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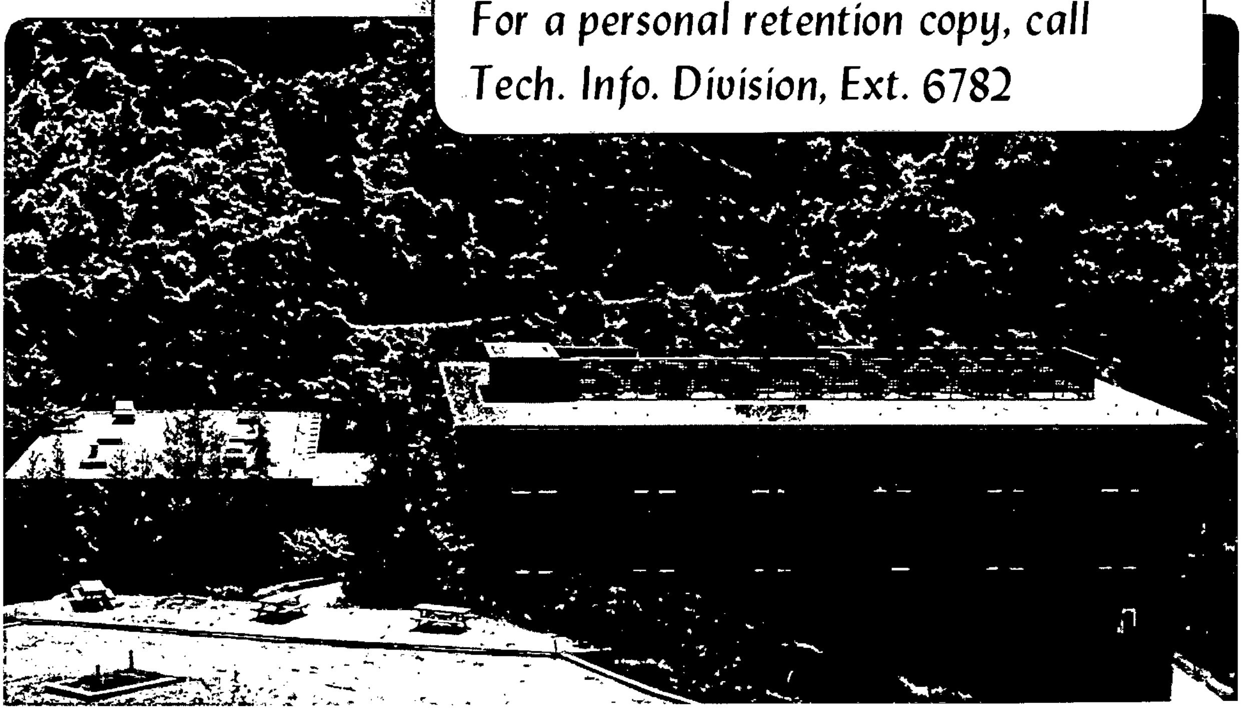

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COMMUNICATION

SOME CONSIDERATIONS ON FATIGUE CRACK CLOSURE AT NEAR-THRESHOLD STRESS INTENSITIES DUE TO FRACTURE SURFACE MORPHOLOGY

R. O. Ritchie and S. Suresh

1. *Introduction:* Recently there has been considerable interest in mechanisms of fatigue crack closure, particularly at very low stress intensities approaching the threshold stress intensity range ΔK_o, below which cracks remain dormant or grow at experimentally undetectable rates[1-13]. Crack closure, as first popularized by Elber[14], was considered to arise from the fact that during fatigue crack growth, material is plastically strained at the crack tip and due to the restraint of surrounding elastic material on this residual stretch, some closure of the crack surfaces occurs at positive loads during the fatigue cycle. This concept, which we term *plasticity-induced crack closure*[4], has proved to be extremely effective in explaining, at least qualitatively, many aspects of fatigue crack propagation behavior including the influence of load ratio[*][14], the role of variable amplitude loading[15], and so forth. It has become clear, however, that such plasticity-induced closure is most prevalent under essentially plane stress conditions,[16,17] and yet, at the ultralow growth rates ($<10^{-6}$ mm/cycle) associated with near-threshold fatigue where plane strain conditions invariably exist, very significant effects of crack closure have been observed[1-13]. To account for such large closure effects in plane

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[*]Load ratio R is defined as K_{min}/K_{max}, where K_{min} and K_{max} are the minimum and maximum stress intensities of the loading cycle.

