MICROMECHANISMS OF FRACTURE - THE ROLE OF MICROSTRUCTURE

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The paper treats the main features of brittle and ductile fracture in metallic alloys, placing particular emphasis on the need to define a length parameter to relate local fracture criteria, such as stress or strain, to global fracture toughness values which can be used in engineering design. For cleavage fracture, it is concluded that the *length* derives from the statistical distribution of brittle ceramic particles, such as carbides, oxides or silicates. For fibrous fractures, the length is usually related to the inclusion spacing, although it may derive directly from the definition of the amount of crack extension associated with initiation. Attention is paid to the production of cleavage fracture ahead of a growing fibrous crack, and it is argued that here it may be necessary to invoke a third "critical distance".

INTRODUCTION

A quantitative understanding of the micro-mechanisms of fracture in the region of the tip of a crack or other sharp stress-concentrator is of value in two respects. First, it enables the materials engineer to explain why a large test piece or engineering component, containing a crack-like defect, breaks at *precisely* the stress that it does, by specifying the *critical* values of crack-tip characterising parameters associated with crack extension. These may be the fracture toughness (K_{Ic}), the J-integral, (J_{Ic}) or the crack-tip opening displacement, CTOD (δ_c). Secondly, it enables the materials scientist to identify those features of the microstructure that limit the resistance of the material to crack extension and, by suitable thermo-mechanical processing, to attempt to produce material with a higher resistance.

The role of microstructure in this respect is to provide a physical basis for the appropriate length dimension, bearing in mind the nature of the local fracture process, the microstructural features associated with this and the dominance of stress or strain with respect to the critical local events. To provide self-consistency, it is necessary to combine a knowledge of crack-tip stress-and strain-fields with detailed metallographic and fractographic observations of the events leading to crack extension. Once a self-consistent model has been developed, the ideal procedure is to test it by modifying the microstructure to alter the distribution of critical components and to predict and observe the effects of such

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changes on behaviour. This is not easy, because changes in one feature of the distribution may produce other consequential changes which affect critical local values of stress or strain. The critical point, however, is that the application of any LOCAL APPROACH model to the characterisation of fracture toughness requires identification of a LENGTH parameter.

Given this requirement it is instructive to pay detailed attention to fracture processes in low-carbon structural steels. Their microstructures are composed typically of ferrite/carbide distributions of different morphologies (ferritic, bainitic, tempered martensitic) and they contain non-metallic inclusions (silicates, sulphides, oxides) which are usually present in low volume fraction, although high volume fractions of manganese sulphides are to be found in free-machining steel. The intriguing feature of such steels is that they undergo a *transition* in behaviour, from tensile-stress-dominated brittle cleavage CRACKING at low temperatures to plastic-strain-dominated VOID COALESCENCE at higher temperatures (1). In wrought materials, the low temperature cleavage is mainly associated with carbide particles (or martensite-austenite-carbide, MAC, products in the heat-affected-zones, HAZs, of weldments), but the higher temperature void coalescence is mainly associated with the non-metallic inclusions. There does not therefore appear to be any reason to suppose that, in such structural steel, the length dimension appropriate to relate local fracture stress to a toughness parameter should be the same as that needed to relate critical local strain to a toughness parameter.

A rather different situation holds, however, for the low-carbon weld-metals used to join structural steels together. Here, there is clear evidence that non-metallic inclusions (weld-pool deoxidation products, such as silicates or oxides) can provide the sites for both low-temperature cleavage microcrack formation and higher-temperature microvoid initiation. Superficially, the same types of inclusions appear to be involved in both processes and it is tempting to speculate that the length dimension might then be the same for both types of micro-fracture process.

A related important question is why the fracture behaviour of these weld-metals changes from cleavage to fibrous as the test temperature is increased. In wrought materials, it is relatively easy to countenance a competition between two independent mechanisms: one, stress-dominated and involving carbides; the other strain-dominated and associated with inclusions. A general treatment of the ductile/brittle transition will be discussed later. The specification of "toughness" in the transition region can sometimes be critical with respect to engineering applications and it is of value to examine the contribution that a knowledge of micromechanisms of failure could make to this area. The particular problem with the weld metals is that the same particles appear to be related to both the cleavage cracking and the microvoid coalescence process, so that the competition between two independent mechanisms is less easy to sustain. Attention needs to be focused more sharply on the nature of the inclusions and the inclusion/matrix interface.

A final point that should be made before considering the micromechanisms in detail relates to the sequence of events followed when carrying out a fracture toughness test on a structural steel testpiece. The pre-crack is produced by subjecting a slotted testpiece to a fatigue loading sequence. If this is carried out at a stress-intensity range of, say, 20 MPam$^{1/2}$ in zero-tension loading and the material's yield strength is 400 MPa, it can be seen from equation 1 (with M = 2) that the value of CTOD before carrying out the fracture toughness test is approx. 2.5 μm, i.e. 10,000 lattice spacings. Even over a range 1,000 - 10,000 spacings, the pre-crack is far from atomically sharp and does not extend in a brittle, cracking manner *from* its tip. It may blunt to advance by ductile microvoid coalescence processes but brittle cracking can be produced *only* by the formation of new, virulent, atomically sharp microcracks in brittle ceramic particles ahead of the crack tip. This observation reinforces the need for micro-mechanistic understanding. A second point concerns the phenomenon of warm-prestressing, WPS, which is observed as an increase in subsequent low temperature fracture toughness if a specimen is first stressed to a high load above the transition temperature. If care is not taken with the (room-temperature) fatigue pre-cracking, it is possible to produce values of K_{IC} at low temperature which have been raised artificially by the WPS effect.

Cleavage Fracture In Structural Steel

The need to initiate cleavage microcracks by the action of mobile dislocation arrays has been established for over 30 years and early nucleation models were based on dislocation pile-ups and interactions. In 1965, McMahon and Cohen (2) demonstrated the importance of grain-boundary carbide films in providing sites for microcrack nuclei in uniaxial tensile specimens and this physical picture was incorporated in Smith's (3) model for the cleavage fracture process, which included a brittle crack nucleus, a dislocation pile-up and the applied tensile stress. The dominant term was the tensile stress; in common microstructures, the dislocation term contributes only some 10% of the driving force to assist crack propagation. Dislocations are essential for nucleation, but the subsequent propagation can be regarded to a fair degree of accuracy as a simple Griffith model, with the crack size equal to the carbide thickness. A somewhat surprising feature of Smith's model is that it appears not to predict an effect of grain size on behaviour but Curry and Knott (4) later showed that there was a direct relationship between grain diameter and grain-boundary carbide thickness for simply as-cooled microstructures, so that coarser grains were associated with thicker carbides and hence a lower propagation stress.

