleavage cracking, the K_{MAX} dependence on growth rate was removed. The effect has also been observed in high strength aluminium alloys (63) where increasing rates of crack propagation with increasing mean stress result from the microcleavage of particles; the mean stress dependence being removed in lower strength aluminium alloys where such fractures cannot occur (36).

This phenomena is analogous when brittle intergranular cracking accompanies striation growth (17,50). Fig. 15 shows the marked effects of mean stress observed in a temper embrittled low alloy steel due to the formation of intergranular facets (Fig. 12), compared with the unembrittled steel which showed mean stress insensitive striation growth.

The effect of mean stress on the fatigue crack propagation rate where the mechanism of growth involves intergranular separation at small values of ΔK is far less clear (50). Certainly at these low growth rates, a strong influence of K_{MAX} and of the microstructure is observed on the crack propagation rate (e.g., 40,41,82). However, under these conditions, it is more probable that the influence of K_{MAX} results from environmental activity rather than due directly to the presence of the intergranular failure mode during striation growth (40).

Many investigations have produced evidence of accelerating rates of fatigue crack growth approaching final failure, but the separate contributions of ΔK and K_{MAX} have not been carefully examined. Fig. 16 shows the influence of K_{MAX} on crack propagation rates in a low alloy weld metal (58) for several values of ΔK . For $\Delta K \leq 40 \text{ MN m}^{-3/2}$, no influence of K_{MAX} was observed, and the growth mechanism was found to be striation formation. At large values of ΔK , however, the rate of crack propagation was found to increase rapidly with increasing K_{MAX} . Under these conditions crack growth was by void coalescence. Similar behaviour has been observed in En24 and En30B quenched and tempered steels (15). It appears, therefore, that both ΔK and K_{MAX} contribute towards the accelerating rates of crack propagation as conditions for failure are approached.

A second consequence of the "static" fracture component during fatigue is the effect of specimen thickness. Fig. 17 shows the rate of crack propagation in 1.5, 8 and 19 mm thick specimens of a pearlitic 1% C steel (15). In contrast to the results for striation mechanisms, an

increase in thickness caused a marked increase in the rate of crack propagation. This is due to an increase in the incidence of microcleavage caused by the higher through-thickness constraint and tensile stress. When full constraint is achieved through the thickness, further increases in thickness do not produce further rate increases. This is shown by the similarity of growth rates for 8mm and 19mm thick specimens in Fig. 18 where there was similar constraint and, therefore, toughness (15).

Faster rates of crack propagation with increasing thickness have also been observed in a coarse grained high-nitrogen mild steel (50,51) where the mechanism of growth involved cleavage (Fig. 7). In a finer grained steel no such cracking was obtained, and the influence of specimen thickness was correspondingly far smaller.

Thus, increased crack growth rates are caused by increasing thickness where the mechanism of crack growth involves cleavage. The difference (up to $2\frac{1}{2}$ times) in rates of crack propagation shown in Fig. 18 are expected to be much higher as K_{MAX} for the thick specimen approaches K_{1C} (15).

There have been few studies of the effect of thickness on fatigue crack growth where the failure mechanisms involved intergranular or void coalescence. It is to be expected, though, that in the case of brittle intergranular cracking, the increased constraint and triaxiality present in thick specimens would certainly promote such cracking and thus accelerate the crack propagation rate. Garrett (37) has found that the growth rates observed in commercial aluminium alloys are faster in 13mm thick specimens than in 2.5mm thick, but only when K_{MAX} approaches final

failure. Under these conditions, the failure mechanisms in such alloys become increasingly dependent on void coalescence.

A further consequence of non-striation mechanisms is that such cracking would be expected to assume more significance in materials of low fracture toughness, because brittle cracks would be easier to produce. Thus, effects of mean stress are expected to prevail particularly in low toughness materials. Evidence for such behaviour can be obtained from data on mild steel (35), 1% C steel (34), low alloy steels (17), and high strength aluminium alloys (36,37). The inference is simply that the occurrence of static modes during fatigue crack growth and associated mean stress effects will predominate as K_{MAX} approaches K_C .

It is clear also that where such monotonic modes occur during striation growth, there may be an increase in the ΔK dependence on growth rate, i.e., an increase in the exponent 'm' of the Paris power law relationship (equation 3). Bursts of brittle cracking would give rise to large accelerations in growth rate leading to a much increased value of 'm'. An extreme case of this is clearly shown in Fig. 15 for embrittled En30A, particularly at $R = 0.50$. Since segments of static fracture will predominate in low toughness materials, it is to be expected that the exponent 'm' should be increased in materials of low static fracture toughness (17). Shown in Fig. 18 are the results from several authors on the variation of 'm' with K_{IC} in steels, aluminium alloys, and titanium alloys (17). Neglecting any influence of the differences in mean stress employed by these authors, which may lead to some of the scatter observed, it is clear that steep slopes ($m \geq 3$) occur almost entirely in materials of low toughness ($K_{IC} \leq 60 \text{ MNm}^{-3/2}$).

Direct experimental evidence for the presence of static fracture components to explain the results in Fig. 18 is not easy, since in most of the data the fatigue mechanisms are not stated. The work of the present authors clearly indicates that the larger 'm' values were associated with microcleavage (34) and intergranular cracking (17) in the low toughness materials. Further, if two separate papers published by Miller (6,83) are compared closely, it may be deduced that his very steep slopes ($m \sim 6-7$) were associated with a total "fatigue" fracture appearance consisting of a combination of a intergranular fracture, fibrous rupture and quasi-cleavage, with little evidence of fatigue striations. For materials which yielded slopes of between 2 and 3, propagation was almost entirely by striation growth. Intermediate slopes were obtained where materials showed only isolated patches of striation growth, particularly at higher values of ΔK . Similarly, the results published by Evans et al. (7) show that their high values of 'm' in low-toughness steels are associated with the presence of intergranular facets.

Further, evidence for the presence of static fracture modes can be found from comparisons of the microscopic growth rates, determined from striation spacing measurements, with the macroscopic rate, obtained from external measurements of crack length. For several high strength steels (83) and aluminium alloys (63,84,85), it has been found that the dependence of ΔK on the microscopic growth rate is significantly less

than that of the macroscopic growth rate. Such a discrepancy can be readily explained if the macroscopic growth rate reflected the contribution from additional fracture modes to striation growth.

6. SUMMARY AND CONCLUSIONS

A study of the fatigue fracture surfaces of metals having a wide variety of microstructures and tested over a wide range of ΔK and K_{MAX} has shown four general mechanisms of growth, namely striation formation, cleavage, void coalescence and intergranular separation. A plot of fatigue crack growth rate against ΔK has been found to be sigmoidal in nature (Fig. 19). In the mid ΔK range (region B) where striation growth occurs, there is little influence of microstructure, mean stress, dilute environment and thickness on crack growth. At high values of $\Delta K/K_{MAX}$, departure from striation growth to include the "static mode" mechanisms leads to higher growth rates (region C). Here, a large influence of microstructure, mean stress and thickness is in evidence. As ΔK is progressively lowered in region A, the crack growth rate diminishes until a threshold ΔK_0 is reached, below which fatigue cracks remain dormant. Very low growth rates are involved just above ΔK_0 and in addition to a sensitivity to mean stress and microstructure, there is an important influence of environment.

The simple growth law

$$\frac{da}{dN} = C \Delta K^m$$

was found to be obeyed with little variation in C and m for the striation mechanism over a wide range of testing conditions (region B, Fig. 19). Modification of this law is necessary for very low (region A) and very high (region C) rates of propagation. For non-striation mechanisms, an equation of the type

$$\frac{da}{dN} = A \left[\frac{\Delta K^4}{\sigma_I^2 (K_C^2 - K_{MAX}^2)} \right]^n$$

provides a useful prediction of the influence of strength, constraint and mean stress on the rate of crack propagation for conditions approaching failure. Further work is necessary before a growth law describing region A can be forwarded with confidence. Under the testing conditions studied, no case of excessive growth rates was observed for the striation mechanism. In designing materials to resist fatigue crack propagation, the metallurgist should avoid structures and situations which give rise to non-striation growth. The remaining useful life of a component containing sharp defects can be assessed using the appropriate crack growth law.

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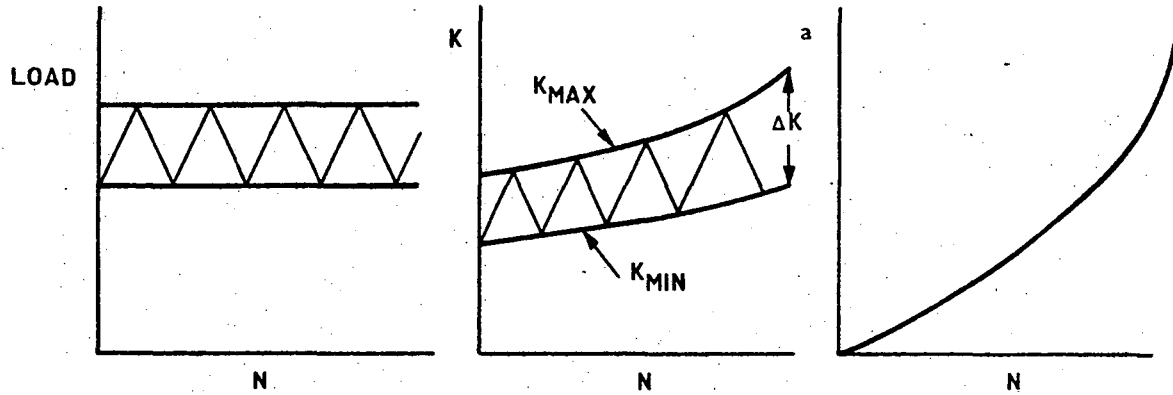
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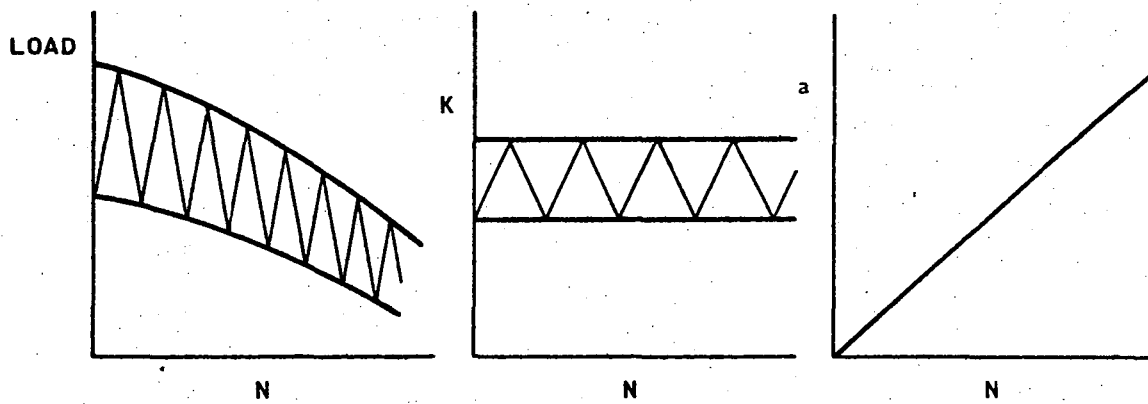
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(a) CONSTANT LOAD TEST

