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FLOW LOCALIZATION AND THE FRACTURE TOUGHNESS OF HIGH STRENGTH MATERIALS

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ABSTRACT

Many alloys with a high flow resistance unfortunately also have poor fracture toughness and poor fatigue characteristics. This is due to flow being localized within very narrow bands, so that the high local strains required for fracture are more readily attained than when the deformation is homogeneous. Relevant previous work is subjected to a critical appraisal, which is used as a basis for the development of a framework understanding of the fracture toughness characteristics of materials that are susceptible to flow localization. Although the paper's emphasis is on the behaviour of high strength materials, a deliberate attempt is made to relate this behaviour to that of low strength materials, which have received fairly extensive consideration by previous workers.

I. INTRODUCTION

Many high yield strength alloys, for example those having a multi-phase microstructure produced by precipitation-hardening procedures, unfortunately have poor fracture toughness and poor fatigue properties in relation to their high yield stresses. Since, moreover, the general trend [1] is for fracture strength to decrease with increasing yield strength, such alloys are limited in their practical applications, where a primary requirement is often a combination of both adequate strength and adequate fracture properties. Recent review papers, [1-3] covering a wide range of alloys, have indicated that the inferior fracture behaviour is due to flow being localized within narrow zones, whereupon the critical amount of local deformation needed for fracture processes to operate, is more readily attained than when the flow is homogeneously distributed. With multi-phase alloys, metallographic observations (for example [4,5]) have shown that flow concentration is usually associated with dislocations cutting the hardening particles (Figures 1 and 2) rather than by-passing them and leaving residual dislocations in their wakes. Particle cutting allows many dislocations emitted by a given source to readily move along a single plane; neighbouring sources can then be activated fairly easily and consequently appreciable dislocation movement will occur within a very narrow zone. A somewhat similar situation arises with the same alloys when the applied loads fluctuate [6]. A common observation (Figure 3) is that when dislocations are able to cut the hardening particles, they can move

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relatively easily in opposite directions within a narrow zone. The necessary damage required for fatigue fracture processes to operate, being more readily attained than when the reverse deformation occurs more homogeneously.

In a comprehensive theoretical study the authors have investigated various facets of the flow concentration problem with particular attention being given to the detailed conditions for flow to concentrate, especially as regards the effect of microstructural parameters, and appreciable success has been attained [7] with respect to correlating microstructure with mechanical behaviour. However, it is equally important to consider the consequence of flow concentration, and the investigation has also encompassed this aspect of the problem, the general philosophy of approach being similar to that adopted for the causes, in that appropriate micro-mechanical modelling procedures have been used extensively. The effort has been concentrated primarily on the monotonic loading situation, although implications with respect to fatigue are briefly discussed.

It might be argued that the relation between microstructure and fracture behaviour under operational conditions could be obtained solely from a wide-ranging data-collecting programme. However, this approach is unsatisfactory in several respects; for example, the reasons for any anomalies in the relations between particular fracture properties and metallurgical structure are better appreciated if the underlying processes are clearly understood. Moreover, since operating conditions are often very complex, the relevant data is impractical to obtain as it is usually impossible to adequately cover all the service variables in laboratory tests. To predict operational performance in many cases, extrapolation of data obtained for simplified situations is therefore essential, and this is possible only if the extrapolation procedure is based on sound models of the controlling mechanisms. These arguments all point to the necessity of understanding the problem in depth, and at the same time suggest that the understanding should be quantitative; these views are the motivating force behind the approach adopted in the investigation. The immediate objectives of the present paper are to critically assess the important factors that affect the fracture toughness of materials that display flow localization, and use this assessment to develop an outline understanding that can be used as a basis for more detailed studies especially of material behaviour in complex operational situations, and also act as a guide for rational alloy development procedures. At this stage, unfortunately, it is impossible to proceed beyond a framework understanding since there is a scarcity of relevant experimental data and associated theoretical work; the relevant gaps are indicated.

The present investigation's basis is outlined in Section II which is essentially a summary of well-accepted general quantitative considerations [8] that focus attention on the principles associated with the development of both high hardness and high toughness values in the same material. Most of the published work [1-3] on the consequences of flow localization, particularly of a theoretical and micro-mechanical modelling nature, has been based on the behaviour of fairly low strength materials, for which experimental evidence clearly shows the major factor controlling fracture toughness to be the growth of voids, formed at inclusions in a process zone immediately ahead of a crack tip, followed by the coalescence of the voids with the crack tip. Flow localization followed by fracture within a narrow zone (glide band decohesion) can occur between the crack tip and an inclusion-type void, thereby truncating the coalescence process, and is in essence a perturbating factor in the crack extension process. Attempts

have been made to explain the characteristics of high strength materials using the behaviour of low strength materials as a basis; Section III critically reviews these approaches, but they have clear limitations since with high strength materials, flow localization is the dominant and not merely a perturbating factor. Thus in considering high strength materials it is obviously desirable to take a fresh look at the problem, and to provide a link with low strength materials, Section IV develops, with the considerations of Section II as a basis, a very simple and sufficiently general model that enables the behaviour of both high and low strength materials to be discussed. The intention throughout the paper is to highlight the important implications of the work and suggestions for further theoretical and experimental work are indicated (Section V); these suggestions, if pursued, should enable the outline framework developed in the present paper to be secured, and increase our general understanding of the mechanical behaviour of high strength materials.

II. GENERAL BASIS FOR DISCUSSING THE EFFECT OF FLOW LOCALIZATION ON CRACK EXTENSION

The basis of all modern discussions of crack extension is the classic Griffith relation [9]:

$$\sigma = \left[\frac{4\gamma S_M}{\pi(1-\nu)a}\right]^{1/2} \tag{1}$$

where 2a is the crack length, σ is the applied tensile stress, S_M is the shear modulus, ν is Poisson's ratio and γ is the surface energy; this relation is applicable to a perfectly elastic material and gives the stress required to extend a two-dimensional crack that is contained within an infinite solid deforming under plane strain conditions. Orowan [10] and Irwin [11] generalized this relation to encompass the situation where plastic deformation occurs in the vicinity of a crack tip, provided the deformation is sufficiently localized that the plastic zone size at crack extension is small compared with the crack size. The form of the Griffith relation is maintained, but γ_p replaces 2γ when relation (1) may be re-written as

$$K^2 = E\gamma_p/(1-\nu^2) \tag{2}$$

where $K = \sigma\sqrt{\pi a}$ is the stress intensification factor; a crack therefore extends when the stress intensification factor K attains a critical value $K_{IC} \equiv [E\gamma_p/(1-\nu^2)]^{1/2}$.