These concepts have also been employed to explain the fracture behaviour of blunt-notched bars tested at low temperatures (5) and (6). Here, it was possible to establish that cleavage fracture was produced, in a low-strain region just inside the plastic/elastic zone, when the maximum local tensile stress $\sigma_{11(max)}$ attained a critical value σ_F, which is not usually strongly dependent on temperature. The magnitude of $\sigma_{11(max)}$ is a multiple of yield stress, σ_Y, the multiplying factor deriving from the level of triaxial stress associated with a given size of plastic enclave at the notch root. At low temperatures, where σ_Y is large, the fracture criterion can be satisfied with a small plastic zone. At higher temperatures, σ_Y decreases and the plastic zone increases in size until general yielding precedes fracture. Eventually, σ_Y becomes so low that it is easier to nucleate and grow voids *from* the notch root rather than fracture from microcrack nuclei *below* the notch root.

The σ_F concept has been applied to grain boundary carbides, to spheroidal carbides, to a variety of tempered carbides and MAC products and, as will be discussed in more detail later, to inclusions in weld metals. The concept also rationalises effects of strain-rate and neutron irradiation on fracture behaviour (7).

The experimental values of σ_F imply values of the effective work of fracture, 2γ, in the range 9 - 14 Jm^{-2}, compared with the value for the elastic work of fracture, γ, (where γ is the surface energy) of 2 Jm^{-2}. It should be noted that the "γ_p" term *must* be greater than "γ" in the particle for fracture to be propagation controlled and that for pure zinc, containing no brittle particles, cleavage fracture occurs at a critical value of shear strain (high local stresses are produced at the ends of dislocation arrays) (8). It has been argued that the "γ" value for the elastic work of fracture is a lower limit, based on the pure thermodynamic argument of "before" and "after" states. If a local rearrangement of atoms at the tip of the microcrack is necessitated to allow the crack tip pair of atoms to separate sufficiently for fracture to occur, extra energy may be expended in creating an "activated state". Movements of a few lattice spacings on either side of the crack tip are sufficient to account for the appropriate amount of extra energy (9).

The situation envisaged at present is that the initiation/propagation sequence should be regarded as a dynamic sequence. Dislocations initiate a sharp micro-crack in a stressed brittle ceramic particle and this microcrack accelerates across the particle. Its terminal velocity at the far particle/matrix interface is dependent on the particle's work of fracture and the applied tensile stress. Assuming well-bonded interfaces, the system then attains a "GO/NO-GO" situation. *Either* the impacting velocity is such that cleavage of the ferrite planes can occur before there is time to operate dislocation sources *or* there is time to operate sources to blunt and arrest the microcrack. Dislocation pinning is important in this process and it has been shown that removal of pinning by dissolving intragranular carbon/nitrogen by solution treating at temperatures of *ca.*700°C and quenching from these temperatures gives an increase in σ_F (10). Subsequent aging surprisingly gives rise to a further increase: this might be attributed to a reduction in the efficacy of dislocation pile-ups as a result of a fine distribution of carbide/nitride precipitates, but the observation has not, as yet, been fully explained.

The conclusion from these studies is that low-temperature cleavage fracture ahead of a notch is characterised by the propagation of microcracks which have been initiated in brittle ceramic (carbide) particles. Plastic strain is necessary to initiate microcracks, but the critical stage is the propagation from carbide particle to ferrite matrix. This takes place at a critical value of maximum tensile stress, σ_F, which is often, at most, weakly dependent on temperature. In very-low-carbon auto-tempered martensite microstructures, there is evidence that the critical stage in propagation may be from "packet" to "packet" and in steel possessing fine-scale pearlitic microstructures, it appears that the plastic strain contributes significantly to the propagation criterion (11).

A particularly simple microstructure is that of spheroidal carbides in a fully-tempered, ferritic matrix. Here, there is a reasonably good relationship between the measured values of σ_F and the inverse square root of the radius corresponding to the 95th percentile of the carbide distribution, implying a γ_p value of 14 Jm^{-2} . There is relatively little scatter in the σ_F values because the test volume at high stress in a blunt-notched bar is sufficient to sample practically the full range of the carbide distribution (12).

The σ_F criterion for cleavage fracture in blunt-notched testpieces for a certain range of ferrite/carbide micro structures therefore has a plausible theoretical base, backed up by experimental results. We now examine its application to the characterisation of cleavage fracture in pre-cracked testpieces and to its relation to the K_{Ic} toughness values via a "critical length" or "critical distance" parameter.

General treatments of micro-scale fracture processes tend to emphasise the importance of the value of maximum tensile stress with respect to the propagation of microcracks, or that of maximum plastic strain (or deviatoric stress) with respect either to the initiation of microcracks or microvoids or to the growth and coalescence of voids by plastic flow processes. Diffusion-controlled growth is of importance at high temperatures and is conventionally dependent on tensile stress. There are, however, situations both at low temperature and high temperature in which *combinations* of plastic strain and maximum tensile stress must be considered. At low temperatures, a brittle microcrack cannot propagate (under tensile stress control) until it has been initiated (by plastic strain). At high temperatures, the features of diffusion-controlled growth may involve climb (a tensile-stress dominated process) of dislocations which are produced to accommodate plastic strain. A point of importance here is the effect of far-field applied tensile stress compared with that of local tensile stress associated with the dislocation arrays. It is however, possible to identify failure processes which are *dominated* by either tensile stress or plastic strain and these provide useful model systems to attempt to relate the micro-mechanical mechanisms to macroscopic toughness parameters and to features of the microstructure.

The first general point that emerges is that there is a basic difference in the DIMENSIONALITY of the parameters that would be chosen to treat micro-scale fracture: stress $[ML^{-1} T^{-2}]$ and strain (dimensionless); and those used to characterise the onset of crack extension in bodies containing crack-like defects. Consider the following set of identities, holding for small-scale yielding

$$K^2 = E'J = M\sigma_Y\delta \qquad \text{.......................... 1).}$$

where E' is the Young's modulus, E, in plane stress and $E/(1 - \upsilon^2)$ in plane strain, υ is the Posson's ratio, σ_Y is the yield stress and M is a constant, equal to unity in plane stress, and to a value close to 2 in plane strain. If K is extracted as a critical parameter, this has dimensions of stress (length) $^{0.5}$; i.e $[ML^{-0.5} T^{-2}]$; J (defined here *per unit thickness*) has dimensions of stress x length $[MT^{-2}]$, although in plates of finite thickness its dimension are MLT^{-2} the same as those of force (hence "crack extension force"; "energy release rate (or differential)"; the CTOD, δ, simply has dimensions of length, [L]. It becomes clear that any attempt to relate local stress and strain values to macroscopic toughness parameters must involve a suitable length dimension: for example, K has dimensions of stress $(\text{length})^{0.5}$; δ has dimensions strain x length.

