Effects of Microstructure and Stress-state on Ductile Fracture in Metallic Alloys

J. F. KNOTT

Department of Materials Science and Metallurgy, University of Cambridge, Pembroke Street, Cambridge CB2 3QZ, UK

ABSTRACT

The paper introduces effects of strees-state on ductile fracture in a qualitative manner and then considers in detail the initiation of fracture ahead of a sharp crack and ahead of a blunt notch. The importance of flow localisation is emphasised, with respect to both the micromechanisms of initiation and the relationship between blunt-notch and sharp-crack initiation values. A brief description is given of the effect of crack depth on initiation and the implications of this with respect to the assessment of toughness are discussed.

KEYWORDS

Ductile fracture, Cracks, Notches, Crack-opening-displacement, J-integral, Triaxiality, Assessment of fracture toughness.

INTRODUCTION

The dictionary definition of the word "ductility" refers to the ability of a material to be drawn out into fine threads. In terms of metal-working processes, the production of fine wire is accomplished by drawing through tapered dies, so that the metal is deformed in a stress-state which combines uniaxial elongation with radial compression. Although a single drawing pass is theoretically limited to a maximum reduction in cross-sectional area of 63%, a sequence of passes enables the material to undergo elongations of several hundred percent without fracture. Ductile fracture, when it occurs, takes the form of arrowhead fractures, referred to as "cuppy" fig. 1. In contrast, the same material, pulled as a uniaxial tensile specimen, is likely to exhibit macroscopic necking after a tensile strain of some 10-30%, and may have a total elongation at failure of order 20-60%. The reduction-of-area at failure is sensitive to the material's inclusion content: for very pure material, it may approach 100%; for many engineering alloys failing in a "ductile" manner, it is likely to lie in the range 50-75%. Fracture is of the "cup-cone" type. Fig.2.

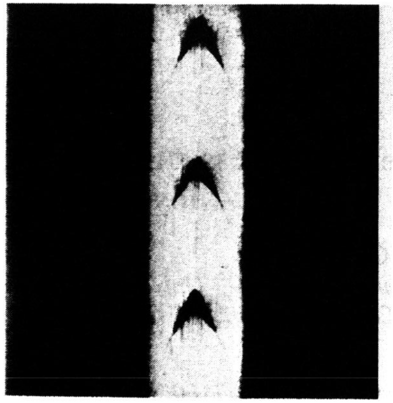

Fig.1 "Cuppy" Fracture in copper wire

Fig.2 "Cup-cone" Fracture in a mild steel tensile specimen.

From these observations, two main features of high-strain deformation and ductile fracture are apparent. The first is that the condition for the onset of a macroscopic 'neck' is affected by stress-state. The second is that the process of material separation - fracture - is separately affected by stress-state. Both these features are of importance in assessing the effects of the different stress-states encountered in metal-working processes on the ability of material to continue to deform in a uniform manner and a number of "abilities" are used to describe the material's response : malleability (being beaten into thin sheet), formability, drawability (the capacity of sheet to be "deep-drawn" into cup-like shapes), bendability and so forth.

A third point refers to the question of whether or not the onset of localised plastic deformation, or "neck", leads to immediate instability in the overall system. Some working processes are load-controlled, so that localised thinning causes the rate of loss of cross-section to increase and catastrophic, unstable failure therefore ensues rapidly. The classic example of this is the failure of a violin string when it is overtightened. In a uniaxial tensile test under load control, it corresponds to catastrophic failure just as the U.T.S. is first reached. On the other hand, it is possible to envisage working processes and tensile tests under displacement-control, where the macroscopic neck simply enables deformation to be continued with smaller loads. Failure is now associated either with a macroscopic condition, where the rate of load decrease becomes steeper than the rate of elastic unloading from the system, or with a microscopic condition, perhaps associated with a dramatic increase in the area fraction of voids present across the minimum cross-section.

In engineering structures, where overall strains are confined to elastic values, ductile fractures under normal operating conditions can only be associated with large stress-concentrators, to enable the fractures to initiate at low applied stress. In a load-controlled structure, it is conceivable, for non-hardening material, that such initiation could give rise to instability, but the loading points must be free to move to maintain

constant load conditions. Most engineering alloys exhibit work-hardening, and the load to promote crack growth is higher than that for initiation. In a displacement-controlled (redundant) system, a ductile crack is unlikely to grow under normal operating loads, unless a low-strain mechanism of ductile fracture is available. One such situation occurs in the ripping of thin sheet. Plane strain crack extension by ductile separation in a pure metal should be a "non-cumulative" process (Cottrell 1963), but it is an experimental finding that materials such as high strength aluminium alloys can fail by ductile fracture under plane-strain "linear elastic" conditions, in conventional fracture toughness testpieces, with K_{IC} values of order 30-50 MPa$\sqrt{m}$. In experiments, a value of 45 MPa$\sqrt{m}$ was maintained down to an initial crack size, a = 2 mm, in a testpiece of width W=25 mm (Wiltshire and Knott 1981).

