INTERRELATION AMONG MICROSTRUCTURE, CRACK-TIP BLUNTING, AND DUCTILE FRACTURE TOUGHNESS IN MILD STEELS

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ABSTRACT

After a review of various types of models proposed up to now to illustrate the nucleation of ductile fractures from sharp cracks, the mechanical and microstructural material parameters controlling the phenomenon are discussed at the light of the mechanisms of crack-tip blunting and void nucleation at second-phase particles that precede real crack advance. A link is established between sharp and blunt notch specimen behavior. Finally, a new mathematical model incorporating previously identified critical parameters is derived and verified by comparing calculated and experimental values of fracture toughness of a number of mild steels.

KEYWORDS

Ductile fracture nucleation mechanisms; microstructure; crack-tip blunting; effective root radius; blunt notch fracture criterion; inclusion content and distribution - influence on K_{Ic} values; fracture toughness mathematical models; fracture toughness of C-Mn, ferritic-pearlitic steels.

INTRODUCTION

Recently a review of plane strain ductile fracture toughness models has been compiled by the authors (Firrao and Roberti, 1982b). It is now summarized to demonstrate that for a full understanding of the process that leads to fully ductile fracture initiation and propagation in the triaxial stress field ahead of a sharp crack, it is needed to model completely the complex interplay between the metal-matrix flow and strength characteristics and the microstructural features that can act as bases for void formation at various levels of stress.

Several models have been developed on the basis that for ductile fracture to occur a critical strain has to be reached over a certain distance ahead of the crack-tip, as proposed by Rice and Johnson (1970). Such a distance is of the same order of magnitude as the critical crack-tip opening displacement (COD). The critical COD in plane strain, δ_{Ic}, can be correlated to the fracture toughness, K_{Ic}, via the usual correlation,

$$\delta_{Ic} = m \cdot K_{Ic}^2 / E \cdot \sigma_y \qquad (1)$$

where E is the Young's modulus, σ_y the yield strength and "m" a numerical constant varying between 0.425 and 0.717, according to various theories. The above distance is a function of the microstrucure of the material, i.e. of what has been later defined as a microstructural characteristic distance (Ritchie and Horn, 1978).

Hahn and Rosenfield (1975) assumed such a distance equal to the average spacing between large second-phase particles (s̲) and substituting s̲ for δ_{Ic} obtained the following formula, where m is assumed equal to 0.5,

$$K_{Ic} = \sqrt{(\sigma_y \cdot E \cdot \underline{s}/m)} \qquad (2)$$

The above way of reasoning had already been laid down in Krafft's (1964) earliest model which assumed the existence of a small "process" zone ahead of the crack-tip, with zone elements, idealized as circular tensile ligaments of diameter d_T, clinching the throat of the crack until they are drawn out to the point of the tensile instability, where their rupture became inevitable. The conseguing Krafft's equation read;

$$K_{Ic} = E \cdot N \sqrt{(2 \cdot \Pi \cdot d_T)} = E \cdot \varepsilon_i \sqrt{(2 \cdot \Pi \cdot d_T)} \qquad (3)$$

N is the strain-hardening exponent, considered equal to the uniform elongation strain at the instability of a tensile specimen, ε_i. d_T should be set equal to s̲, as also demonstrated later by Birkle, Wei and Pellissier (1966).

Other models emphasized that the fracture toughness of a material should be more strongly dependent on the plane strain ductility, $\varepsilon_{f,ps}$ (Clausing, 1969), than the axisimmetric ductility, since the former reflects more closely the stress state conditions at the crack-tip. Hahn and Rosenfield (1968) suggested that $\varepsilon_{f,ps}$ could be obtained from the true fracture strain in uniaxial tension, ε_f, setting $\varepsilon_{f,ps} = \varepsilon_f/3$. Hypothesizing a relationship between the critical COD and the plastic zone width, which was supposed to be a function of the square of the strain-hardening coefficient of the material, they proposed the semi-empirical equation,

$$K_{Ic} = N \cdot \sqrt{(2 \cdot E \cdot \sigma_y \cdot \varepsilon_f \cdot 0.0254/3)} \qquad (4)$$

