REVIEW OF FRACTURE MICROMECHANISMS AND A LOCAL APPROACH TO PREDICTING CRACK RESISTANCE IN LOW STRENGTH STEELS

A. Pineau

Centre des Matériaux
Ecole Nationale Supérieure des Mines de Paris
B.P. 87 - 91003 EVRY CEDEX
ERA CNRS N° 767

ABSTRACT

This paper is concerned with the application of the local micromechanisms of failure to the prediction of the macroscopic fracture toughness properties of steels. This approach is based essentially on the incorporation of micromechanistic criteria of failure to analytical or numerical crack tip stress-strain field solutions which are shortly reviewed. The two modes of transgranular failure, i.e. cleavage fracture and ductile rupture, are considered. In each case the emphasis is laid on the experimental results susceptible of providing quantitative criteria for the various stages of failure. For instance, several possible criteria for the three elementary stages of ductile rupture are examined. In the application to the prediction of cleavage fracture toughness it is shown that two types of local approaches - i.e. the concept of a critical cleavage stress and the Weibull theory - can account for the temperature and strain rate dependence of K_{IC} in ferritic steels and in low alloy A508 and A533 steels. The interest of the local approach for the assessment of the cleavage fracture toughness corresponding to complex load temperature history is equally illustrated by reviewing recent studies on the warm-prestress effect. In low strength steels the two important macroscopic aspects of ductile rupture, i.e. crack initiation and stable crack propagation are examined. For crack initiation a critical cavity growth criterion which incorporates the statistical distribution of MnS inclusions is applied to predict the critical crack tip opening displacement. It is shown that this model accounts for experimental results obtained on various heats of A508 steel. Finally, several possible criteria for stable crack growth are shortly discussed in the light of recent numerical calculations and experimental observations.

KEYWORDS

Ductile rupture - Cleavage fracture - Micromechanisms. (Cleavage stress, Weibull theory, Cavity nucleation, growth, coalescence.) - Crack resistance prediction.

INTRODUCTION

The assessment of the integrity of any flawed mechanical structure requires the development of approaches which can deal not only with simple situations (symmetrical Mode I loading in small scale yielding) but also with much more complex situations (large plasticity, non symmetrical loading, non isothermal loading). Several approaches to deal with this problem are possible. It is convenient to divide the possible approaches into two main types. The "global approach" which results directly from linear elastic fracture mechanics (LEFM) or elastic-plastic fracture mechanics (EPFM) phrases the fracture resistance in terms of global parameters such as K_{1C}, J_{1C}, C.O.D. This approach applies most easily to Mode I isothermal loading conditions. Another possible approach relies upon the fact that it is possible to model macroscopic fracture behavior in terms of local fracture criteria. Actually the two approaches are more complementary than contradictory. This paper is concerned essentially with the "local approach". An attempt is made to show to what extent it is possible to relate macroscopic measurements of fracture resistance in steels to the local micromechanisms of failure.

The development of this methodology requires that at least two conditions be fulfilled.

1. Micromechanistically based models for a given physical process of fracture must be developed. In this paper which is mainly concerned with the fracture behavior of structural steels the two main transgranular micromechanisms of failure are considered in turn, i.e. cleavage fracture (lower shelf temperatures) and ductile rupture (upper shelf temperatures).
2. Since, in the local approach, it is assumed that fracture takes place when critical stress or/and strain criteria are obtained at a crack tip, this requires a perfect knowledge of the stress-strain field in front of stationary and propagating cracks. This has been made possible by the advent of analytical and numerical solutions, the main results of which are shortly summarized in the first part of this paper.

The recent progress in the numerical analysis methods has largely contributed to the development of the local criteria based approach. In the third part of this paper which is concerned with the application of local criteria to the prediction of fracture resistance either in the cleavage regime or in the ductile domain, it is indicated, in particular, how it has been possible to account for the macroscopic measurements of fracture toughness of steels corresponding to complex load temperature histories through the use of finite element calculations.

CRACK TIP STRESS-STRAIN FIELD

The advent of elastic-plastic stress distributions ahead of sharp stationary cracks, namely the HRR crack tip analysis by Hutchinson (1968) and Rice and Rosengren (1968) in conjunction with finite element calculations (Levy, Marcal, Ostergren, Rice (1971), Tracey (1976) enabled us to relate local criteria to macroscopic fracture behavior, at least when the situation of small scale yielding is prevailing. In Fig. 1 the results of two finite element calculations can be compared. This figure contains recent results corresponding to axisymmetric cracked specimens obtained

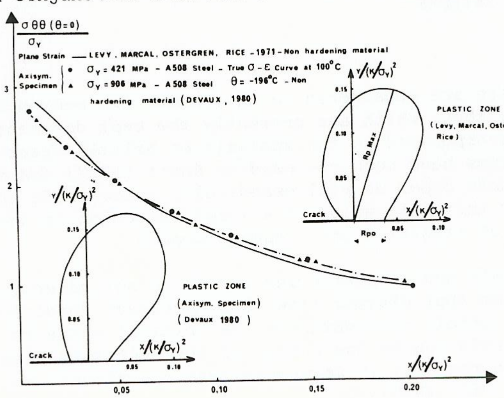

Fig. 1. Examples of numerical solutions for the stress-strain field ahead of a crack tip in small scale yielding.

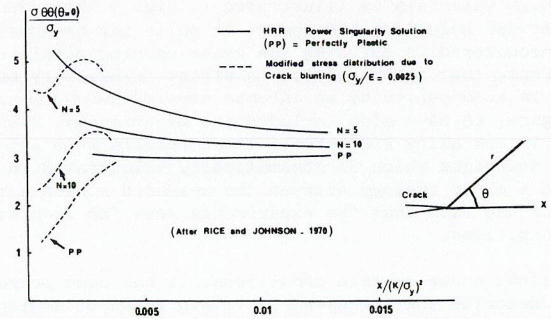

Fig. 2. Maximum principal stress $\sigma_{\theta\theta}$ ahead of a crack tip. HRR asymptotic solution and crack tip blunting solution.

by using different stress-strain curves representative of the behavior of A508 steel at two extreme temperatures, in the ductile regime (θ = 100°C) and in the brittle domain (θ = 196°C) (Devaux, 1980). The comparison between the two types of calculation shows very similar stress distribution in spite of the fact that different constitutive equations and different stress conditions (plane strain vs axisymmetric deformation) were used. In both cases it is observed that the elastic-plastic interface is located at a distance $X/(K/\sigma_y)^2 \simeq 0.03$ ahead of the crack tip. Well inside the plastic zone the HRR solution in addition to the Rice and Johnson (1970) model which accounts for crack blunting shows that the maximum tensile stress can reach values as much as 3 to 5 times the yield stress depending on the work-hardening of the material (Fig. 2). For a material which obeys the stress-strain law written as :

$$\varepsilon_p = \alpha(\sigma/\sigma_y)^{N-1}.(\sigma/E) \qquad (1)$$

it must be remembered that the HRR solution gives the maximum tensile stress in front of the crack tip as,

$$\sigma_{\theta\theta}(\theta=o)/\sigma_y = f(\alpha,N)\, r^{-\frac{1}{N+1}}\, (K_1/\sigma_y)^{\frac{2}{N+1}} \qquad (2)$$

where the values of f(α,N) can be found in the study by Shih (1974). It has equally

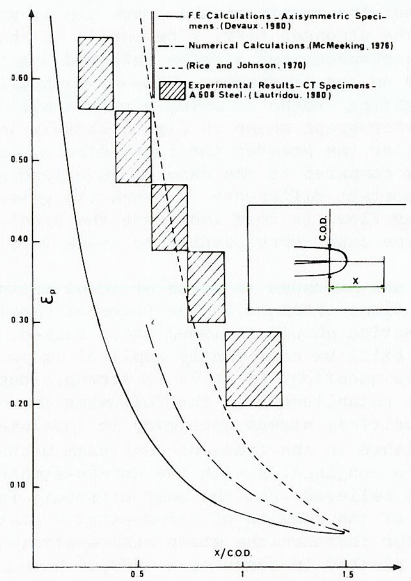

Fig. 3. Strain distribution ahead of a blunted crack. Numerical, analytical and experimental results.

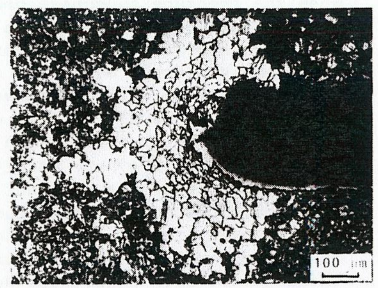

Fig. 4. A508 Steel. Illustration of heavy deformation associated with crack blunting effect. Recrystallization technique (Lautridou, 1980).

been shown by Rice and Johnson (1970) that crack blunting does not only affect the stress distribution but also the strain field. This aspect which is very important for the ductile rupture of tough materials is illustrated in Fig. 3 where the results of three numerical and theoretical solutions are shown. In spite (or because) of the difficulties which are encountered in defining the crack opening displacement, C.O.D. all these solutions indicate that the reduction in stress triaxiality produced by the large geometry change is accompanied by an intense strain concentration ahead of the crack tip. In this figure, we have also included the experimental results obtained by Laudridou (1980) in low alloy A508 steel. These results were obtained by using a recrystallization technique which is schematically illustrated in Fig. 4. It is noted that there exists a clear analogy between the measured strain and the calculated results in spite of the fact that the experiments were far from satisfying small scale yielding conditions.

In large scale yielding, at least under certain conditions, it has been proposed that the HRR field can still describe the stress-strain field ahead of stationary cracks (e.g. McMeeking and Parks, 1978). However the experimental conditions for J controlled crack initiation are not yet fully established. In the case of low strength tough materials, numerical solutions in conjunction with local fracture criteria similar to those mentionned hereafter might be very useful for the characterization of crack initiation.