The incentive to develop a deeper understanding of ductile fracture processes in structural alloys is reinforced by the consideration given by regulatory and licensing authorities to fault scenarios, in which it is envisaged that a component is subjected to loads much greater than its design load, perhaps as a result of failure of other componenets in the structure. Here, stresses and strains well above yield are contemplated and the failure process is assessed in terms of both crack advance and plastic instability of the net cross-section (the ligament, (W-a) ahead of the crack tip). There is clearly an interaction between these two limiting assessments and means of trying to allow for the interaction have been proposed in recent documents addressed to the safety of electrical power plant (Marshall 1982).

The plan of this paper is first to describe some of the main features of ductile fracture at stress-concentrations from the point of view of micromechanisms of crack initiation and growth, and then to comment briefly on the relevance of ductile fracture mechanisms to the integrity of engineering structures. It is pertinent to consider the processes of ductile fracture ahead of two types of stress-concentrator: a sharp crack and a blunt notch (e.g. a Charpy V-notch of root radius 0.25 mm) or a narrow slot.

<u>Ductile Fracture at Crack Tips</u>

The model situation for a sharp crack is shown in fig. 3, in which it is supposed that there is an initial line of inclusions of radius, R_o, and spacing, X_o, ahead of the crack. As the plastic strain in the crack tip

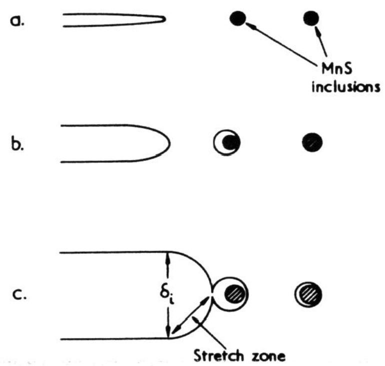

Fig.3. The process of ductile fracture at a crack tip.

region increases, the crack blunts and its tip moves forward, whilst a void, initiated on the inclusion nearest to the crack tip, expands under the combined influence of the local strain field and the hydrostatic tension. INITIATION is deemed to have occurred when the blunting crack tip first coalesces with the expanding void. The displacement of the sides of the crack at the position of the original tip is termed the crack (tip) opening displacement, C(T)OD, δ, and its critical value at initiation is denoted by δ_i. The clean slip steps produced at the crack tip are referred to as the stretch zone. For the situation shown in fig. 3 the projected stretch zone width (SZW) is related to δ_i through the expression $\delta_i = \sqrt{2}$ SZW.

If these events were to occur in a cracked body of size such that the plastic zone required to accommodate the C.T.O.D. of ihterest were much smaller than the length of crack, a, or size of ligament (W-a), it would be possible to relate δ to the stress intensity factor, K, or to the J-integral, J, using the relationships:

in plane stress, $\delta = K^2/\sigma_y E = J/\sigma_y$ ......(1)

in plane strain, $\delta = K^2/m\sigma_y E = J/m\sigma_y$ ......(2)

where σ_y is the yield stress, E is Young's modulus, and m is a constant, approximately equal to 2 for elastic/perfectly plastic material under these quasi linear-elastic conditions (Rice 1976). In plane stress, under conditions of more extensive yielding, the value of δ is related to the applied stress, σ_{app}, through the analytical expression (Burdekin and Stone 1966):

$\delta = (8\sigma_y/\pi E)\ a \ln\{\sec(\pi\sigma_{app}/2\sigma_y)\}$ ....(3)

and this has been used to define an "effective" value of K from $K^2 = E\sigma_y\delta$ (equation 1) with δ substituted from equation 3). Equivalent expressions for both plane stress and plane strain have been derived for relationships between δ and J, assuming conditions of power-law hardening, with the hardening exponent, n, appearing in the equations (Shih and Hutchinson 1976). Values of K_{eff}, defined from $K^2 = EJ$. may also be derived. Under conditions of full plasticity in bending, finite-element analysis has been used to derive a value for m (equation2) of 2 for perfectly plastic material and 3 for linear-hardening material (McMeeking and Parks 1977).