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strain, several "microscopic" mechanisms have recently been proposed, based on the role of crack flank corrosion deposits[4-6] and fracture surface roughness or morphology[2,6,10,12,13] (Fig. 1). The former effect, termed *oxide-induced crack closure*[4], follows from the fact that during near-threshold crack growth at low load ratios in moist environments, corrosion products of a thickness comparable with the size of crack tip opening displacements can build up near the crack tip, thus providing a mechanism for enhanced closure such that the crack is wedged-closed at stress intensities above K_{min}.[4-6] For lower strength steels, tested in moist air (30% relative humidity) at $R = 0.05$, oxide films have been observed on near-threshold fracture surfaces with a maximum thickness some twenty times larger than films naturally formed on metallographically-polished samples exposed to the same environment for similar time periods[6]. The closure effect promoted by such deposits, which are considered to arise via a mechanism of fretting oxidation[18], aided by plasticity-induced closure and significant Mode II displacements[19], has been substantiated by Auger measurements of oxide thicknesses[6-8] and direct experimental closure measurements using ultrasonics techniques[8]. As discussed in detail elsewhere[4-8], this model has provided a very plausible explanation for many of the somewhat surprising observations on the role of environment (i.e., hydrogen, water vapor, inert gas, vacuum, etc.) in influencing near-threshold fatigue behavior in steels.

In analogous fashion, significant effects of crack closure can arise from an irregular or rough fracture surface morphology[2,6,10,12,13]. As first reported by Walker and Beevers[2] and subsequently modelled by McEvily and others[10,13,7], where the size-scale of the fracture surface roughness is comparable with crack tip opening displacements and where significant Mode II displacements exist (a situation found at near-threshold levels at low load ratios), crack closure can again be promoted since the crack may be wedged open at discrete contact points along the crack faces[6] (Fig. 1). This latter effect, termed *roughness-induced crack closure*[7], can provide an important contribution to the role of microstructure in influencing near-threshold crack growth[3,6,7,11,12] and further may have wider significance to a larger range of engineering materials[13].

The purpose of the current note is to discuss the origin and significance of this latter mechanism of roughness-induced crack closure in the light of recently published data[2,6-8,10-13].

2. *Mechanistic Aspects:* Initial attempts to quantify the effect of fracture surface roughness on fatigue crack closure were reported by Purushothaman and Tien[21], who simply suggested that the closure stress intensity $K_{c\ell}$ could be estimated by equating the change in fracture surface asperity height (taken as a function of true fracture ductility) to the crack tip opening displacement. However, the model did not incorporate the role of crack tip Mode II displacements which are clearly very relevant to the extent of fracture surface interference at near-threshold levels. This can be appreciated from the replica studies on fatigue cracks in titanium[2] and mild steels[13] which show the mismatch (arising from crack tip shear displacements) of the serrated fracture surface profile at near-threshold levels (c.f. Fig. 1), compared to a much more planar profile at higher growth rates.

This distinction between behavior at near-threshold levels and at higher growth rates is consistent with the different modes of crack advance in these two regimes[13]. As pointed out by Tomkins[22], fatigue crack propagation can be considered in terms of intense localized shear deformation in flow bands near the crack tip which results in the creation of new crack surface by shear decohesion at the tip (Fig. 2). However, when the crack and its associated local plasticity is contained within a few grain diameters, growth is confined to one shear direction, with one primary slip system in the direction of growth and a tensile component normal to this (Fig. 2a). Such crack propagation, called Stage I in Forsyth's original terminology[23], proceeds under a combination of Mode II plus Mode I displacements. Support for such Stage I growth at near-threshold levels is evident from work on low carbon steel[24] which showed a marked transition from a tensile mode striation growth above $\sim 10^{-6}$ mm/cycle to an intense shear mode of fracture close to ΔK_o. Further, stereoimaging studies[19] have clearly confirmed a strong Mode II component to crack growth at near-threshold levels. Conversely, at higher stress intensities where the extent of crack tip plasticity encompasses many grains, crack advance proceeds by the operation of two slip systems, either simultaneously or alternately[25-27], resulting in the formation of ductile striations (Fig. 2b). Such crack growth, termed Stage II[23], is more planar in nature, and occurs in pure Mode I perpendicular to the principal tensile stresses.

Thus, at near-threshold stress intensities where the maximum plastic zone size (r_y) is typically smaller than the grain size (d), fatigue crack growth takes place by the Stage I mechanism (Fig. 2a), and the resulting serrated or faceted fracture surface morphology coupled with crack tip

Mode II displacements would be expected to generate high closure loads via the mechanism depicted in Fig. 1. At higher growth rates, on the other hand, as the plastic zone size increases, the transition to Stage II propagation in Mode I results in a more planar fracture surface morphology such that closure loads would be expected to diminish.