Pre-Cracked Trestpieces and the "Critical Distance"

Values for "critical distance" for the cleavage fracture process were determined by Ritchie, Knott and Rice (RKR) in 1973 (13). Experiments were carried out on mild steel in a simple annealed condition, such that the micro-structure was composed of equiaxed ferrite grains with cementite films at the grain boundaries. Measurements were made of values of σ_F in blunt-notched testpieces and of K_{Ic} in pre-cracked testpieces over a range of low testing temperatures. Elastic/plastic finite element analysis was employed to calculate local stresses ahead of the pre-cracked tips in the K_{Ic} testpieces and it was demonstrated that the K_{Ic} values could be reconciled with σ_F values if the critical distance were of order two grain diameters (approx. 120 μm in the steel studied). The later work of Curry and Knott (14) demonstrated that there was no fixed relationship between critical distance and grain diameter in annealed steels, and it was suggested that a statistical approach needed to be adopted. This was developed for a microstructure of ferrite grains containing a distribution of spheroidal carbides. Similar statistical models have subsequently been developed by BEREMIN (15) and by Lin, Evans and Ritchie (16). The appropriate value of X_o, however, has to be obtained by inference, rather than by direct measurement.

More recently, studies have been made of cleavage fracture processes in low-carbon weld metals, which contain a distribution of brittle oxides or silicates, formed as deoxidation products in the weld-pool (17). Typically, these range in size up to 2-3 μm diameter, although occasional, exogenous inclusions up to 10 μm in size can be found if unfused silicates from the binder on the welding rod enter the melt. The important feature is that the site of cleavage fracture initiation can often be located in these weld metals, by tracing back river-lines to a focal point. It has been shown clearly that, in many cases, the cleavage fracture has been triggered by the fracture of a single inclusion, Fig. 1. Identification is made easier if the inclusion is located in coarse, grain-boundary ferrite and studies made on brittle cleavage during low-temperature fatigue suggest that the cleavage is associated not only with a large inclusion, but also with coarse ferrite (18). McRobie has made a detailed study of the cleavage fracture of weld metals in notched bars and has shown that the value of σ_F ($\sigma_{11\ (max)}$ *at* the site of initiation) varies in a linear fashion with inverse square root of particle diameter, implying a work of fracture of approx., 9 Jm^{-2} (19).

Some limited information is also available from low temperature fracture toughness tests made on McRobie's material. Single-site initiation has been detected ahead of a fatigue-crack tip in four specimens (T,U,V,W) of a two-pass submerged arc weld. The size of the initiating inclusion is then used to deduce an appropriate value for σ_F and the RKR model can then be applied to calculate a value for X_o which can be compared with the distance measured directly on the fractograph. The results are given in Table 1.

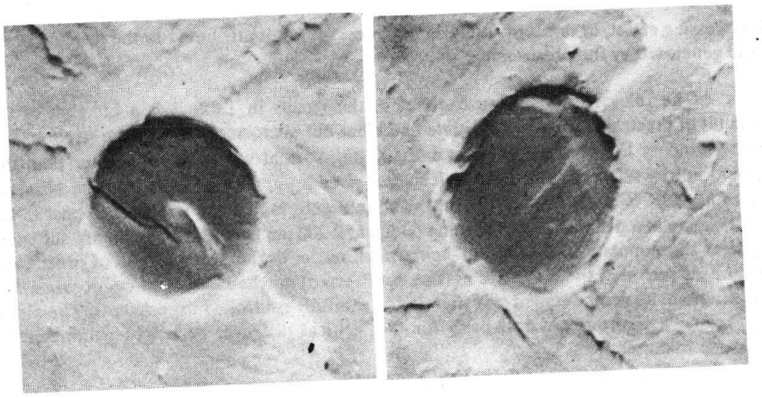


Fig. 1. a).Cracked silicate inclusion approx. 3μm in diameter which provided the microcrack nucleus for total cleavage failure.
b). Matching half of inclusion on the other side of the cleavage fracture. (Courtesy Dr. M.B.D. Ellis).

Table 1. Values of Critical Distance (after Dr. D.E. McRobie)

Specimen	Temperature, K	K_{Ic},MPam$^{1/2}$	X_o(obs),μm	X_o(RKR),μm
T	100	44	34	188
U	123	49	56	220
V	153	55	57	217
W	173	63	190	204

It can be seen that there are wide discrepancies between the observed values and the calculated values, except at the highest temperature, 173K, where agreement is close.

The reason for the discrepancies is not clear, but could relate to the possible effect of the (room-temperature) fatigue precracking procedure in terms of "warm prestressing" behaviour. For a maximum stress intensity of 20 MPam$^{0.5}$ in the fatigue precracking, and a flow stress of 400 MPa the maximum extent of the plane strain plastic zone, 0.16 $(K/\sigma_Y)^2$ would be 400 μm and the minimum extent 100 μm. A strong possibility is that the plastic deformation put in at room temperature initiates microcracks in brittle inclusions, but that these blunt out and are rendered inoperative at the low test temperatures. This effect is not relevant to the notched bars on which values of σ_F are measured, because these are not prestrained plastically at room temperature before testing at low temperature. There is evidence to support the effect from work carried out by Read to study warm prestressing in PWR weld-metals (20). Here, the fractography indicated that the initiating inclusions in warm prestressed material were smaller than the largest in the population and that voids could be observed around larger inclusions. It should be noted, however, that the warm

prestressing effect, in terms of fracture toughness values, is attributed to the residual stress field generated by the prestressing.

The relevance of any such effect to the results in Table 1 is a function of the amount of prestrain required to initiate and blunt out microcracks. The plane strain plastic zero boundary lies between 100 μm and 400 μm, so that values of X_o greater than this should not be affected. The RKR calculation gives X_o approx. 200 μm and the observed value at 173K is much smaller (57 μm). The inference is that the strains that exist in the plastic zone at a distance between some 60 μm and 200 μm from the crack tip are sufficient to alter the population of "virulent" microcrack nuclei in the process zone of precracked specimens such that the σ_F values in (non-prestressed) notched bars are inappropriate, but the phenomena need detailed investigation in carefully controlled experiments. The aim is to provide a rigorous physical basis for the "critical distance" or length parameter for cleavage fractures.