These relationships between δ and J, which can be closely defined for different testpiece geometries and material hardening properties, mean that either parameter may be used equally to characterise a given state of crack tip deformation, provided that this can be totally described by a single-valued parameter. For ease of measurement and for some engineering applications, it may be preferable to use values of J. To relate crack-tip ductility to features of microstructure or to assess the toughness of a small region in a complex configuration of varying properties (such as a weld and associated heat-affected-zone), critical values of δ are more straightforward to interpret. It must be noted that critical values of J or δ are utilised to characterise the resistance to crack extension of tough material: if the material were brittle, linear elastic parameters could be used directly.

Experimentally, the values are measured on rather small testpieces, which have usually undergone general yielding prior to ductile fracture initiation. Either a set of nominally identical specimens, or sequential crack monitoring (e.g. by unloading compliance) is employed to derive values

of J or δ as a function of crack extension, Δa. These crack growth resistance curves may be extrapolated to $\Delta a = 0$ or to the "blunting line", $\Delta a = \delta/2$ (from the geometry in fig. 1), to derive values of J_i or δ_i. Alternatively, provided that growth is "J-controlled" (or "δ-controlled"), it is possible to obtain a value for a fixed amount of crack growth, e.g. $J_{0.2}$ corresponds to 0.2mm of growth. Resistance curves for a number of steels and for an aluminium alloy in different states of ageing are shown in figs. 4 and 5. It can be seen that the initiation values and the slopes of the curves vary markedly with alloy composition and heat-treatment.

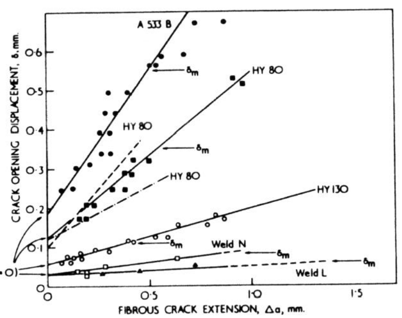

Fig.4 Crack growth resistance curves for a number of steels.

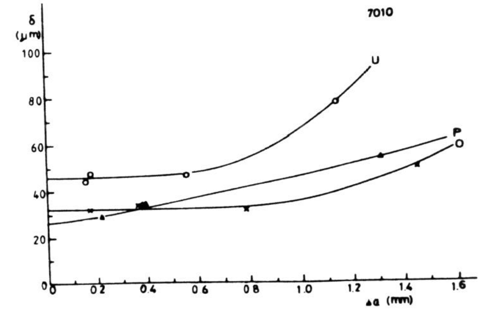

Fig.5 Crack growth resistance curves for 7010 aluminium alloy in underaged (U) peak-hardened (P) and overaged (O) conditions.

The first interpretation of these results refers to the ideal model shown in fig. 3. Here, the value of δ_i is clearly related to the distance, X_o, at which the first inclusion is found ahead of the crack tip, and to R_o, the radius of that inclusion: if R_o is large, correspondingly less displacement is required to produce coalescence. The thoretical predictions of Rice and Johnston (1970) are shown in fig. 6, which plots δ_i/X_o vs X_o/R_o, together with experimental results for a number of steels. It can be seen that some data points - those for steels with relatively clean ferrite matrices and high strain-hardening capacities - do indeed approach the theoretical curves, which may be taken to represent an upper limit to crack-tip ductility (for initially non-bonded inclusions of circular cross-section). An immediate inference would be that materials, such as weld metals, with closely-spaced inclusions are likely to exhibit low values of δ_i, but it must be noted that inclusions in welds are predominantly oxides or silicates which may require the application of a significant plastic strain before voids initiate.

The main feature of fig. 6 however, is that the crack-tip ductilities of many steels are very much lower than the theoretical limit and we recall that high-strength aluminium alloys may fail by ductile fracture under

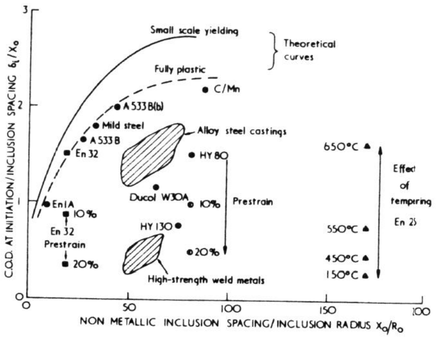

Fig.6 Crack tip ductilities in steels

Fig.7. Microvoid formation in upper bainite (courtesy Dr P Bowen).

"plane strain" LEFM conditions in testpieces of modest size. There are two main factors which contribute to this, either separately or in combination.

