MODELLING OF FRACTURE MICROMECHANISMS

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ABSTRACT

Fracture processes occurring on the scale of the microstructure ("micromechanisms") have only been modelled in physically simple cases such as microvoid coalescence (MVC). Other locally plastic modes such as quasi-cleavage (QC) and the "tearing topography surface" (TTS), and transitions among these modes and to intergranular fracture, have received little modeling attention. An approach to such modeling through use of quantitative fractographic information is proposed, with emphasis on the local "microroughness" *M*. Application of *M*-based analysis to MVC, QC and TTS fracture modes is discussed, as are issues arising from "blocky" fracture surfaces and inter-mode transitions. Additional work needed to extend the use of *M* is identified.

KEYWORDS

Fracture micromechanisms; fracture modeling; quantitative fractography; microstructure-property relations; fracture modes; quasi-cleavage; ductile fracture.

INTRODUCTION

Fracture can be regarded as occurring on a wide range of scales, from the macroscopic to the atomic. For most structural materials, an important scale on which fracture must be understood is that of the microstructure. The term "micromechanism" refers to fracture processes on this scale, i.e. of order 1 μm. This is an important scale because material modifications commonly take the form of microstructural changes, and understanding the fracture consequences of such modifications must be couched in terms of micromechanisms.

There have been numerous efforts to describe (and model) some of the micromechanisms, but most such efforts have neglected an important source of information: fracture surface topography. For the (physically) simple case of ductile fracture by a micromechanism called microvoid coalescence or MVC, it has been found (Thompson, 1983; Thompson and Ashby, 1984) that fracture surface micro-roughness in terms of the void aspect ratio can be described by a parameter *M*:

$$M = h/w, \tag{1}$$

where h and w are the microvoid height or depth, and the width, respectively. The definition is illustrated in Fig. 1. It appears from modeling (Thompson, 1983) that M can be connected in a predictive way with changes in ductility as experimental conditions are changed, provided that the fracture mode remains 100% MVC. Since it appears (Thompson and Ashby, 1984) that M may also be a useful parameter in the description of other micromechanisms, some comments are included below on the magnitude and trends of M values in MVC. These are then extended to modeling needs for other micromechanisms.

MICROVOID COALESCENCE

It is evident from the definition of M, Equation (1), that as $M \rightarrow 0$ the microvoid halves or dimples become vanishingly shallow. If M is indeed near zero, the fracture surface can be treated (locally) like a metallographic section, and the usual quantitative relations (Underwood, 1977) then apply. As pointed out below, however, even very small M values still contain important topographic information. When M is large, dimples are relatively deep and the fracture surface is quite unlike a metallographic section (Widgery and Knott, 1978; Chermant and Coster, 1979; Thompson, 1979; Thompson and Bernstein, 1982). As discussed elsewhere (Thompson and Ashby, 1984), M values are typically of order 0.5 – 1.0 in tensile and bend (notched or pre-cracked) specimens. It should be emphasized that microvoid shapes are in reality quite complex and are not simply ellipsoids, as Fig. 2 illustrates.

After microvoids nucleate in a tensile specimen, their growth in length (prior to necking) simply reflects local strain (LeRoy and others, 1981). Small nucleation strains therefore tend to increase M. In the presence of triaxial stresses, as in the neck, voids grow laterally as well as longitudinally (McClintock, 1978; Tracey, 1971); their length effectively continues to reflect local strain, so the value of h (or h/d, where d is the diameter of the original nucleating particle) is a measure of local longitudinal strain (Thompson and Ashby, 1984). Thus large total strains also increase M. On the other hand, w increases in value after necking (though growth is highly non-linear (McClintock, 1978; Tracey, 1971) with strain), and accordingly tends to reflect the extent of triaxiality of stress which develops. Thus h/d and w/d can be used separately as measures, respectively, of local strain and local triaxiality, provided information is available about microvoid nucleation behavior (Thompson, 1979, 1983; LeRoy and others, 1981).