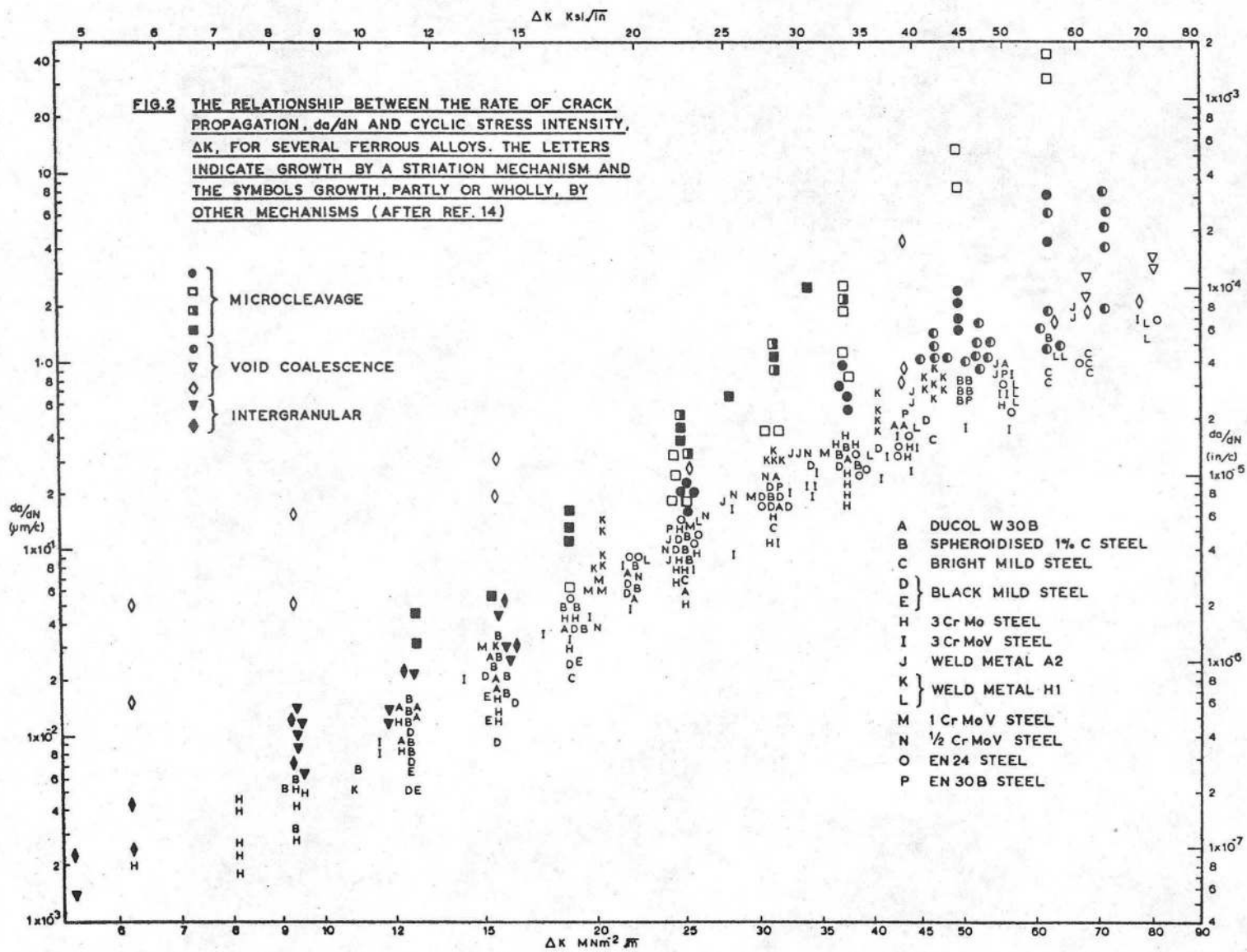


(b) CONSTANT K TEST

XBL 755-6263

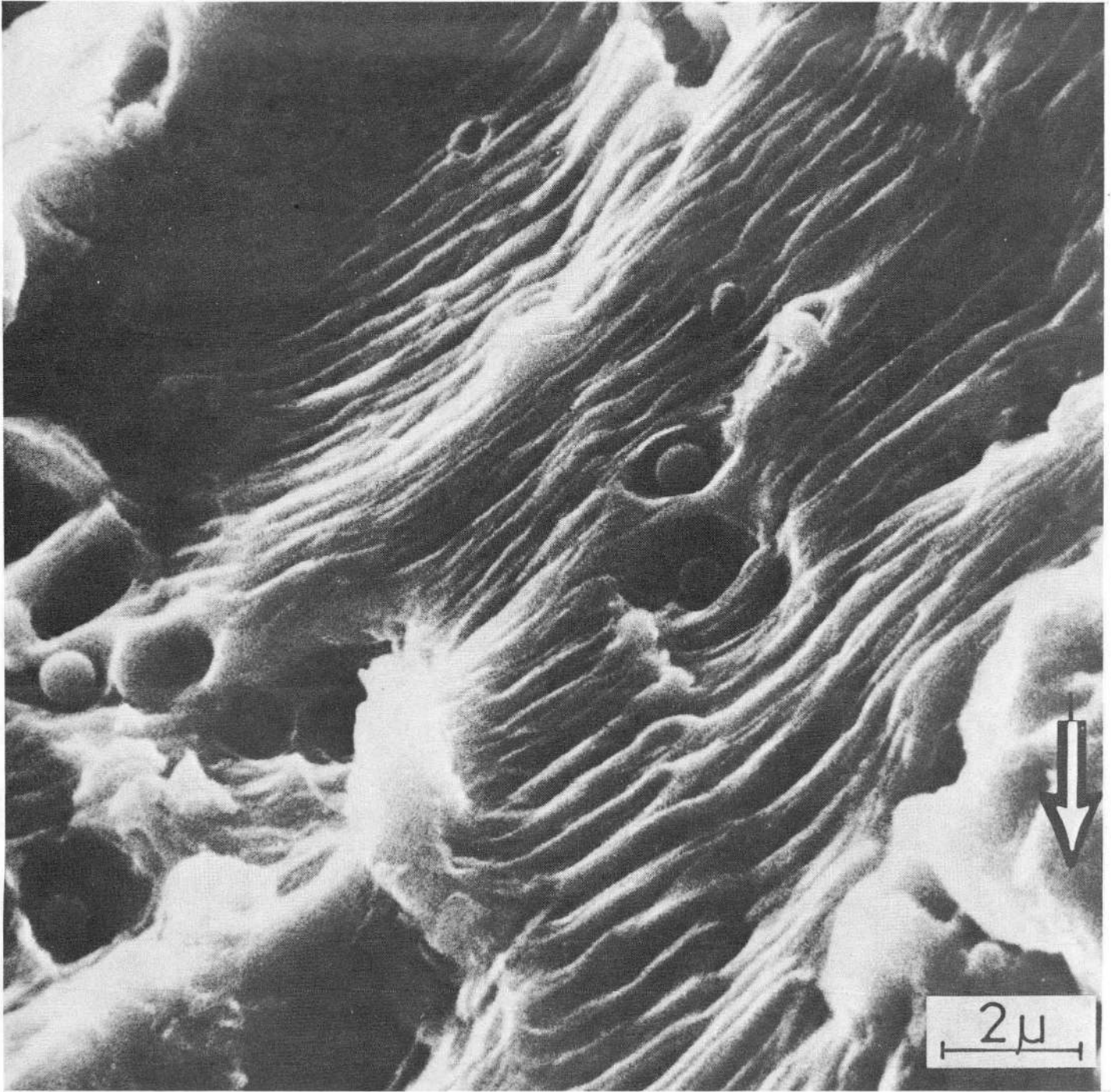
FIG. 1 TWO COMMON TYPES OF FATIGUE CRACK PROPAGATION TEST

(a - CRACK LENGTH; N - NUMBER OF CYCLES; K - STRESS INTENSITY)



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00004207416



XBB 755-3465

Fig. 3. (a.) Region of Ductile Striation growth during fatigue of mild steel at $\Delta K = 30 \text{ MNm}^{-3/2}$. Arrow indicates general direction of crack propagation (after Ref. 35).