The first is that the material may contain a population of second-phase particles, smaller than inclusions, and perhaps initially well-bonded to the matrix. In steels, the inclusions are likely to be sulphides, silicates or oxides and the second-phase particles are carbides. In aluminium alloys, the inclusions are composed predominantly of iron and silicon, and the second-phase particles are dispersoids: intermetallics containing aluminium and elements such as manganese, chromium or zirconium. Consider now the region between the blunting crack tip and the expanding void in fig. 3. As displacements are applied to the ends of the testpiece, the strains in this crack-tip region increase rapidly. If this region now contains a rather closely-spaced distribution of second-phase particles, the increasing strain may reach a value sufficient to initiate micro-voids which provide an "easy linkage" path between the inclusion-centred void and the crack tip. The value of δ_i is then less than would be expected from the simple coalescence model. The precise value of δ_i depends on the strain required to initiate microvoids. For spheroidal carbides in a fully tempered steel, local strains of order unity may be required to decohere the interface, but these may be reduced by the segregation of trace impurity elements to the interface, which reduces cohesion (King and Knott 1981). Plate-like carbides in bainitic or lightly tempered martensitic microstructures may crack at rather low strains to provide micro-void nuclei (see fig.7) and dispersoids in aluminium alloys may also crack or decohere at low strains. (Garrett and Knott 1978, Slatcher and Knott 1981).

The second main factor of importance is that certain microstructures and heat-treatments may permit flow to localise, so that all the strain is confined to a narrow band, rather than being distributed uniformly. A critical value of strain is then achieved with a smaller opening displacement. This effect is demonstrated clearly by examining some experimental results for the effect of a uniform prestrain on the values of δ_i and the slope of the resistance curve for free-cutting mild steel. (Thompson and Knott 1986). In the normalised condition, δ_i was 0.07 mm and

the slope was 0.3 mm/mm; for a 20% prestrain, δ_i was reduced to 0.02 mm and the slope to 0.01 mm/mm. Micro-hardness measurements and metallographic sections through the fracture showed that, for normalised steels, significant plastic strains were developed at large distances on either side of the fracture path and these were associated with observable void openings. For the prestrained condition, any detectable increase in strain was confined to a distance of about an inclusion spacing behind the fracture surface and the only expanded voids were those comprising the fracture path (fig. 8). Macroscopically the crack tended to run at steep angles to the tensile axis or zig-zag corresponding to the directions of crack-tip slip-lines (directions of maximum shear stress or shear strain). Similar effects would be expected in material subjected to high doses of neutron irradiation in which the clustering of point defects allows plastic flow to channel. As-quenched aluminium-zinc-magnesium alloys tested at very slow strain-rates (in the serrated yielding region) also tend to exhibit flow localisation (King, You, and Knott 1981; Lewandowski and Knott 1985). It appears that if the rate is sufficiently slow, the overall plastic deformation can be carried by a single band of high dislocation density, whilst the matrix dislocations are pinned by magnesium atoms.

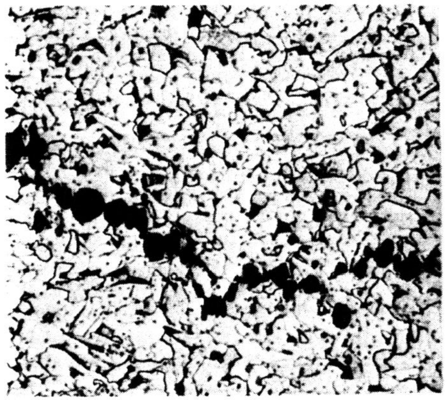

Fig.8 Ductile fracture in free-cutting, prestrained mild steel x 80 (courtesy Dr R.F. Smith)

An interaction between the two effects occurs because the initiation of microvoids is a very good method of allowing flow to localise. Conversely, a system in which flow localises tends to facilitate microvoid initiation at low crack-opening displacements. In lightly-tempered steels, it is rather difficult to distinguish between the effects of transformation dislocation density in allowing flow to localise or in stressing particles and those of particle shape, but the net effect is that the crack tends to follow the zig-zag path associated with sharp crack slip-lines. In a more heavily tempered steel, the tip first blunts, ideally to a semi-circular profile, and decohesion then occurs along a logarithmic spiral path, associated with the slip-lines ahead of a round notch. A similar path is exhibited for such material in blunt-notched bars, as described in the following section.

Ductile Fracture in Notched Specimens

The discussion here is limited to notches of rather high severity in bending, such that ductile fracture initiates at, or just below, the notch root. A typical severity would be equal to, or greater than, that of a Charpy "V" -notch, which comprises a 45^{o} "V" with root radius, ρ = 0.25 mm, and depth ratio, a/W = 0.2. This ratio is associated in three-point bending with a general yield slip-line field of "plastic hinges" confined to the ligament and it should be noted that a/W must be increased to approx. 0.33 to obtain a similar field in four-point (pure) bending. Less severe notches and tensile loading can give rise to ambiguities in the location of the initiation. Notches of mild severity are similar to necked, initially smooth bars in which initiation occurs in the centre of the testpiece: more severe notches are associated with initiation at the notch root.