Such notions are in agreement with experimental closure measurements on steels, aluminum and titanium alloys which in general show a decrease in closure stress intensity $K_{c\ell}$ with increasing ΔK[13,20,28]. In particular, the recent closure measurements on 2048 aluminum by Asaro et al[20] show a distinct transition from a region of high closure ($K_{c\ell}/K_{max} \sim 0.5$) to one of low closure ($K_{c\ell}/K_{max} \sim 0.2$) with increasing growth rates. By examining several coarse and fine-grained microstructures, these authors concluded that the transition could not be fully accounted for by changes in specimen compliance but rather occurred when the computed maximum plastic zone size (r_y) was fully extended over at least one grain (Fig. 3). This is entirely consistent with the roughness-induced closure concepts presented above since it is to be expected that high closure loads would be observed when the plastic zone is small compared to the grain size as a consequence of faceted Stage I crack growth. Furthermore, at larger plastic zone sizes, the transition to a Stage II striation growth mechanism would be expected to result in a corresponding reduction in closure.

The distinction between crack growth behavior at near-threshold levels and higher stress intensities is further supported by extensive fractographic evidence in the literature. In fact, near-threshold crack growth has been termed "microstructurally-sensitive"[29] owing to the presence of planar facets on fracture surfaces which are intergranular in ferritic steels and transgranular in austenitic stainless steels and alloys of titanium, aluminum, copper and nickel*. However, the proportion of facets is found to diminish with increasing growth rates and to be largely absent once the maximum plastic zone roughly exceeds the grain size[33]. Correspondingly, at higher growth rates, crack propagation mechanisms have been termed "microstructurally-insensitive". The salient features associated with these modes of fatigue crack growth are summarized in Fig. 4.

3. *Implications:* The implications of the Stage I mechanism of near-threshold crack advance with its associated large roughness-induced closure contributions are particularly relevant to the effects of grain size on fatigue crack growth behavior[6]. Whereas refining grain size can be beneficial in raising the fatigue limit or endurance strength of (planar slip) materials[34], the effect of grain size on crack propagation has in general been found to be negligible in most studies at intermediate growth rates[31,34,35]. However, at lower near-threshold growth rates, most reported observations show decreasing growth rates (and higher thresholds) in coarse-grained materials[31]. Although several microstructural mechanisms have been proposed for this effect (see ref. 31 for a review), it is significant that where behavior has been examined at both low and high load ratios, the beneficial effect of coarse grain sizes is absent at the higher load ratios (Fig. 5)[11,36]. This is strong evidence for a prominent role of crack closure, and is analogous to the absence of environmental effects at

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*Such faceted fractures have been described by a number of terms including "crystallographic fatigue" in nickel alloys and "cyclic cleavage" in titanium and stainless steels. For a more complete description see, for example, refs. 30-32.

high load ratios during near-threshold corrosion fatigue where oxide-induced crack closure is important[6]. In terms of the current roughness-induced closure mechanism, such behavior is to be expected, since at near-threshold levels coarser-grained microstructures will promote rougher (more serrated) fracture surfaces. Further, such microstructures delay the transition to the more planar Stage II growth until higher ΔK levels, where the plastic zone size exceeds the grain size. At high load ratios, on the other hand, where closure effects in general are minimal, this mechanism would be insignificant and accordingly no influence of grain size would be expected. Recent studies[11] in pearlitic rail steels have demonstrated this phenomenon particularly clearly in that coarser microstructures tested in room air displayed markedly lower growth rates at $R = 0.1$, yet at $R = 0.7$, behavior was virtually identical for both coarse and fine structures (Fig. 5). Moreover, an increase in fracture surface roughness, together with a corresponding increase in corrosion debris, was observed in the coarse-grained material suggesting that rougher fracture surfaces may additionally promote oxide-induced crack closure (in moist environments), presumably by enhancing abrasion (fretting oxidation) between mating crack faces.

4. *Summary:* It is noted that at near-threshold levels, in addition to the role of plasticity- and oxide-induced crack closure, fracture surface roughness or morphology may promote significant closure effects in plane strain, as similarly noted by Minakawa and McEvily[10,13]. This is considered to result from the fact that, where maximum plastic zones sizes are small compared to the grain size, fatigue crack growth proceeds by a single shear decohesion mechanism (Stage I) with associated Mode II + I displacements. The resulting serrated or faceted fracture surfaces

("microstructurally-sensitive growth") coupled with Mode II crack tip displacements thus induce high closure loads (i.e., $K_{c\ell}/K_{max} \sim 0.5$) by wedging the crack open at discrete contact points. At higher growth rates where the plastic zone encompasses many grains, striation growth via alternating or simultaneous shear mechanisms (Stage II) produces a more planar fracture surface, with pure Mode I displacements, and a corresponding reduction in closure loads. Such concepts of roughness-induced closure are shown to be consistent with observations of the role of coarse grain sizes in reducing near-threshold crack growth rates at low load ratios and of the absence of this effect at high load ratios.