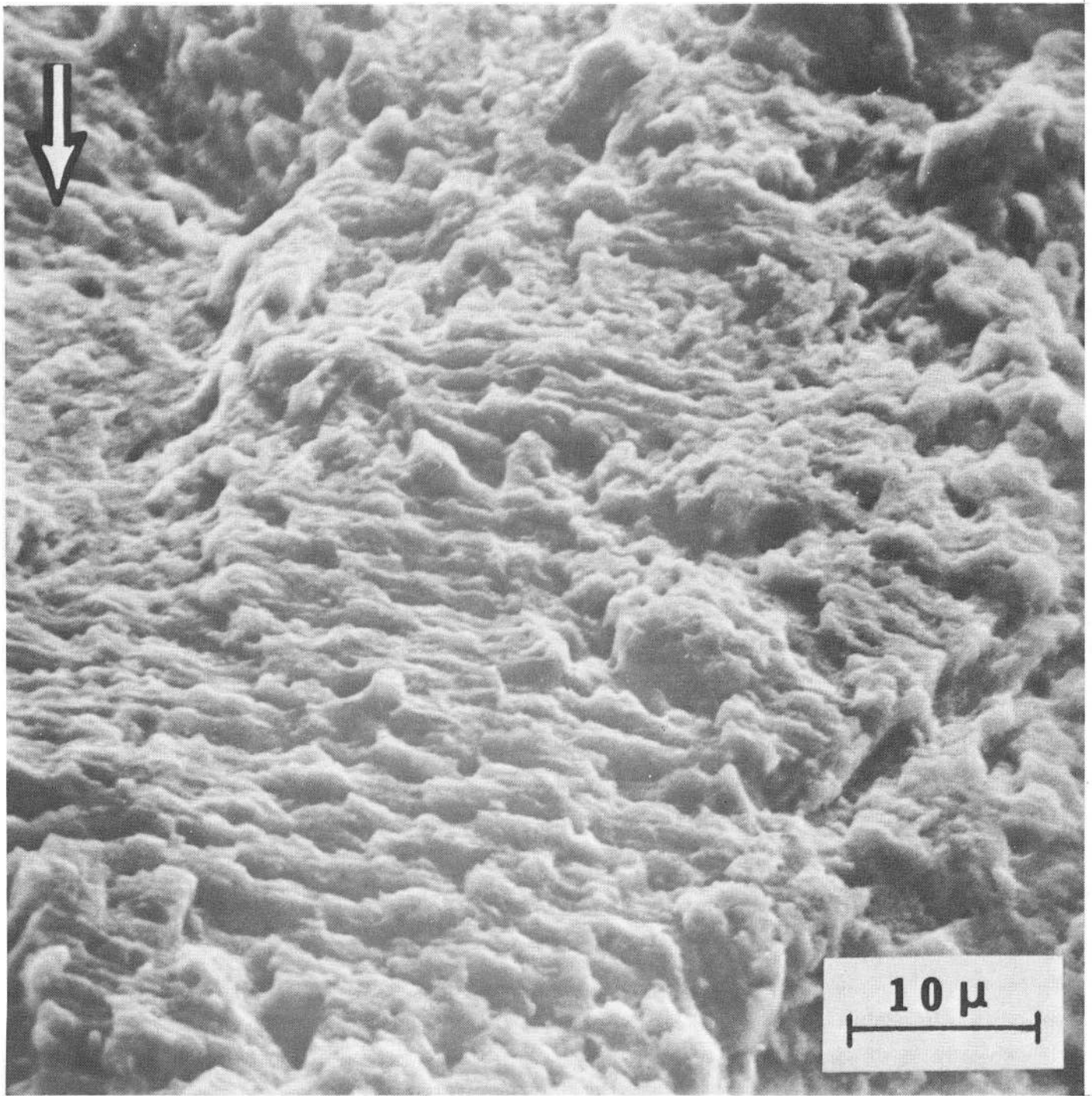
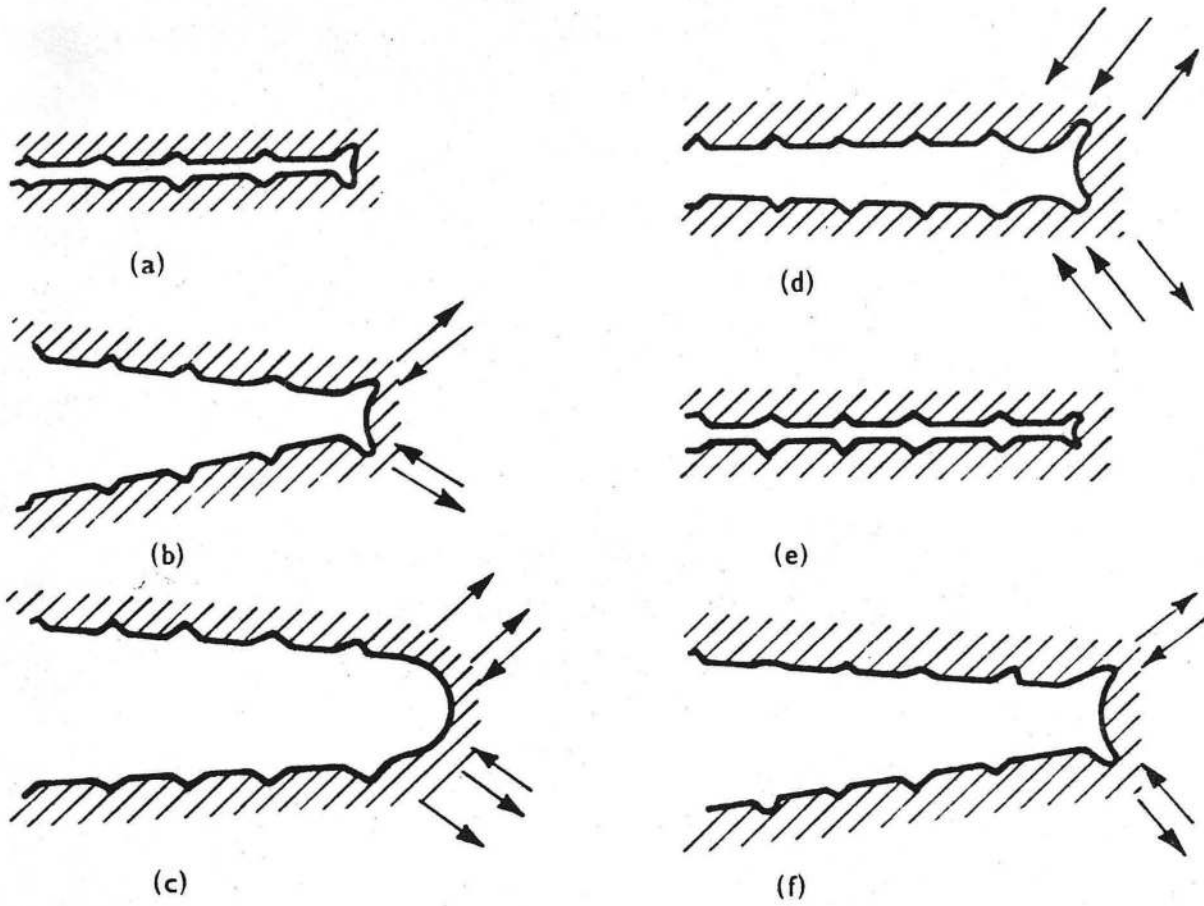


Fig. 3

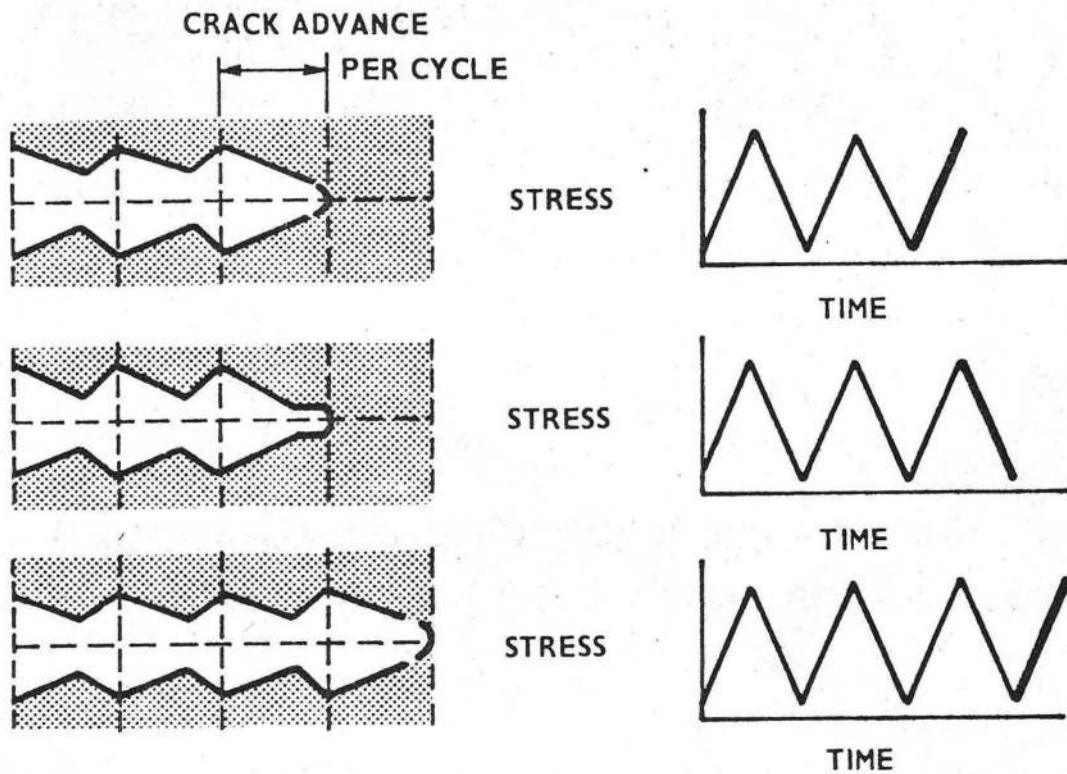
XBB 755-3468

(b.) Ductile striation growth through tempered martensite during fatigue of unembrittled low alloy steel. $\Delta K = 45 \text{ MNm}^{-3/2}$, $K_{\text{MAX}} = 47 \text{ MNm}^{-3/2}$ (after Ref. 50).



XBL 755-6267

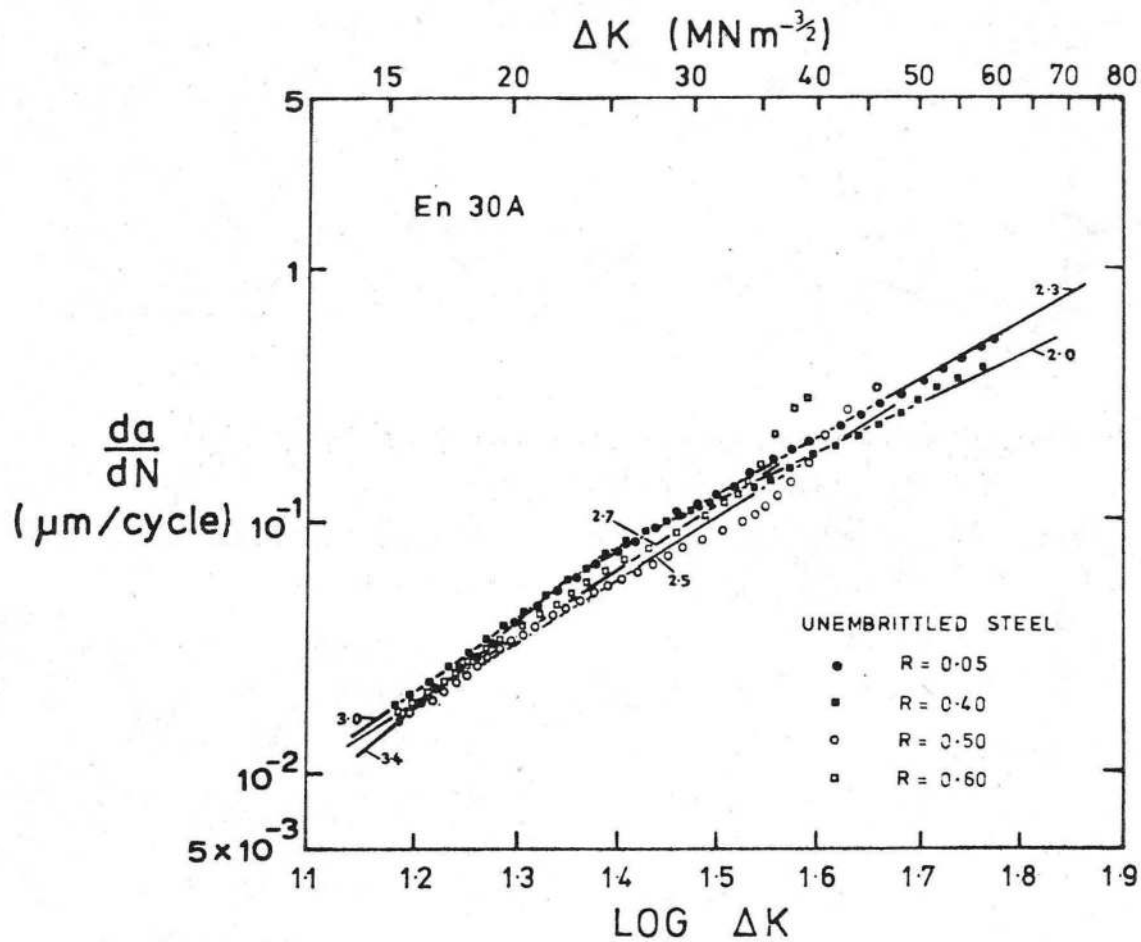
**FIG. 4(a) DIAGRAMMATIC REPRESENTATION OF STRIATION
FORMATION BY A PLASTIC BLUNTING PROCESS
(C. LAIRD)⁽³¹⁾**



CRACK PROPAGATION SEQUENCE

XBL 755-6266

FIG. 4(b) DIAGRAMMATIC REPRESENTATION OF THE FORMATION OF STRIATIONS
OF "SAW-TOOTH" PROFILE, J.C. McMILLAN AND R.M.N. PELLOUX)⁽³³⁾



XBL 755-6260

FIG. 5 VARIATION OF CRACK GROWTH RATE (da/dN) WITH ALTERNATING STRESS INTENSITY (ΔK) FOR RANGE OF STRESS RATIOS $R = 0.05-0.60$ FOR UNEMBRITTLLED En 30A STEEL. (NUMBERS INDICATE SLOPES (m) OF REGRESSION LINES DRAWN THROUGH DATA POINTS) (AFTER REF. 17)