For material of high strain-hardening capacity, the sequence of plastic deformation and fracture at a notch root proceeds as follows. Initially, yielding begins at a point on the notch root surface where there is an irregularity in profile or some other stress-concentrator. As the plastic strain increases, this region hardens and it becomes easier to deform at another location on the surface. The process continues until the whole "gauge-length" at the notch root becomes equally hardened, and the sequence is repeated continuously to produce quasi-uniform hardening (and hence strain-levels) over the gauge-length. Just beyond general yield, finite element analysis gives the gauge length (for a deep Charpy V notch in pure bend) as 1.2ρ (Griffiths and Owen 1971) and experiments give a value of approx. 1- 1.5ρ at higher strains. (The value of ρ changes as deformation proceeds).

If a number of inclusions are distributed at uniform intervals along the gauge length, it follows that when "initiation" (necking down between the surface and an inclusion-centred void) occurs at one location in the gauge length, it has all but occurred simultaneously at all the other locations associated with inclusions in the gauge length: see fig. 9. The notch- (or slot-) opening displacement at initiation would then be expected to be δ_i for a sharp crack multiplied by the number of inclusions in the gauge length.

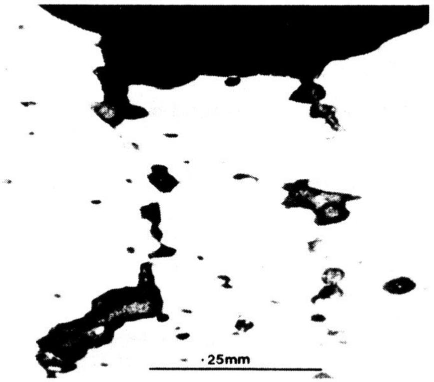

Fig. 9 Ductile Fracture Initiation in Normalised Mild Steel

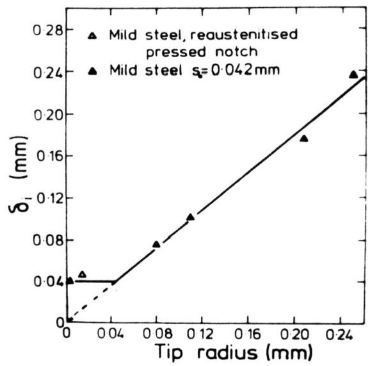

Fig.10 Variation of C.O.D with root radius

This principle has been tested for both slots and notches. Values of initiation C.O.D. have been determined for different slot widths or notch radii and have been shown to follow the expected linear dependence down to a limiting cut-off value, corresponding to the "gauge length" of a sharp fatigue crack (Smith and Knott 1971, Chipperfield and Knott 1975), see fig.10. For a free-cutting mild steel, cut normal to the rolling direction to give a uniform distribution of circular inclusion cross-sections, this "gauge-length" was found to be approximately equal to the average inclusion spacing (based on the average centre: centre distances between circumscribing cells, rather than mean free paths).

For non-hardening material, the sequence of events is dramatically different. If plastic flow is initiated at some minor discontinuity at the notch root, there is no hardening as the local strain increases and so the strain level can continue to increase, virtually without limit until fracture ensues. The effect is demonstrated clearly by comparing figs. 9 and 11, which show ductile fracture initiation at a notch root for mild steel in a normalised condition (with high strain-hardening) and in a prestrained condition (with low strain-hardening). In the first case, a number of voids have formed around the notch: in the second case, the fracture path follows a single line of inclusions initially lying along the logarithmic spiral macroscopic slip-line, which corresponds to the locus of maximum shear strain ahead of the notch.

Fig.11 Ductile Fracture Initiation in Prestrained Mild Steel

Fig. 12 Ductile Fracture Initiation in As-Quenched 7010 alloy

Another example of similar flow localisation leading to "shear" fracture is shown in fig. 12, for an as-quenched aluminium alloy, tested at slow strain-rate. Macroscopically, the fractures follow paths of maximum shear stress or strain, but the fracture criterion is not one simply of shear, because a notch loaded in compression to a notch closing displacement of magnitude equal to the opening displacement associated with shear cracks in tension, does not exhibit any sign of fracture, even though kinks were observed on the notch surface (Slatcher 1983). One suggestion that has been made to explain this effect is that to promote flow localisation in a shear band, it is necessary not only to decohere the interface around a second-phase particle, but also to allow the particle to rotate. If the particle/matrix interface is not perfectly smooth and circular, this rotation is facilitated by tension across the slip band, but hindered by compression across the band, through a "dry friction" effect (Knott 1981). Alternatively, the formation of microvoids may require a double slip mechanism, with the senses of the shears such as to open a cavity under tensile stress, as exhibited by the aluminium alloy, fig. 12. This cavity would not open if the stress were compressive.