"BLOCKY" FRACTURES

For fractures which are locally inclined to the direction of maximum stress, microvoid dimensions are unchanged from those observed in areas prependicular to the maximum stress, *provided* (Thompson, 1979; Thompson and Bernstein, 1982) that observation and measurement are conducted (for example, in a scanning electron microscope or SEM) with the viewing direction parallel to the direction of maximum stress. It is of interest to consider, however, "blocky" fractures, which in the present context would be defined as in Fig. 3. In this situation there are changes in fracture surface height on a scale which could be called "regional", i.e. over tens of microvoid or tear-ridge spacings. Such topographies reflect a local scale of plasticity which is larger than the $4h$ value appropriate (Widgery and Knott, 1978; Thompson, 1979, 1983; Thompson and Ashby, 1984) for MVC. This can be characterized by a regional M, called M_r, defined from Fig. 3 as H/W. Note that

$$H = \tfrac{1}{2}\sum^{i}|h_i'| \qquad (2)$$

For the situation, local strain is still given by h/d but the scale of fracture-process plasticity is more nearly H/d.

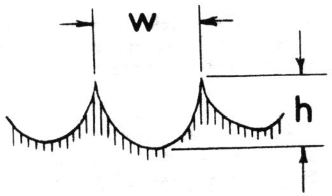

Fig. 1. Schematic cross-section through fracture surface dimple or microvoid, defining depth h and width w. From Thompson (1983).

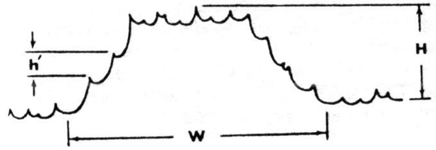

Fig. 3. Schematic depiction of section through a region of blocky fracture (one "block" shown), defining parameters H,W, and h'.

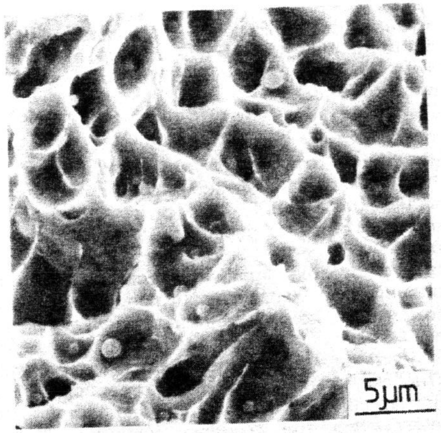

Fig. 2. Example of microvoid coalescence fracture surface, illustrating complexity of dimple shapes. Spherical particles are MnS inclusions; steel is HY-100. Photograph courtesy L. Christodoulou.

For asymmetrical microvoids in general, the mean value of h (whether or not the fracture is blocky in the sense of Fig. 3) is given by

$$\bar{h} = \frac{1}{2n}\sum^{i}|h_i'|, \qquad (3)$$

where n = number of microvoids along a measurement trace. When H is calculated from Equation (2), $\bar{h} = H/n$.

In addition to providing a means of characterization for blocky MVC fractography, the foregoing formulation has the advantage that it can also be applied to "tearing topography surface" or TTS fractures (Thompson and Chesnutt, 1979). Such fractures are often observed to be blocky (Thompson, 1979; Gray and others, 1983; McLaren and Thompson, 1983). The fractographic units in TTS fracture are not dimples, but are irregularly shaped tear-ridge arrangements, so that the number per unit area, N_A, cannot be taken as approximately $4/\pi w^2$, as it can in MVC fractures (Thompson, 1983; Thompson and Ashby, 1984) in spite of the complexities of Fig. 2. Modeling of the micromechanism(s) responsible for TTS largely remains to be done. One stimulus to such work would be the collection of quantitative fractographic data on TTS fractures.

QUASI-CLEAVAGE

By definition (Beachem, 1965; Beachem and Pelloux, 1965), the primary distinguishing characteristic of quasi-cleavage (QC) fracture is that the features which resemble "river lines" are tear ridges, and that these match peak-to-peak on the mating fracture surfaces. By contrast, river lines in true cleavage are steps, not ridges, and they interlock on mating surfaces. Accordingly, fractographic analysis has been focused on these ridges, particularly their spacing (Thompson and Bernstein, 1982). In the present context, it is appropriate to choose dimensions as in Fig. 4, to preserve the meaning of M given in Equation (1). The ridge spacing is w, giving a point of comparison to earlier data on spacings.