An experimental correlation between the crack-tip strain, ε_t, and δ_i, the COD at onset of crack propagation was proposed by Smith and Knott (1971): $\varepsilon_i = \varepsilon_t \cdot \underline{l}$, where l̲ can be considered as the gauge length along the crack contour or notch-tip contour over which the strain can be considered approximatively constant; l̲ = ρ or l̲ = 1.2ρ in the case of a blunt notch with end radius equal to ρ (Griffiths and Owen, 1971). For sharp cracks in mild steels l̲ has to be set equal to the inclusion spacing, s̲ (Chipperfield and Knott, 1975) so that the following equations could be derived, upon conservatively setting δ_{Ic} equal to δ_i:

$$K_{Ic} = \sqrt{(2 \cdot \sigma_y \cdot E \cdot \varepsilon_t\ \underline{s})} = \sqrt{(2 \cdot \sigma_y \cdot E \cdot \varepsilon_{f,ps} \cdot \underline{s})} \qquad (5)$$

Another series of models take into account the blunting of the crack-tip upon loading. A numerical solution for the strain distribution in front of the blunted crack, coupled with the above reported approximate failure criterion by Rice and Johnson (1970), lead Schwalbe (1977) and Ritchie, Server and Wullaert (1979) to relationships of the type,

$$K_{Ic} = \text{constant} \cdot \sqrt{(\varepsilon_{f,ps} \cdot E \cdot \sigma_y \cdot h \cdot \underline{s})} \qquad (6)$$

with h assuming a value of the order of the unity for Schwalbe or a value that could reach 6 for the latter authors.

Other models make a linkage between the plastic zone and the applied stress intensity factor. Schwalbe (1974, 1977) hypothesizes a strain distribution within the plastic zone, analogous to the one proposed by Rice (1967) for shear strain upon Mode III loading; then setting again that the plane strain fracture strain has to be reached over a distance equal to the interparticle spacing, he proposes;

$$K_{Ic} = \frac{\sigma_y}{1-2\nu} \sqrt{\left[\underline{s}\, \Pi (1+N) \left[\frac{\varepsilon_{f,ps}}{\sigma_y} E \right]^{1+N} \right]} \qquad (7)$$

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The above reported formulas show that K_{Ic} has been linked to 4 tensile quantities (σ_y, E, N, and some type of fracture or instability strain) and to a microstructural parameter, s̲, although only Eq. 7 incorporates all of them. It has to be noted that calculations reported by Schwalbe (1977), as well as others that can be effected using data listed by him, show that the agreement between experimental values of fracture toughness and the ones that can be computed via the above listed formulas can be considered only fair in the best cases and decidedly bad in the others. A critical analysis of the role of the above parameters is then needed.

Obviously, the quantities whose values are most affected by error are s̲ and the value to be inserted in place of the crack-tip strain. Regarding the former, the uncertainty arises from the phenomenological observation that not all the second-phase particles within a material have associated voids at the same level of stress. It is generally accepted (Hahn and Rosenfield, 1975) that larger particles crack or decohere at lower levels of stress or strain than the intermediate or fine ones. Void formation at large inclusions is promoted by the triaxial state of stress that develops in plane strain conditions in the vicinity of the stress concentration. Then, the critical stages in fracture initiation when loading a pre-cracked specimen are (i) the linkage of the crack-tip with some closely located voids at different positions along the crack-front and (ii), most of all, the linkage of the crack-front protrusions generated in the above way. These coalescence process are the ones that control the real critical crack-tip advance and conversely COD at initiation.

Under the above assumptions, the resistance to the onset of crack-growth, once voids are generated, is controlled by the capacity of the matrix material between voids associated with large particles to further straining under in creasing stress. Such a resistance is enhanced by the ability of the material to strain-harden and is foreshortened by the tendency to strain localization in narrow slip bands, i.e. to shearing instability. These characteristics are not independent of one another (Spretnak and Firrao, 1980): materials with higher N values show a greater resistance to the development of shearing instabilities than those with lower associated N values. The matrix material between first generated voids comprises intermediate and fine particles which favour strain localization, and therefore tend to limit fracture deformation (Hahn and Rosenfield, 1975). Consequently, their role can be stated as largely detrimental to fracture toughness.

From what precedes, it can be inferred that the s value which controls fracture toughness is the one related to the spacing of large inclusions, especially in mild alloys where large plastic zones develop at the crack-tip. Instead the volume fraction of all the inclusions, large, intermediate, and fine determines the maximum strain that can be sustained in the region close to the crack-tip before the onset of critical growth.