This generalization of the Griffith relation has been confirmed by formal plastic-elastic analyses (for example Bilby, Cottrell and Swinden [12]) appropriate for the restricted yield situation, and γ_p has been given a physical interpretation. Thus if $\sigma_y(u)$ is the stress in a material element ahead of the crack tip, and u is the relative displacement, it being assumed for this initial general discussion that yield occurs within an infinitesimally thin zone immediately ahead of the crack tip,

$$\gamma_p = \int_0^\infty \sigma_y(u)du \tag{3}$$

In order to proceed in a simple manner, Cottrell [8] assumed $\sigma_y(u)$ to decrease almost to zero beyond a certain "critical fracture displacement" u_f so that very approximately

$$\gamma_p = \sigma_y u_f \tag{4}$$

where σ_y is now a "characteristic stress" associated with the failure mode by which the crack extends. For a tough material K_{IC} and thereby γ_p should have a high value, i.e. $\sigma_y u_f$ must be large for the operative failure mode. With high yield strength materials, such as appropriately aged precipitation-hardened alloys, σ_y correlates with the yield stress and is obviously high; unfortunately, however, u_f is low and this means that γ_p may not be large, certainly not in relation to the high σ_y values that such materials possess. In very general terms a low u_f value reflects the ease with which the material fractures at a crack tip. Specifically with regard to the materials that are the main subject of the present investigation, u_f reflects the ease with which glide band decohesion occurs within a localized flow band and how this failure mode is influenced by inclusion-type voids. The objective of the next two sections is to build upon these basic principles with a view to providing a framework understanding of the fracture toughness behaviour of materials that are susceptible to flow localization.

III. CRITICAL APPRAISAL OF PREVIOUS DISCUSSIONS OF CRACK EXTENSION IN MATERIALS DISPLAYING FLOW LOCALIZATION

The most recent detailed discussion pertaining to crack extension in materials susceptible to flow localization, is that due to Green and Knott [13]; this discussion is based essentially on the work of Rice and Johnson [14] and extends the Battelle group's considerations [1-3]. The basic model is that of a crack tip deforming in a plane strain mode, the yielding being small-scale in that the extent of spread of plasticity is small compared with the crack size. The slip line field solution (Figure 4), based on the assumption of small geometry changes, is characterized by regions A and B in which the fans are centred and the lines are straight; this solution therefore predicts no strain concentrations ahead of the crack tip although there are intense shear concentrations both above and below the tip, where the lines are curved (region C). When large geometry changes are taken into account [14], however, the crack tip blunts so that the fan region at the tip becomes non-centred, a logarithmic spiral of shear lines focussing at a point D that is at a distance 1.9 δ_t ahead of the crack tip where δ_t is the crack opening (Figure 5); intense strain concentrations are found between D and the blunted crack tip. Green and Knott were primarily concerned with low strength materials for which, as indicated in the Introduction, there is good experimental evidence [1-3] to show that the major feature in crack extension is the growth of voids, formed at inclusions in a process zone immediately ahead of a crack tip, followed by the coalescence of the voids with the crack tip. In applying the blunting crack tip model to such materials, therefore, the region ahead of the tip must be strained sufficiently for a void to nucleate, grow and coalesce with the crack tip.

With non-hardening materials, it is argued [13] that as the external stresses are progressively increased, the plastic strain field first reaches an inclusion situated a distance X_o from the crack tip when $\delta_t \sim$ 0.02-0.1 X_o the precise value depending on the material's yield stress. With many materials, for example steels containing typical inclusions, void nucleation at inclusions is relatively easy and void growth in the stress field of the blunting crack tip proceeds while δ_t increases from 0.02-0.1 X_o to ~0.5 X_o (i.e. 1.9 δ_t = X_o) when the logarithmic spirals

first "envelop" the void in the sense that D coincides with the inclusion centre. Failure is then presumed to occur between the crack and void by localized flow and decohesion along the logarithmic spirals, and there is direct experimental evidence to support this behaviour pattern (Figure 6). It is assumed that this final stage of the failure process produces little increase in δ_t, and therefore the critical value $\delta_t = \delta_i$ for crack extension should be approximately equal to 0.5 X_o for non-hardening materials, a prediction for which Green and Knott provide experimental support in the nature of results for a pre-strained free-machining mild steel, whose work hardening rate is low because of the pre-straining treatment. The critical value δ_i for initiation of crack extension under small-scale yielding plane strain deformation conditions, may be translated into a critical fracture toughness value K_{IC} by associating δ_i with u_f (see previous section) when relations (2) and (4) give

$$K_{IC} \sim \sqrt{E\sigma_y\delta_i} \tag{5}$$

σ_y being the material's yield stress.

With a work-hardening material, the strain distribution near the crack tip is more widespread, since each strain increment causes a hardening that encourages the next increment to occur in a different position. Predicted values of δ_i then exceed 0.5 X_o, the precise value depending on the initial inclusion size; experimental support for this predicted trend is presented by Green and Knott.

The initiation of crack extension may conveniently be discussed in terms of two regions: (a) a small process zone immediately ahead of the crack tip, and within which the strains are large, being approximately equal to the fracture strain; (b) a much larger macroscopic plastic zone within which the strains are smaller, the size of this zone being sufficiently large to accommodate the displacements associated with the process zone. It should be emphasized that initiation of crack extension and unstable crack extension essentially coincide when the crack tip deforms under small-scale yielding plane strain conditions, [14,15] although there are isolated cases, for example maraging steels, where appreciable stable crack growth is possible under such conditions. Any stable crack advance, referred to as the "stretch zone", [16] is usually limited to the advance associated with the tip blunting process itself, and consequently initiation of crack extension is the controlling event in the overall fracture process. However, if a specimen is insufficiently thick to maintain plane strain conditions throughout a substantial proportion of its thickness, appreciable macroscopic stable crack growth can occur because shear lips at the surface retard crack extension in the central portion of a specimen; in such circumstances, crack extension initiation is not the critical event in the fracture process.

Green and Knott use the preceding behaviour pattern for low strength materials to make some observations concerning high strength materials, arguing that their low fracture toughness is due primarily to a very low work-hardening rate. Strong experimental support for this correlation between toughness and work-hardening rate is the free-machining mild steel result, already mentioned, that pre-straining, which reduces the work-hardening rate, causes crack extension initiation when $\delta_t \sim 0.5\ X_o$. With $\delta_i \sim 0.5\ X_o$ equation (5) gives the fracture toughness as

$$K_{IC} \sim \sqrt{0.5E\sigma_y X_o} \tag{6}$$

indicating that K_{IC} ought to increase with yield stress and also increase indefinitely as the inclusion volume fraction is decreased to zero. However the general tendencies, which are particularly emphasized in the Battelle group's discussions [1-3], are that toughness decreases with increasing yield stress, and increases towards a finite limit as the inclusion volume fraction decreases to zero. Consequently serious difficulties arise if Green and Knott's arguments, which in many respects extend those of the Battelle group, are applied to high strength materials. These workers clearly recognize the importance of flow concentration and glide band decohesion in such materials, arguing that they truncate the growth of voids and their linking with the crack tip. Such truncation becomes easier when a material has a high yield stress, since this is generally accompanied by a low work-hardening rate when flow concentration and glide band decohesion are relatively easy. However, with non workhardening materials, linkage between a void and the crack tip is predicted to occur when the logarithmic spirals first "envelop" the void (i.e. $\delta_i \sim 0.5\ X_o$) and increasing the yield stress beyond some limit ought not to affect δ_i; thus the toughness should increase whereas numerous experimental results [1-3] clearly show that this is not the case.