The average fracture plane in QC fractures has in some cases been proven to be a low-index crystallographic plane. In at least two cases, however, that plane differed from the usual cleavage plane in the same material (Nakasato and Bernstein, 1978; Araki and Kikuta, 1980): {110} rather than {100} in iron. It is possible that the difference arose from the presence of hydrogen in those tests, although there is no specific evidence for the uniqueness of hydrogen in this regard. But regardless of the role played by hydrogen, the local plasticity involved in the QC process (Beachem, 1965) should, as it decreases in magnitude (for example, as temperature is lowered), cause the tear ridges to become vanishingly small (Beachem, 1976; Thompson and Ashby, 1984), i.e. $M \rightarrow 0$. Whether the resulting low-index plane is necessarily that of cleavage or, as in iron, is a slip plane, cannot yet be answered in general because so few data exist. The slip plane possibility is especially interesting because of evidence (Nakasato and Bernstein, 1978) that at least some fractographic markings are the result of slip intersecting the fracture surface. It has also been suggested that fracture along martensite lath boundaries in steels (Costa and Thompson, 1981; Kim and Morris, 1983) could account for the {110} fracture orientation (Kim and Morris, 1983).

From the modeling perspective, the plastic process zone size should scale with h (probably $2h$), as in MVC. Moreover, since the width of tear ridges is typically about h, a comparison between h values and slip band widths would be of interest, as would a comparison between slip band spacings and w (Nakasato and Bernstein, 1978). Alternatively, microstructural units such as martensite laths or lath packets may control (Costa and Thompson, 1981) the spacing of tear ridges (w), so that lath or packet widths would be compared to w.

INTER-MODE TRANSITIONS

In the fractography literature, it is conventional to discuss fracture modes as though they are separate and distinct phenomena, with each mode having its own characteristics. In practice, however, one not only sees modes which are mixed on an intimate scale, but also transitions from one mode to another. These transitions are of particular interest for possible identification of underlying similarities between modes. For example, it is often observed that under conditions of limited local plasticity, when the dimple structure becomes shallow and tear ridges are the predominant features (Beachem, 1965; Thompson and Bernstein, 1977; Thompson and Williams, 1977), the MVC mode begins to resemble other fracture modes, such as quasi-cleavage (as discussed by Thompson and Bernstein, 1977) and the "tearing topography surface". One implication of such observations is that the tear ridge spacing w may reflect the spacing of nuclei for fracture in each of these modes. This could provide one basis for the assertion (Thompson and Chesnutt, 1979) that nucleation appears very closely spaced in some TTS fractures. TTS is also a propagation mode. Fig. 5 illustrates a "facet" of TTS fracture nucleated at a titanium carbo-sulfide inclusion. Then M would be of interest as a direct measure of the scale of plastic growth from the nucleus, as would h/d as a measure of local strain, for all three of these modes.

Intergranular fracture is in some cases very brittle, for example in strongly temper embrittled steels, but more commonly there is evidence of at least some plasticity near the grain boundary path of fracture. In the limit, the local mode may be observed to be MVC, evidently nucleated at grain boundary particles, so that the intergranular path merely reflects the location of the largest or most dense particles. There has also been observed a transition between MVC and intergranular fracture (Thompson, 1977), with tear ridges as the connecting feature. As mentioned above, this may reflect a

connection between particle or other nuclei at the grain boundary, and the spacing to which plastically-driven microvoids can grow. There is as yet no detailed microstructural or micromechanical rationale for any of these limited-local-plasticity fracture modes (Thompson and Ashby, 1984), but the interpretation of M should remain the same in such cases. Moreover, since M gives a measure of the scale of the local plasticity, it may be a significant parameter in efforts to model these modes (Beachem and Pelloux, 1965; Thompson and Bernstein, 1981; LeRoy and others, 1981; Knott, 1983; Thompson, 1983; Thompson and Ashby, 1984). The paramount need now is for the collection of M data in varied circumstances.

ROLE OF MICROSTRUCTURE

It has been suggested (Thompson and Bernstein, 1981) that modelling of micromechanisms must begin with an understanding of operative fracture nuclei and fracture paths, or at least with assumptions about those factors. As an illustration, a diagram similar to Fig. 6 was presented for a particular assumption about fracture nuclei, namely that they are particles or inclusions. This is not intended to exclude the possibility (or probability) that other microstructural nuclei exist; indeed, other reasonable assumptions can be and have been made (Thompson and Bernstein, 1977; Hirth, 1980; Thompson and Bernstein, 1981). There is ample evidence, however, that nucleation can and does occur at particles.