Very few analytical results exist for the difficult problem of extending cracks. Recent theoretical studies devoted to the analysis of the strain field at an extending crack tip (Slepyan, 1974; Rice and Sorensen, 1978 ; Amazigo and Hutchinson, 1977) have shown that in small scale yielding the strain singularity in front of a propagating crack is smaller than that corresponding to a stationary crack. Results in the ideally plastic case indicate that the strain at the crack tip is governed by a (Ln.r) singularity as apposed to the stronger (1/r) singularity in the stationary case. This difference in strain distribution can also be inferred from the experimental work by Lautridou (1980) based on the assessment of the strain profile ahead of stationary blunted cracks and propagating cracks by using a recrystallisation technique. A close examination to the micrograph shown in Fig. 5 clearly indicates that the mean grain size, which is smaller the greater the local strain, is smaller in the vicinity of the blunted crack as compared to the case of an extending crack. This micrograph illustrates also an important difference in crack tip geometry between a stationary crack and an extending flaw. In this last case the blunting effect is almost absent as opposed to the case corresponding to crack initiation.

When the fracture toughness properties are assessed in terms of local criteria in conjunction with stress-strain distributions obtained in the frame of continuum mechanics a difficult but important question should be immediately raised. At what scale can continuum mechanics concepts still be confidently applied? It is felt that there is no definite answer to this question. First it is strongly dependent on the type of local criteria which are established. In the following the critical cleavage stress (σ_c) criterion or the critical stress necessary to initiate cavities from inclusions (σ_d) are mainly established in the frame of continuum mechanics. They can therefore be coherently used in conjunction with the stress-strain fields which have been mentionned above. It is believed that the most difficult facet of this problem lies in the applicability of the results of stress-strain distributions. In particular it is clear that, for instance, the steep stress-strain gradients which derive from analytical solutions plotted in terms of $x/(K/\sigma_y)^2$ have a doubtful physical meaning when the distance x is smaller than a characteristic distance which is a function of the problem considered. For instance in the case of cleavage fracture this characteristic distance under which continuum mechanics concepts cannot be applied without some adjustement might be the grain size in ferritic steels or the packet size in bainitic-martensitic quenched and tempered microstructures. An implication of this limitation in the direct use of the stress-strain fields at a crack tip is mentionned hereafter in the study of the temperature and strain rate effect on brittle fracture toughness.

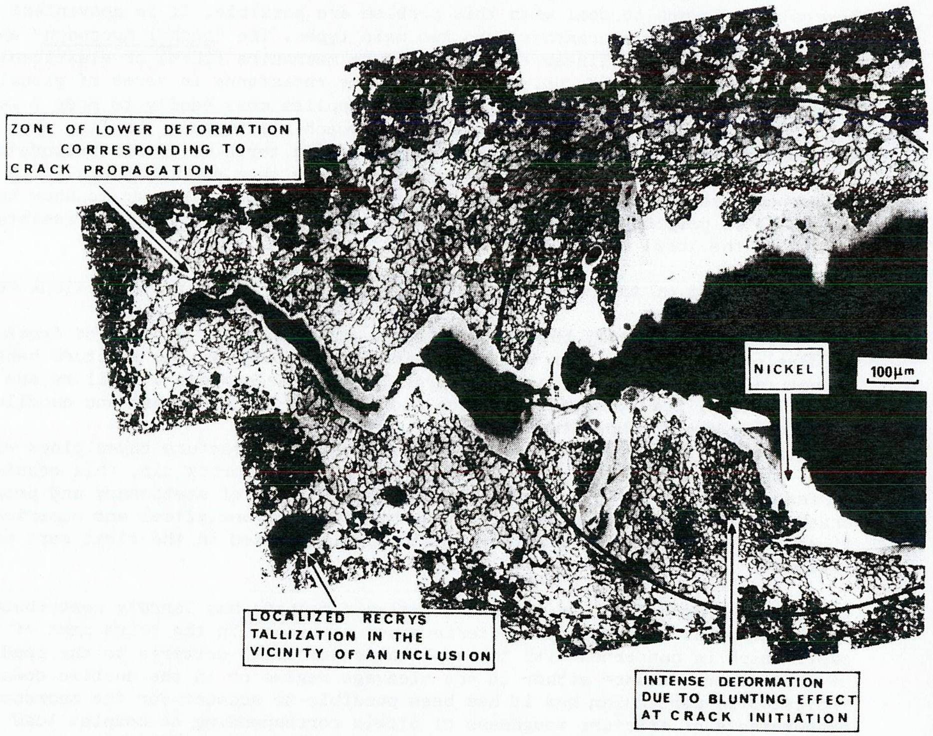

Fig. 5. Micrograph of a sectioned CT30 specimen - A508 steel. The limits corresponding to crack initiation and crack propagation are drawn approximately. They were obtained by a recrystallization technique and they correspond to $\varepsilon \simeq 20\%$. Note the difference in grain size in the two regions. Note also highly strained regions around one inclusion.

MICROMECHANISMS OF FAILURE

Cleavage micromechanisms

In the two modes of failure which are considered in this paper the mechanisms of cleavage fracture are certainly those which are presently the best documented and which can be the most easily incorporated in the analysis of brittle crack extension. The cleavage micromechanisms have been reviewed by Knott (1977) and more recently by Curry (1979a). In this paper we will essentially concentrate on the aspects which are important for the application of a local criterion to the prediction of the fracture toughness of steels in the cleavage domain.

It is well established that metals and in particular steels do not generally exhibit true brittle elastic fracture and that plastic flow is a necessary condition for cleavage fracture. Hence under certain circumstances the critical stage in the brittle fracture behavior of steels can be the advent of the necessary plastic deformation. This has been shown recently in warm-prestress experiments which will be shortly described later (Beremin, 1979).

Several models based on dislocation mechanics have been proposed to account for the stress elevation produced by the inhomogeneity in the slip distribution at the microscopic scale. The first model (Stroh, 1954) assumed that cleavage fracture was nucleation controlled. In some materials, especially in zinc, it is recognized that this is probably the case. A recent investigation by Curry, King and Knott (1978) dealing with the effect of hydrostatic tension on cleavage fracture of pure polycristalline zinc tends to support this analysis. More recently it has been shown by Lemant and Pineau (1980) that a criterion combining a critical resolved normal stress, σ_n and a critical resolved shear stress, τ, applied to the basal slip plane which is also the cleavage plane in this material could account for the mixed mode (K_I, K_{II}) fracture behavior of strongly textured sheets of impure zinc sheets.

In mild steel the situation seems to be different from that in zinc since it has been established that the critical stage is propagation controlled (Knott, 1966). It is now accepted that cleavage fracture of mild steel obeys a critical tensile stress criterion (Knott, 1966). This critical stress corresponds to the propagation of Griffith type cracks nucleated in some brittle second phase such as carbides (Knott, 1977 ; Curry, 1979a). Smith (1966) has proposed a theoretical model of cleavage fracture in mild steel based on the crack nucleation mechanism illustrated in Fig. 6. In this model it is assumed that a grain boundary carbide particle is cracked by an impinging dislocation pile-up. The defect so formed subsequently propagates as a Griffith crack under the stress field produced by the dislocation pile-up and the applied stress. Smith calculated the critical stress to propagate the crack by examining the change in energy when the crack length is increased. Therefore this micromechanistic model is based on an energy balance similar to that employed in linear fracture (macro) mechanics for the establishment of the equivalence between the critical fracture toughness, K_{IC} and the energy release rate, G_{IC}. This appears clearly when the effect of dislocation pile up is neglected since in this case the critical stress is given by :

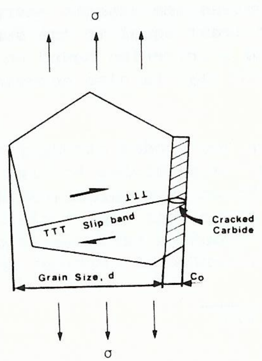

Fig. 6. Smith's model for cleavage fracture.

$$\sigma_c = 2\sqrt{\frac{E\gamma_p}{\pi(1-\nu^2)C_o}} \qquad (3)$$

where γ_p is the effective surface energy of ferrite. This model predicts that the only microstructural parameter affecting the cleavage stress is the carbide thickness, Co. This result appears to be contradictory to the well-know fact that the grain size in mild steel strongly influences the fracture toughness. This contradiction is only apparent when it is considered that small carbides are generally associated with small grain sizes. Curry and Knott (1978) compiling results from several studies have shown that there exists a general relationship between ferrite grain size and largest observed carbide thickness.

From these observations it can be concluded that the critical event in cleavage fracture of mild steel is related to the propagation of a carbide microcrack into the ferrite matrix. Knott (1966) investigating the influence of tensile stress on cleavage fracture by testing at different temperatures notched mild steel specimens showed that, at least when the plastic deformation occurs only by slip, the cleavage stress was independent of the temperature. This result greatly facilitates the analysis of brittle crack extension. However there are several observations which

could indicate that this critical tensile stress criterion in relation with carbide microcracking could be an oversimplification of the micromechanisms of cleavage fracture. In this respect two remarks can be made.

The first is related to the fact that very frequently grain sized microcracks have been observed in the vicinity of the fracture surface of mild steel specimens. These microcracks instead of those associated with carbides could provide the Griffith crack nuclei. If it were the case the cleavage stress should be proportionnal to $d^{-1/2}$, where d is the grain size. Curry and Knott (1978) and Curry (1978) have compiled the experimental results of many workers and have shown that the variation of σ_c with grain size in mild steels follows more closely a law of the type :

$$\sigma_c = k_f\, d^{-1/4} \qquad (4)$$

It is therefore concluded by these authors that the propagation of ferrite grain size microcracks does not generally control cleavage fracture in mild steels.