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1. A. J. McEvily: Metal Sci., 1977, vol. 11, p. 274.

2. N. Walker and C. J. Beevers: Fat. Eng. Mat. Struct., 1979, vol. 1, p. 135.

3. C. J. Beevers: Metal Sci., 1980, vol. 14, p. 418.

4. R. O. Ritchie, S. Suresh, and C. M. Moss: J. Eng. Mat. Tech., Trans. ASME Series H, 1980, vol. 102, p. 293.

5. A. T. Stewart: Eng. Fract. Mech., 1980, vol. 13, p. 463.

6. S. Suresh, G. F. Zamiski, and R. O. Ritchie: Met. Trans. A, 1981, vol. 12A, p. 1435.

7. R. O. Ritchie: in Fatigue Thresholds, Proceedings 1st Intl. Conf., Stockholm, J. Bäcklund, A. Blom and C. J. Beevers, eds., EMAS Publ. Ltd., Warley, U.K., 1981.

8. S. Suresh, D. M. Parks, and R. O. Ritchie: ibid.

9. B. L. Freeman, P. Smith, and A. T. Stewart: ibid.

10. K. Minakawa and A. J. McEvily: ibid.

11. G. T. Gray, A. W. Thompson, J. C. Williams, and D. H. Stone: ibid.

12. I. C. Mayes and T. J. Baker: Fat. Eng. Mat. Tech., 1981, vol. 4, p. 79.

13. K. Minakawa and A. J. McEvily: Scripta Met., 1981, vol. 15, p. 633.

14. W. Elber: in Damage Tolerance in Aircraft Structures, ASTM STP 486, p. 280, 1971.

15. E. F. J. von Euw, R. W. Hertzberg, and R. Roberts, in Stress Analysis and Growth of Cracks, ASTM STP 513, p. 230, 1972.

16. T. C. Lindley and C. E. Richards: Mater. Sci. Eng., 1974, vol. 14, p. 281.

17. B. Budiansky and J. W. Hutchinson: J. Appl. Mech., Trans. ASME Series E, 1978, vol. 45, p. 267.

18. D. Benoit, R. Namdar-Tixier, and R. Tixier: Mater. Sci. Eng., 1981, vol. 45, p. 1.

19. D. L. Davidson: Fat. Eng. Mat. Tech., 1981, vol. 3, p. 229.

20. R. J. Asaro, L. Hermann, and J. M. Baik: Met. Trans. A, 1981, vol. 12A, p. 1135.

21. S. Purushothaman and J. K. Tien: in Strength of Metals and Alloys, Proc. ICSMA5 Conf., Pergamon Press, New York, vol. 2, p. 1267, 1979.

22. B. Tomkins: Metal Sci., 1979, vol. 13, p. 387.

23. P. J. E. Forsyth: in Crack Propagation, Proc. Symp., Cranfield, College of Aeronautics, Cranfield Press, 1962, p. 76.

24. A. Ohtsuka, K. Mori, and T. Miyata: Eng. Fract. Mech., 1975, vol. 7, p. 429.

25. B. Tomkins and W. D. Biggs: J. Mater. Sci., 1969, vol. 4, p. 544.

26. R.M.N. Pelloux: Eng. Fract. Mech., 1970, vol. 1, p. 697.

27. P. Neumann: Zeitschrift f. Metallkunde, 1967, vol. 11, p. 780.

28. P. C. Paris and L. Hermann: Proc. Intl. Cong. Theor. Appl. Mech., Delft, W. T. Koiter, ed., North-Holland, Amsterdam, 1977.