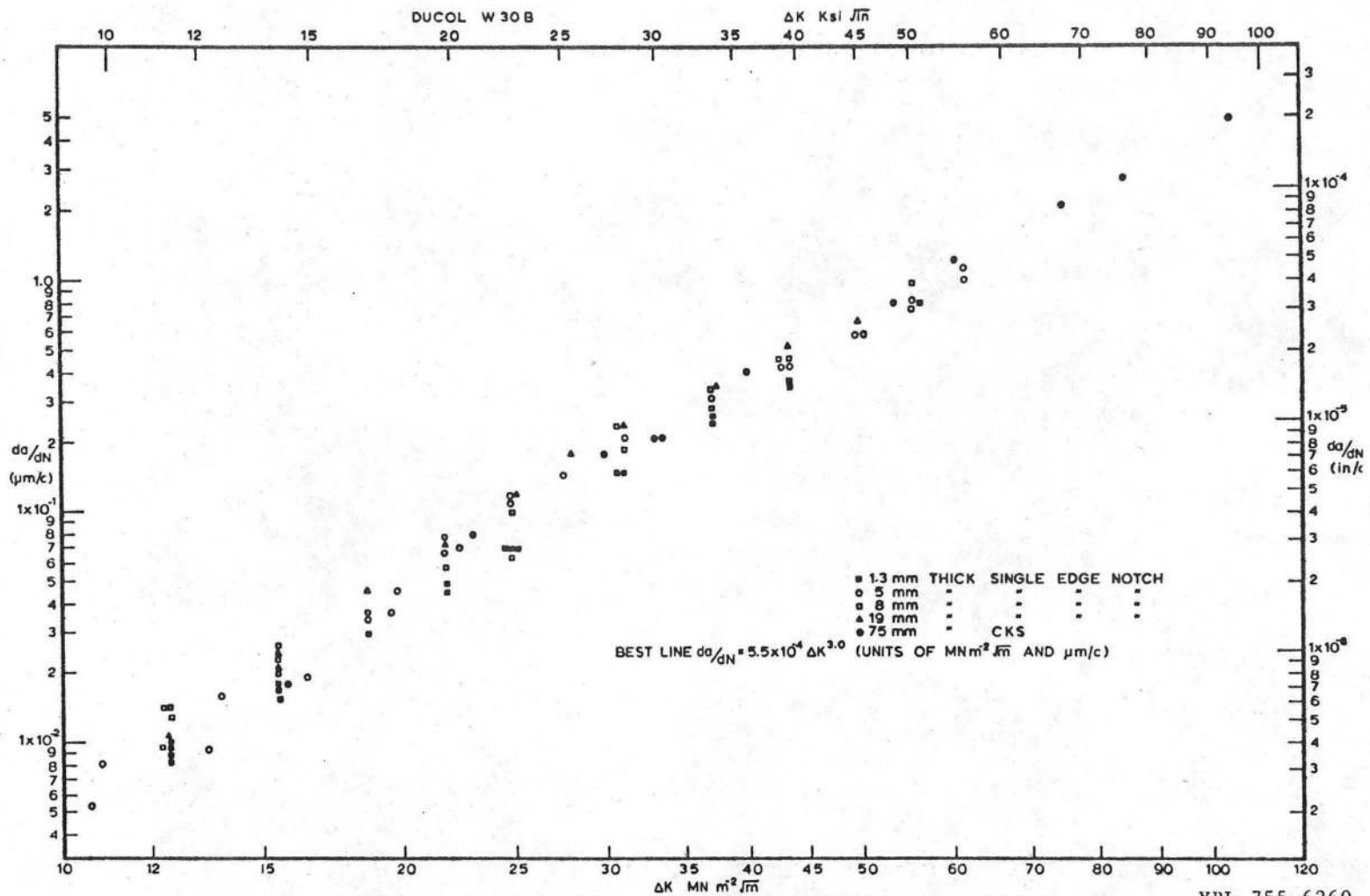
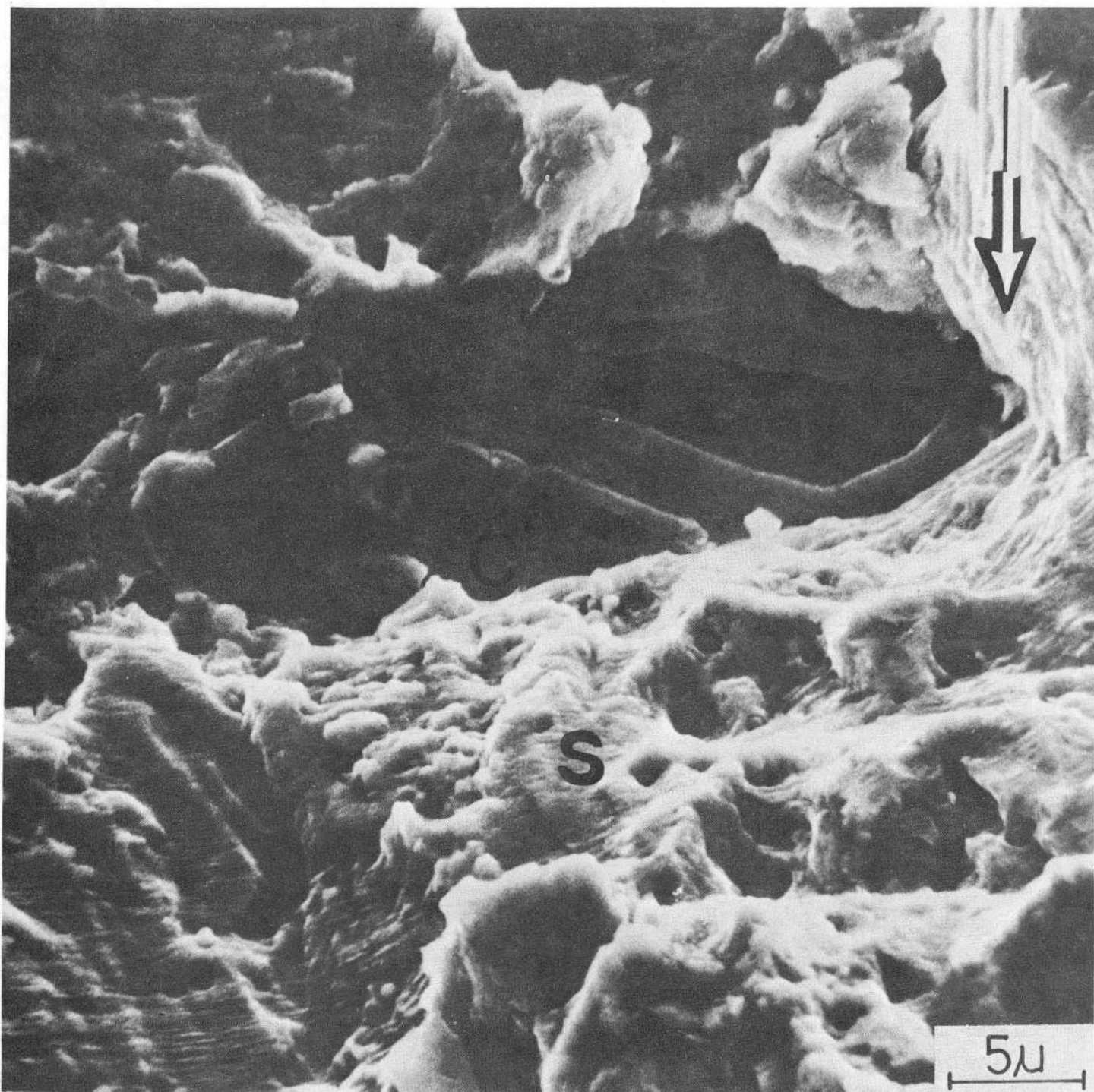


FIG. 6 THE INFLUENCE OF THICKNESS ON CRACK PROPAGATION RATES IN DUCOL W30B (AFTER REF.15)