Fibrous Rupture and Length Parameters

Fibrous rupture is accomplished by separation processes bounded by two extreme mechanisms: *void coalescence* and *fast shear*. The advance of a crack by a pure void coalescence mechanism proceeds as shown in Fig. 2. Here, the configuration is that of a crack embedded in a material containing a uniform distribution of weakly bonded, non-metallic inclusions. A good model system is that of a free-cutting mild steel, from which specimens have been extracted transverse to the rolling direction. The inclusions here are MnS and the cross-sections approximate to circles, or ellipses of axial ratio close to unity. Application of tensile strain causes the crack tip to blunt and advance by movement of dislocations. The combination of plastic strain and hydrostatic tensile stress ahead of the tip initiates a void (at virtually zero strain) which expands, as remote tensile strain is applied, until the blunting tip coalesces with the expanding void. At this point, ductile fracture *initiation* is deemed to have occurred and the process is continued as the ductile fracture grows (Fig 3). Note that the increment in CTOD (or J) per unit of advance ($\Delta\delta/\Delta a$) or ($\Delta J/\Delta a$) is less than (although, often, proportional to) the value of CTOD (or J) at "initiation" (the first unit), because "precursor" strains have already been generated for "unit two" as a result of the strain distribution required for "unit one" (see Fig. 4).

Some work on ductile fracture has been carried out on slotted or notched bars, rather than on precracked specimens, and, to rationalise the values of δ_i obtained, use has been made of the concept of "gauge length". This is, however, a length *normal* to the line of the crack, and is, in concept, not relevant to a critical length parameter, $X_o = X_D$, which refers to the spacing of void-initiating inclusions in a direction which extends the line of the crack (even though the inclusions may, in reality, lie to one side or the other of this line). The concept of "gauge length" is, however, of importance if energy measurements made on blunt-notched testpieces, such as Charpy specimens, are correlated empirically with K_{Ic} values. Such correlations are often made use of to estimate effects of neutron irradiation on fracture toughness, via Charpy surveillance specimens, but the point has been made that any correlation established for initially heat-treated material (with high work-hardening rate) *cannot* hold for heavily irradiated material, because the localisation of strain associated with a low work-hardening rate alters the effective gauge length in the notched specimen (21).

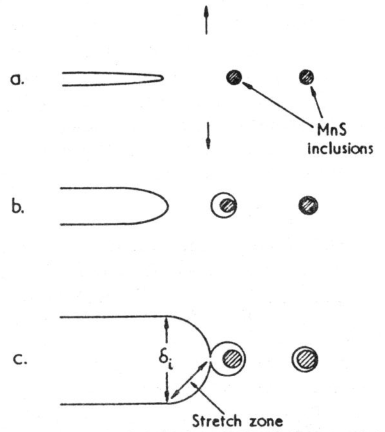


Fig.2 Schematic diagram of fibrous fracture initiation.

Fig.3 Crack tip region associated with fibrous crack growth in free cutting mild steel.

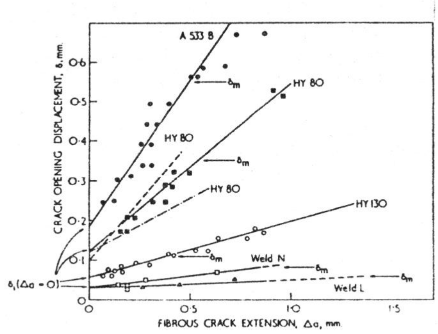


Fig.4 Dependence of CTOD on fibrous crack growth for a number of structural steels.

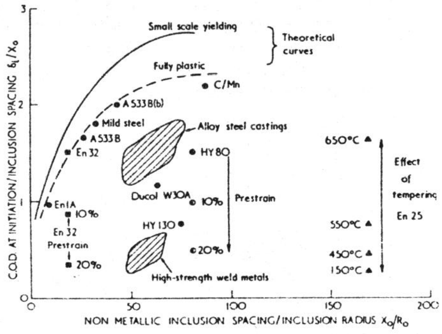


Fig.5 Dependence of δ_i on inclusion size and spacing.

The length parameter required to characterise the initiation value of CTOD in a precracked specimen is essentially the inclusion spacing, $X_o = X_D$. A model, derived initially by Rice and Johnson (22), treats void coalescence by internal necking of the ligaments between inclusion and predicts a variation of (δ_i/X_o) with (X_o/R_o) where R_o is the inclusion radius as shown in Fig. 5. Experimental results for systems containing inclusions in a "clean" matrix of high work-hardening capacity conform fairly well to the predictions, but, in many cases, the values of δ_i are smaller than predicted, because "premature" linkage occurs by the formation of a subset of micro-voids around tempered carbide particles. This is shown in Fig. 6 and 7. Cold prestrain also produces large decreases in the value of δ_i because all the strain can remain localised along a narrow fracture path, Fig. 5 (and ref. 21)

Reductions in δ_i due to prestrain simply reflect a decrease in work-hardening characteristics and could, in principle, be predicted by changes in the constitutive stress-strain relationship employing the same value of critical distance, X_o. The appropriate value to use when a subset of microvoids is produced requires more detailed consideration. In a moderately clean steel, it is observed that the sequence followed is that voids form first around inclusions (for which the initiation strain is virtually zero) and that when microvoid initiation and coalescence occurs, it does so in a shear band which links the crack tip to the nearest inclusion-initiated void ahead of the tip. Rapid decohesion occurs when the strain in this band reaches a critical level for inhibition of a microvoid at a carbide. Although the details of final separation are intimately linked to the features of the carbide population, it is arguable that the best way to characterise the CTOD at initiation, δ_i, or the increment in CTOD, $\Delta\delta$ corresponding to a unit of crack advance, Δa, is to employ the *same value of critical distance* $X_o = X_D$, but to terminate the internal necking process abruptly at an inter-inclusion strain corresponding to the initiation of microvoids at the carbide particles. This strain may be able to be measured independently in torsion tests, or even in plane-strain tension or uniaxial tension tests if the final global instability in these tests is induced by the onset of microvoiding.

The inter-inclusion strain in an assumed 2-D plane-strain stress state is given simply by

$$\varepsilon = ln(X'_o / X_F) \qquad \text{.............................2)}$$

where X'_o is the original distance *between* the void circumferences ($X'_o = X_D - D$) and X'_F is the final distance between expanded voids ($X'_F = X_D - D'$) where D' is the expanded void diameter.