29. R. J. Cooke and C. J. Beevers: Mater. Sci. Eng., 1974, vol. 13, p. 201.

30. C. J. Beevers: Metal Sci., 1977, vol. 11, p. 362.

31. R. O. Ritchie: Int. Metals Rev., 1979, vol. 20, p. 205.

32. G. R. Yoder, L. A. Cooley, and T. W. Crooker: Met. Trans. A , 1978, vol. 9A, p. 1413.

33. J. D. Frandsen and H. L. Marcus: Scripta Met., 1975, vol. 9, p. 1089.

34. R.M.N. Pelloux: in Ultrafine-Grain Metals, Syracuse Univ. Press, p. 231, 1970.

35. A. W. Thompson and R. J. Bucci: Met. Trans., 1973, vol. 4, p. 1173.

36. E. K. Priddle: Scripta Met., 1977, vol. 11, p. 49.

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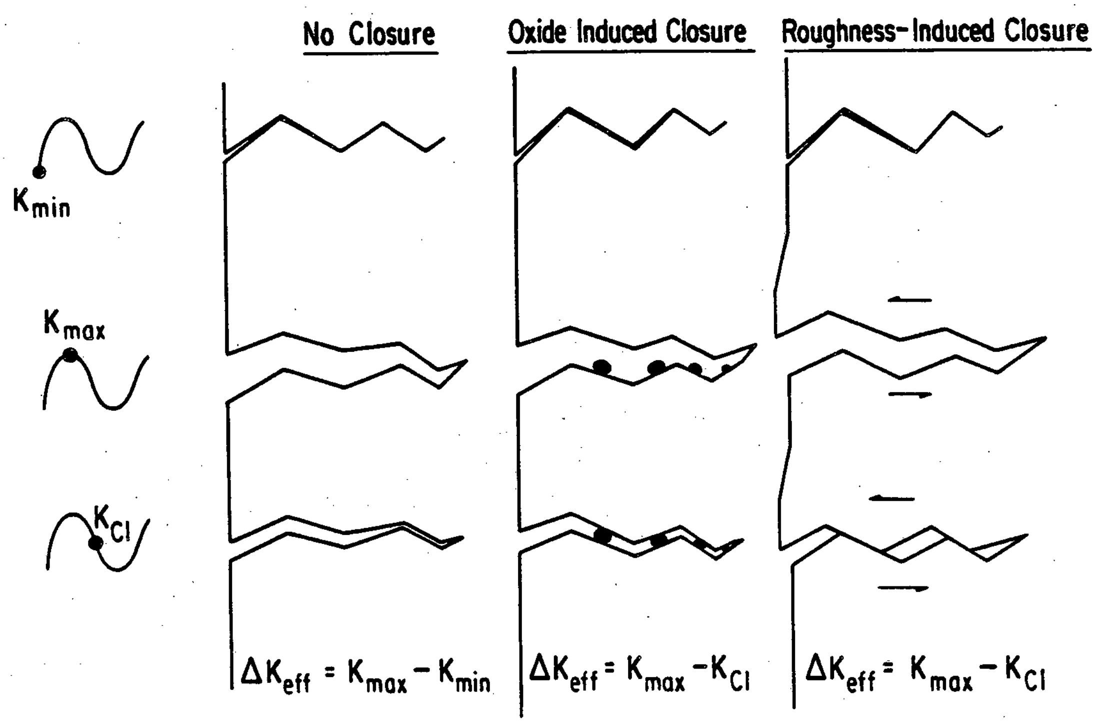


Fig. 1. Schematic illustration of possible mechanisms of fatigue crack closure at near-threshold stress intensities. ΔK_{eff} is the effective stress intensity range, given by $K_{max} - K_{c\ell}$, where K_{max} is the maximum stress intensity and $K_{c\ell}$ is the stress intensity to close the crack ($K_{c\ell} \geq K_{min}$, the minimum stress intensity).