The second remark is more indirect and deals with the effect of a predeformation on cleavage stress. A certain number of results obtained by Groom and Knott (1975) and by Knott (1966, 1967) in mild steels, and by Beremin (1980a) in a A 508 steel are shown in Fig. 7. These results suggest that the formation of cleavage cracks in a deformed lattice is made more difficult by strain. There are several reasons which can be invoked in order to explain these results. The first one is an increase in the effective surface energy γ_p as suggested by the simplified form of the Smith model (Equation 3). The differences which result are difficult to understand. The second possible explanation for the increase in σ_c could be associated with a reduction in local stress concentration around the carbides which contributes to

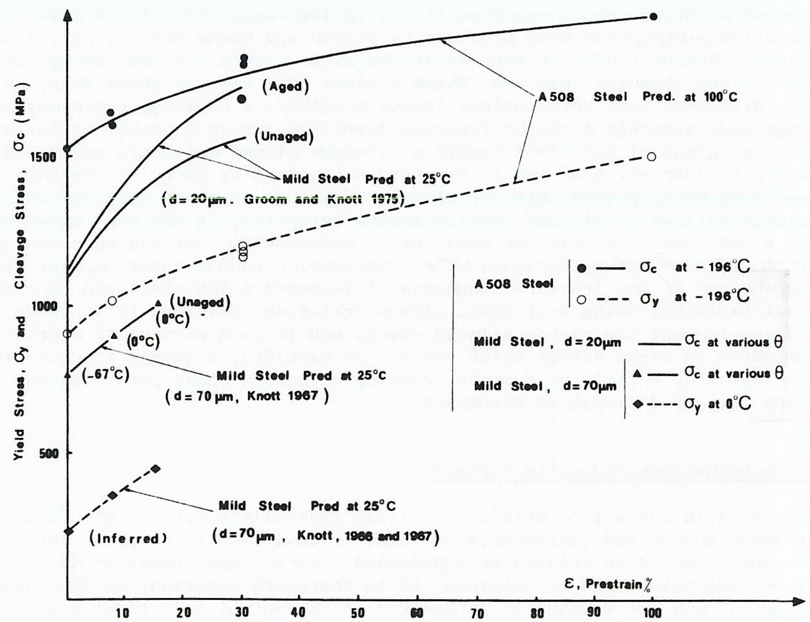

Fig. 7. Influence of a prestrain applied at various temperatures on yield strength, σ_y, and cleavage stress, σ_c.

the propagation of a brittle crack into the ferrite matrix. With this respect, it is interesting to note that in A 508 steel the increase in σ_c with predeformation is more or less equal to the corresponding variation in the yield stress measured at -196°C after a prestrain applied at 100°C.

In spite of these two remarks, it can be stated that, at least for a numerical analysis of crack tip behavior, the concept of a critical tensile cleavage stress can be used. Moreover very similar observations have been made on other microstructures found in steels. Curry and Knott (1978) have investigated the behavior of spheroidized plain carbon steels and a tool steel. They have equally concluded that the critical event in the cleavage fracture of this type of microstructure can be considered as the propagation of Griffith defects of penny shape crack nuclei produced by the cracking of intragranular carbide particles. Parks and Bernstein (1979) have recently studied the process of cleavage crack initiation in lamellar perlitic eutectoid steel. Detailed observations have shown that initial cracking was largely the result of shear cracking of lamellar perlite which is provided by the formation of localized slip bands in ferrite. It appears that the initiation site for cleavage fracture in pearlitic steels is a fibrous region which grows under the applied stress and initiates an unstable Griffith cleavage crack. Using thin foil transmission electron microscopy these authors concluded that a cleavage facet observed on the fracture surface was an orientation unit where the ferrite matrix and the pearlite lamellae of contiguous colonies share a common crystallographic orientation. The size of this orientation unit which is equal to the cleavage facet was found to be controlled by the prior austenite grain size.

This crystallographic aspect of cleavage fracture is equally important in the understanding of cleavage fracture behavior of quenched and tempered steels with microstructures such as lath martensite and bainite. In these microstructures it seems that cleavage fracture is largely controlled by the packet size. This has been discussed by Brozzo and co-workers (1977) in low carbon bainitic steels. Further detailed evidence has been provided by Naylor and Krahe (1975), Naylor and Blondeau (1976), Naylor (1979) in martensitic-bainitic steels for the concept of packet size controlled cleavage fracture. These authors have clearly shown that, in spite of the different lath orientations inside a packet, a cleavage crack may adopt an average approximately straight fracture direction across a packet by following a particular group of different possible cleavage planes which are separated by low angle boundaries. Although it has been emphasized by Curry (1979a) that it is not yet completely proved that the cleavage fracture of these quenched and tempered steels follows a critical tensile stress criterion, it has been shown recently that in A 508 steel the fracture behavior of axisymmetric notched specimens is consistent with such a criterion (Beremin, 1979). Moreover it is shown later that the temperature dependence of the fracture toughness of tempered A 508 steel can be reasonnably well explained using a critical stress criterion. However, it is clear that, especially, in very low carbon alloyed steels and in lath martensite microstructures produced in Fe-Ni alloys which contain no carbides, cleavage fracture is no longer initiated by brittle or ductile cracking of second phase particles but instead by some form of dislocation mechanism.

Micromechanisms of ductile rupture

The ductile rupture of metals and alloys generally occurs by the formation and subsequent growth and coalescence of voids or cavities. It is clear that if cavity formation could be delayed or suppressed a large improvement in ductility and fracture toughness could be obtained. It is therefore important to determine the metallurgical and the mechanical parameters which control this first stage of ductile rupture. Cavity growth and coalescence are equally briefly discussed.

Cavity nucleation. Cavity initiation sites within grains are generally associated with second phase particles or non-metallic inclusions. However there is some evidence to indicate that at least under certain conditions cavities can be homogeneously formed. This is perhaps the case of single phase Titanium alloys (Thompson and Williams, 1977). However in most technologically important materials cavity formation takes place either by the separation of the interface between the matrix and second phase particles or by the cracking of particles.

A recent paper by Goods and Brown (1979) has reviewed a number of theoretical and experimental results dealing with this problem. Cavity nucleation invariably results from the inhomogeneity in deformation between the matrix and the inclusions. Several approaches have been proposed which are based either on dislocation theory or on pure continuum mechanics or which combine both approaches. As pointed out by Goods and Brown it is clear that the applicability of one approach or the other is mainly a function of the particle size. According to these authors the critical radius of a particle above which continuum mechanics could apply is approximately 1-2 µm depending on the work-hardening rate. Particles of this size can be found as carbides in spheroidized steels or non-metallic inclusions, whilst the second type of particles are associated with precipitation strengthening in steels and aluminium alloys. Cavity formation cannot occur unless the elastic energy released from the particle by interfacial separation is at least equal to the surface energy created (energy criterion). On the other hand a criterion based on a critical stress at the interface or inside the particle (σ_d) is also necessary (critical stress criterion).

For small particles ($\lesssim$ 1 µm) which most often are bonded strongly to the matrix, it is necessary to nucleate voids by subjecting the particles to high stresses from dislocations tangled around them. Goods and Brown have calculated the local stress due to the local dislocation density. It is clear that a particle size dependence exists since the rate of dislocation storage around a particle is dependent upon its size. For an uniaxial tensile test, the interfacial critical stress can be written approximately as :

$$\sigma \approx \mu\sqrt{\varepsilon_d\ b/R} \qquad (5)$$

where µ is the shear modulus and R the radius of particles and b the Burgers vector. This equation shows that in the domain of very small particles the initial strain to cause cavitation, ε_d, is a linear function of the particle size. The application of this model to small particles yields a satisfactory correlation although several studies have shown that for particle size larger than about 1 µm (e.g. spheroidized Fe_3C particles in steel (Gurland, 1972) it is generally agreed that particles cracking occurs preferentially in the larger precipitates. The application of the dislocation approach to cavity formation shows equally that except for very small particles the energy criterion is always satisfied before the stress criterion.

For bigger ($\gtrsim$ 1µm) and widely spaced particles a continuum mechanics approach can be applied to calculate the stresses inside the inclusions and in the neighbouring matrix when the material is subjected to plastic deformation. It is clear that a stress criterion based on this approach - at least when particle interaction is not taken into account - leads to a critical strain independent of particle size. Several types of calculations have been proposed to determine the local stress-strain field as a function of the far field applied to the material. The theory of inclusions and inhomogeneities by Eshelby (1961) has been used by Tanaka and co-workers (1970). This theory is based on elastic analysis. Hence it is strictly valid only when the applied strains are not too important. A nucleation model based on a critical stress derived essentially from continuum plasticity has been formulated by Argon and co-workers (1975). Their model incorporates also certain microstructural features of the deformation process in dispersion strengthened alloys which is a dislocation punching mechanism proposed by Ashby (1966). Furthermore for high particle concentrations Argon and co-workers account for particle interaction. Finally since it is well established that the hydrostatic component of the applied stress field has an effect on cavity formation (see e.g. the experimental results by French and Weinrich on spheroidized steel (1974)), their criterion incorporates this effect. It can be written as :

$$\sigma_d = \sigma_{eq} + \sigma_m \quad (6)$$

where σ_{eq} the equivalent Von-Misès stress and σ_m the hydrostatic stress.

In this expression the inhomogeneity in plastic deformation between the matrix and the inclusions does not appear explicity. This effect is directly related to the difference $(\sigma_{eq}-\sigma_y)$, where σ_y is the yield strength. In a recent study dealing with cavity formation from MnS inclusions in a A 508 steel a stress criterion incorporating more explicitily the yield stress has been proposed (Beremin, 1980b). This expression can be written as :

$$\Sigma_1 + k(\sigma_{eq}-\sigma_y) = \sigma_d \quad (7)$$

where Σ_1, is the maximum principal stress whilst k and σ_d are temperature independent material parameters which are a function of particle shape. It is worth noting that this expression was derived from an extension of Eshelby theory to plastically deformed materials by Berveiller and Zaoui (1978).

Typical values for σ_d which have been determined are 1700 MPa for Fe_3C particles in a spheroidized steel (Argon and co-workers, 1975). There is at least one interesting point concerned with these values and which is related to the ease with which cavities nucleate around particles. The interfacial strength between second phase particles and the matrix is dependent on the local chemical composition. The segregation of impurity elements such that those which induce intergranular embrittlement can reduce the interfacial resistance. Hydrogen induced ductility losses in low strength steels could also be at least partly explained in this way. A reduction in σ_d by a factor of about 2 between uncharged and hydrogen charged specimens has been reported recently by Cialone and Asaro (1979). The common observations of dimple size reduction with hydrogen (Thompson, 1979) represent also possible large increases in cavity formation.