Besides the inclusion volume fraction, the stress state acting at the crack-tip at the moment of fracture nucleation also controls such a limiting strain (Ritchie, Server, and Wullaert, 1979). For such a reason, the suggestion by Hahn and Rosenfield (1968) or Osborne and Embury (1973), that the strain to be taken into account is some fraction of the uniaxial fracture strain, can be considered as leading to only rough approximations. Better attempts to reproduce the stress state at a stress concentration are those made by the use of Clausing's bars (Ritchie and Horn, 1978; Schwalbe, 1977). However, it is to be noted that, with such a method, it is possible to duplicate the plain strain situation acting at the stress concentration without indeed copying the same small gauge length or the steep stress gradients encountered in actual blunted crack-tips. In fact, results in predicting experimental K_{Ic} values have met with limited success.

The best procedure envisioned up to now, to identify the strain acting at the crack-tip at the moment of fracture nucleation, is the one proposed by Ritchie, Server, and Wullaert (1979) by the use of circumferentially notched specimens, with various values of the notch-end radius, in order to determine the variation of the blunted crack-tip strainat fracture as a function of the stress state. Unfortunately, it is a lengthy one and, furthermore, the results reported by the authors indicate that a precise prediction of K_{Ic} is not always possible, owing to the lack of a clear-cut relationship between the microstructural characteristic distance and the interparticle spacing valid for all metal alloys.

CRACK-TIP BLUNTING AND FRACTURE TOUGHNESS IN DUCTILE RUPTURES

The researches performed in recent years on cracked specimens to obtain accurate variations of applied J-integral values as a function of crack extension (J_R R-curves) have demonstrated that the early stages of crack-growth stem from the blunting of the crack under increasing loads. True crack-propagation

intervenes only after the tip has reached a finite value of the radius of curvature, which will be hereafter defined as ρ_{eff}.

Chipperfield and Knott (1975) (Fig. 1), Lereim and Embury (1978), Roberti and collegues (1981) (Fig. 2) have reported that fracture toughness data obtained with notched specimens with ρ values smaller than a finite limiting radius were identical to those obtained employing pre-cracked samples. It was also seen that, prior to fracture advance, both types of specimens originate analogous "stretch zones" (Firrao and Roberti, 1983a). It was then inferred that either pre-cracked test-pieces or small notch samples reach the onset of crack advance by the same ductile rupture nucleation mechanism, i.e. in both types the radius of curvature at the tip of the stress concentration enlarges up to ρ_{eff} before fracture proceeds as a consequence of the maximum tip strain reaching the limiting value appropriate for the actual metal alloy. Therefore, ρ_{eff} can be identified as the maximum notch-end radius that causes a notched specimen to fail at the same level of fracture toughness as pre-cracked test-pieces of the same material.

Furthermore it has been demonstrated that blunt notch specimens of a given material with $\rho > \rho_{eff}$ reach a ductile fracture initiation stage because a constant limiting strain $\varepsilon^*_{max,f}$ is achieved at the notch root. In fact Firrao and others (1979, 1980, 1982) as well, as Roberti and co-workers (1981) have demonstrated that with these medium size notches applied J-integrals at onset of fracture vary linearly with ρ (e.g.,Fig. 2), thus proving the validity of a ductile fracture initiation model from blunt notches originally proposed by Begley, Logsdon, and Landes (1977). The model took into account the equation derived by Rice (1968) to relate the applied J-integral, the notch-end radius, and the strain hardening properties of a material to the maximum strain acting at the notch root, ε^*_{max};

$$\varepsilon^*_{max} = \varepsilon_y \left[\frac{(N+1/2)(N+3/2)\cdot\Gamma(N+1/2)}{\Gamma(1/2)\cdot\Gamma(N+1)} \frac{J}{\sigma_y \varepsilon_y \rho}\right]^{1/(1+N)} \quad (8)$$

where Γ is the mathematical gamma function. Introducing $\varepsilon^*_{max,f}$ in place of ε^*_{max} and indicating as $F(\Gamma(N))$ the first fraction in Eq. 15, one derives (Firrao and Roberti, 1982b),

$$J_A = \sigma_y^{(1-N)} \varepsilon^{(1+N)}_{*max,f} E^N \rho/F(\Gamma(N)) \quad (9)$$

with J_A being the J-integral applied at fracture initiation in a notched sample with $\rho > \rho_{eff}$. Eq. 9 either adequately interpret the resistance of blunt notch specimens to ductile rupture nucleation, or can be employed to calculate an accurate value of the maximum strain active at a notch root before the onset of crack growth. For instance, values of $\varepsilon^*_{max,f}$ equal to 0.991, 0.876 and 0.474 can be obtained for steels 1, 2, and 3 of Fig. 2 (Firrao and Roberti, 1983b). From what has been said before, when ρ reduces to ρ_{eff}, J_A coincides with J_{Ic}; it is then possible to derive that $J_A/J_{Ic} = \rho/\rho_{eff}$. The experiments carried out by the authors on low carbon, ferritic-pearlitic, C-Mn steels with different microstructures (Fig. 2) allow to prove that ρ_{eff} is of the same order as s, which is to be taken as the spacing between major inclusions in accordance with previous considerations. Thus, it is possible to write the relationship,