The effect of flow localisation can be demonstrated further by examining the effect of cold prestrain on initiation displacements in mild steel specimens containing slots of width 0.15 mm. For the normalised condition, the initiation C.O.D. was 0.12 mm, about three times that for a sharp fatigue crack. For the heavily prestrained condition, the initiation C.O.D. for the slot was 0.02 mm, equal to that for a sharp fatigue crack in prestrained material, because the fracture was confined to a single line of inclusions.

It is pertinent now to discuss these sequences of ductile fracture initiation at a notch and at a crack to the general problem of the assessment of the toughness of material subjected to neutron irradiation in nuclear reactors. Particularly for the older reactors, the surveillance specimens employed are Charpy bars, sufficiently small to be accommodated in the core regions of reactors, so that they can be subjected to accelerated dosages. Charpy specimens cannot by themselves give quantitative toughness data which enable critical defect sizes to be related to applied stresses and so a correlation is established between Charpy values and K_{IC} fracture toughness values, measured on large testpieces. Naturally, this correlation can be established only for the unirradiated condition. Charpy values are then measured for irradiated samples and the SAME CORRELATION is employed to deduce appropriate fracture toughness values and defect sizes. A point to note is that concern is expressed if the upper shelf values drop to levels of order 70J (50 ft lbs). At these levels and above, initiation is almost certainly a ductile fracture process.

Consider now the validity of employing the same correlation for non-irradiated and irradiated conditions in terms of the observations made earlier concerning ductile fracture initiation at notches and cracks. The non-irradiated condition is equivalent, in terms of hardening capacity, to a normalised condition. The irradiated condition is more closely approximated by prestrained material. Any correlation between notch tip ductility and crack tip ductility for non-irradiated (normalised) material is a function of the number of void-initiating inclusions in the notch's gauge length, whereas the notch tip ductility of irradiated (prestrained) material involves only one inclusion : contrast figs. 9 and 11. The correlation derived for non-irradiated material therefore CANNOT in general hold for irradiated material, and it is not immediately apparent as to whether or not the assumption of an identical correlation is conservative or non-conservative, because the Charpy values are expressed as energies, rather than C.O.D. values.

A rough estimate of the effect can be made, using figures for mild steel, and assuming that energy values are proportional to $\sigma_y\delta$ where σ_y represents the yield stresses for the normalised and prestrained conditions

respectively. For free-cutting mild steel, containing a notch of radius 0.25 mm, an experimental value of δ_i was 0.2 mm (Chipperfield and Knott 1975) : for a sharp crack it was 0.04 mm; for a 0.15 mm wide slot it was 0.12 mm (Smith and Knott 1971). Comparison with other results (Thompson and Knott 1986) suggests that for a sharp crack, the value of δ_i would drop to 0.01-0.02 mm after heavy strain and similar values were indeed obtained for a 0.15 mm wide slot. It may be assumed that the localisation effect would give a similar result for a notch of 0.25 mm radius. It is difficult to estimate precisely what value of σ_y to assume, because it is increased, in normalised material, by strain-hardening during the test and, in prestrained material, by strain-hardening during the prestrain sequence. If changes in σ_y are ignored, we see that the correlation δ_i (notched)/δ_i(precracked) for normalised (non-irradiated) material would suggest that "Charpy" values should be divided, <u>inter alia</u>, by five, to give a "sharp crack" value, whereas they would be equal to "sharp crack" values for prestrained (irradiated) material. Use of the "factor-of-five" correlation derived for non-irradiated material might then be unduly pessimistic. The importance of this conclusion is such that further, detailed experimental work is warranted. It may be noted that the effects of prestrain on δ_i in A533B steel are similar in magnitude to those in mild steel.

EFFECTS OF STRESS STATE

In the SEN-bend configurations discussed in this paper, there are two methods by which stress-state can be effectively controlled for through-thickness cracks:

a) by varying specimen thickness or by utilising side-grooves;
b) by varying the crack depth, or a/W ratio.

Additionally, if semi-elliptical cracks are employed, a difference in stress-state is obtained on the top surface, where the geometry approximates to that of a centre-cracked panel, loaded by the uniform fibre stress. The findings from SEN-bend specimens are reinforced by studies made on a wider range of cracked testpiece geometries (Hancock and Cowling 1980).