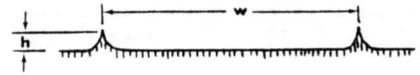

Fig. 4. Schematic section through quasi-cleavage tear ridges, with definitions of h and w consistent with Fig. 1.

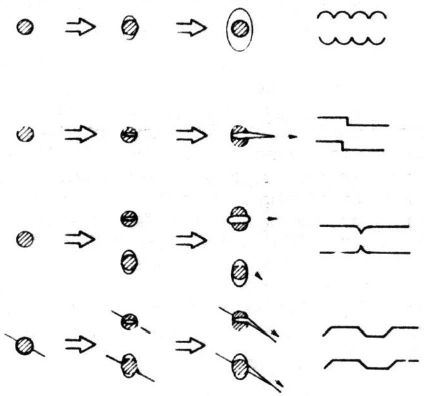

Fig. 6. Schematic illustration of fracture mode development from a single type of fracture nucleus (particles or inclusions), discussed in text. From Thompson and Bernstein (1981).

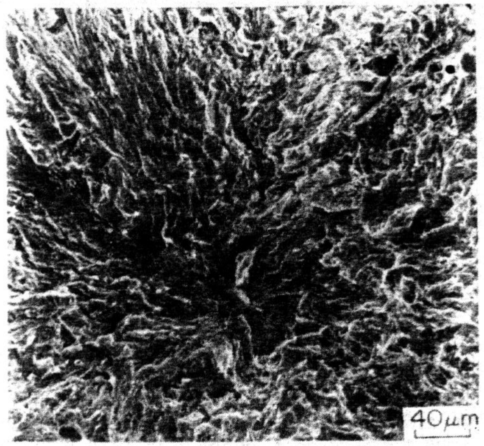

Fig. 5 SEM photograph of TTS fracture facet in low-alloy steel (0.1 C, 1.07 Mn, 0.43 Ti, all wt. pct.) charged with hydrogen, evidently nucleated at Ti(C,S) inclusion at center. Photograph courtesy of M.F. Stevens.

Each individual sketch in Fig. 6 shows a possible nucleation event at the left, an indication in the center as to how propagation from the nucleus may begin, and at the right a rather schematic profile of the ensuing fracture surface. At the top of Fig. 6 is illustrated the MVC case. The next case below that is cleavage, here depicted as nucleating from cracking or fracture of a particle, a process for which particular stress-distance conditions may have to be met (Ritchie, Knott and Rice, 1973). At the right is shown the interlocking character of the "river lines". Following the cleavage case is an illustration of QC fracture, which is here shown as possibly arising from *either* a cracked particle or from plastic microvoid growth on a more restricted scale than for MVC (Beachem and Pelloux, 1965; Thompson and Bernstein, 1981). The local plasticity depicted here, which gives rise to the typical small-scale tear ridges characteristic of QC (Beachem, 1965; Beachem and Pelloux, 1965) and shown at the right in Fig. 6, may be of a similar kind to that proposed as operative in TTS fractures (Thompson and Chesnutt, 1979). Finally, at the bottom of Fig. 6, intergranular fracture is depicted as originating by either cracking or void growth at a grain boundary particle, the latter case corresponding to what is usually called "ductile intergranular fracture."

Fig. 6 provides an example of the development of a viewpoint in which the various fracture modes can be understood to develop from particular microstructural elements under particular degrees of local plasticity, interfacial strength, and similar parameters. Fig. 6 itself represents, as mentioned (Thompson and Bernstein, 1981), a specific viewpoint concerned only with particle-nucleated fracture, and thus cannot be general; but it does illustrate, for this particular assumption about nucleation, understandable processes for the propagation of fracture. Moreover, it is evident that these nucleation and fracture path concepts can benefit (in development of models) from the additional information available from measurement of *h, w,* and *M* (Thompson, 1983; Thompson and Ashby, 1984). It is intended that such data be collected from ongoing programs in our laboratory as model inputs.

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