The dilemma created by extending the low strength material approach arises because it focusses attention primarily on the role of voids and secondly on the effect of flow localization and glide band decohesion. When considering high strength materials, however, it is more logical to concentrate primarily on flow localization and glide band decohesion and assess the behaviour of "clean" material, i.e. material that does not contain void-forming inclusions, followed by a discussion of the modifying effect of voids. Indeed the ideal procedure is to develop a sufficiently general approach that the special behaviours of both high and low strength materials evolve as a natural consequence, with their respective emphases on glide band decohesion and void-forming inclusions; this would enable the behaviours of the two types of material to be linked, the unique behaviour of each arising quite logically as special cases. The development of a general, albeit idealised, approach is the objective of the next section.

IV. OUTLINE OF A GENERAL MODEL REPRESENTING THE EFFECTS OF FLOW CONCENTRATION, GLIDE BAND DECOHESION, AND INCLUSION-TYPE VOIDS ON THE INITIATION OF CRACK EXTENSION.

Against the background of the preceding section's discussion, and with the behaviour of high strength materials particularly in mind, attention will initially be concentrated on the behaviour of "clean" material, following which the effect of inclusion-type voids will be discussed. An appropriate simple model is illustrated in Figure 7, infinitesimally thin plastic zones emanating from a crack tip in a symmetric manner. There is assumed to be no plastic relaxation immediately ahead of the crack tip, and consequently no process zone exists if the terminology of the preceding section is strictly applied. If the process zone definition is widened to encompass the region in the immediate vicinity of the crack tip where fracture processes are operative, the model does not discriminate between the process and plastic zones. A major problem is to describe the flow behaviour within the concentrated flow zones, but since there are insufficient experimental results available to enable an accurate quantitative picture to be formed, it is only possible to speculate. Thus the shear resistance-relative displacement curve is likely to consist of four separate regions (Figure 8):

(a) A hardening region corresponding to the initial movement of dislocations, these passing through the hardening particles with two-phase materials.

(b) A softening region associated with, for example, the degradation of particle-matrix interfaces; such a softening process is conceivably a necessary condition for flow concentration [7].

(c) A further hardening region corresponding to a substantial increase in the dislocation density, which is generally recognized to be a characteristic feature of concentrated flow bands.

(d) A softening region associated with the operation of glide band decohesion processes within a concentrated flow-zone.

For an initial discussion it is justifiable to simplify the material behaviour by using idealised shear resistance-relative displacement laws, such as the variations shown in Figure 9. For both variations, σ_y represents the material's yield stress while w_f is a measure of the critical relative displacement required for fracture within a concentrated flow band. The Figure 9(a) variation is obviously more amenable to mathematical analysis; however, with either variation, the fracture toughness K_{IC} is of the form

$$K_{IC} = \alpha\sqrt{E\sigma_y w_f} \tag{7}$$

where α is a constant of order unity. For such a relation to be applicable, and indeed for the operation of glide band decohesion processes, flow must be sufficiently concentrated; thus for a low K_{IC} value, flow must concentrate sufficiently and additionally fracture must occur readily within a concentrated flow band (i.e. w_f must be low). Conversely, for a high K_{IC} value, either flow concentration must be difficult, or if flow does concentrate, the operation of glide band decohesion processes must be difficult. With regard to relation (7), the significance of σ_y is easily understood in that it represents the yield stress, but w_f is far more difficult to interpret since it has a composite character, reflecting both the extent to which flow concentrates and also the ease with which fracture occurs within a concentrated flow zone. w_f and thereby the fracture toughness can be increased either by making flow homogeneous, or accepting that flow does concentrate, by making the operation of glide band decohesion processes difficult. There is little specific experimental evidence to confirm this behaviour pattern for "clean" materials, and it is therefore necessary to look for general evidence.

In this respect there is reasonably good experimental evidence regarding the correlation between w_f and flow homogeneity. For example Zinkham et al [18] and Ronald and Voss [19] show that ageing high strength aluminium alloys to peak hardness, thereby causing particle cutting and pronounced flow concentration, produces a marked reduction in K_{IC}. Moreover Knott and his collaborators have clearly shown that premature failure is encouraged by a reduction in the macroscopic work-hardening rate of a material, which allows flow to concentrate more readily; with a low alloy steel HY 80 [20] significant reductions in δ_i were observed as the amount of pre-strain increased, and $\delta_i/X_o = 0.3$ for a 20% pre-strain. These reductions were accompanied by a transition in the crack extension mode from link-up between the crack tip and an inclusion-type void immediately ahead of the tip, to decohesion along planes inclined to the original crack plane leading to a zig-zag type fracture.

There is also experimental evidence which shows that w_f increases when glide band decohesion processes become more difficult. Low et al [21] by appropriate heat-treatment procedures obtained a substantial increase in K_{IC} with a quenched and tempered high strength steel, without altering the flow characteristics, and this they interpreted as being due to a reduction in the size of the dispersion-hardening second phase particles, which delayed the onset of glide band decohesion. Again, Chipperfield and Knott [22] have shown that varying the heat-treatment procedures for a high strength steel HY 130 can produce marked changes in δ_i, without changing the fracture mode from the zig-zag type; since it is presumed that the maximum degree of flow concentration is attained or otherwise zig-zag fracture would not occur, this result is probably best explained in terms of the ease with which glide band decohesion processes operate.

In summary, therefore, with regard to the initiation of fracture at a crack tip in "clean" high strength materials, it is doubtful whether it is possible to be more specific about w_f, beyond stating that it represents in a composite manner, both the degree to which flow concentrates and also the ease with which glide band decohesion proceeds. The preceding discussion and the supporting experimental evidence suggests that high K_{IC} values can be attained (a) by making flow more homogeneous, and (b) if flow remains inhomogeneous, by making glide band decohesion more difficult. The former is the most direct and probably the more promising approach, but if it is unsuccessful, then prevention of easy glide band decohesion should be the aim. It is also worth mentioning, in the context of possible alternative simple tests correlating with crack opening displacement or fracture toughness tests, that the simple tension test and particularly the axisymmetric test, are far removed from the plane strain deformation state, characteristic of a crack tip, and which is so conducive to flow concentration. Indeed, the crack tip situation is also different to Clausing's plane strain tension test [23], in that growth of a deformation zone at a crack tip is restrained by elastic material ahead of the zone tip. These differences emphasize that the crack tip situation is special, and that it is difficult to apply results for supposedly more simple situations to assist the understanding of the crack tip situation.

So far, this section has concentrated on developing a framework understanding of the fracture toughness of high strength materials that are susceptible to crack extension by the glide band decohesion mode, the discussion being restricted to the case where inclusion-type voids play no role in the crack extension process. This discussion will now be extended to the case where voids are present, and might therefore play a significant role in the fracture process; in this respect inclusion-type voids are viewed as perturbing the behaviour of "clean" material. At the outset, it is worth noting that with the shear resistance-relative displacement law of Figure 9(b), the concentrated flow zone length at the initiation of crack extension is $R \sim S_m w_f/\sigma_y$. With high strength materials $S_m/\sigma_y \sim 10^2$, and even when the toughness is very low, $w_f \equiv \delta_i$ is markedly in excess of 1 μm. Thus R >> 10^2 μm and unless the inclusion content is exceptionally low, there will always be an opportunity for some interaction between inclusion-type voids and the plastic zone at a crack tip. It is therefore of vital importance to ascertain the strength of this interaction, and to consider the extent to which inclusion-type voids affect the crack extension process.