The effect of testpiece thickness is similar in general form to that on LEFM toughness. Above a critical thickness, some square (quasi "plane strain") fracture is observed in the centre of the testpiece and the absolute widths of the shear lips remain constant as the total thickness is increased. For mild steel, with a (sharp crack) δ_i value of 0.037 mm, the widths of the shear lips were observed to be 2mm. By analogy with LEFM, it is possible to estimate a "radius" of plane stress plastic zone and, using the model of 45° through-thickness slip-lines entering the material from the extremities of the zone, derive the extent of the thickness over which shear lips might be expected to occur. The calculation is strictly improper for a generally yielded body, but the result is given for interest:

$$r_y = (1/2\pi)(E\delta_i/\sigma_y) \quad ..... (4)$$

If σ_y is taken as a strain-hardened flow stress, say 500 MPa, substitution of $E = 200$ GPa, $\delta_i = 0.037$ mm gives $r_y \approx 2.5$ mm, which is close to the observed thickness of shear lips. As testpiece thickness, B, is increased, the shear lips become a smaller and smaller fraction of the total thickness and eventually the load instability coincides with the attainment of δ_i (Green and Knott 1975a). A similar effect can be obtained in thinner testpieces by employing side-grooves, which increase the hydrostatic tensile component and eliminate the shear lips. (Green and Knott 1975b).

It is of interest in this context to examine the "J-validity" criterion with respect to testpiece thickness, B. The "J-controlled growth" arguments are usually derived for plane problems, but standard specimen designs are such that the criterion is often expressed in terms of thickness B (W = 2B; a/W = 0.45 - 0.55), e.g.

$$B > 25\text{-}50\ J/\sigma_y \quad \ldots\ldots (5)$$

for "J-controlled" growth. This might, however, be seen as a criterion to ensure that some region of "plane strain" exists in the centre of the testpiece. Re-writing J as $2\sigma_y\delta$ for plane strain we have B > 50-100 δ. For the C.O.D. results treated earlier, this would imply that, for δ = 0.037 mm. B > 1.85 - 3.7 mm. The higher figure is very close to the value of two "shear lip" thicknesses and would be in general agreement with the arguments based on through-thickness yielding. If equation (5) does, in fact, refer to "plane strain" conditions in the thickness direction, rather than to in-plane J-controlled growth, it would seem that a semi-quantitative physical model is available, but that the multiplying factor in equation (5) is likely to be a function of (E/σ_y).

The effect of varying the (a/W) ratio on the value of δ_i is strongly dependent on the value of δ_i, and the strain-hardening characteristics of the material. In bending, the effect is primarily one of relaxation of hydrostatic stresses by gross-section yielding. If δ_i for a deep-cracked specimen is relatively large, a modest decrease in a/W causes a dramatic increase in δ_i : if δ_i in the deep-cracked specimen is small, effects of a/W will be noticed only when a/W is small. In normalised free-cutting mild steel, for example, δ_i for a/W = 0.5 was 0.07 mm : for a/W = 0.2, it increased to 0.16 mm. In the prestrained condition, in contrast, δ_i was 0.02 mm for both a/W = 0.5 and a/W = 0.2, because the amount of general testpiece deformation required to accommodate a δ value of 0.02 mm was such as only to require yielding on the net section, even for the shallower crack (Thompson and Knott 1986). This is an important point again with respect to correlations between Charpy values (a/W = 0.2 in three-point bend) and K_{IC} values for non-irradiated ($\equiv$ normalised) and irradiated ($\equiv$ prestrained) material.

The effect has been observed also in testpieces in which the value of δ_i was modified by heat-treatment (You 1984). For HY80 tempered at 650°C the values of δ_i were: 0.2 mm for a/W = 0.5; 0.43 mm for a.W = 0.2; 0.56 mm for a/W = 0.1. It is observed that there is a steady increase as a/W decreases. For HY80 tempered at 350°C, however, the values were: 0.05 mm for a/W = 0.5; 0.05 mm for a/W = 0.2; 0.16 mm for a/W = 0.1. Here there is a decrease in δ_i for the deep-cracked testpieces, due to the microstructural features of dislocation density and carbide morphology described previously, but the point of present interest is that the effect of decreasing a/W is only observed for a/W = 0.1, because it is only for such a shallow crack that gross yielding is required, and relaxation of hydrostatic stress does not occur until this gross yielding takes place. In passing, it may be noted that centre-cracks in plane strain or the top-surface of a semi-elliptical crack in bending are associated with low hydrostatic tensile stresses, and that ductile crack extension is correspondingly difficult (You and Knott 1986).