Cavity growth and coalescence. Considerable progress in the understanding of hole growth has been made through the theoretical models by Berg (1962), McClintock (1968) and Rice and Tracey (1969). In those models the voids grow under the combined effect of the applied plastic strain and that of the mean stress (σ_m) or stress triaxiality (σ_m/σ_{eq}). These models are based upon a certain number of assumptions which do not necessarily reflect the actual behavior of cavity growth in real materials. In particular they assume that no interaction takes place between two neighbouring cavities. In the Rice and Tracey model the kinetics of cavity growth is given by :

$$dR/R = 0.28\, d\varepsilon_{eq}^{\infty} \exp(\tfrac{3}{2}\,\sigma_m/\sigma_y) \quad (8)$$

where R is the size of holes and $d\varepsilon_{eq}^{\infty}$ the increment of plastic deformation.

This relation was derived for a spherical cavity located inside an infinite perfectly plastic material. Recent experiments conducted on A 508 steel have shown that the critical cavity growth at failure (R_c/R_o) was weakly dependent on stress triaxiality. In a first approximation (R_c/R_o) was found to be almost constant in the range $0.33 < \sigma_m/\sigma_{eq} < 1.50$ (Beremin, 1980b). With this assumption the integration

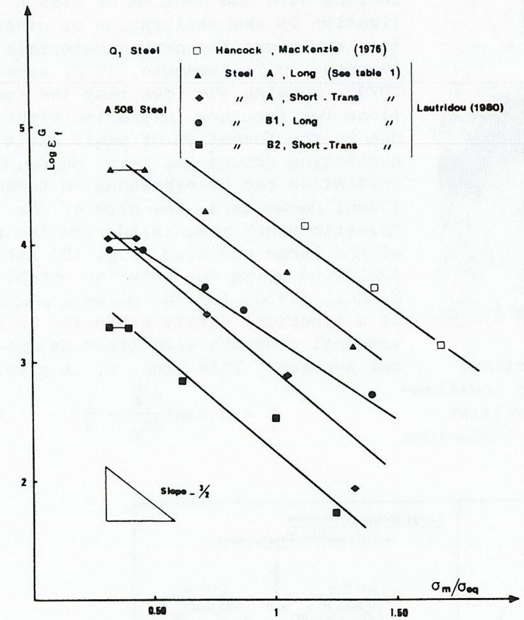

Fig. 8. Strain corresponding to cavity growth at failure ε_f^G as a function of stress triaxiality (σ_m/σ_{eq}).

of the Rice and Tracey expression implies that the ductiliy corresponding to hole growth, ε_f^G, where $\varepsilon_f^G = \varepsilon_f - \varepsilon_d$, ε_f and ε_d being the ductility at failure and the strain necessary to nucleate cavities from inclusions, respectively, should be strongly function of stress triaxiality. This is shown in Fig. 8 where the results obtained by Hancock and MacKenzie (1976) and those obtained on various heats of A 508 steel (Lautridou, 1980) are given. In the results reported by Hancock and MacKenzie ε_d was not taken into account since it had not been determined by these authors, whilst in the results on various heats of A 508 steel, ε_d was substracted from ε_f to obtain ε_f^G. These results indicate that in most cases the slope of the Log ε_f^G vs σ_m/σ_{eq} curve is consistent with the Rice and Tracey model.

Very few experimental studies have been conducted in order to verify more directly the applicability of these theoretical models. However it is worth mentionning McClintock (1968) experiments carried out on plasticine and other experiments conducted on metals and alloys (e.g. Beremin (1980c) on A 508 steel; Cox and Low (1974) on high strength steels; Floreen and Hayden (1970) on Maraging steel; Perra and Finnie (1977) on a coarse grain copper matrix). As a general rule, it is found that the predictions derived from the theoretical models tend to underestimate the observed hole growth. The reasons for this discrepancy are not yet fully understood. One possible reason could be the fact that most engineering materials contain at least two populations of particles nucleating cavities. Cavity formation takes place much more easily at large non metallic inclusions than at small precipitates formed either by carbides in steels or by intermetallic compounds in some steels or in aluminium alloys. Some metallographic evidence suggest that hole growth from large inclusions in those materials does not occur by the simple continuum mechanics of individual cavity growth but also by the coalescence with other smaller voids nucleating at carbides or intermetallic precipitates. As pointed out by Hancock and MacKenzie (1976) in their study of ductile failure in high strength materials these small cavities can grow rapidly under the stress-strain field associated with the growth process of large voids. Simply stated, this means that the local workhardening rate which is an important parameter in the continuum mechanics models for cavity growth can be reduced by the presence of the small cavities formed around large holes nucleating at large inclusions.

Cavity coalescence leading to fracture initiation is a phenomenon which is still poorly understood. It has been observed by many investigators (e.g. Cox and Low, 1974; Hancock and MacKenzie, 1976) that at failure uniform hole coalescence does not occur but that microcracking or some type of local flow instability occurs between the holes. An example of such an instability is shown in Fig. 9. One way of

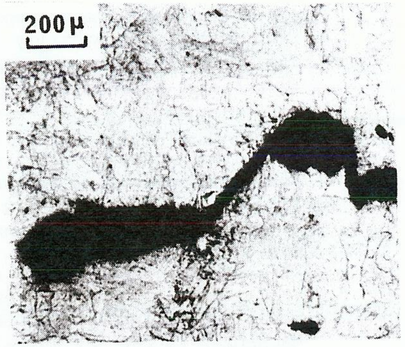

Fig. 9. A 508 steel-S.T. direction. Illustration of strain localization between gowing cavities formed from large MnS inclusions (Lautridou, 1980).

dealing with the problem of strain localization is the utilization of constitutive equations for porous materials (Gurson, 1977; Yamamoto, 1978; Rousselier, 1979). Pursing the idea that the conditions for fracture initiation might be due to the formation of small cavities nucleating around the large holes, failure initiation can be expressed in terms of global parameters. The size of the "plastic zone" associated with the growth of the large cavities (Fig. 10) can be evaluated using the relation established by Hill (1950) for the plastic deformation of a spherical cavity subjected to an internal pressure with minor adjusment to our problem. This size, c, is given by :

$$c = R.\exp\left(\frac{\sigma_m^\infty}{2\sigma_y} - \frac{1}{3}\right) \qquad (9)$$

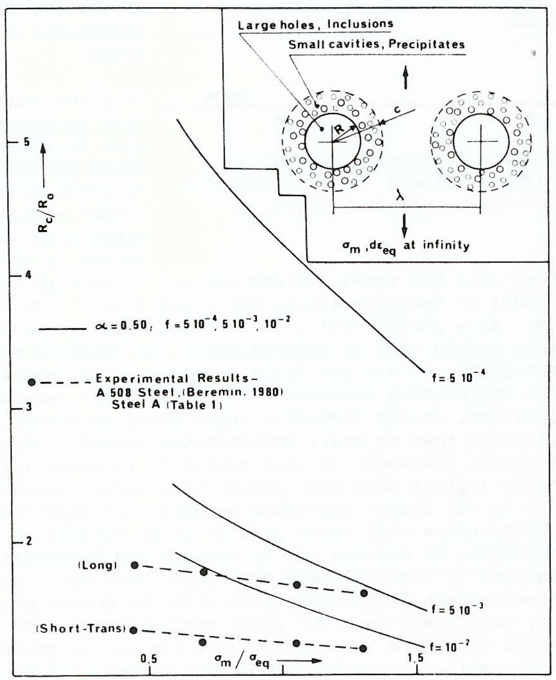

Fig. 10. Schematic diagramm showing "plastic zones" associated with cavity growth. Variation of calculated critical cavity growth R_c/R_o as a function of stress triaxiality and for various values of inclusion content (f). Comparison with experimental results obtained on A 508 steel (Beremin, 1980b).

If it is assumed that the instability occurs when a critical proximity condition is fulfilled, i.e. $\alpha = 2c/\lambda$, this relation can be written in terms of a critical growth as :

$$R_c/R_o = 0.80\,\alpha f^{-1/3}\exp\left(\frac{1}{3} - \frac{\sigma_m^\infty}{2\sigma_y}\right) \qquad (10)$$

where the volume fraction of large second phase inclusions has been expressed as $f = 4\pi/3.R_o^3/\lambda^3$. The critical hole growth derived from this relation is plotted in Fig. 10 for a typical value of the volume fraction of MnS inclusions which can be found in steels, i.e. $f = 5.10^{-4}$ and for $\alpha = 0.50$. It is noted that this simple model predicts values of R_c/R_o which are much higher than those which have been recently reported in a A 508 steel, i.e. $R_c/R_o \simeq 1.70$ in the longitudinal direction and $R_c/R_o \simeq 1.20$ in the short-transverse direction (Beremin, 1980b). It is felt that this large difference reflects the fact that in failure initiation the distribution of inclusions, which is not taken into account in this simple analysis is very important. In particular it is a common observation that failure initiation invariably takes place at the weakest location where the "local" volume fraction of inclusions is much higher than the "mean" volume fraction. In Fig. 10, it is observed that the "local" volume fraction must be as high as 10 to 20 times the "mean" volume fraction in order to obtain values for R_c/R_o consistent with the experimental results. This statistical aspect of ductile failure initiation is very important, especially with respect to the problem of ductile crack tip extension. Another aspect which is also associated with the statistical distribution of inclusions can be inferred from the observation of Fig. 10. It is felt that the difference in slopes between the calculated curves and the measured values of (R_c/R_o) vs (σ_m/σ_{eq}) can partly be attributed to the fact that in the analysis of the experiments on bulk specimens the size effect has not been taken into account. The incorporation of this size effect should lead to a lower slope of the calculated curves.

APPLICATION OF LOCAL CRITERIA FOR THE ASSESSMENT OF FRACTURE TOUGHNESS PROPERTIES

Application to cleavage crack extension

The application of the local approach to brittle crack tip extension is certainly the most developped at this time. In this note an attempt is made to indicate how this approach can be useful for the treatment of two interesting problems which are respectively the effect of temperature and strain rate on the cleavage fracture toughness is steels and the application to warm-prestress effect.