Assuming crack extension proceeds by the glide band decohesion mode, the simplest possible model describing the effect of a void distribution ignores orientation differences between the starting crack and concentrated

flow zones (Figure 10). An infinite solid contains a crack of length 2a and is subject to an applied shear stress σ, while ahead and coplanar with the crack tip, there is a series of cracks of length 2h, with their centres a distance s apart. The series of cracks represent inclusion-type voids formed within a concentrated flow zone at the crack tip, and a simple averaging procedure shows that the material in the flow zone behaves as if it has a yield stress $\sigma_y[1-(2h/s)]$ where σ_y is the yield stress of the material, which is assumed to be non work-hardening. The considerations of Section II therefore give the opening displacement at the crack tip as being inversely proportional to $\sigma_y[1-(2h/s)]$, and in order for voids to produce significant effects at a crack tip, the void density must not be unduly low. This very simple model and its associated approach have obvious inadequacies, the major one being that the voids do not enlarge as a consequence of plastic deformation within a flow band. A model which accounts for void growth is illustrated in Figure 11. The resistances to flow within the various intervals of the flow zone are as indicated, σ_y being the material's yield stress; those regions with zero resistance stress represent growing holes. The holes are assumed to grow from zero size, the initiation sites being a distance s apart, where s is representative of the inclusion spacing, and the size of any given hole is assumed to be proportional to the displacement at its centre. If the degradation of the material ahead of the crack tip is averaged, the force law for the flow zone becomes

$$\text{Resistance} = \sigma_y \left[1 - \frac{kw}{s} \right] \tag{8}$$

where w is the relative displacement in the flow zone and k is a constant; this relation (Figure 12) is valid for $w < w_f$ the critical value of w at which glide band decohesion occurs at the crack tip. Equations (1)-(3) of Section II give the relation between the applied stress and the displacement w_t at the crack tip as

$$\frac{\pi(1-\nu)\sigma^2 a}{2S_m} = \int_0^{w_t} \sigma_y \left[1 - \frac{kw}{s}\right] dw = \sigma_y w_t \left[1 - \frac{kw_t}{2s}\right] \tag{9}$$

a result shown in Figure 13. Giving k the reasonable value of unity, w_f will be small and less than s/k for high strength materials, where flow concentration and glide band decohesion are both easy. Glide band decohesion is therefore the controlling event in crack extension, which occurs when $w_t = w_f$, the fracture toughness being

$$K_{IC} \sim \left[E\sigma_y w_f \left(1 - \frac{w_f}{2s}\right)\right]^{1/2} \tag{10}$$

This means that

$$\frac{K_{IC}(\text{CLEAN}) - K_{IC}}{K_{IC}(\text{CLEAN})} = \frac{w_f}{4s} \tag{11}$$

when $w_f/4s$ is small, where K_{IC}(CLEAN) is the value of K_{IC} obtained from equation (10) by putting S = ∞. Marked effects of void-forming inclusions on fracture toughness are therefore unlikely, and they become progressively less likely the higher the yield strength, presuming that this correlates with a lower w_f value. This prediction is high-lighted in the Battelle group's considerations [1-3], where the importance of clean matrix material behaviour is emphasized for high strength materials. There is good experimental evidence to support this predicted behaviour pattern. The data of Birkle et al [24], on the effect of sulphide inclusion spacing on the crack opening displacement for a Ni-Cr-Mo steel, show that increases in toughness due to inclusion removal become increasingly less pronounced as the strength level is raised by lowering the tempering temperature. Moreover experience with maraging steels has shown that large K_{IC} increases are obtained by removal of Ti(C,N) particles from grain boundaries when the strength level is less than 1750 M/Pa [25], but there is no change in K_{IC} due to purification when the strength is 2070 M/Pa [26].

If glide band decohesion is not so easy, w_f can exceed s/k and crack extension will be associated with a toughness value

$$K_{IC} \sim \left[\frac{E_y s}{2}\right]^{1/2} \qquad (12)$$

with k = 1; in this case crack extension is controlled by a type of instability, which is manifested by the absence of w_f in the fracture toughness expression [12]. However, in Section III, where the primary concern was crack extension by void growth immediately ahead of a crack tip followed by the linkage of a void with the tip, this failure mode was shown to be operative when $w_t \sim 0.5s$, and is therefore likely to precede the instability associated with relation (12). Crack extension should therefore occur either by glide band decohesion on an inclined flow plane (high strength materials), or by the linking-up of the crack tip with voids lying ahead of the tip (low strength materials), but not by a void growth instability mode that operates within an inclined flow bend. This discussion therefore focusses on the distinction between high strength materials where flow concentration and glide band decohesion are the dominant features of the crack extension process, and low strength materials where void growth is the major feature. We may conclude with the former, that flow concentration and glide band decohesion both play a role, and to influence K_{IC} the ability to concentrate flow and the ability of the material to fracture within a concentrated flow band must be modified. With low strength materials, however, we conclude that the controlling event is the growth of inclusion-type voids and their coalescence with the crack tip, and to influence K_{IC} the main objective is to modify the inclusion content and the work-hardening rate so as to affect the coalescence processes, but attempts to change K_{IC} by modifying the glide band decohesion mechanism are unlikely to prove rewarding.

The focussing of attention on these conclusions is a key feature of the present paper, and because of their importance it is worth emphasizing that the paper's general aim has been to strive for broad generalizations while recognizing that several uncertainties remain. For example, with some materials, fracture of clean material within a concentrated flow zone may be a fairly gradual process, when the Figure 9(a) force law will be more appropriate than that of Figure 9(b). However, an analysis based on the Figure 9(a) force law suggests that the conclusions are essentially unchanged; nevertheless even this force law is idealized, as indicated earlier in this Section. Because of this and other uncertainties associated with the model's simplicity, coupled with the conclusions' importance, it is clearly necessary to explore the problem in greater detail using the existing model (Figure 11) and discussions as a basis. In this context,

a particular inadequacy of the present model is the averaging procedure adopted in the simple analysis, and it is obviously more satisfactory to approach the problem directly. A purely analytical approach appears out of the question, but numerical procedures, similar to those used [7] to discuss the causes of flow concentration, should enable this difficulty to be readily by-passed. Pending the obtaining of appropriate numerical results, there is a good case for additional use of more simplified models, so as to provide further support for the views expressed in this Section. Furthermore, it would also be useful to consider, along the lines of Rice and Johnson [14], the interaction between a blunting crack tip and a single growing void that is situated off the crack plane; such an investigation would complement the proposed numerical work on void distributions, and be extremely valuable in securing the position that has already been outlined.

V. DISCUSSION

Existing theoretical and experimental work has been critically appraised with the object of providing a basis for the development of an understanding of the fracture toughness of materials that are susceptible to flow localization. Such materials generally have a high yield strength since this is usually accompanied by a low macroscopic work-hardening rate which often correlates with flow localization. The paper has therefore been primarily concerned with high strength materials recognizing, of course, that it is difficult to define high and low strength in strict terms. Indeed, there is some merit in viewing the problem in mechanistic terms, with high strength referring to materials that display flow localization, and low strength referring to materials where flow is homogeneous, without specifying a discriminating strength level.