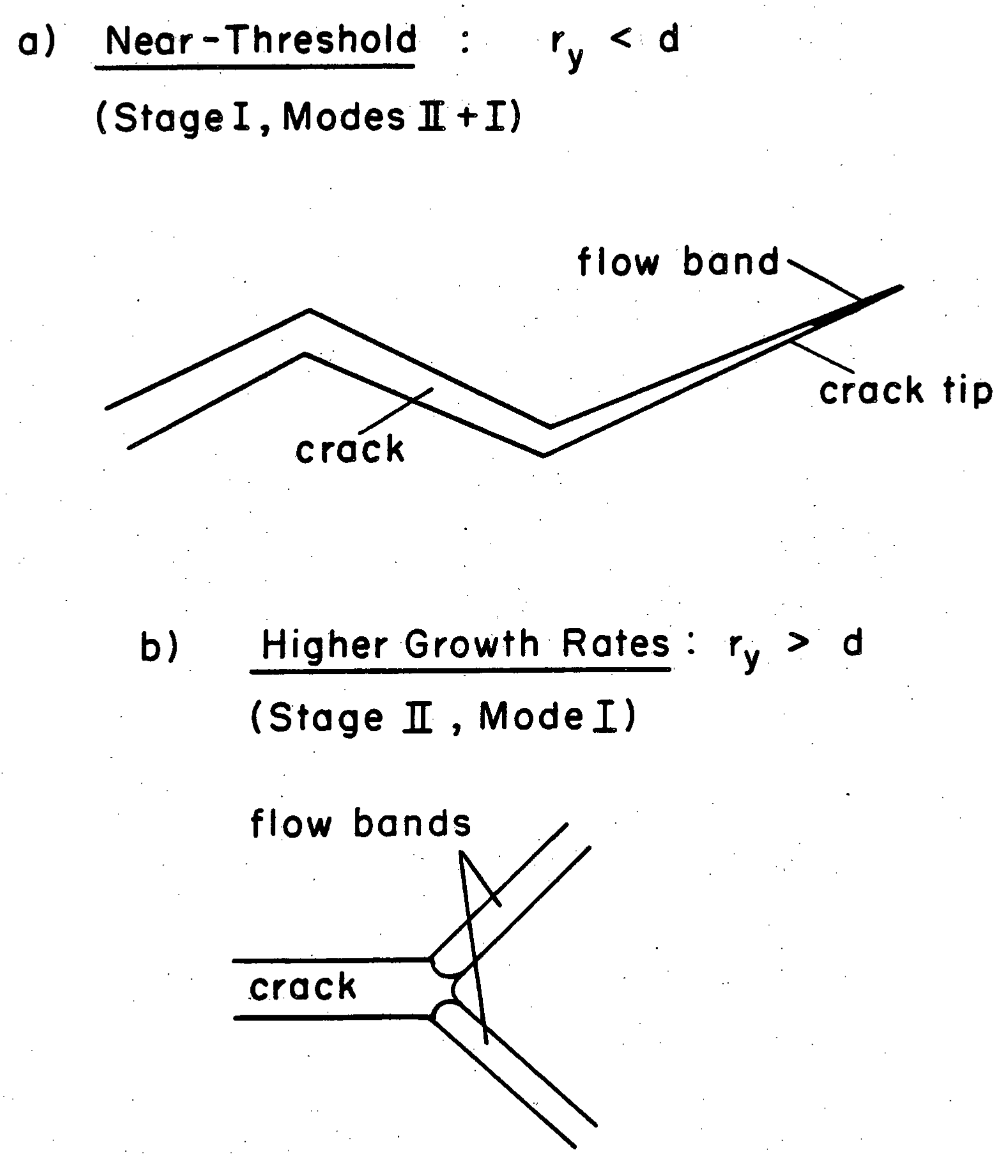


Fig. 2. Crack opening profiles corresponding to a) near-threshold (Stage I) and b) higher growth rate (Stage II) fatigue crack propagation (after ref. 22).