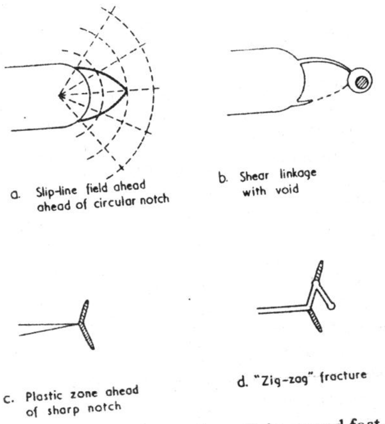
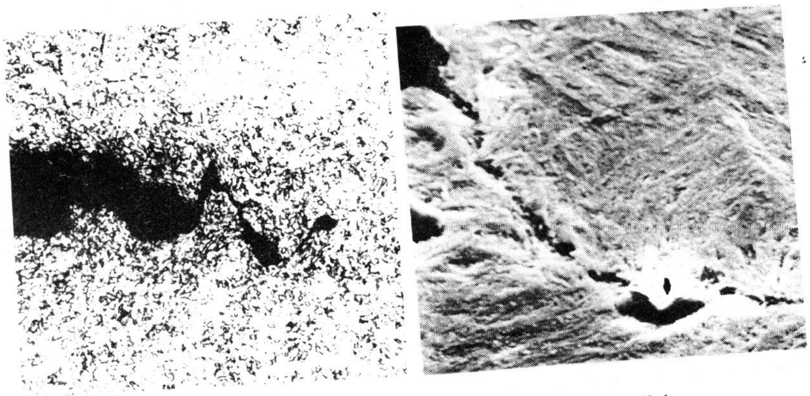


Fig.6 Schematic diagrams illustrating shear linkage and fast-shear failure



a) Illustration of shear fracture in HY steel

b) Details of shear linkage
(Courtesy Dr. C.P. You)

Fig.7

When ε reaches a critical value (e.g. unity for spheroidal carbides) the ligament is deemed to have failed. The CTOD is smaller than that predicted by the internal necking model, *but the critical distance* remains constant. Observations have been made on systems in which the critical value of ε for microvoid initiation was deliberately reduced by heat-treatments which were designed to reduce the decohesion strain by segregating phosphorus to carbide/matrix interfaces (refs 23 and 24). Values of δ_i (or J_i) were reduced accordingly and it was clear that, in segregated samples, less expansion of inclusion-initiated voids had occurred, so that the inter-inclusion strains were reduced.

The question that remains is the definition of critical distance in a steel which contains no inclusions around which voids can initiate easily, but a distribution of fine, tempered carbides which decohere at a critical strain. An analogous situation would be that of a high-strength aluminium alloy containing no inclusions, but a distribution of "dispersoid" intermetallics of size range 0.05 - 0.2 μm. The characteristic distance can now no longer be related to an inclusion spacing but would be thought to relate to a carbide or dispersoid spacing. The problem is that, even in model systems, the particles must be treated as spheres distributed randomly in 3D and some must lie at, or very close to, the fatigue crack tip , where the strains are very high. A small "packet" of fibrous crack then initiates and grows at very low applied strain. As the applied strain is increased, sequentially higher fractions of the fatigue crack front link with microvoids initiated in the high strain region ahead of the tip, until the fibrous crack front extends across the full (plane-strain) thickness of the testpiece (25). The front is still likely to be irregular on the fine micro-scale (of order equal to the inter-carbide spacing). The important feature is, however, that in the absence of inclusion-initiated voids, which allow shear fractures to advance in a discontinuous fashion, the growth of the fibrous fracture initiated on carbides is of the non-cumulative type and can only proceed under increasing load, so that catastrophic failure does not take place before general yield of the uncracked ligament . "Initiation" and "critical distance" now relate to arbitrary definitions. As for the 0.2% proof strength compared with a lower yield strength in a steel which demonstrates a sharp yield drop, we have to *define* a value of Δa corresponding to "initiation". (Even if the back-extrapolation method is used, a degree of subjective judgement has to be used to define the edge of the stretch zone). If Δa is chosen as, say 0.2 mm, or the summed area $B\Delta a$ is equated to B x 0.2 mm (as would be recorded by e.g. the potential drop technique), then we have effectively defined the critical distance as 0.2 mm and the value of $\delta_{0.2}$, (or $J_{0.2}$) is simply given by the integration of strains (or work-rate) in the plastically-deformed region required to achieve critical carbide/matrix decohesion strains over a distance of 0.2 mm ahead of the crack tip. A different definition would imply a different critical distance.

It then becomes of interest to establish the level at which inclusions have significance with respect to the value of critical distance for fibrous fracture. If the strain associated with void formation around inclusions is zero, the inclusion will have some effect if the plastic zone interacts with it before "initiation" occurs *at* the crack tip by microvoiding. Definitions in both cases are somewhat arbitrary, but a rough estimate may be obtained as follows. The plastic zone, in plane strain, has a maximum extent $R_{IY} = 0.16 (K/\sigma_Y)^2$ and minimum extent $r_{IY} = 0.04\ (K/\sigma_Y)^2$. Taking an average value $R = 0.1(K/\sigma_Y)^2$ and combining this with $\delta = K^2/ 2\sigma_Y E$, we have

$$R \sim 0.2(E/\sigma_Y)\delta \qquad \text{.................................3)}.$$

For HY80 (QIN) in the fully heat-treated condition, typical values are σ_Y=500 MPa, δ_i=0.12 mm, E = 200 GPa, hence R ~ 10 mm. Similar values would hold for A533B/A508 in the fully heat-treated condition and it is clear that the inclusion content will never practically be reduced to a level at which inclusions do not influence the critical distance. Even in HY 130, where σ_Y = 1000 MPa and δ_i = 0.05 mm R must be greater than 2 mm. It is possible to heat-treat the HY steels to a condition of high yield strength and low resistance to ductile crack growth, e.g. by quenching and tempering at temperatures in the range 250°- 450°C. Typical values here are σ_Y= 1350 MPa, δ_i= 0.02 mm, hence R ~ 0.6mm. For a simple, square 2-D array, the number of particles per unit area, N_A, is given by $1/R^2$: hence, if R ~ 1 mm N_A is 1 mm^{-2}. For 5 µm diameter particles, this implies an area fraction of approx. 0.002%: for 10 µm diameter, approx. 0.01%. For MnS inclusions, the figures should be multiplied by 0.4 to give the sulphur levels that have to be achieved to obtain such low inclusion levels (large values of R). If figures are re-worked for 3-D distributions of spheres, even lower inclusion levels are required. It appears that, in most practical cases, the inclusions are likely to have an effect on the critical toughness values and that the best way to specify CTOD or J will be by a combination of a distance related to inter-inclusion spacing and a critical inter-inclusion strain.