Most of the previous mechanistic work, both experimental and theoretical, has been mainly concerned with low strength materials where the major role in the crack extension process is played by inclusion-type voids which form and grow in the process zone ahead of a crack, flow localization and glide band decohesion truncating the void coalescence process, and therefore acting essentially as perturbating factors. As demonstrated in Section III, developing an understanding of high strength material behaviour from the basis of low strength material experience can lead to serious difficulties. Consequently, this paper has attempted to adopt a fresh approach in a sufficiently general manner that both high and low strength material behaviours are incorporated within a single pattern, albeit in outline form.

In the first instance a very simple model (Figure 7) formed the basis for a discussion of "clean" material behaviour, and this was then modified (see models in Figures 10 and 11) to allow for the possible effect of void-forming inclusions. Important conclusions have arisen from these considerations and, as indicated in the paper, these are in reasonable accord with the limited available experimental data for high strength materials. The main conclusions regarding fracture toughness are

(a) With high strength materials, void-forming inclusions play only a minor role, and they have an increasingly less effect as the flow strength increases, assuming that there is a correlation between the latter and a greater tendency for flow localization and glide band decohesion.

(b) To attain high toughness, the aim should be to prevent flow localization, but if this cannot be achieved, the aim should be to make glide band decohesion processes more difficult to operate. There is no advantage to be gained in striving for undue cleanliness with regard to void-forming inclusions assuming, of course, that these are too small to act as pre-existing cracks themselves; there is therefore a marked contrast with the behaviour of low strength materials, for which the primary means of attaining a high toughness value is to reduce the void-forming inclusion content.

Regarding (a), the minor role of void-forming inclusions refers to the extension of a pre-existing crack along a plane inclined to the initial crack. However, as indicated in the preceding section, a zig-zag type of fracture can arise [20] from this crack extension mode, and inclusions although not affecting the toughness, do determine the fracture surface profile in that inclusion spacing determines the step lengths of the zig-zag pattern.

With respect to (b), some comments have already been made concerning factors affecting flow localization and glide band decohesion processes. In addition to these earlier comments, it should be noted that aluminium alloys contain particles that are intermediate in size between the particles responsible for strength and the inclusions that cause the formation of major voids in the sense of this paper's discussions. Metallographic observations [1] suggest that these intermediate particles rupture during linking-up of the larger voids, after deformation has become localized between these voids. Furthermore it is argued [1] that the inherent cracking resistance of these particles is retarding the formation of localized flow bands; if this interpretation is correct, judicious use of such particles could provide a fairly general method for increasing the toughness of high strength materials. It is also desirable to understand the effect of grain boundaries on flow localization, but since the extensive literature reviews [1-3] make no special mention of grain boundaries in this context, it is likely that their effect is small, but obviously it is important to confirm this view. With regard to making glide band decohesion processes more difficult to operate when flow localization cannot be prevented, the long-term aim should be to base practical approaches on an understanding of decohesion mechanisms; since there are various possible sites for fracture nucleation, for example the hardening particles, intermediate particles and grain boundaries, there is considerable scope for further research on this particular facet of the problem.

Since the conclusions (a) and (b) are extremely important both from an alloy development view-point, and also because they could form the basis for predicting high strength material behaviour in complex operational conditions, it is clearly desirable to secure the arguments that lead to these conclusions. Thus, as indicated in Section IV, more detailed consideration should be given to the interaction of inclusion-type voids with a crack tip by: (a) employing numerical procedures to study the effect of void distributions, thereby avoiding the uncertainties associated with the averaging procedure used to analyze the model of Figure 8, and (b) considering the effect of a single void that is situated off the crack plane. It is also important to confirm the generality of the flow concentration mode, i.e. flow along inclined planes, that has been implicitly assumed for clean high strength material behaviour throughout the preceding sections. In this context, Rice and Johnson [14] have noted that the theoretical solution for the stress relaxation at a crack tip in non-work hardening material is not unique, but that it can proceed either by flow along inclined planes or by the blunting of the tip into a smooth arc as discussed extensively in Section III. If, in practice, flow concentration

proceeds along the logarithmic spirals in preference to inclined planes, this added complication would need to be incorporated within the framework behaviour pattern developed in this paper.

Regarding the implications of the current fracture toughness considerations, the Introduction has already briefly mentioned their importance with regard to reverse loading situations. A considerable amount of experimental data has shown that Stage II growth of a pre-existing crack by fatigue processes, when expressed in terms of the cyclic stress intensity factor ΔK, follows the schematic behaviour pattern illustrated in Figure 14, with a power law relationship

$$\frac{da}{dN} = A(\Delta K)^n \tag{13}$$

being valid for most alloys over the intermediate range, where crack extension is due to the exposure of fresh surface by plastic relaxation mechanisms. The lower the fracture toughness K_{IC}, the more limited is the intermediate range, and this causes accelerated growth at an earlier stage; it can also give effective values of n in excess of the values 2-4 which are usually associated with the intermediate range [27,28]. Indeed, fatigue crack extension by mechanisms other than plastic relaxation has been observed [29], and this leads to higher n values; moreover, an extensive survey [27] has shown that the effective value of n increases as the fracture toughness decreases, the reason being that other failure mechanisms, for example glide band decohesion, provide a larger contribution to the amount of crack extension. An obvious means of developing a material with good Stage II fatigue crack growth resistance is therefore to strive for a high fracture toughness value, but in this context and against the background of this paper's considerations of high strength materials, there is little to be gained in striving for undue cleanliness with respect to void-forming inclusions. However, it must be emphasized that Stage II growth is only one part, albeit an important part, of the overall fatigue behaviour pattern. Fatigue crack initiation and growth by the Stage I mechanisms are both clearly favoured in materials that display flow localization, and although void-forming inclusions might not affect Stage II growth in high strength materials, they certainly play a role in fatigue crack initiation. Thus there is a general correlation between poor overall fatigue characteristics, low fracture toughness and the ease with which flow concentrates; there is scope for detailed consideration of specific aspects of this correlation, based on modern models of the fatigue crack growth process, such as for example the model recently outlined by Tomkins [30].