Temperature and strain rate dependence of the cleavage fracture toughness. The local criterion which is the most widely used for predicting brittle crack extension in steels is certainly the concept involving a critical stress (σ_c). This concept has been succesfully applied by Ritchie, Knott and Rice (1973) (RKR) for modelling the variation of K_{1c} of a mild steel with temperature (T). It was shown by these authors that a cleavage crack propagates in an unstable manner when the maximum principal stress σ_{yy} ahead of the crack tip exceeds a critical value (σ_c) over a certain distance (X_o). Finite element calculations indicate that the stress distribution ahead of a propagating crack is almost stationary (d'Escatha and Devaux, 1979). The σ_c, X_o, criterion therefore predicts an unstable propagation. The characteristic distance Xo can be considered as an adjustable parameter. Recently Curry (1980a) has shown that this model can be successfully applied to predict the temperature and strain rate dependences of K_{1c} in a ferritic steel which had been investigated by Shoemaker and Rolfe (1971). If it is assumed that σ_c and X_o are independent on $\varepsilon^{\cdot}$ and T and that the work-hardening parameters (α,N) are also not affected by $\varepsilon^{\cdot}$ and T, a straightforward application of the HRR field indicates that the fracture toughness dependence K_{1C} ($\varepsilon^{\cdot}$;T) should be directly related to the variation of σ_y with $\varepsilon^{\cdot}$ and T. The application of equation (2) leads to :

$$K_{1C}(\varepsilon^{\cdot};T).\ \sigma y(\varepsilon^{\cdot};T)^{\frac{N-1}{2}} = \sigma_c^{\frac{N+1}{2}}\ f^{-\frac{N+1}{2}}\ X_o^{1/2} = \text{cste} \qquad (11)$$

As stated previously the physical meaning of the characteristic distance X_o is not yet clear since Curry and Knott (1978) found no simple relationship between a metallurgical length, such as for instance the grain size, and X_o. Curry and Knott (1979) have proposed a statistical based model which assumes that the crack nucleus size distribution is related to the carbide particle size distribution. Curry (1980b) has shown recently that this last approach is entirely compatible with the RKR model. Although originally proposed for cleavage fracture in mild steel the RKR approach has been shown by Parks (1976) and by Ritchie, Server and Wullaert, (1979), to be applicable to low alloy quenched and tempered steels, as well.
Although these models imply that the fracture toughness is a statistical based property they give no insight into the scatter of the experimental results which is an important aspect well known by all the investigators. One way of tackling this problem is to apply the Weibull theory for brittle fracture (Weibull, 1939). The importance of the size effect which is taken into account into this theory has been shown recently in an experimental study on A 508 steel (Beremin, 1980a). The introduction of the Weibull expression at a crack tip gives a cumulative distribution function P which can be expressed as :

$$P = 1-\exp\left\{-\frac{1}{Vu\sigma u^m}\int \sigma_{yy}^{\ m}\ dV\right\} \qquad (12)$$

In this expression, V is defined in a small sector of angle θ (Fig.11), Vu, σu and m are empirically determined parameters. In the Weibull expression it has been assumed that the cutt-off parameter is null. It can be considered that Vu and σu play similar roles to σ_c and Xo in the RKR model. In particular it should be noted that in both cases two parameters, σ_c and X_o in one case, and Vu σu and m in the other case, are considered. The analytical HRR solution and the numerical results indicate that the stress distribution ahead of a crack tip can be expressed as $\sigma_{yy} = \sigma_y\ g(x/(K/\sigma_y)^2)$. Therefore, it can be easily shown that the relationship (12) in front of a crack can be written as :

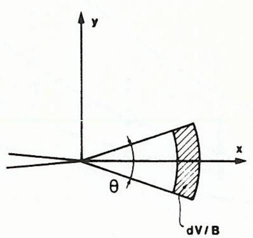

Fig. 11. Definition of angle θ . B is the thickness of the specimen.

$$\text{Log } 1/1-P = \theta B/Vu\ \sigma u^m\ (\sigma y)^m (K/\sigma y)^4 \int g(u)udu$$

where B is the specimen thickness and u = $x/(K/\sigma y)^2$. In the second member of this expression the integral is constant for a given material whatever the temperature and the strain rate if it is also assumed that the work-hardening parameters are independent of T and $\varepsilon^{\cdot}$. Therefore the application of the Weibull theory for a given probability P leads to a variation of K_{1C} as :

$$K_{1C}(T,\varepsilon^{\cdot}).\sigma y(T,\varepsilon^{\cdot})^{\frac{m}{4}-1} = \text{cste} \qquad (13)$$

This expression is very similar to that obtained previously. Moreover the scatter can be estimated between for instance P=0.10 and P=0.90 as $K_{1C}(P=0.90)/K_{1C}(P= 0.10)=2.16$.

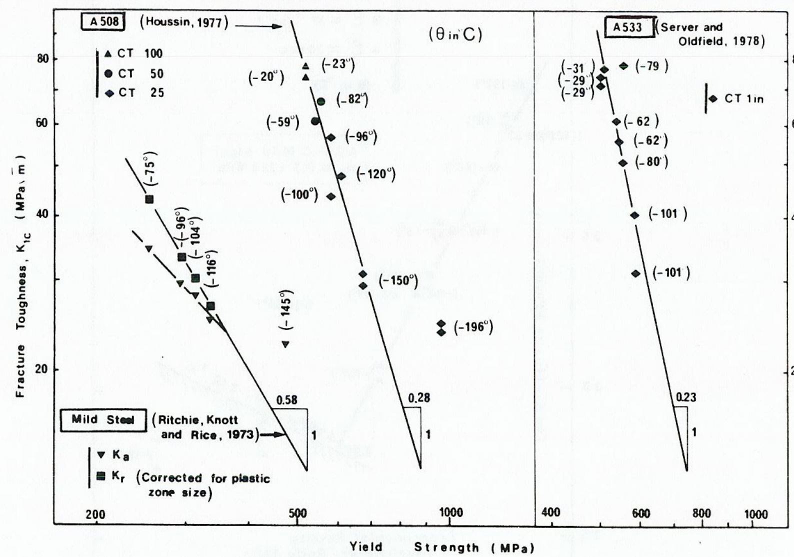

Fig. 12. Variation of the fracture toughness, K_{1C}, as a function of yield strength at various temperatures in three materials (mild steel, A 533 steel, A 508 steel).

In Fig. 12 the temperature dependence of K_{1C} via the variation in yield strength with temperature is plotted for three materials, the mild steel investigated by RKR, A 508 steel (Houssin, 1977) and A 533 steel (Server and Oldfield, 1978). In the mild steel and in A 508 steel it is noted that, except at the lowest temperatures i.e. -145°C and -196°C, respectively, a relationship between K_{1C} and σ_y similar to that suggested either by the RKR model or by the application of the Weibull theory is found. The slope of the K_{1C}-σy curves which can be more confidently defined in the case of the two low alloy steels where more data points are available gives m$\simeq$ 21 or N = 9.5 which are reasonnable values for this type of materials. In Fig. 13 the same representation is applied to the results published by Shoemaker and Rolfe (1971). These authors have investigated the static and the dynamic low temperature K_{1C} of several structural steels and in particular the fracture toughness of a mild steel, referred as ABS-C steel. In this figure it is also noted that, except at the lowest temperatures (-147°C and -196°C), the variation of K_{1C} with temperature and strain rate is directly related to the corresponding variation in yield strength, whilst the slope of the K_{1C}- σy curve leads to values of m or N very similar to those measured in pressure vessel steels.

The fact that the results obtained at the lowest temperatures do not fit the K_{1C}-σy curves is believed to be largely due to the overestimation of the actual stress distribution by continuum mechanics calculations. It is worth noting that, for example in A 508 steel, the plastic zone size at failure at -196°C is only of the order of 2 or 3 grain sizes. In Fig. 13 a limit is drawn which shows that the results do not obey the K_{1C} -σ_y relationship when the plastic zone ahead of the crack tip ($\simeq 0.04\ (K_{1C}/\sigma_y)^2$) is smaller than approximately 20 μm, this value being of the order of the grain size. As stated previously this observation indicates the

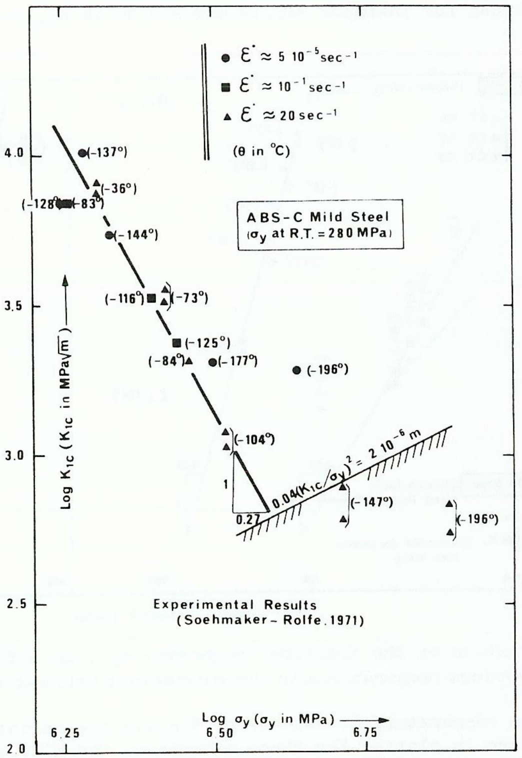

Fig. 13. Static and dynamic fracture toughness K_{1C} as a function of yield strength. The strain rates and temperatures are indicated.