Assessment of ductile crack growth resistance, using through-thickness cracks in testpieces of a/W = 0.5 therefore tends to give unduly pessimistic values for initiation J or δ values, because defects in service are often of low a/W and semi-elliptical in form. Some methods used to compensate for

this pessimism are arbitrary in derivation and may, for example, contemplate as much as 2 mm of ductile crack growth. There is a good case to be made for establishing a proper physical understanding of effects of triaxiality: some information is available (Hancock and Cowling 1981) and testing procedures are being considered, but it is important to make sure that these concepts reach the design-code stage.

The final point concerns the testing of high-strength materials which fail by ductile fracture, such as aluminium alloys and marageing steels. Here the effect of decreasing a/W is relatively simple: for SEN bend tests, the fracture condition is essentially characterised by K_{IC} until a/W is so small that yielding occurs on the top surface of the testpiece before the load required to give the K_{IC} value is attained (Wiltshire and Knott 1981). Since top-surface yielding in a bend bar occurs at a load only two-thirds of that required to produce general yield (ignoring the effects of the small crack) it is possible to test some of the fracture assessment diagrams, using relatively small specimens and carefully controlled alloys. A comparison of some results with a form of the R6 curve is shown in fig.13.

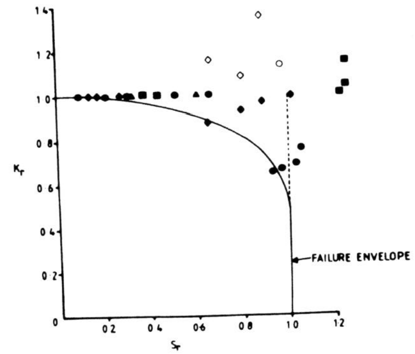

Fig.13 Experimental Validation of R6 curve (Points for marageing steels and aluminium alloy)

The use of the curve is substantiated, as would be more modern versions of the failure locus. It is of interest that failure points do not necessarily hug the failure locus (because loss of triaxiality at low a/W values enhances the toughness), but that the locus does provide a sensible fail/safe lower bound.

ACKNOWLEDGEMENTS

The author wishes to thank Prof. D. Hull, F.Eng for provision of research facilities, and members of his research group, past and present, for useful discussions and for permission to use micrographs and quote unpublished results.

REFERENCES

Burdekin F.M. and Stone D.E.W. (1966) Jnl. Strain Analysis, 1, p.145.
Chipperfield C.G. and Knott J.F. (1975) Metals Technology, 2, p.45.
Cottrell A.H. (1963) Proc. Roy. Soc. A 276, p.1.
Garrett G.G. and Knott J.F. (1978) Met. Trans, 9A, p.1187.
Green G. and Knott J.F. (1975a). Jnl. Mech. Phys. Solids, 23, p.167.
Green G. and Knott J.F. (1975b). Metals Technology, 2, p.422.
Griffiths J.R. and Owen, D.R.J. (1971). Jnl. Mech. Phys. Solids, 19, p.419.
Hancock J.W. and Cowling M.J. (1980) Metal Science, 14, p.293.
King J.E. and Knott J.F. (1981) Metal Science, 15, p.1.
King J.E., You C.P. and Knott J.F. (1981) Acta. Met., 29, p.1553.
Knott J.F. (1980) Metal Science, 14, p.327.
Knott J.F. and You C.P. (1984) Proc 3rd Intl. Conf. on Numerical Methods in Fracture Mechanics. ed. Owen D.R.J. and Luxmoore A.R. Pineridge Press. p.1.
Lewandowski J.J. and Knott J.F. (1985). Proc. 7th Intl. Conf on Strength of Metals and Alloys ed McQueen et al. Pergamon, p.1193.
Marshall W. (1982) "An Assessment of the Integrity of PWR Pressure Vessels" UK Atomic Energy Authority.
McMeeking R.M. and Parks D.M. (1977). ASTM STP 668, p.175.
Rice J.R. (1976) "The Mechanics of Fracture" ed. F. Erdogan ASME p.23.
Rice J.R. and Johnson M.A. (1970) "Inelastic Behaviour of Solids" ed. M.F. Kanninen et al McGraw Hill. p.641.
Shih C.F. and Hutchinson J.W. (1976). Trans ASME Jnl. of Eng. Mat. and Tech. p.289.
Slatcher S (1983). "The Fracture Toughness of Forging Steels" PhD. thesis, University of Cambridge.
Smith R.F. and Knott J.F. (1971) "Practical Application of Fracture Mechanics to Pressure-Vessel Technology" Inst. Mech. Engrs. p.65
Thompson H.E. and Knott J.F. (1986) "Fracture Control of Engineering Structures" Proc. ECF6 EMAS p.1737.
You C.P. (1984) "Mechanisms of Fracture in High-Strength Structural Steels" PhD. thesis, University of Cambridge.