It is possible to obtain somewhat lower values of CTOD in quenched-and-tempered forging steels or in maraging steels, but, in highly dislocated, lightly tempered microstructures it is not always clear that the apparently fibrous fracture does not contain small, micro-scale components of brittle fracture (even intergranular in form) which occur in response to levels of local tensile stress and confuse the interpretation of critical distance based on purely fibrous fracture concepts. Slatcher (Ref. 26) has, however, made observations on NiCrMo forging steel, tempered at different temperatures in a range from 200°C to 600°C which support conclusions which can be drawn from calculations based on equation 3. At low tempering temperatures, corresponding to high σ_Y and low δ_i, not much difference was observed between the toughness of a vacuum melted version of the steel (low volume fraction of inclusions, large value of R) and an air-melted version (smaller value of R) because fracture initiation was dominated by low-strain initiation of micro-voids around carbides. For higher tempering temperatures (lower value of σ_Y, higher value of δ_i) there was a factor of two difference between the toughness of the two melts for the same yield stress. (see Table 2).

Table 2 Effect of Air-Melting (A) vs. Vacuum-Melting (V) on the Fracture Toughness of 2.7 Ni O.7Cr 0.6 Mo, (En25), Forging Steel (courtesy Dr. Slatcher).

	Yield Stress MPa	δ_i mm	J_i kJm^{-2}
En 25 A	1140	0.038	43
En 25 V	1170	0.044	48
En 25 A	1060	0.057	59
En 25 V	1020	0.105	129

An extreme example of this effect occurs, when mild steel fractures by cleavage at low temperatures. Typical values are K_{Ic} = 30 MPam$^{-0.5}$, σ_Y = 800 MPa; from $R = 0.1(K/\sigma_Y)^2$, R = 0.14 mm. For a 2-D array of 5 µm diameter particles, the critical level of easily decohered sulphide inclusions is of order 0.1%; for a 3 - D array of 10 µm diameter spheres, it is approx. 0.01%. In wrought steels, for which cleavage is initiated on carbides, it is uncommon to see any signs of inclusion-initiated voids on cleavage fracture surfaces: the fracture is of course confined to the cube plane, and the crack does not deviate to "seek out" voids.

The final microstructure of interest is that of a weld metal deposit, which contains a high volume fraction of inclusions. Mostly, these are deoxidation products, typically oxides and silicates, but, in practice, many of them have a duplex character, with an oxide/silicate "core" and a sulphide "shell" which may well form, at least partially, in the solid-state as sulphur segregates to the inclusion/matrix interface during cooling in the austenite range. A distinction between these inclusion types is that the thermal expansion coefficients of relevant oxides and silicates are less than those of the steel matrix, so that the steel "clamps" the inclusion, whereas MnS has a greater thermal expansion coefficient, so that the inclusion tends to shrink away from the matrix. The oxides/silicates may then be expected to serve as good cleavage crack initiators, because continuity is provided between dislocation arrays in the matrix and the brittle particle, whereas MnS is likely to be a void initiator (17).

Experimental observations of fibrous fracture in weld-metals of the C/Mn type show that the process is associated with internal necking between inclusions and values of δ_i/X_o) can approach the Rice-Johnson predictions in micro structures with high work-hardening rates. In a given set of tests, it appears that there is no differentiation between inclusion types in terms of contribution to the fibrous fracture process, but this could result simply from the fact that constant deposition conditions imply a constant ratio of sulphide "shell" to oxide/silicate "core" so that all inclusions have equal void-forming potential. The necessity for plastic strain to initiate voids has not been established with generality, and this clearly needs to be carried out in conjunction with detailed micro-analysis of the inclusions themselves. In higher strength weld metals, the fine voids tend to occur as part of a coarser "zig-zag" fracture path, similar to that observed in high-strength wrought structural steels. Here, the zig-zags are associated with large, inclusion-initiated voids, but the presence of any such large voids has not been established in the weld-metals.

The definition of "critical distance" for fibrous fracture initiation in weld metals does not yet have a firm physical basis, due mainly to a lack of understanding of the details of the fracture process. For the internal necking model, it is clearly related to the average spacing of small deoxidiation products; although the determination of δ_i by lack-extrapolation requires support from extremely careful fractography. If initiation strains are involved, it may be better to use the arbitrary 0.2 mm definition, as described above for carbides. The problem is that the choice of the more appropriate model may depend in a very sensitive manner on the details of the micro-chemical composition of the inclusion/matrix interface, since this controls the ease of void initiation.

An interesting point concerns the observation of inclusion-initiated voids on the cleavage fracture surfaces of weld-metals. These are observed commonly in warm prestressed testpieces and it is deduced that the voids formed during the application of warm prestressing blunted out potentially virulent microcracks. A second, related point is, however, that in cleaved, non-warm-prestressed testpieces, the *river-lines* often seem to seek out inclusions. This can be explained as follows. A river - line is formed by shearing material between two cleavage cracks propagating on parallel {001} planes. If the cleavage propagates more rapidly than the shear, the configuration assumes that of a thin, plate - like ligament i.e. a cracked plate in plane stress, which fails, essentially in mode III by through - thickness shear. This is accommodated by a packet of screw dislocations, running ahead of the crack tip. Any void in the vicinity of this packet of dislocations exterts an image-force attraction (due to the free surface) which can cause the screw dislocations to cross-slip and produce the "shearing-off" on a different plane. Hence the river-line "bends" towards the inclusion. There is evidence that slip on {112} planes is possible around crack tips in steel at low temperatures and such slip can provide a particularly simple crystallographic configuration for the above process. Imagine the main crack to be propagating on two parallel, closely - spaced (001) planes in the [110] direction, such that the line of the crack front is $[1\bar{1}0]$. Shearing-off is postulated to occur on the $(1\bar{1}2)$ plane by means of screw dislocations with Burgers vectors a/2 $[\bar{1}11]$ (and parallel line vectors), moving in the (110) direction. The applied stress is parallel to [001]. If there is a void near to the packet of dislocations, cross-slip can occur, on (211) and/or $(12\bar{1})$ causing a deviation in the direction of the river-line (segments of $[\bar{1}20]$ and $[\bar{2}10]$ or their opposites adding to $[\bar{1}10]$ but the shearing apart still occurs in the [111] direction.