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REFERENCES

1. HAHN, G. T. and ROSENFIELD, A. R., Metallurgical Transactions, 6, 1975, 653.
2. HAHN, G. T., KANNINEN, M. F. and ROSENFIELD, A. R., Annual Review of Materials Science, 2, 1972, 381.
3. ROSENFIELD, A. R. and HAHN, G. T., Colloquie sur la Rupture des Materiaux, Grenoble, France, Societe Francaise de Metallurgie, Section Regionale du Sud-Est, 1972.
4. LUTJERING, G. and WEISSMANN, S., Acta Met., 18, 1970, 785.
5. BLACKBURN, M. J. and WILLIAMS, J. C., Trans. A.S.M., 62, 1969, 398.
6. CALABRESE, C. and LAIRD, C., Mats. Sci. Eng., 13, 1974, 141.
7. COOK, T. S., RAU, C. A. and SMITH, E., ASME Conference on Micro-mechanical Modelling of Flow and Fracture, Troy, N.Y., U.S.A., Paper No. 75-Mat-7, June 1975.
8. COTTRELL, A. H., Proc. Roy. Soc. A282, 1964, 2.
9. GRIFFITH, A.A., Phil. Trans. Roy. Soc. A221, 1921, 1963.
10. OROWAN, E., Rep. Prog. Phys., 12, 1948-9, 214.
11. IRWIN, G. R., Trans. Amer. Soc. Met., 40, 1948, 147.
12. BILBY, B. A., COTTRELL, A. H. and SWINDEN, K. H., Proc. Roy. Soc. A272, 1963, 304.
13. GREEN, G. and KNOTT, J. R., ASME Conference on Micromechanical Modelling of Flow and Fracture, Troy, N.Y., U.S.A., Paper No. 75-Mat-10, June 1975.
14. RICE, J. R. and JOHNSON, M. A., Inelastic Behaviour of Solids, eds. M.F. Kanninen et al., McGraw-Hill, New York, 1970, p.641.
15. RICE, J. R., Paper presented at B.S.C. Conference on "Mechanics and Mechanisms of Crack Growth", Churchill College, Cambridge, 1973.
16. HARRIS, D. and ELLIOTT, D., Paper 12, The Mechanics and Physics of Fracture, Institute of Physics and Metals Society Conference, Churchill College, Cambridge, January 1975.
17. BILBY, B. A. and SWINDEN, K. H., Proc. Roy. Soc. A285, 1965, 22.
18. ZINKHAM, R. E., DEDRICK, J. H. and JACKSON, J. H., Proceedings of Fifth International Leichtmetallagung Leoben Austria, Aluminium-Verlag GMBH - Dusseldorf, 1968.
19. RONALD, T. and VOSS, D. P., unpublished research quoted in reference 1.
20. CLAYTON, J. Q. and KNOTT, J. F., unpublished work referred to in reference 13.
21. LOW, J. R., Prospects of Fracture Mechanics, Proceedings of International Conference held at Delft University, Netherlands, Noordhoff, 1974, 35.
22. CHIPPERFIELD, C. G. and KNOTT, J. F., Metals Technology, 1975, 45.
23. CLAUSING, D.P., Int. J. Frac. Mech., 6, 1970, 71.
24. BIRKLE, A. J., WEI, R. P. and PELLESIER, G. E., Trans. ASM, 60, 1967, 275.
25. SPAEDER, G. J., Met. Trans., 1, 1970, 2011.
26. CARTER, C. S., Air Force Materials Laboratory TR-70-139, 1970.
27. KNOTT, J. F., Fundamentals of Fracture Mechanics, Butterworths, London, 1973.
28. TOMKINS, B., Paper presented at B.S.C. Conference on "Mechanics and Mechanisms of Crack Growth", Churchill College, Cambridge, 1973.
29. RITCHIE, R. O. and KNOTT, J. F., Mats. Sci. Eng., 14, 1974, 7.
30. TOMKINS, B., ASME Conference on Micromechanical Modelling of Flow and Fracture, Troy, N.Y., U.S.A., June 1975.

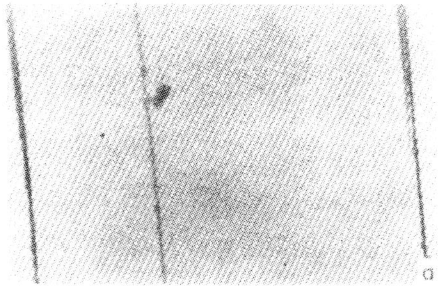

Figure 1 Flow concentration in Ti-16.8%Al alloy, with the slip distribution revealed by optical microscopy [4]. (x 400)

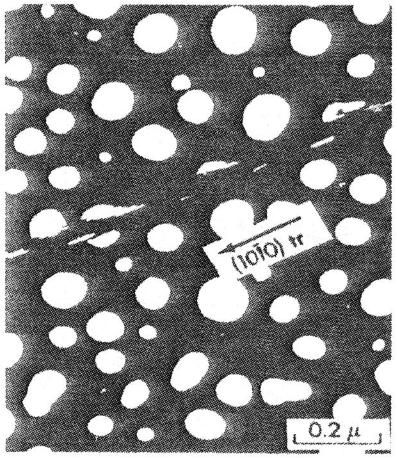

Figure 2 Electron micrograph illustrating the cutting of ordered Ti_3Al precipitates in Ti [5].

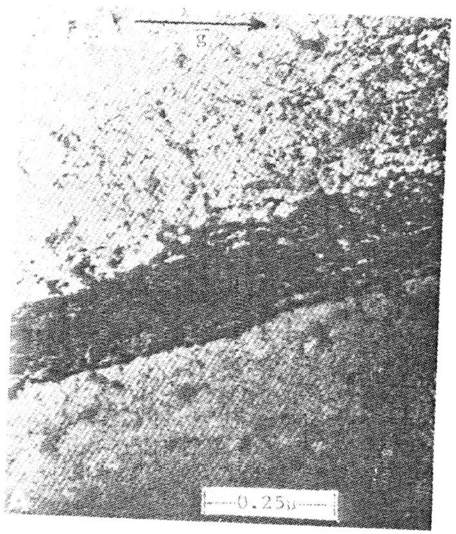

Figure 3 Transmission electron micrograph showing an intense deformation band in age-hardened Al-4%Cu cycled at ±1% plastic strain.

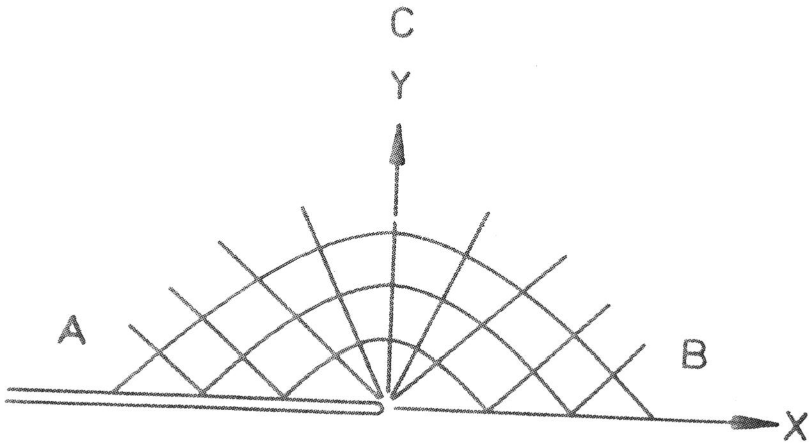

Figure 4 Slip line field for a sharp crack tip deforming in a plane strain mode; small geometry changes are assumed.

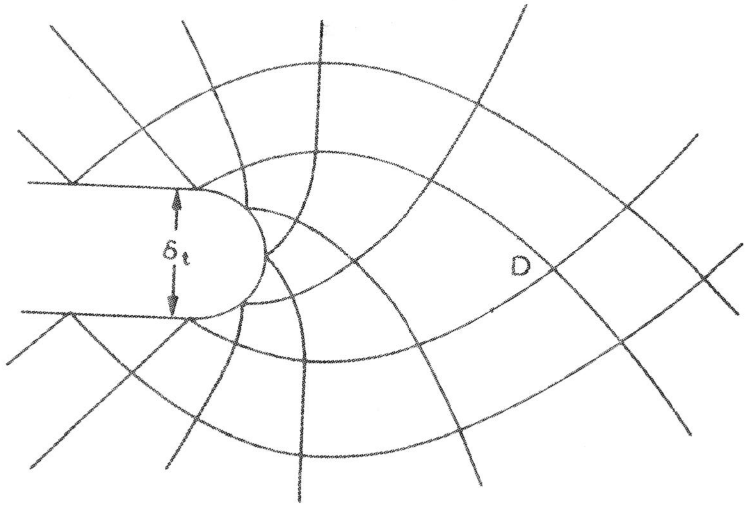

Figure 5 Slip line field for a blunted crack tip deforming in a plane strain mode; large geometry changes are taken into account.