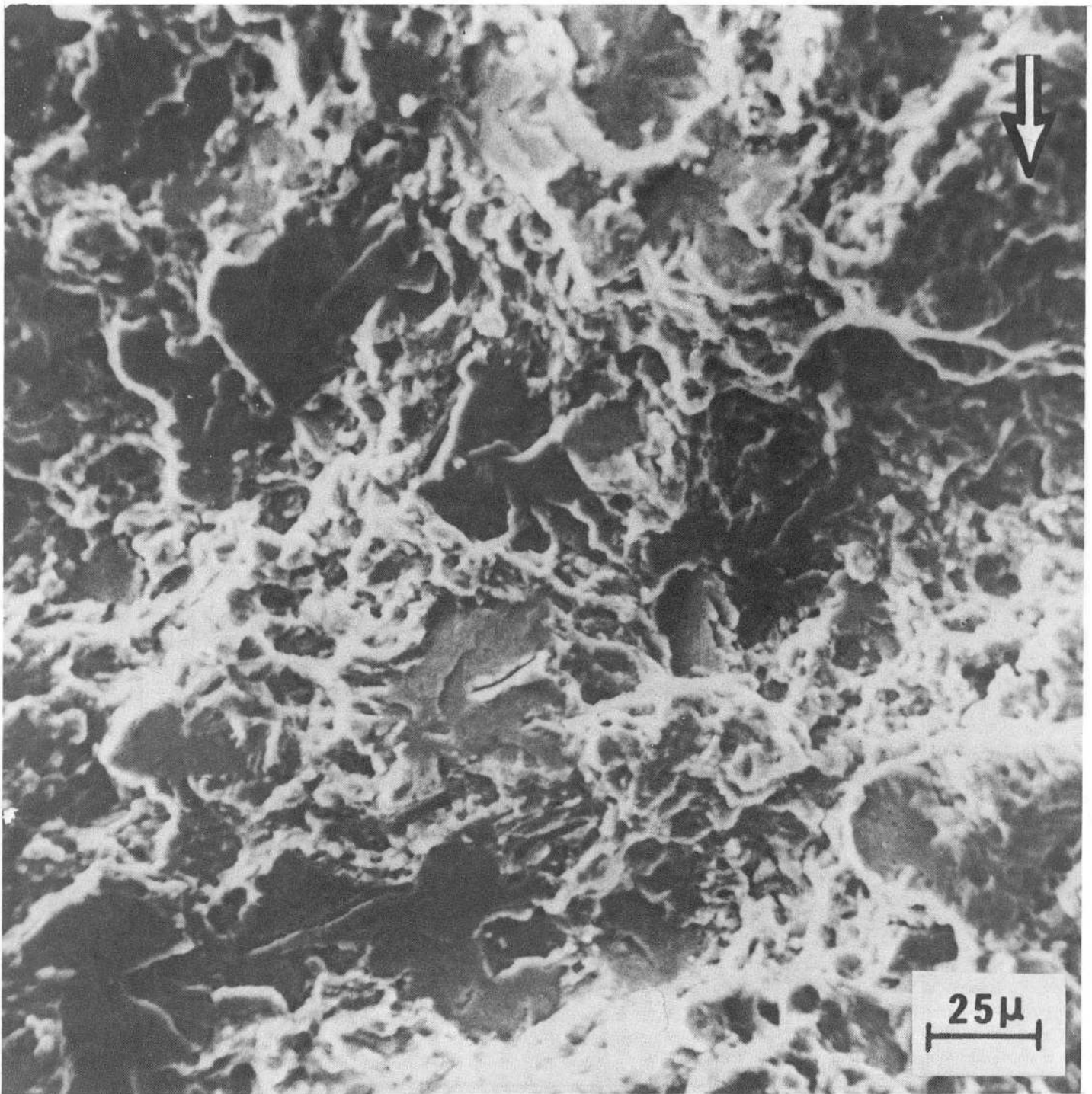
XBL 755-6269

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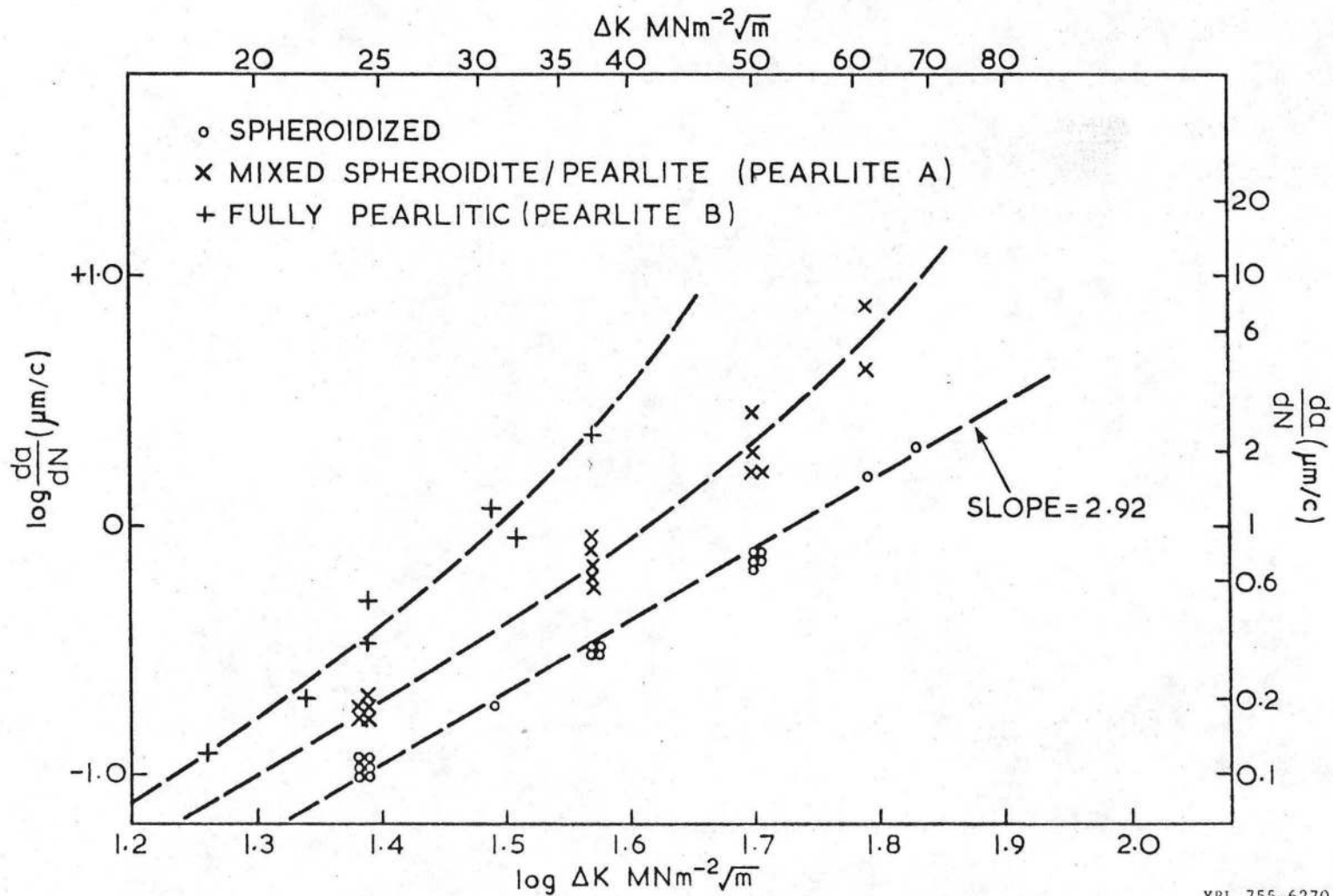
XBB 755-3466

Fig. 7. Isolated cleavage cracking (c) during striation growth (s) in mild steel at $\Delta K = 19 \text{ MNm}^{-3/2}$, $K_{\text{MAX}} = 22 \text{ MNm}^{-3/2}$. (after Ref.35).



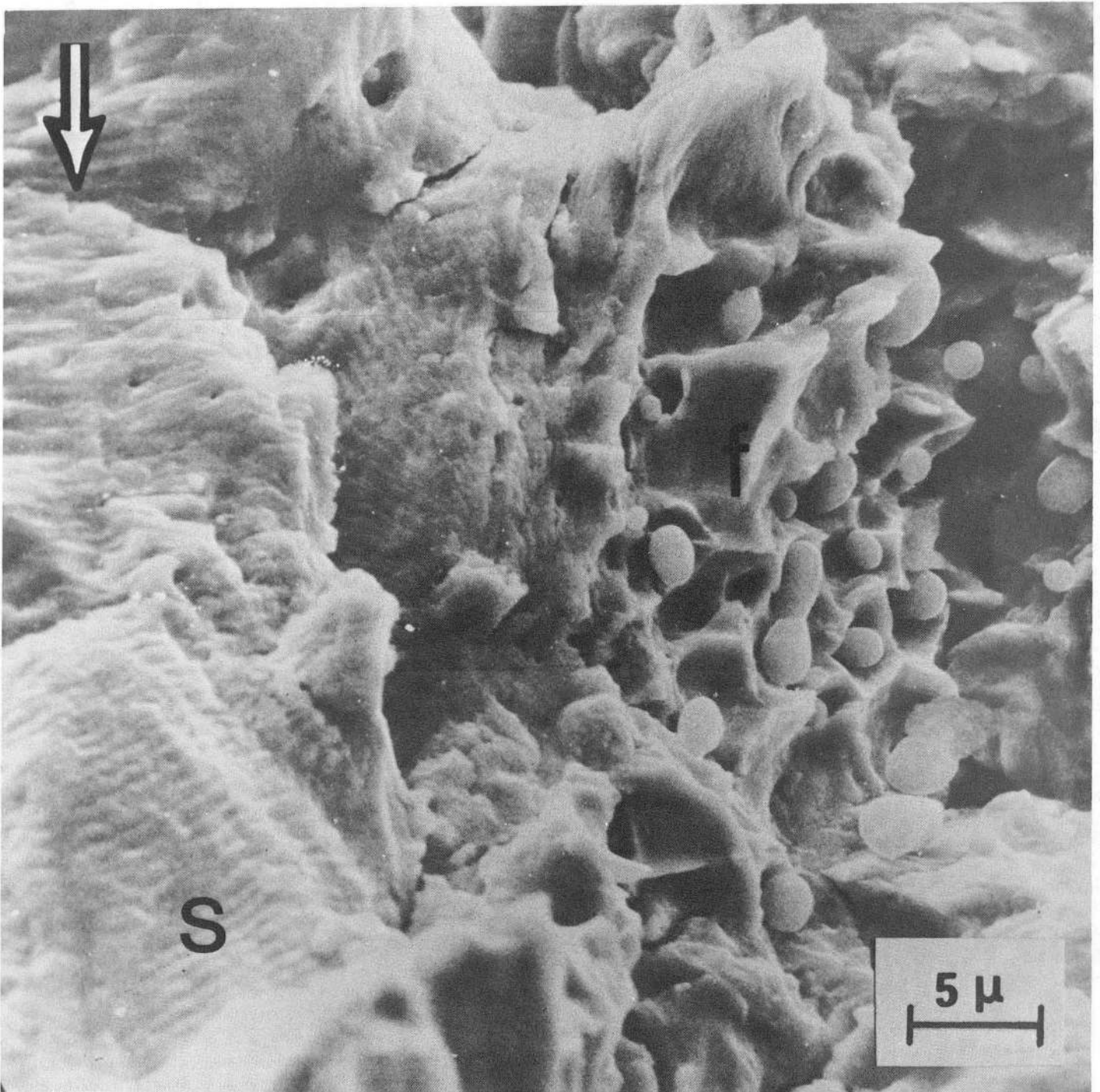
XBB 755-3470

Fig. 8. Areas of microcleavage in a coarse pearlitic structure, (1% carbon steel): $da/dN = 8 \mu\text{m}/\text{cycle}$; $\Delta K = 62 \text{ MNm}^{-3/2}$, $K_{\text{MAX}} = 80.7 \text{ MNm}^{-3/2}$ (after Ref. 34).



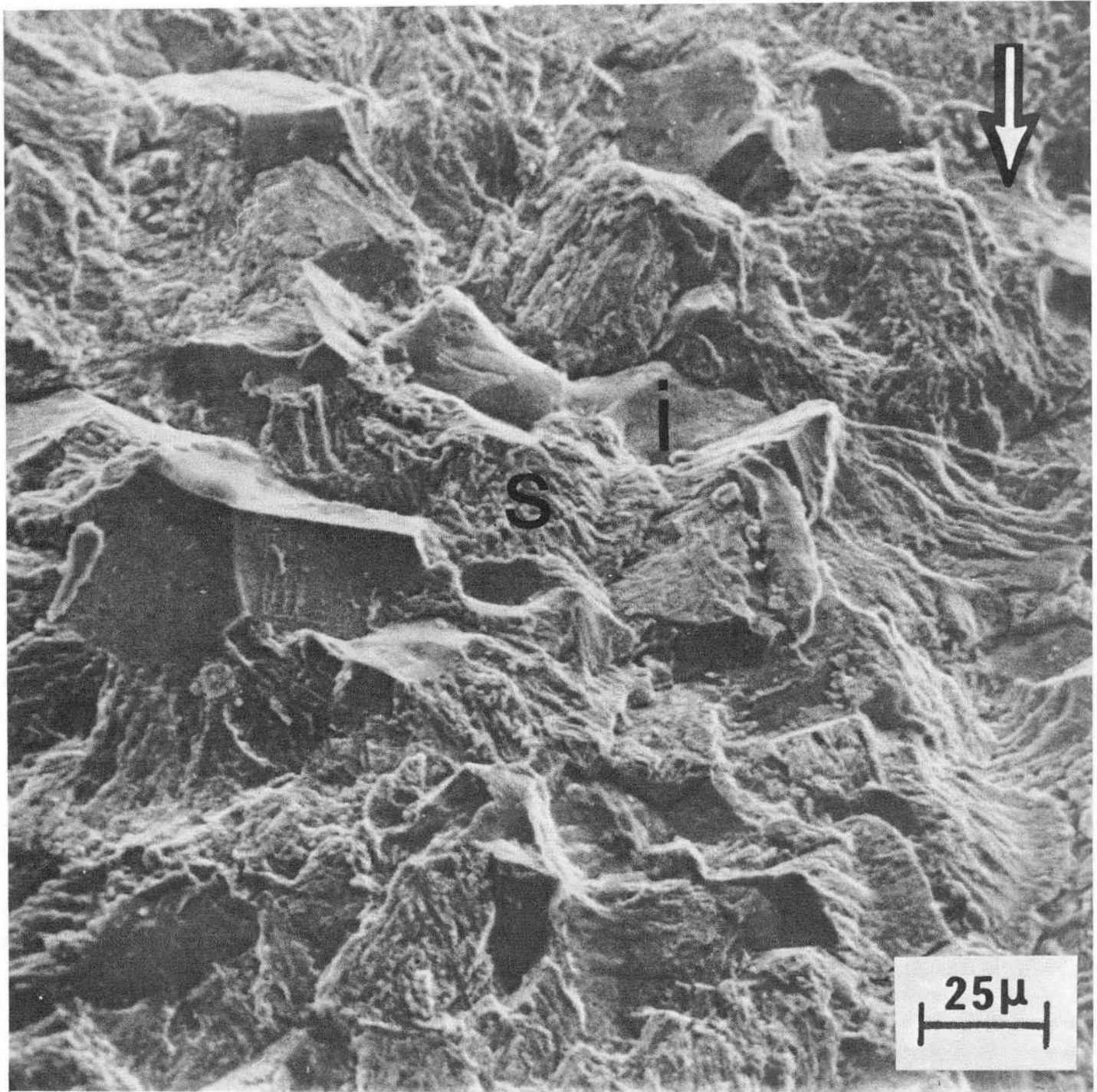
XBI. 755-6270

FIG.9 INFLUENCE OF ΔK ON FATIGUE CRACK PROPAGATION RATES IN SPHEROIDIZED AND PEARLITIC HIGH CARBON STEEL (AFTER REF.34)