The Fibrous/Cleavage Transition and The Critical Distance

The previous sections have been concerned with values of critical distance for independent cleavage and fibrous fracture modes. For cleavage, the distance relates to the statistical distribution of brittle ceramic particles, such as carbides or oxide/silicate deoxidization products. For fibrous fracture, it relates to the spacing of void-forming inclusions or, in clean steels, to the value of Δa chosen to characterise "initiation". The point is that, in general, different values of critical distance must be employed, and so different values must hold below and above the fibrous/cleavage transition temperature. The reason for the transition, based on the RKR model, is that even in a fully constrained plane strain stress state, increase in test temperature, and the corresponding decrease in yield strength, will result in a state in which the maximum in the tensile stress distribution coincides with the critical distance (for cleavage). Above this temperature, insufficient stress is available to propagate microcracks, and plastic strains associated with work-hardening serve only to blunt microcrack nuclei, converting them to small voids. In the typical annealed mild steel microstructure these voids form in lath-like grain-boundary carbides. The RKR model simply predicts that cleavage cannot occur and would envisage that the macroscopic pre-crack would continue to blunt indefinitely, if fibrous initiation did not occur. In this rather idealised form of the transition, the "lower shelf" cleavage predictions simply stop at T* and the value of K_{Ic} rises in a discontinuous manner to a higher value, associated with fibrous initiation above T*. This fibrous fracture relates to a different value of critical distance.

The point has been made by Ritchie and Thompson (27) in their analysis of K_{Ic} results for A533B steel. Below T* they use the RKR model and show the effect of different values of critical distance pertaining to cleavage: 50 μm, 75 μm, 100 μm. Above T*, they employ a fibrous fracture model based on Rice and Johnson, with critical distances of either 300 μm or 350 μm, to correspond to inclusion spacings. It is important that any micro-modelling of the fibrous/cleavage transition recognises the different nature of the two fracture processes and chooses different values of critical distance accordingly.

The differences in fracture modes and criteria have recently been demonstrated in an elegant manner by Bhattacharjee, (28) following earlier work by Maccagno (29). He has examined fracture behaviour throughout the transition range for HY 130 steel, tempered at low temperatures (250 - 450°C) in mixed Mode I/Mode II loading. At low test temperatures, he obtains cleavage fracture which runs normal to the maximum tensile stress. At room temperature, he obtains shear fractures which run parallel to the maximum shear strain. In the transition range, the mode of fracture is a function of the "mixity" of the loading. If the Mode I criterion is met before the Mode II criterion, the steel fails by cleavage. If the Mode II criterion is met first, the steel fails by shear. The results demonstrate that fracture behaviour must be related to separate micromechanistic criteria. No single, global macroscopic parameter (such as strain energy) could explain this type of transitional behaviour.

The approach followed by Ritchie and Thompson is based on an ideal system, in fully constrained plane strain, and assumes single-valued, deterministic values of fracture toughness. In practice, the transition is often less sharply defined, because "premature" loss of constraint may occur as a result of through - thickness yielding or gross-section yielding in a testpiece of finite dimensions. The problem is particularly acute when trying to relate transitional behaviour in small specimens, such as Charpy "V"-notched bars, to that in large fracture toughness testpieces. (This has to be done when assessing affects of neutron irradiation or long-term aging, on properties). The basic principles hold, but detailed 3-D elastic/plastic stress analysis has not yet been combined with detailed fracture criteria to predict the onset of "premature" relaxation and its effect on toughness. The fact that structural materials possess a degree of inhomogeneity also means that toughness values are not single-valued. Neville (30) has demonstrated that low-temperature K_{Ic} values can be maintained to approx. ± 1% (the level of random experimental error) in careful tests made on homogeneous material. In tests made with equal care on inhomogeneous (two-phase) material, the spread in K_{Ic} values was greater than ± 10% (The work, incidentally contains a consistent method for calculating a lower bound from a data set of K_{Ic} values, based on a three-parameter Weibull fit). This is of particular concern, when treating multi-pass welds, which contain patches of relatively tough, fine-grained material and relatively brittle, coarse-grained material. There are therefore, several circumstances under which the deterministic approach has to be replaced by statistical fitting to data sets and the important role that micro-modelling has to play is as a guide to the most appropriate form of statistical curve-fitting.

The RKR model leads to predictions of a small increase of fracture toughness with temperature in the cleavage range, terminated abruptly at a temperature T*. Experimental data show scatter, with K_{Ic} values generally lying above the RKR low-temperature extrapolation, and a less well-defined transition. It needs to be established that no "premature " loss of constraint has occurred, but it is arguable that, as T* is approached in a large specimen, a number of fracture events need to occur at roughly the same load to cause catastrophic failure of the testpiece, whereas only one is needed at low temperature. If this is the case, the K_{Ic} value as such would be higher than the RKR prediction although failure would have been preceded by isolated cleavage microcracks. There is evidence of such behaviour in 9% nickel steel, tested by Ishikawa in his low-temperature dynamic COD test (31). Under these conditions (ensuring full constraint or allowing for less than full constraint by an appropriate adjustment to T*), it is appropriate to take account of the higher K_{Ic} values, in terms of cleavage toughness, but the value of T*, which deterministically corresponds to the onset of fibrous initiation as well as to the termination of the RKR cleavage mechanisms, should, in principle, remain unchanged.

There are, however, some problems with the practical definition of T* in this way, because there will be some variation in (back-extrapolated) δ_i values or in the values of CTOD associated with 0.2 mm of fibrous crack extension. In a steel of yield strength 500 MPa, a useful guideline (from equation 1) is that a CTOD of 0.2 mm corresponds to a K_{Ic} value of 200 $MPam^{1/2}$. When scatter is admitted in both the cleavage and fibrous toughness data, problems arise. A number of safety cases for engineering plant have been based on the requirement to demonstrate that operation under full load should be under "upper shelf" conditions. If T* is defined as the intersection of the lower - bound cleavage curve with the lower bound fibrous curve, it is a *lower temperature* than if the upper bound fibrous curve is used, (T' say), yet the (cleavage) *initiation toughness values* will be higher. There is also concern regarding the situation in which fracture is initiated by fibrous mechanisms, but then changes to cleavage as it propagates (see Fig. 8).