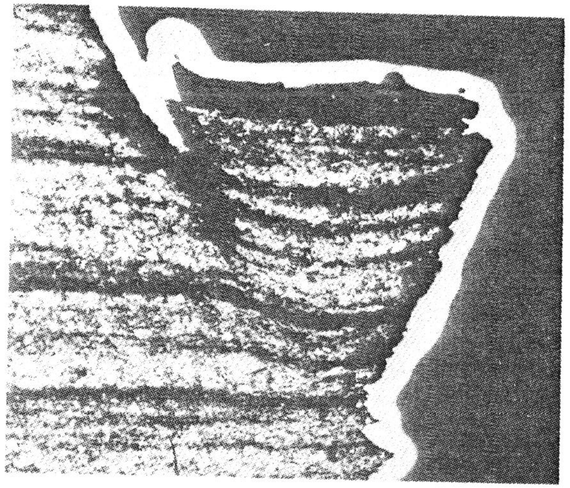

Figure 6 Section through a crack tip in a pre-strained, and therefore a low work-hardening, steel (HY 80); the deformation follows the logarithmic spiral of shear lines and is directed towards voids at the spiral focus. (x 100).

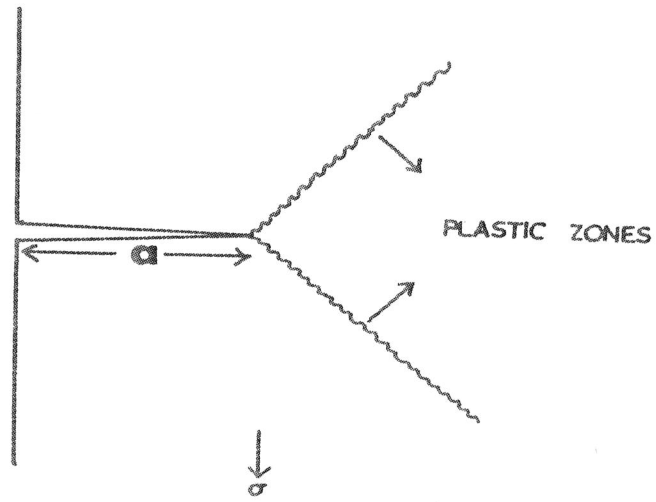

Figure 7 Model used to represent plastic relaxation and crack extension in "clean" material.

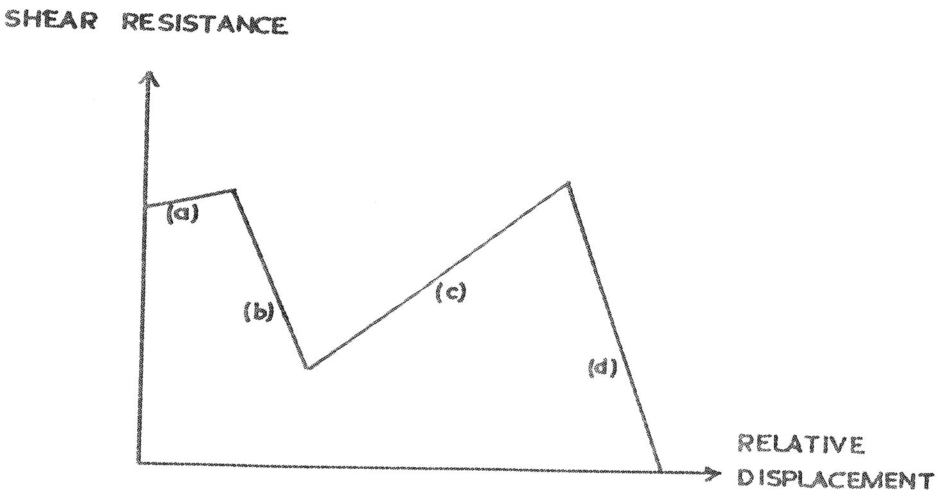

Figure 8 Schematic flow behaviour of material within a concentrated flow zone; see text for description of various stages.

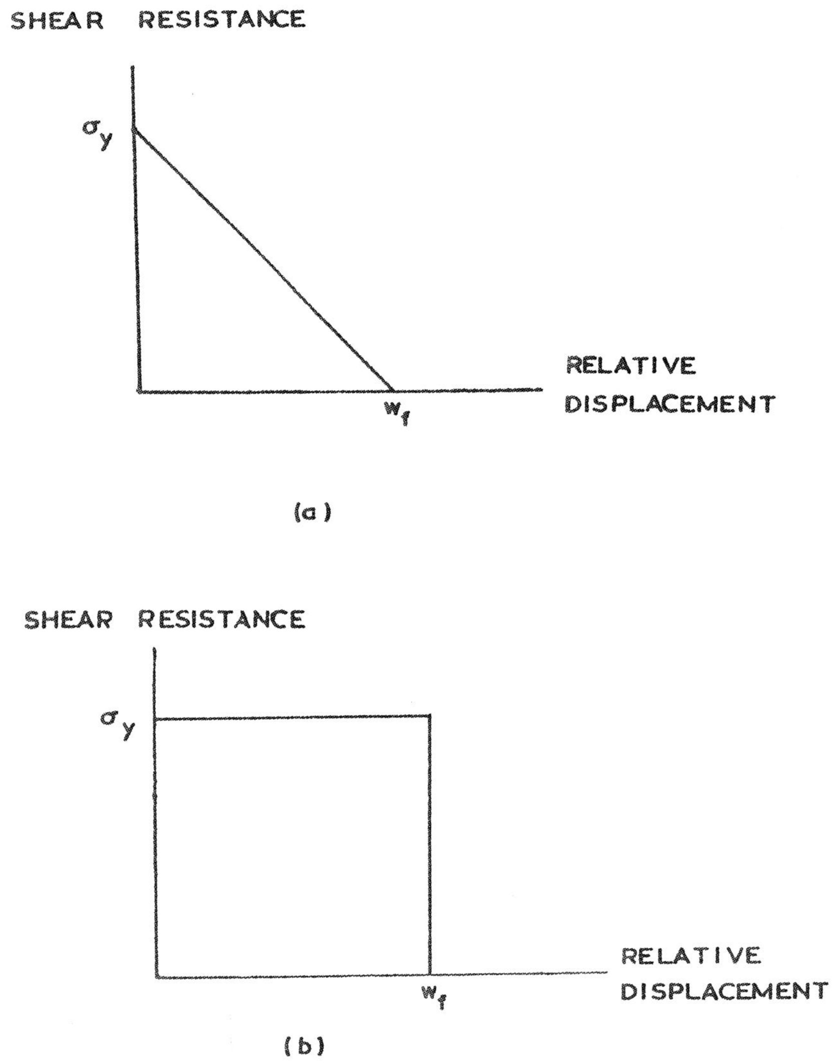

Figure 9 Simplified shear resistance-relative displacement curves for material within a concentrated flow zone.