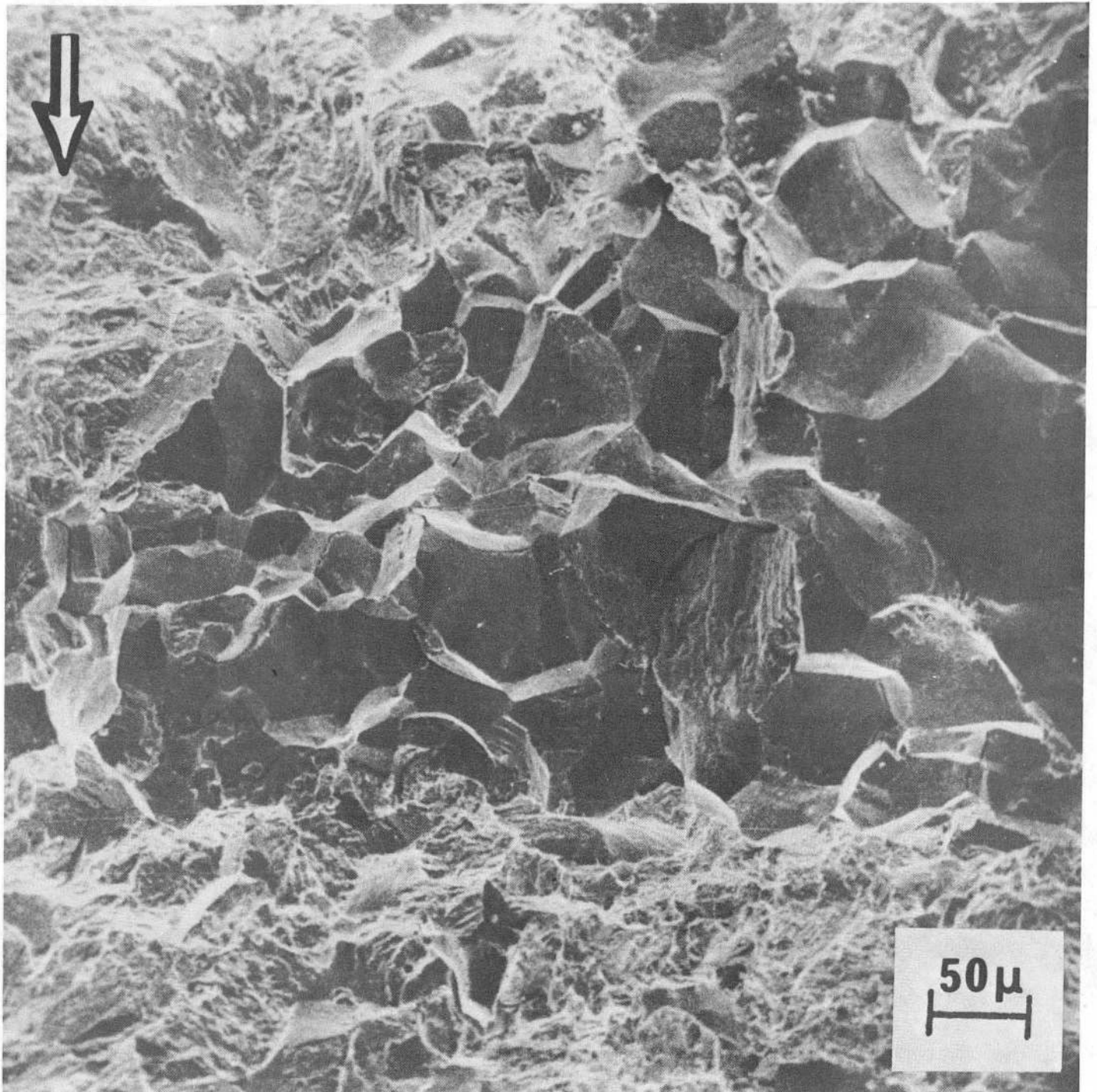
XBB 755-3467

Fig. 10. Occurrence of fibrous fracture (F) during striation growth (s) in a stainless steel weld metal. Voids form around deoxidation products. $\Delta K = 30 \text{ MNm}^{-3/2}$, $K_{\text{MAX}} = 43 \text{ MNm}^{-3/2}$ (after Ref. 48).



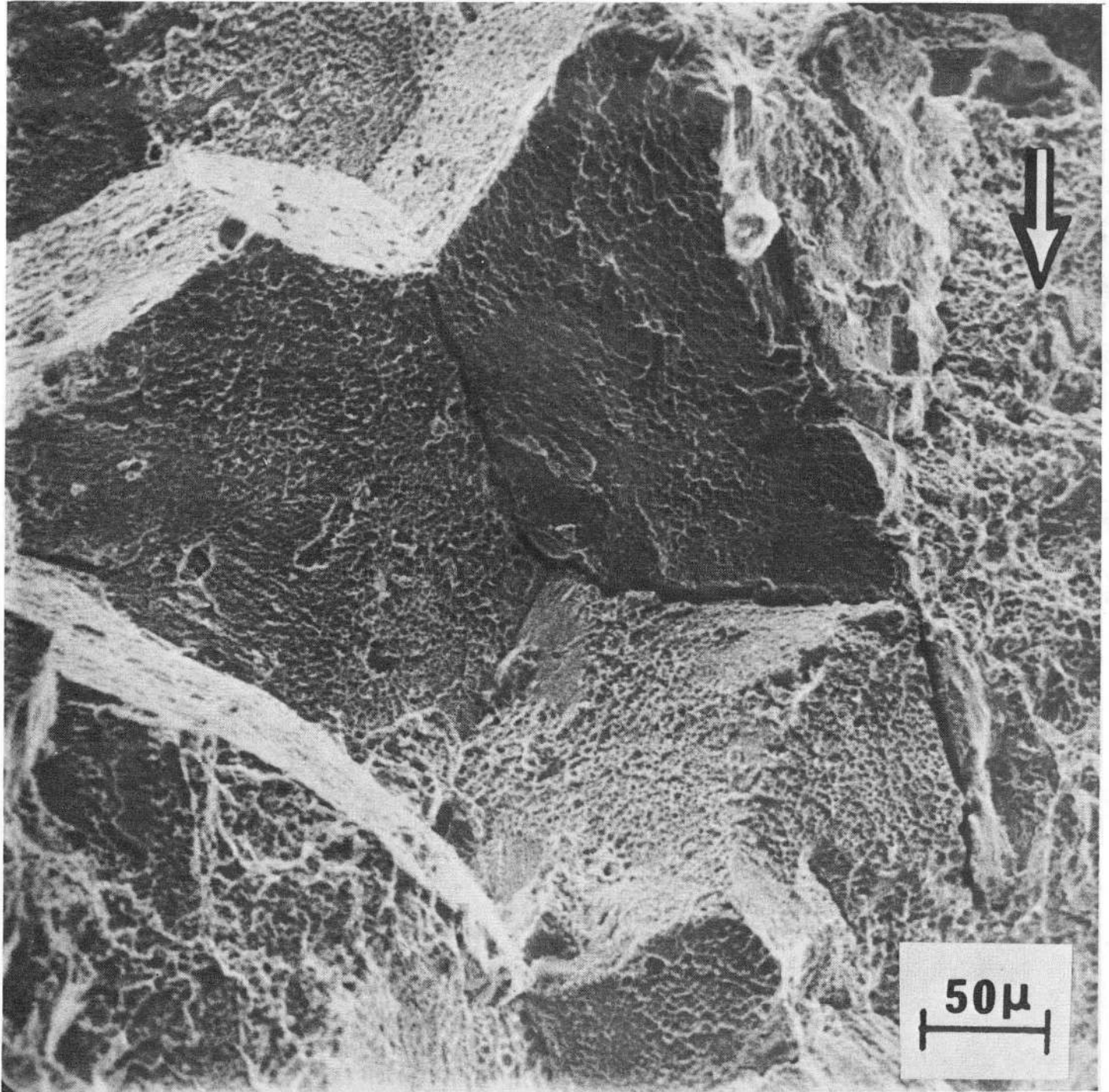
XBB 755-3469

Fig. 11. Isolated intergranulated facets (I) during striation growth (s) in embrittled low alloy steel at lower stress intensities. $\Delta K = 15 \text{ MNm}^{-3/2}$, $K_{\text{MAX}} = 16 \text{ MNm}^{-3/2}$ (after Ref. 17).