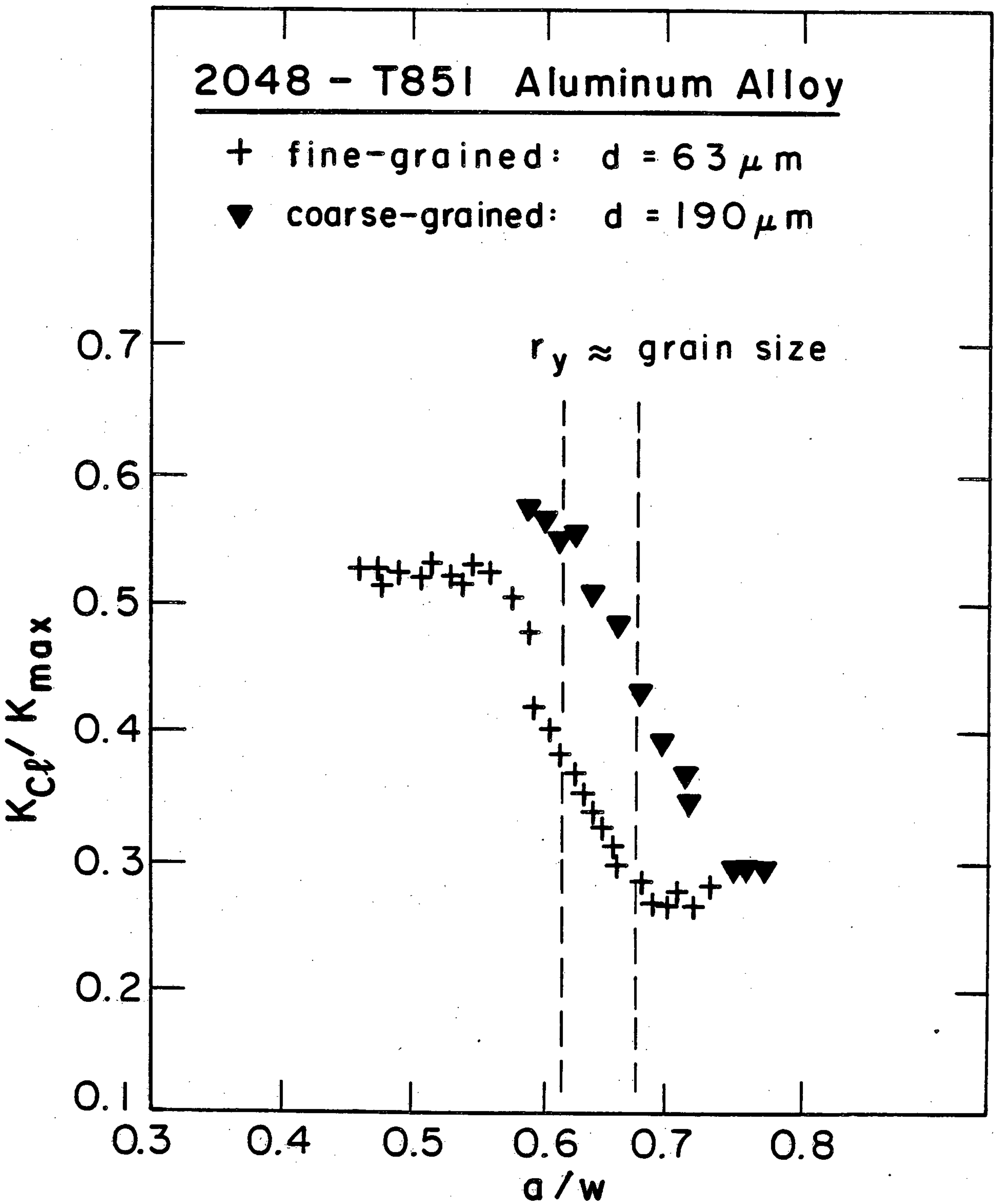
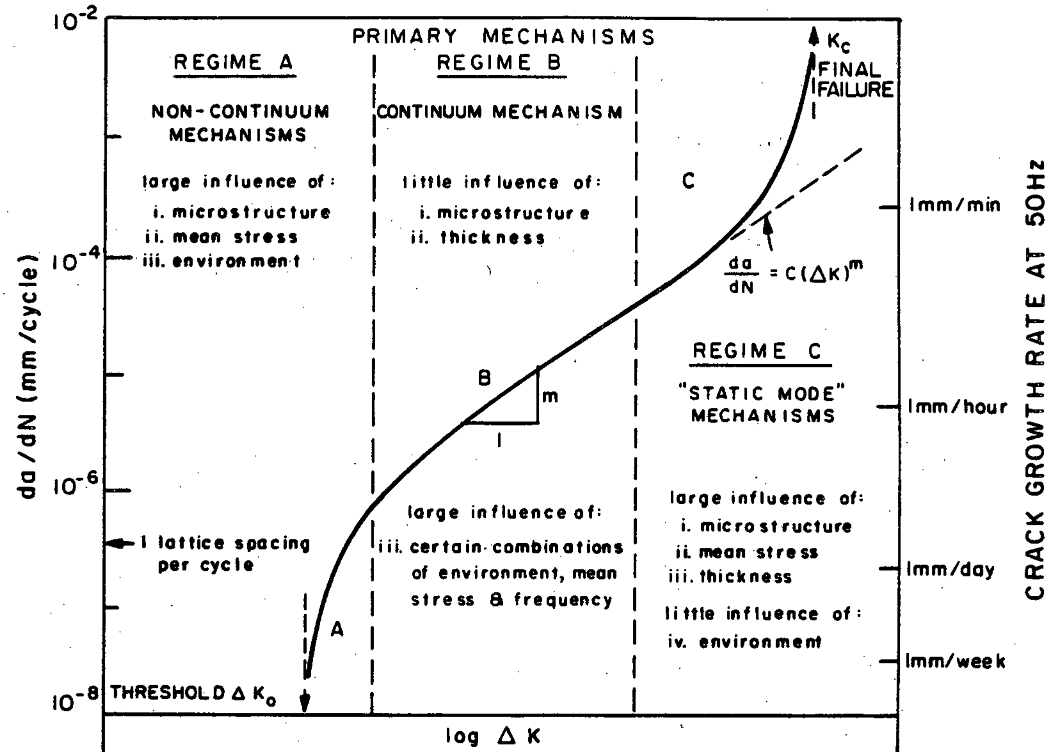


Fig. 3. Measurements of fatigue crack closure in coarse and fine grain 2048-T851 aluminum alloy, represented in terms of the ratio of closure stress intensity to maximum stress intensity ($K_{c\ell}/K_{max}$). Note transition from high closure to low closure once the maximum plastic zone size (r_y) exceeds roughly the grain size (d) (after ref. 20).



REGIME	NEAR-THRESHOLD	MID-GROWTH RATE	HIGH-GROWTH RATE
TERMINOLOGY	Stage I	Stage II	—
MECHANISM	Single Shear	Alternating or Simultaneous Shear on Two Slip Systems (Striation Growth)	Additional Static Modes
CRACK TIP DISPLACEMENTS	Modes II + I	Mode I	Mode I
NATURE OF FRACTURE SURFACES	Serrated or Faceted	Planar (Striations)	Additional Cleavage, Intergranular or Microvoid Coalescence
CLOSURE LEVELS	High $\left(\frac{K_{cl}}{K_{max}} \gtrsim 0.5\right)$	Low $\left(\frac{K_{cl}}{K_{max}} \lesssim 0.2\right)$	—
OCCURRENCE	Plastic Zone Size $\lesssim$ Grain Size	Plastic Zone Size $\gtrsim$ Grain Size	Plastic Zone Size >> Grain Size

Fig. 4. Schematic illustration of the three regimes of fatigue crack propagation behavior and their corresponding mechanisms and characteristics.