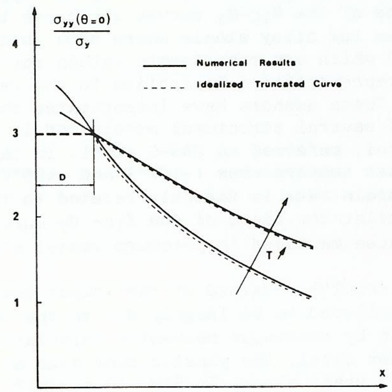

Fig. 14. Schematic diagramm showing the variation of maximum tensile stress ahead of a crack at a given distance as a function of temperature. The maximum stress is assumed to be constant for X <D (truncated curve).

limit of the strict applicability of the results of continuum mechanics. On a physical viewpoint it is probably more realistic to assume that below a certain critical distance (1 or 2 grain sizes) the actual stress gradient is lower than that one derived from the calculations. Mudry (1980) adopting this viewpoint assumed that the stress ahead of the crack tip was constant over a distance D as shown shematically in Fig. 14. It is clear that this modification has essentially an effect in the low temperatures range where the stress gradient is very steep. Moreover, Mudry integrated spatially the stress distribution obtained by finite element calculations. The results of these numerical calculations applied to A 533 steel are shown in Fig. 15. In this figure the data points are taken from the recent publication by Ritchie, Server and Wullaert (1979) who studied the applicability of the RKR model to the fracture toughness data obtained from the EPRI data bank on nuclear pressure vessel steels (Server and Oldfield, 1978). The parameters used for the application of the Weibull theory are given in Fig. 15. i.e. σ_u = 2530 MPa, Vu = 1.70 10^5 μm^3 and m = 21. Moreover the distance over which the stress was assumed to be constant was taken as D = 60 μm. Although the values adopted for the parameters of the Weibull expression are empirically determined it should be noted that the value of σ_u is reasonnable as compared to the value used for Vu which corresponds to about 10 grain sizes in this material. Fig. 15 shows that most of the experimental results can be included between the two limiting curves corresponding to P = 0.10 and P = 0.90. Moreover it also worth noting that the curve derived from the application of the Weibull theory predicts a larger temperature dependence of K_{1C} than the curve obtained with the RKR model. This tendency is clearly closer to the experimental results.

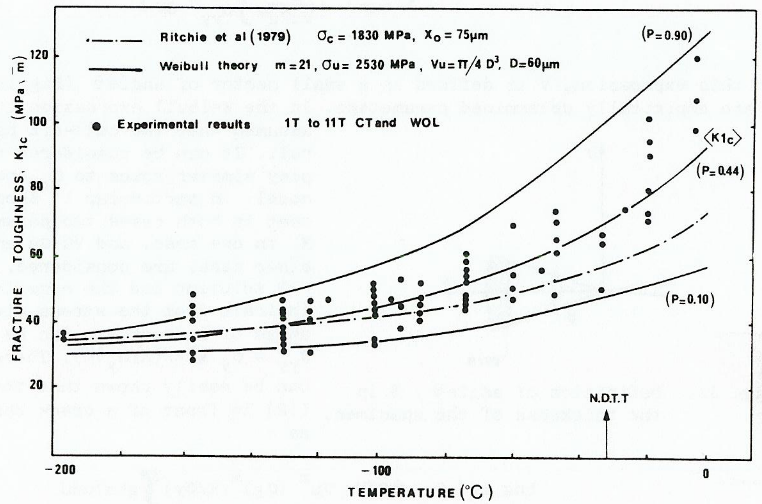

Fig. 15. Comparison of critical stress model (RKR) predictions and of Weibull theory predictions with lower shelf static fracture toughness values for A 533 steel.

Warm prestress effects. Warm prestressing (WPS) is a special case of over-stressing, a procedure which has been well documented and discussed (e.g. Steigerwald, 1961; Brothers and Yukawa, 1963; Bewitt, Cowan and Stott, 1964). Recently the practical importance of WPS in the safety of pressurised water reactors has been recognised (Loss, Gray and Hawthorne, 1978). In W.P.S. the load applied to a structure is

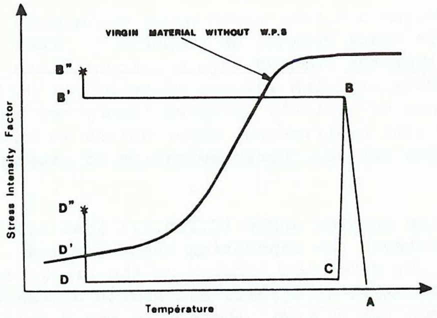

Fig. 16. Warm-prestress effect.

raised above its operating value to a level beyond the brittle-ductile transition. The structure is then partially or completely unloaded, this unloading being not necessarily isothermal. Experimentally it is found that this procedure increases the resistance to fracture of a cracked structure loaded at lower temperature. Fig. 16 illustrates typical effects observed. For example a complete unloading ABC leads to an apparent fracture toughness DD" which is higher than the fracture toughness of the virgin material, DD'. The history ABB'B can lead to still higher values for the apparent fracture toughness. Recently it has been shown that a warm prestress applied at 100°C on A 508 steel specimens up to ductile crack initiation raises the fracture toughness measured at -196°C by a factor of 4 (Beremin, 1979).

W.P.S. is an interesting example for the possible application of a local approach to cleavage fracture in terms of a critical stress σ_c over a characteristic distance X_o. This approach has been used recently by Curry (1979b) in his study on W.P.S. effect in a mild steel and by Beremin (1979, 1980a) in their investigation on A 508 steel. The application of a local approach to this problem necessitates the investigation of the different factors which can contribute to the increase in K_{IC} produced by W.P.S. These factors can be classified into three main groups.

a. Production of a cold-worked structure which is metallurgically more resistant to cleavage fracture. The benefical influence of a predeformation on the cleavage stress of A 508 steel has already been mentionned (see Fig. 7).

b. Modification of the crack tip geometry which becomes blunted after preloading. This blunting effect modifies the stress-strain distribution at the head of the crack tip. It is clear that, when the characteristic distance is of the same order of magnitude as the C.O.D., allowance must be made for the elevation of the cleavage stress due to prestrain in addition to the effect of crack blunting.

c. Introduction of residual compressive stresses. This factor is certainly the most important especially for W.P.S. corresponding to moderate preloadings since in this case the effects of the two previous factors are weaker .

The relative importance of these factors differ in the two mentionned studies in the sense that Curry's work is limited to situations where the results of small scale yielding analysis can be applied, whilst Beremin investigated loading conditions involving large plastic deformation and hence large C.O.D. Curry obtained the stress distribution associated to a given loading history by using the principle of superimposition of stress fields characterized in terms of K and corresponding to the various steps of the loading history. In his experiments, amongst the three mentionned factors, the effect of residual stresses is certainly the most important. Beremin used the results of finite element calculations to obtain the stress distributions since most of their experiments corresponded to loadings for which small scale yielding fracture mechanics cannot be applied. Moreover they took into account the influence of prestrain on σ_c and the crack blunting effect. In both studies it was found that this type of local approach for cleavage fracture can account for the beneficial effects of WPS which were observed. Finally it is worthwhile to mention that at least under certain loading conditions (loading ABB'B", Fig. 16) the study on A 508 steel suggested that the critical event controlling brittle fracture

was no longer the attainment of σ_c over X_o but the advent of plasticity at the crack tip which is a prerequisite for cleavage fracture.

Application to ductile crack extension

Ductile crack extension in low and medium strength steels is most often characterized in terms of global parameters (J_{IC}, COD, dJ/da approaches). In this section an attempt is made to illustrate how the results derived from a local approach can be applied to the study of the two steps of ductile crack extension, which are crack initiation and stable crack growth.

Ductile crack initiation. The main difficulty encountered in the application of a local approach to predict ductile crack initiation is associated with the fact that very steep stress and strain gradients are present at the crack tip (see Fig. 3). Several approaches can be proposed to deal with this problem.

The first approach uses the concept of a process zone over which average values for the stresses and the strains are used more or less explicity. In order to model the physical process of ductile rupture this process zone must have a dimension which can be compared to a physical length such as the mean distance between inclusions (Δ). Several criteria like a critical $(C.O.D.)_c$ at crack initiation which is compared to Δ (e.g. Green and Knott, 1976) or a critical void growth at a critical distance (Rice and Johnson, 1970; McMeecking, 1976) have been proposed. Recently the effect of stress state on the plastic strain to initiate ductile failure in axisymmetric notched specimens has been used by Mackenzie, Hancock and Brown (1977) to predict crack initiation and by Ritchie, Server and Wullaert (1979) to model the fracture toughness of A 533 steel at the upper shelf temperature. As mentionned by Mackenzie, Hancock and Brown, one source of error with this analysis lies in the fact that the stress-strain history in front of a crack tip and inside an axisymmetric notched specimen are quite different. In particular at a given position in front of the crack tip the stress triaxiality σ_m/σ_{eq} does not remain constant but reduces as C.O.D. increases.

Another difficulty with this approach lies in the definition of the critical distance, Δ. In particular, it is clear that there are many inclusion spacings depending on the type of particles which are considered. This aspect is particularly important in the case where the inclusions are not always spherical which is the situation in most wrought materials. As an exemple the results obtained recently by Lautridou (1980) can be mentionned. This author measured $(C.O.D.)_c$ and J_{IC} at crack initiation in 4 heats of A 508 steel which contained various amount of MnS inclusions and which were given different hot working schedules. This resulted in a strong anisotropy effect as noted in Table 1. These results indicate that the $(C.O.D.)_c$ measured from metallographic observation of the crack tip correlate much better with the distance to the nearest neighbour in a plane normal to the crack front (Δ_2) than with the mean distance in the volume (Δ_3). The fact that the original Rice and Johnson model (1970) based on the void growth process gives only a limited agreement, since the ratio $(C.O.D.)_c/\Delta$ is found to cover a considerable range (see e.g. McMeeking (1976)) could be due to inaccurate statistical estimation of the model parameter. Moreover this model can only apply to materials with loosely bonded particles since the strain necessary for cavity formation is neglected.