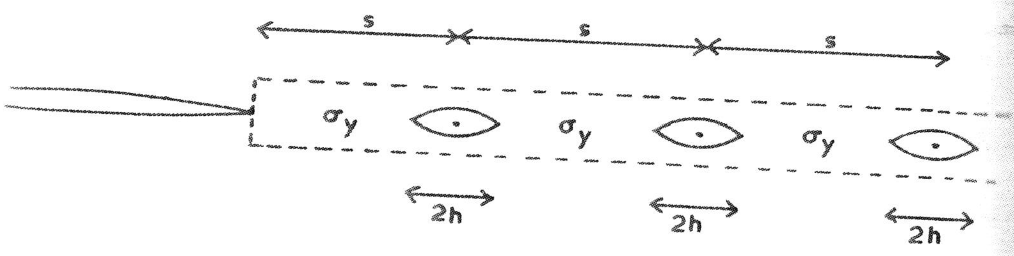

Figure 10   Very simple model describing the effect of void distribution on crack extension when this proceeds by the glide band decohesion mode.

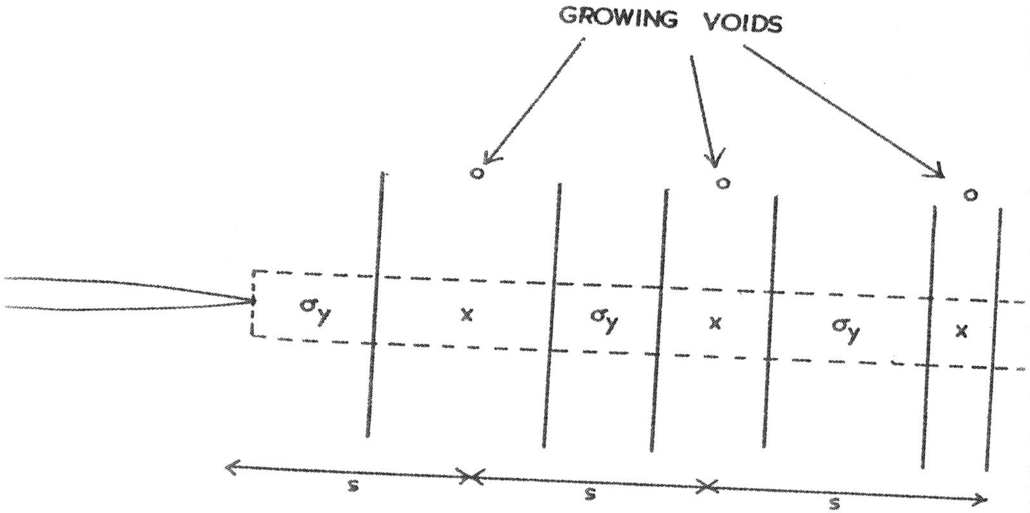

Figure 11   Model describing the effect of void distribution on crack extension, with allowance being made for void growth within a concentrated flow zone.

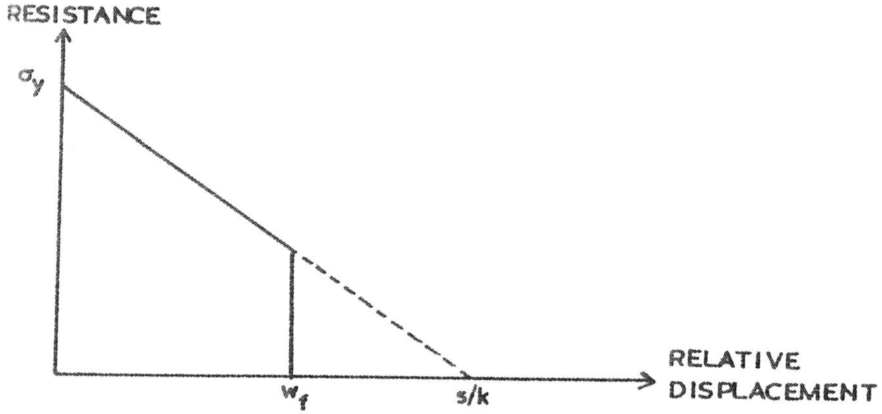

Figure 12   The concentrated flow zone force law as obtained by the averaging procedure.

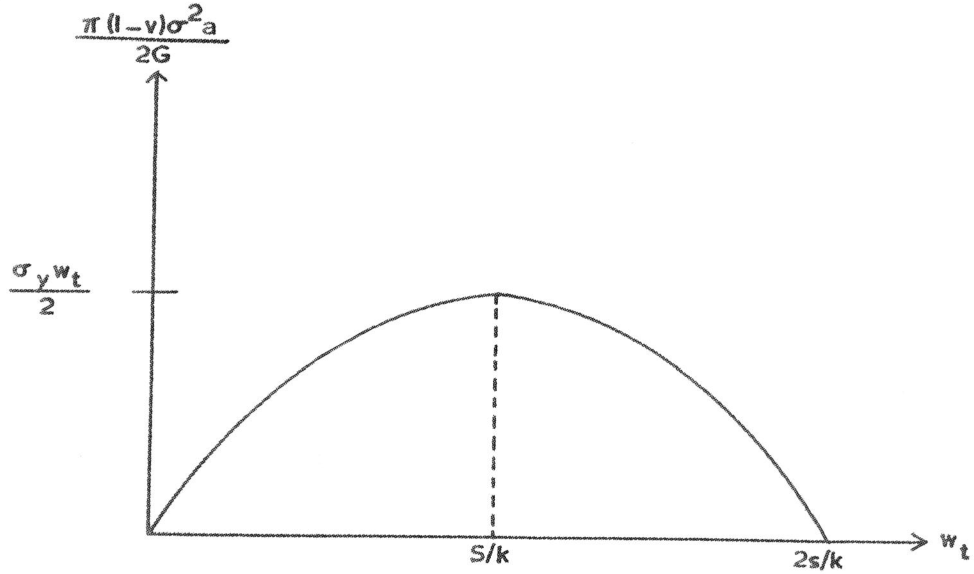

Figure 13   The relation between the applied stress σ and the relative displacement w_t at the crack tip.

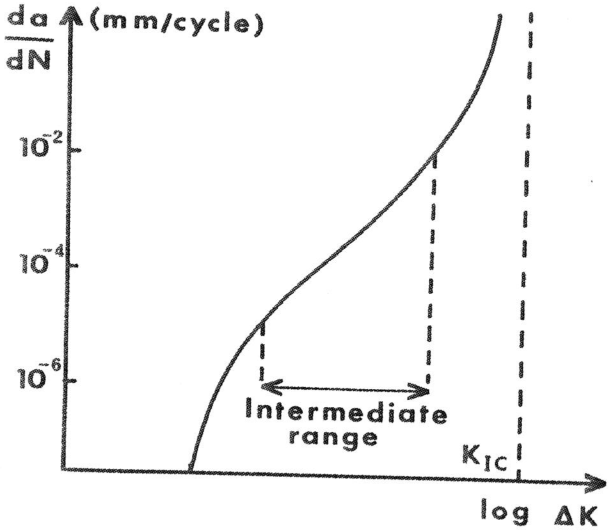

Figure 14 The variation of Stage II fatigue crack growth rate (da/dN) with the alternating stress intensity (ΔK); K_{IC} is the material's fracture toughness.