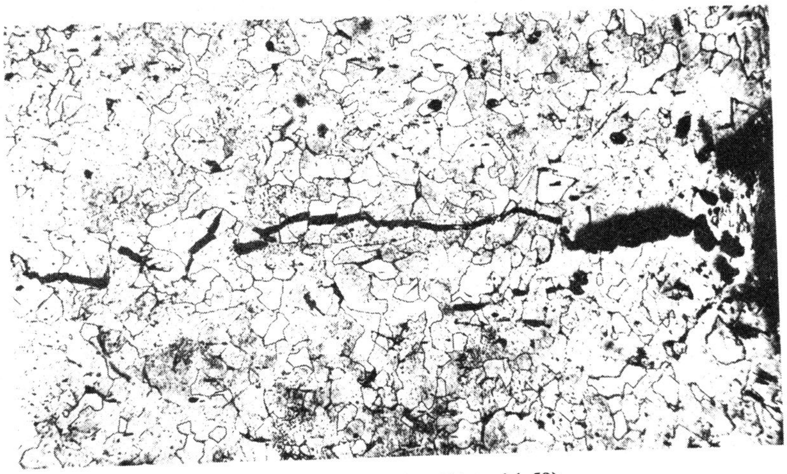


Fig.8 The fibrous /cleavage transition in mild steel (x50).

There are three reasons why this change can occur: two relating to quasi-homogeneous microstructures and one more specifcally to heterogeneous microstructures. In the first case, an appealing argument is that the *fibrous crack accelerates* as it advances. This raises the local strain-rate ahead of the advancing crack tip so that the local yield stress is increased and this raises local tensile stresses such that cleavage can be induced. Experiments carried out by Smith, (32) in displacement-rate-controlled tests made on rather small testpieces, having a/W values < 0.36, showed, however, that a small initial acceleration was followed by retardation as a result of the "drag" of the shear lips. He attributed the reason for the transition to the *increase in constraint* produced by the higher (a/W) value associated with crack growth. This was shown convincingly in his experiments, but does not address the question of the transition in specimens having initial (a/W) values of 0.5 or greater. This topic is currently being studied by Zhang (33) who has employed plain, side-grooved and prestrained specimens to minimise effects of shear-lips. He finds that the fibrous crack accelerates during the first mm or so of growth and that it can continue to grow rapidly if the shear lip "drag" is eliminated. Experiments need to be carried out also under load control. There is interest also in examining the behaviour of specimens containing semi-elliptical cracks because fibrous crack growth rates in these are strongly affected by "plane stress" relaxation at positions where the crack tip intersects the free surface. A further reason for the transition in inhomogeneous material, such as a multi-pass weldment, in that the crack tip encounters a patch of brittle material as it grows. This is an important effect in practice, but cannot be advanced as a general reason for the fibrous/cleavage transition, since this is observed also in quasi-homogeneous material.

The concept of "critical distance" requires careful consideration when cleavage is produced ahead of a growing fibrous crack. In a standard K_{Ic} test (with, say, K_{Ic} = 40 $MPam^{0.5}$ and σ_Y= 800 MPa), the CTOD at fracture is approx. 5 μm. For fibrous initiation, δ_i is perhaps 0.2 mm, and even though the incremental $\Delta\delta$ is smaller than this value, the local crack tip openings ahead of a growing fibrous crack are of order 50 μm, some order of magnitude greater than that corresponding to the measurement of K_{Ic} and the corresponding calculation of critical distance. The strains ahead of the fibrous crack tip are larger and more far-reaching, so that voids have opened up around inclusions and microcracks initially formed in carbides have blunted to form microvoids. The new, virulent microcrack is therefore likely to be located at a position relative to the crack tip very different from that corresponding to the K_{Ic} test (see Fig. 9). A *different critical distance is anticipated*.

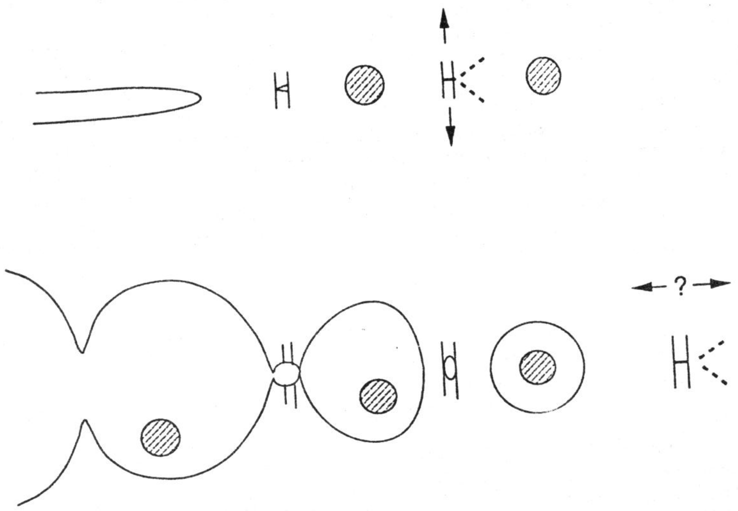


Fig.9 Contrast between cleavage initiation ahead of a fatigue precrack (top) and ahead of a growing fibrous crack (bottom).

It is of interest to try to establish this point experimentally, perhaps using weld metals. In a sense, the phenomenon relates to that of warm prestressing (WPS), if the increase in yield stress due to increase in strain-rate equates to that due to decrease in temperature. In the WPS situation, the application of prior plastic strain clearly blunts out microcracks in large inclusions and renders them inoperative. Since the subsequent (low-temperature) microcrack-initiating inclusion population has been altered, it is unlikely that the statistical summation can lead to the same critical distance. This argument would also apply to the case of cleavage ahead of an accelerating fibrous crack. Values of σ_F would also be affected.

Final Comments

The paper has summarised effects of microstructure on cleavage and fibrous fracture in terms of the appropriate value of critical distance. Two independent values are needed: one (for cleavage) related to the spacing of potential microcrack nuclei, the other (for fibrous fracture) related to the spacings of microvoid nucleating sites. The point is made also that, if cleavage is produced ahead of a growing fibrous crack it is likely that a different critical distance is appropriate.

Similar arguments hold for conventional intergranular fracture in quenched and tempered steels, because micro-crack nuclei still form on tempered carbides. There are, however, two forms of embrittlement which lead to fundamentally different types of "critical distance". One concerns hydrogen-induced cracking, for which it appears that hydrogen, transported into inclusion - initiated voids, diffuses into the strained regions at the tips of the voids and embrittles the void-tip region. The increment of crack advance is then a complex function of applied stress level, pressure of hydrogen in the void and degree of embrittlement (34). An embrittlement, superficially similar in general effect, but different in detail, is produced by the segregation of sulphur, at temperatures of approx. 500°C to grain boundaries in the region of maximum triaxial stress ahead of a notch or crack (35, 36). The segregation embrittles the boundary and the crack grows incrementally by linkage back to the original tip.

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