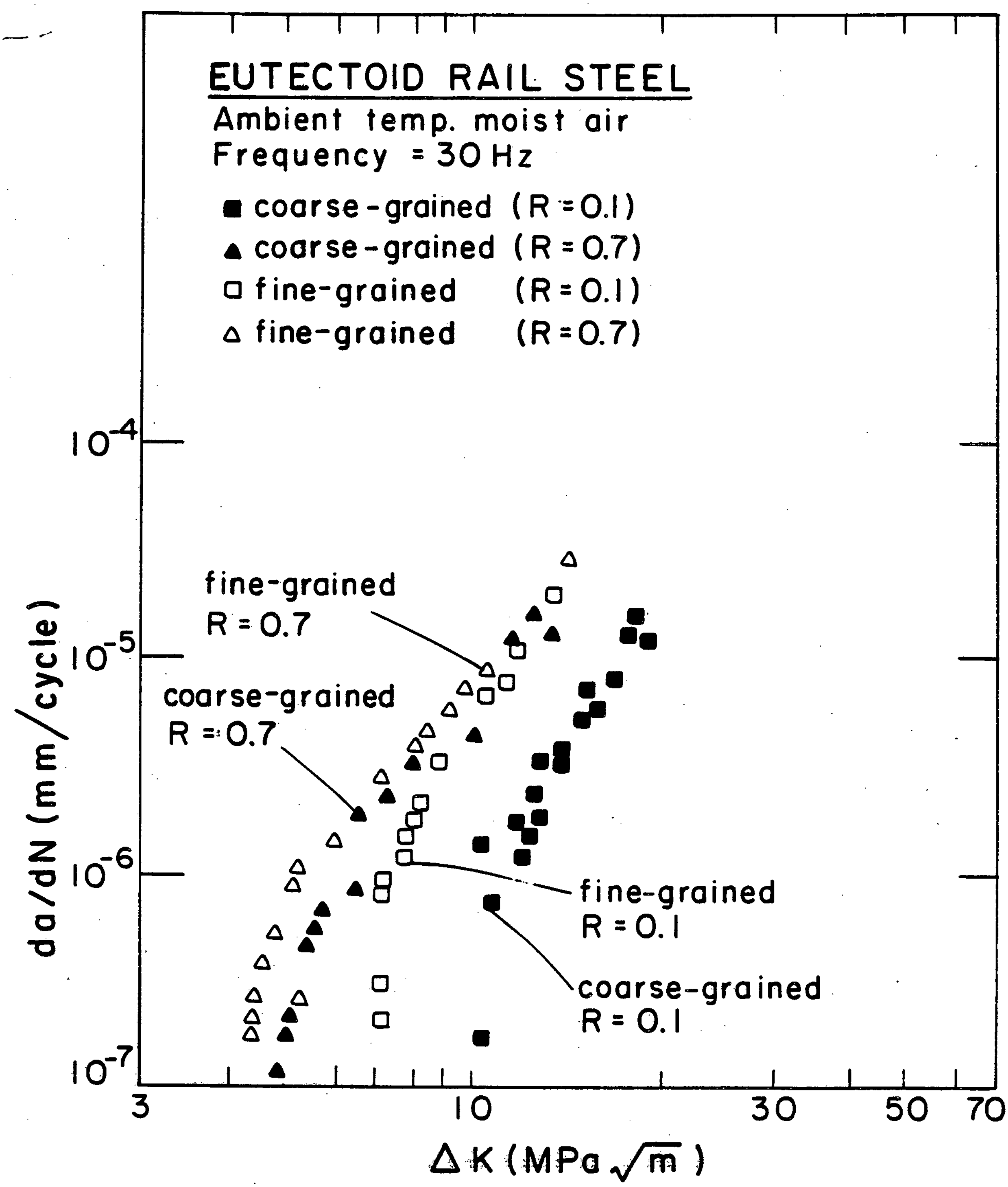


Fig. 5. Fatigue crack propagation in fully pearlitic hot-rolled rail steel (R = 0.1 and 0.7), showing lower growth rates and higher threshold values in the coarse-grained microstructure (d = 130μm) compared to the fine-grained structure (d = 25μm), at low load ratios only (after ref. 11).

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