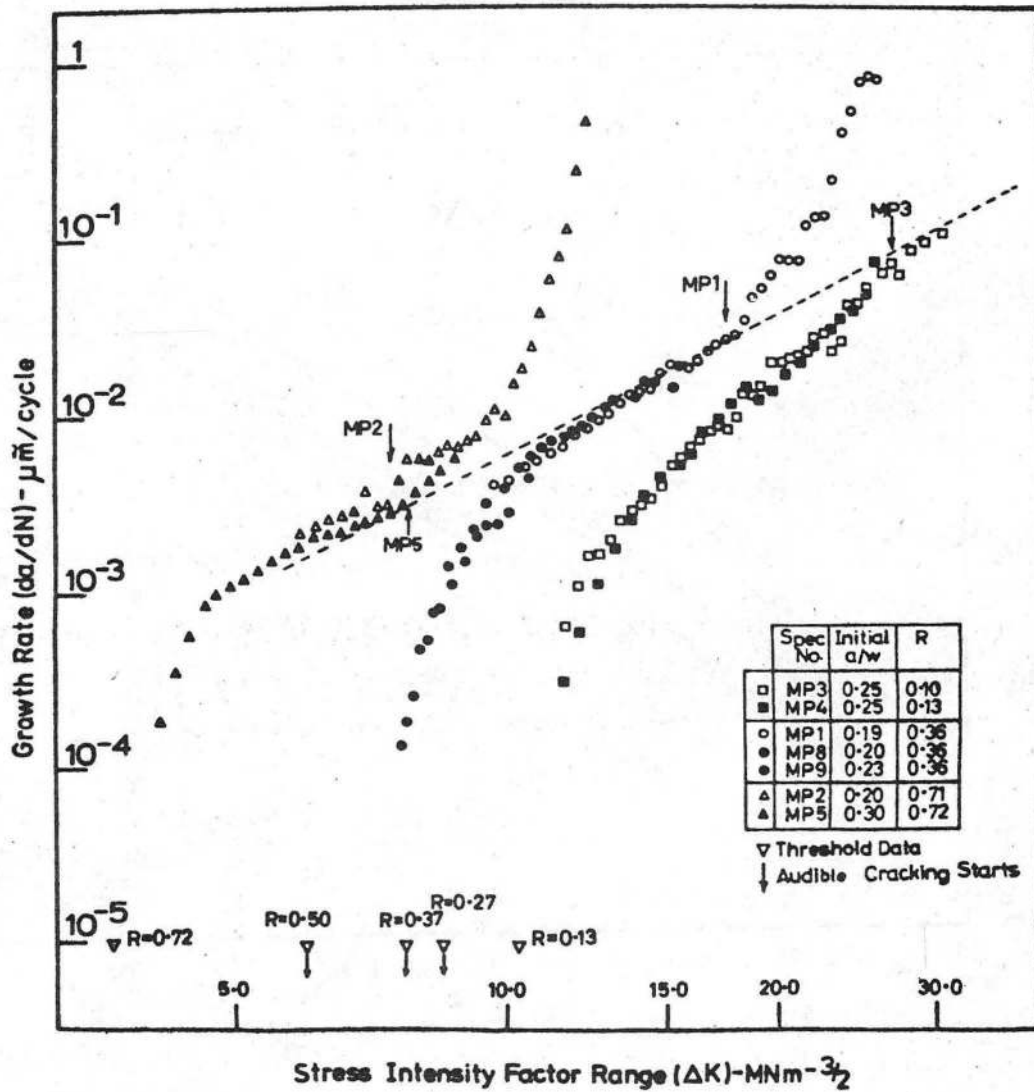
XBB 755-3471

Fig. 12. Burst of brittle intergranular cracking during striation growth in fatigue of embrittled low alloy steel. $\Delta K = 26 \text{ MNm}^{-3/2}$, $K_{\text{MAX}} = 43 \text{ MNm}^{-3/2}$ (after Ref. 17).



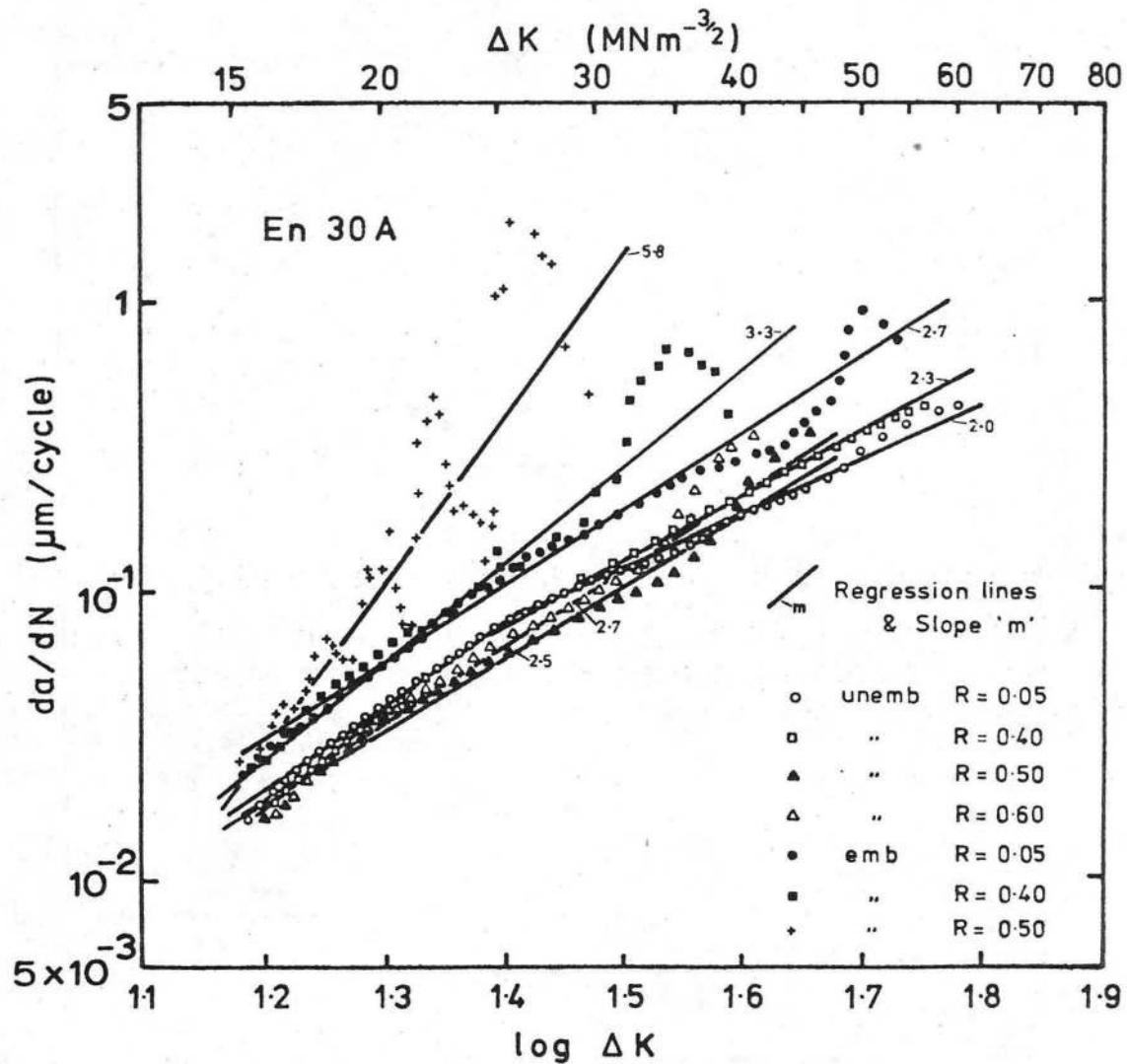
XBB 755-3472

Fig. 13. Intergranular fibrous failure during fatigue of "overheated" low alloy steel En30A at $\Delta K = 30 \text{ MNm}^{-3/2}$, $K_{\text{MAX}} = 60 \text{ MNm}^{-3/2}$ (after Ref. 57).



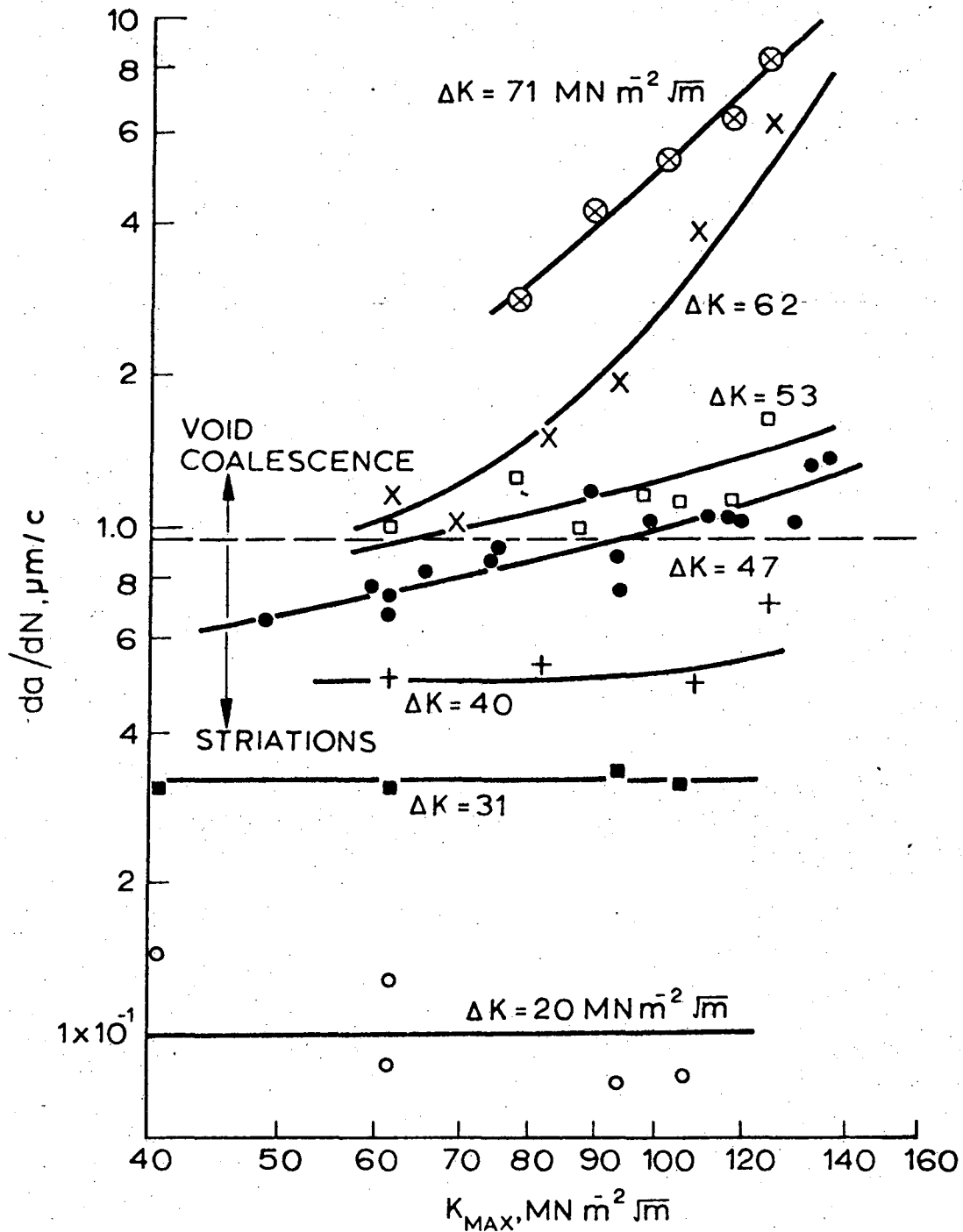
XBL 755-6261

FIG. 14 EFFECT OF MEAN STRESS ON THE GROWTH RATE FOR A HIGH MANGANESE STEEL WHERE CLEAVAGE CRACKING ACCOMPANIES STRIATION FORMATION AS THE MECHANISM OF GROWTH (AFTER REF. 81)