Very few experimental studies have yet been undertaken to apply to the situation at a crack tip the criteria which allow the description of the three steps in ductile rupture to be made. An attempt using cavity growth combined with statistical analysis has been made recently to predict crack initiation in A 508 steel (Beremin, 1980b; Lautridou, 1980). In this material the application of relationship (7) shows that at a crack tip where the stress triaxiality is very high a very small strain is required to initiate voids from MnS inclusions. Therefore in this material it is quite appropriate to neglect this first step in ductile rupture and to take into account only the process of hole growth and hole coalescence. Hole growth was calulated using the results of Rice and Johnson (1970) or those of McMeeking (1976) for the stress-strain distribution in conjunction with the Rice and Tracey expression (equation 8). In order to take into account the effect of inclusion distribution, a mean value for the cavity growth in front of the crack tip was calculated as :

$$\langle R/R_o \rangle = \int_0^{\infty} R/R_o(x/COD,\theta).P(x,\theta)\ dV(x) \qquad (14)$$

where P is the probability of finding one inclusion in the volume dV defined in Fig. 17. The angle 2θ of the sector was taken equal to 90 degrees whilst the elementary length h parallel to the crack tip was taken as $h = (L_1.L_2.L_3)^{1/3}$, where L_1, L_2 and L_3 are the three mean dimensions of the inclusions in the x,y,z directions. Equation (14) was integrated up to the critical C.O.D. which was measured by Lautridou (Table 1). In steel A, these results led to a mean value of 1.6 in the longitudinal direction and 1.25 in the short transverse direction. These two values are very similar to those determined in notched specimens (Fig. 10). Similar results were obtained in the other materials. The reason why these values were retained rather than those which could be obtained by extrapolation of the experimental (R_c/R_o) vs (σ_m/σ_{eq}) curve to higher values of stress triaxiality has been discussed elsewhere (Beremin, 1980b) and is associated to the fact that the size effect was not taken into account, as already mentionned. The results corresponding to the 4 heats of A 508 steel are included in Fig. 18 where a reasonnable agreement between the calculated C.O.D. and the experimentally determined values is observed. In particular it is noted that the anisotropy effect between the longitudinal and the transverse direction is largely taken into account. However it is recognised that the results are still too limited to conclude that the incorporation of the statistical aspect in the void growth model definitively improves this type of local approach applied to ductile crack initiation.

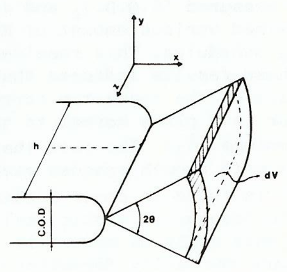

Fig. 17. Definition of 2θ,h, C.O.D. and dV for the statistical approach to ductile crack initiation.

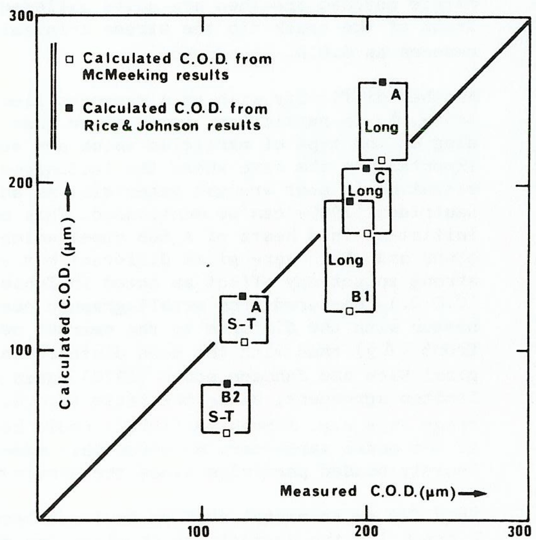

Fig. 18. Comparison of calculated C.O.D. with experimental values. A 508 steel (Table 1).

Ductile crack propagation. Less attention has been given by metallurgists to the understanding of the process of stable crack growth, as compared to crack initiation. As stated previously it seems that the reduced strain concentration accompanying propagating stable cracks partly explains why the slope of the J- Δa resistance curve is positive. In the absence of a widely accepted theory for stable growth we will concentrate mainly on some experimental observations which tend to single out certain candidate parameters for the characterization of propagating cracks.

One possible parameter is the crack tip opening angle (C.T.O.A.) (see e.g. Rice and Sorensen (1978), which is the angle between the separating crack surfaces. There is some difficulty in defining the C.T.O.A. since the tangent at the crack tip is vertical. However the difficulty does not exist at a small but finite distance if the C.T.O.A. is defined away from the crack tip by δ/Δl, where δ is the displacement at a distance Δl characteristic of the material. It has been shown by Rice and Sorensen (1978) that the slope of the J-Δa curve could be directly related to δ/Δl. It seems logical that δ should be related to the ductility of the material, whilst Δl should be equivalent to the process zone size, a clear definition of which is not yet definitively accepted. There is clear evidence indicating that the slope of the J-Δa curve is more or less directly related to the ductility of the material. For instance in A 508 steel it was confirmed that the slope of the J-Δa curve was much smaller in the short transverse direction as compared to the longitudinal one (Table 1). In spheroidal graphite cast iron it has equally been shown (François, 1979) that the increase under pressure-which is a parameter strongly affecting the ductility- of the slope of the J-Δa curve tends to support a general relation between this parameter and the ductility of the materials.

TABLE 1 Mechanical Properties, Inclusion Parameters (Δ_2,Δ_3) and Sulfur content in 4 heats of A 508 steel

Steel	Direction of loading	σ_y (MPa)	Σ (%)	(C.O.D.)c (μm)	J1c 10^3Pam	dJ/da MPa	Δ_2* (μm)	Δ_3** (μm)	S wt %
A	Long. Short-Trans.	466 476	75 65	210 125	220 140	520 200	228 145	58 58	0.010 0.010
B1	Long.	290	72	190	165	305	256	59	0.020
B2	Short-Trans.	225	40	115	80	170	107	38	0.013
C	Long.	432	74	200	220	350	157	43	0.005

* $\Delta_2 = 0.50/\sqrt[2]{N_A}$, where N_A is the number of inclusions per unit area in a plane normal to the crack line.

** $\Delta_3 = 0.554/\sqrt[3]{N_v}$, where N_v is the number of inclusions per unit volume.

Another possible way of dealing with this problem is the utilization of the criteria characterizing the three elementary stages in ductile rupture, in conjunction with finite element calculations which use the node release technique to model crack propagation. This approach would seem logical since there are clear indications showing that the physical processes involved either in crack initiation or in crack propagation are very similar. With this respect an attempt has been made by d'Escatha and Devaux (1979) in their numerical model of crack propagation. These authors have shown that the main features of stable crack propagation could be obtained using for the conditions of node release a criterion based on a constant value for void growth (R_c/R_o). In their calculations a value $R_c/R_o = 1.40$ was chosen. It is worth noting that this specific value is very close to those which were measured in axisymmetric notched specimens of A 508 steel (see Fig. 10). This gives a strong indication of a possible local approach similar to critical void growth model which can be very useful for predicting crack initiation and crack propagation as well. It is clear that, as for crack initiation, the effect of inclusion distribution on stable crack growth properties is very important. For instance in the 4 heats of A 508 steels given in Table 1, it was observed that the slope of the resistance curve could change by a factor of about 3 between the longitudinal direction in steel A and the short-transverse direction in steel B2. From an examination of the sulfur content it is noted that dJ/da is not only dependent on the volume fraction of inclusions. A closer examination to the particle size showed that it was possible to relate the crack growth properties to a parameter $R_s\sqrt[3]{N_v}$ which is the product of the mean size of inclusions projected onto the fracture surface (R_s) and cube root of the number of inclusions per unit volume (N_v). This parameter is proportionnal to the ratio between the initial cavity size and their mean volumic distance (Δ_3). The correlation is shown in Fig. 19 where it is noted that the parameter characterizing crack propagation was taken as $(C.O.D.)_c/J_{1c} \cdot dJ/da$. This figure suggests that it is possible to relate the macroscopic parameters describing stable crack propagation to the parameters characterizing the inclusion distribution.

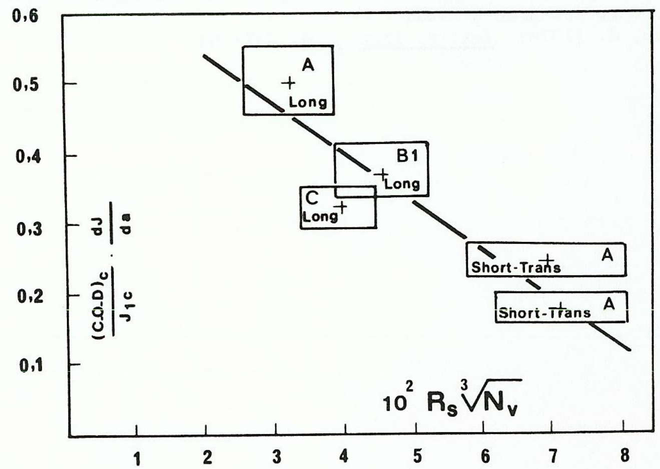

Fig. 19. Ductile crack extension in several heats of A 508 steel (Table 1).

CONCLUSIONS

Hopefully this paper has shown that much insight into the assessment of the fracture resistance of steels can be gained by the application of local micromechanism based criteria. It is clear that the relative simplicity of the criteria established for brittle fracture - either the concept of a critical stress over a characteristic distance or a modern form of the Weibull theory - may largely contribute to the understanding of the main macroscopic features of cleavage fracture. The practical implications of the local approach for the prediction of brittle fracture toughness are extensive since, with a limited number of experiments which are conducted not necessarily on cracked specimens, it is in theory possible to establish

not only the strain-rate and temperature dependence of the fracture toughness but also the scatter in the results, as shown in ferritic steels. Comparatively less progress has been made in the development of criteria for ductile rupture. The reasons for this slower development are twofold. The physical processes of ductile rupture are relatively more complex as compared to cleavage fracture. This paper seeks to underline that much insight into ductile crack initiation can be obtained by using a simplified form of critical void growth rate measured on bulk notched specimens in conjunction with the appraisal of the statistical distribution of inclusions. Future research in this area will be directed towards a full statistical approach to ductile rupture. In this respect, extensive examinations of sectionned specimens prior and immediately after ductile crack extension would seem to be very useful in order to assess the applicability of various failure criteria. The second main difficulty encountered in the study of ductile rupture of very tough materials is associated with the fact that most of the experiments involve large plasticity and large geometry changes. This situation greatly complicates the analysis since numerical or analytical solutions for the crack tip stress-strain field do not necessarily apply to the real crack tip behavior. In particular, most often, these solutions are based on the idealization of the stress state (plane stress or plane strain assumption) which do not necessarily represent the situation of real cracked specimens, such as widely used CT plate specimens. A proper analysis of these specimen geometries would necessitate at least three dimensionnal numerical calculations which are still very costly. It is therefore suggested that possible future studies in this field might be performed on specimen geometries susceptible of being calculated without too much simplification, such as axisymmetric cracked specimens.