XBL 755-6262

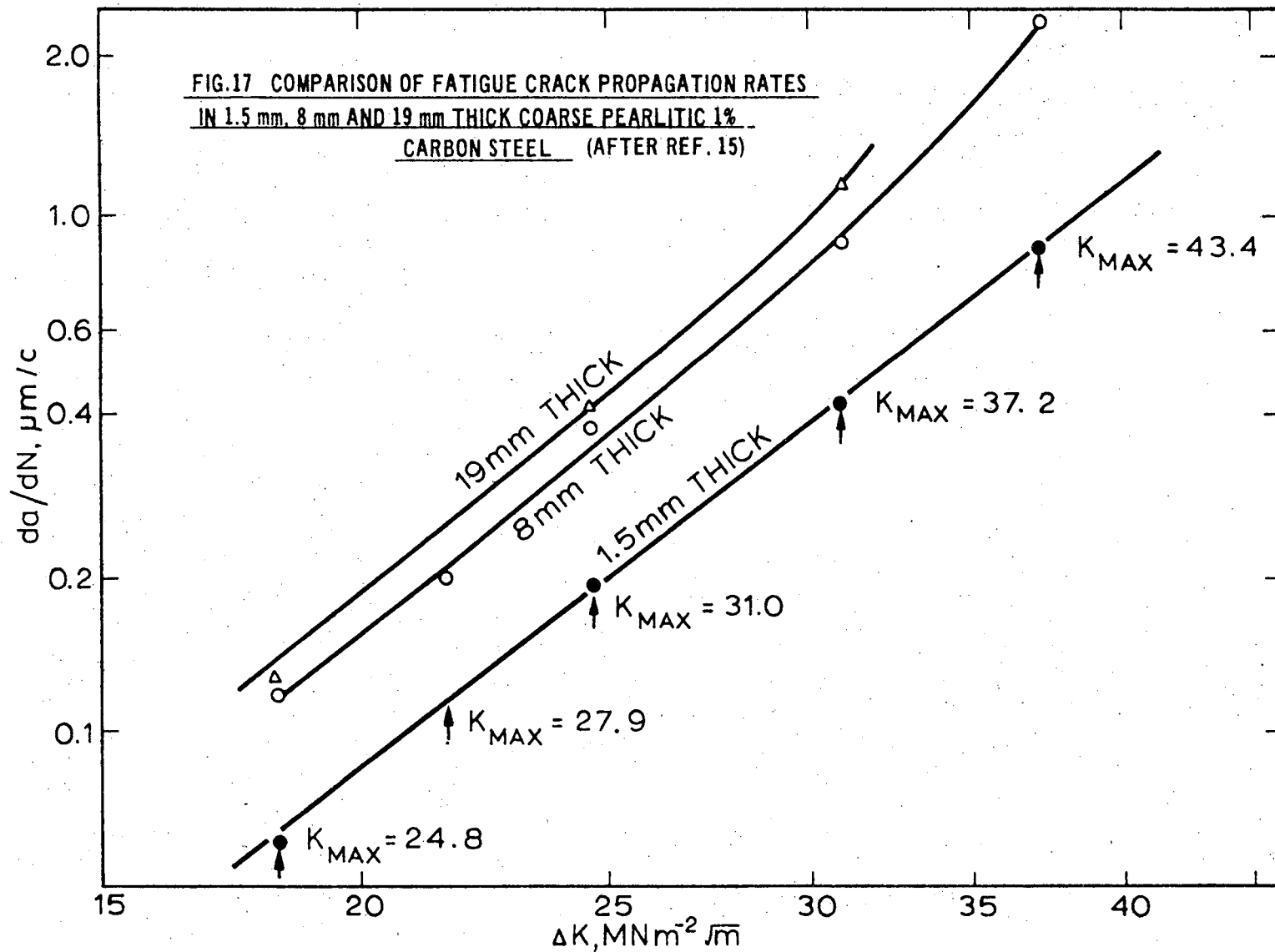
FIG. 15 VARIATION OF CRACK GROWTH RATE (da/dN) WITH ALTERNATING STRESS INTENSITY (ΔK) FOR UNEMBRITLED AND EMBRITLED En 30A OVER RANGE OF R VALUES 0.05-0.60 (NUMBERS INDICATE SLOPES 'm' OF REGRESSION LINES DRAWN THROUGH DATA POINTS (AFTER REF. 17))

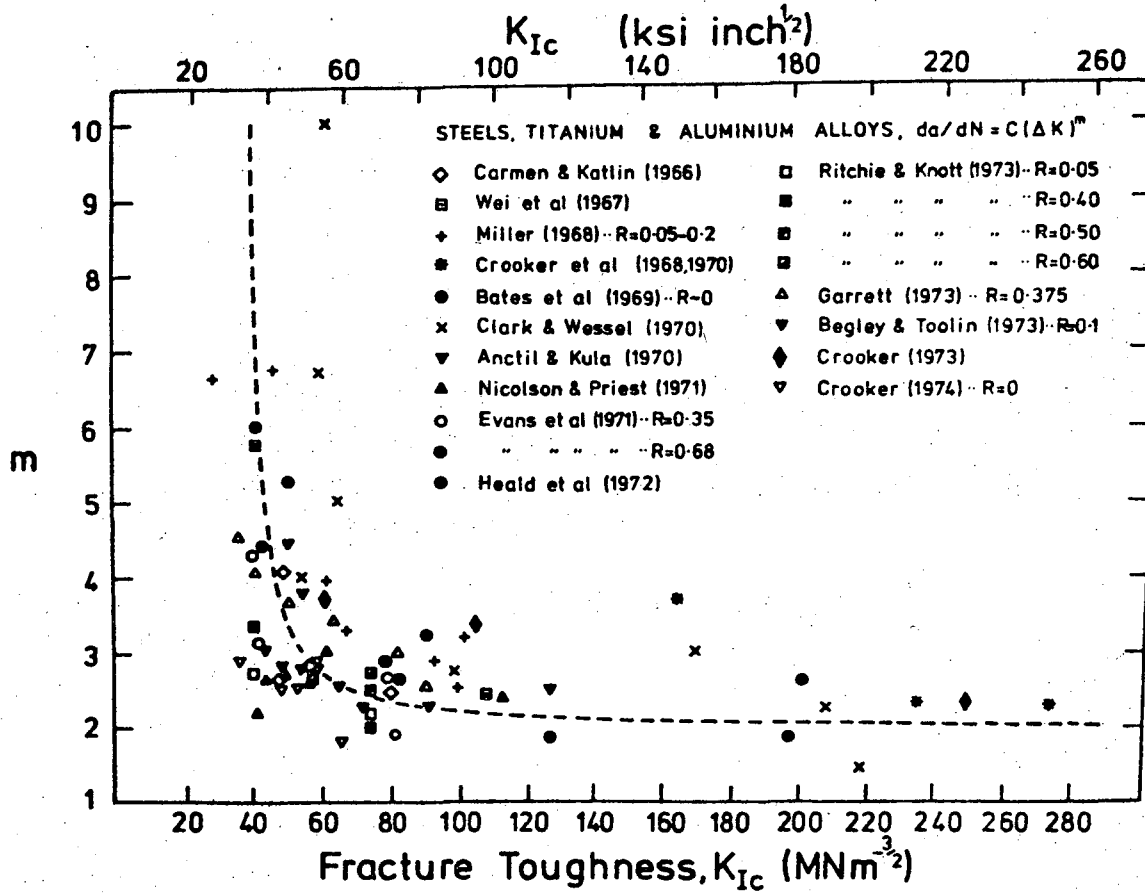


XBL 755-6264

FIG. 16 THE EFFECT OF K_{MAX} ON CRACK PROPAGATION RATES FOR SEVERAL VALUES OF ΔK IN A LOW ALLOY WELD METAL (WELD 1)

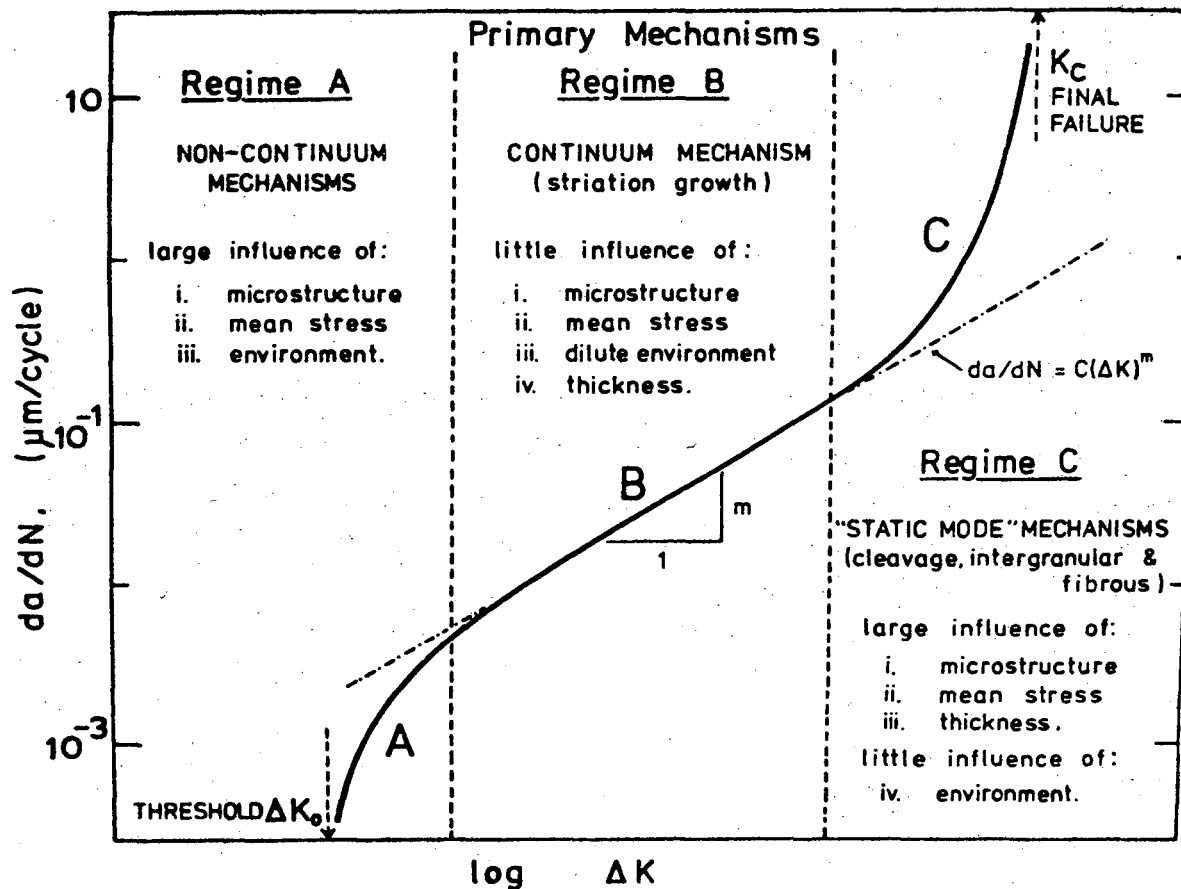
(AFTER REF. 58)





XBL 755-6259

FIG. 18 VARIATION OF THE EXPONENT 'm' IN THE PARIS RELATIONSHIP (EQUATION 3) WITH STATIC FRACTURE TOUGHNESS K_{Ic} FOR A NUMBER OF MEDIUM TO HIGH STRENGTH STEELS, ALUMINIUM ALLOYS AND TITANIUM ALLOYS (AFTER REF. 17)



XBL 755-6258

FIG. 19 SUMMARY DIAGRAM SHOWING THE PRIMARY FRACTURE MECHANISMS ASSOCIATED WITH THE 'SIGMOIDAL' VARIATION OF FATIGUE CRACK PROPAGATION RATE da/dN WITH ALTERNATING STRESS INTENSITY ΔK . ΔK_0 IS THE THRESHOLD STRESS INTENSITY FOR CRACK GROWTH AND K_c THE STRESS INTENSITY AT FINAL FAILURE (TERMINAL K) (REF. 81)

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