ACKNOWLEDGEMENTS

The author wishes to thank all the members of Beremin group (Mr. Y. d'Escatha, Mr. P. Ledermann, Mr. J.C. Devaux, Mr. F. Mudry) for many useful discussions. The help of Dr. J.C. Lautridou, Mr. R. Locicero and Mrs Beaugendre in preparing the paper is gratefully acknowledged.

REFERENCES

Amazigo, J.C. and J.W. Hutchinson (1977). J. Mech. Phys. Solids, 25, 81-97.
Argon, A.S., J. IM, and Safoglu (1975). Met. Trans., 6A, 825-837. See also Argon, A.S. and J. IM (1975). Met. Trans., 6A, 839-851.
Ashby, M.F. (1966). Phil. Mag., 14, 1157-1178.
Beremin, F.M. (1979). Study of instability of growing cracks using damage functions. Application to warm prestress effect. C.S.N.I. Specialist Meeting on Plastic Tearing Instability Saint-Louis 25-27 Sept. 1979.
Beremin, F.M. (1980a). Numerical modelling of warm prestress effect using a damage function for cleavage fracture. This issue.
Beremin, F.M. (1980b). Experimental and numerical study of the different stages in ductile rupture. Application to crack initiation and stable crack growth. IUTAM Conf. Dourdan 2-5 June 1980.
Beremin, F.M. (1980c). Study of fracture criteria for ductile rupture of A 508 steel, this issue.
Berg, C.A. (1962). Proc. 4th U.S. National Congress of Applied Mechanics - University of California June 18-21, 1962. Edited by Rosenberg R.M., 2, 885
Berveiller, M. and A. Zaoui (1978). J. Mech. Phys. Solids, 26, 325-344.
Bevitt, E., A. Cowan, and A.L. Stott (1964). J. Brit. Nucl. Eng. Society, Jan. 1964.
Brothers, A.J. and S. Yukawa (1963). Trans. A.S.M.E. J. Basic Eng., 97-104.
Brozzo, P., G. Buzzichelli, A. Mascanzoni, M. Mirabile (1977). Metal Science, 11, 123-129.
Cialone, H. and R.J. Asaro (1979). Met. Trans., 10A, 367-375.

Cox, T.B., and J.R. Low (1974). Met. Trans., 5, 1457-1470.
Curry, D.A. (1978). Nature, 276, 50-51.
Curry, D.A., and J.F. Knott (1978). Metal Science, 12, 511-514.
Curry, D.A., J.E. King, and J.F. Knott (1978). Metal Science, 12, 247-250.
Curry, D.A. (1979a). Cleavage micromechanisms of crack extension in steels. Conf. "Mechanics and Physics of Fracture II, Cambridge, 1980" - C.E.G.B. Report RD/L/N 156/79, Dec. 1979.
Curry, D.A. (1979b). A micromechanistic approach to the warm pre-stressing of ferritic steels. C.E.R.L. Report N° RD/L/N 103/79, to appear in Int. J. Fract. Mech.
Curry, D.A., and J.F. Knott (1979). Metal Science, 13, 341-345.
Curry, D.A. (1980a). Mat. Sci. and Tech., 43, 135-144.
Curry, D.A. (1980b). Metal Science, 14, 78-80.
D'Escatha, Y., and J.C. Devaux (1979). Elastic-Plastic Fracture, ASTM. S.T.P. 668, 229-248.
Devaux, J.C. (1980). Eprouvette axisymétrique fissurée. Rapport Framatome TM/C DC/79.024.
Eshelby, J.D. (1961). Prog. Solid Mech., 11. Edited by I.N. Sneddon and R. Hill.
Floreen, S. and H.W. Hayden (1970). Scripta Met., 4, 87-94.
François, D. (1979). Micromechanisms of slow stable crack growth. Ispra Conf. 2nd advanced seminar on Fracture Mechanics, 2-6 April 1979.
French, I.E., and P.F. Weinrich (1974). Scripta Met., 8, 87-90.
Goods, S.H., and L.M. Brown (1979). Acta Met., 27, 1-15.
Green, G., and J.F. Knott (1976). Trans. AS.M.E. J. Eng. Mat. and Tech., 37-46.
Groom, J. D.G., and J.F. Knott (1975). Metal Science, 9, 390-400.
Gurland, J. (1972). Acta Met., 20, 735-741.
Gurson, A.L. (1977). J. Eng. Mat. Tech, 44, 2-15.
Hancock, J.W., and A.C. Mackenzie (1976). J. Mech. Phys. Solids, 24, 147-169.
Hill, R. (1950). The mathematical theory of plasticity. Edited by Clarendon Press, Oxford.
Houssin, B. (1977). Determination des caractéristiques de ténacité de l'acier SA 508 cl.3. Rapport Framatome TE/M. DC0259, Dec. 1977.
Hutchinson, J.W. (1968). J. Mech. Phys. Solids, 16, 13-31.
Knott, J.F. (1966). J.I.S.I., 204, 104-111.
Knott, J.M. (1967). J.I.S.I., 205, 966-969.
Knott, J.F. (1977). Fracture 1977. ICF4 Conf., 1, 61-92.
Lautridou, J.C. (1980). Etude de la déchirure ductile d'aciers à faible résistance. Influence de la teneur inclusionnaire. Thèse Ingénieur Docteur. Ecole des Mines.
Lemant, F., and A. Pineau (1980). Mixed mode fracture of a brittle orthotropic material. Example of strongly textured zinc sheets. To appear in Eng. Fract. Mech. 1980.
Levy, N., P.V. Marcal, W.J. Ostergren, and J.R. Rice (1971). Int. J. Fracture Mechanics, 7, 143-156.
Loss, F.J., R.A. Gray, and J.R. Hawthorne (1978). Nuclear Engineering and Design, 46, 395-408.
Mackenzie, A.C., J.W. Hancock, D.K. Brown (1977). Eng. Fract. Mech., 9, 167-188.
McClintock, F.A. (1968). J. App. Mech., 35, 363-371.
McMeeking, R.M. (1976). Finite deformation analysis of crack tip opening in elastic-plastic materials and implications for fracture initiation. Brown University Report COO-3084/44. See also J. Mech. Phys. Solids, (1977), 25, 357-381.
McMeeking, R.M., and D.M. Parks (1978). On criteria for J dominance of crack tip fields in large scale yielding. Standford University Report.
Mudry, F. (1980). Private communication.
Naylor, J.P. (1979). Met. Trans., 10A, 861-873.
Naylor, J.P., and R. Blondeau (1976). Met. Trans., 7A, 891-894.
Naylor, J.P., and P.R. Krahe (1975). Met. Trans., 6A, 594-598.
Parks, D.M. (1976). Trans. A.S.M.E. J. Eng. Mat. and Tech., 30-35.
Parks, Y.J., and I.M. Bernstein (1979). Met. Trans., 10A, 1653-1664.
Perra, M., and I. Finnie (1977). Fracture 1977. ICF4, 2, 415-423.

Rice, J.R., and M.A. Johnson (1970). The role of large crack tip geometry changes in plane strain fracture in "Inelastic behaviour of solids" Ed. McGraw-Hill, N.Y. 641-672.
Rice, J.R., and G.F. Rosengren (1968). J.Mech. Phys. Solids, 16, 1-12.
Rice, J.R., and E.P. Sorensen (1978). J. Mech. Phys. Solids, 26, 163-186.
Rice, J.R., and D.M. Tracey (1969). J. Mech. Phys. Solids, 17, 201-217.
Ritchie, R.O., J.F. Knott, and J.R. Rice (1973). J. Mech. Phys. Solids, 21, 395-410.
Ritchie, R.O., W.L. Server, and R.A. Wullaert (1979). Met. Trans., 10A, 1557-1570.
Rousselier, G. (1979). Contribution à l'étude de la rupture des métaux dans le domaine de l'élastoplasticité. Thèse Doctor-ès-Sciences. Ecole Polytechnique.
Server, W.L., and W. Oldfield (1978). EPRI Report N° NP-933. Electric Power Research Institute, Palo Alto, C.A. Dec. 1978.
Shih, C.F. (1974). Fracture Analysis. ASTM STP 560.187-210.
Shoemaker, A.K., and S.T. Rolfe (1971). Eng. Fract. Mech., 2, 319-339.
Slepyan, L.I. (1974). AN S.S.R. Mekhanika Tverdogo Tela, 9, 57-67.
Smith, E. (1966). Pro. Conf. Physical Basis of Yield and Fracture. Inst. Phys., Phys. Soc. Oxford, 36-45.
Steigerwald, E.A. (1961). Trans. A.S.M., 54, 445-455.
Stroh, A.N. (1954). Proc. Roy. Soc., A223, 404.
Tanaka, R. T. Mori, and T. Nakamura (1970). Phil. Mag., 21, 267-279.
Thompson, A.W. (1979). Met. Trans., 10A, 727-731.
Thompson, A.W., and J.C. Williams (1977). Fracture 1977. ICF4, 2, 343-348.
Tracey, D.M. (1976). Trans. ASME. J. Eng. Mat. Tech., 146-151.
Weibull, W. (1939). A statistical theory of the strength of Materials. Roy. Swed. Inst. Eng. Research N° 151.
Yamamoto, H. (1978). Int. J. Fract., 14, 347-364.