$$J_A/J_{Ic} = \rho/\underline{s}, \qquad (10)$$

Upon setting $J_A/J_{Ic} = \delta_A/\delta_{Ic}$ (δ_A is the value of COD at fracture initiation in notched specimens), it can be seen that also Fig. 1 proves such an equation; in fact, according to the authors of these experiments, $\underline{l}$ can be set on the average equal to 1.1ρ.

Eq. 10 can also be employed to calculate J_{Ic} after measuring $\underline{s}$ by metallographic observations and using J_A data obtained by performing J-integral tests on notched specimens with $\rho > \underline{s}$. Table 1 reports experimental and calculated values of J_{Ic} for the C-Mn steels of Figs. 1 and 2. In the case of the steels tested by Chipperfield and Knott (1975), computed values have been derived converting the δ_A/ρ quantities, determined averaging their results, into J_A/ρ values by the usual relationship, $J_A = 2\cdot\sigma_y\cdot\delta_A$. Results listed in Table 1 clearly indicate the close agreement between the values of fracture toughness calculated by the use of Eq. 10 and those measured on pre-cracked samples. Only in one case (steel D) is the difference rather high, which might be ascribed to the inexactness of the unique data (dimension of the sample insufficient to guarantee plain strain) from which δ_A/ρ was calculated. It is interesting to note that the here described ductile fracture nucleation mechanism and the mathematical model based on it yield satisfactory results also in the case of the steel T, where specimens differently oriented in respect to the rolling direction had been tested.

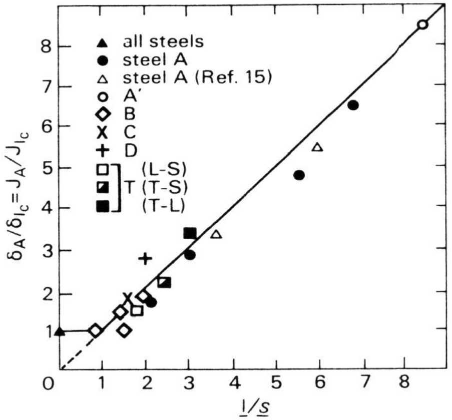

Figure 1 - Fracture toughness results by Chipperfield and Knott (1975) on a series of low C, Mn-steels. Data are presented as adimensioned plots of the δ_A/δ_{Ic} ratio vs $\underline{l}/\underline{s}$. δ_A is the COD at fracture initiation in a notched sample with notch-end radius ρ; $\underline{l} = 1.1\,\rho$.

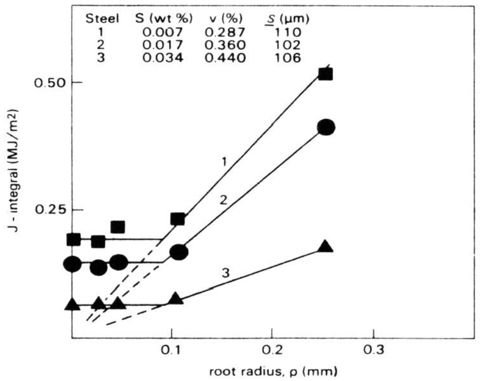

Figure 2 - Applied J-integral values at fracture initiation vs the notch root radius of 3-point bend specimens fabricated with 0.17 C-1.33% Mn steels with various sulphur and inclusion contents (v) and similar inclusion spacing ($\underline{s}$). Cracks and notches were in the TL direction. Roberti and coll. (1981)

TABLE 1 Comparison of Experimental and Calculated Values of the Fracture Toughness (J_{Ic}, MN/m) for the C-Mn steels of Fig. 1 and 2.

Steel	A	A'	B	C	D	T(T-S)	T(L-S)	T(T-L)	1	2	3
J_{Ic} exp.	0.020	0.011	0.029	0.148	0.154	0.064	0.119	0.041	0.194	0.145	0.06[illegible]
J_{Ic} calc.	0.020	0.012	0.028	0.172	0.262	0.067	0.128	0.051	0.229	0.163	0.07[illegible]

For the sake of comparison with equations 2 to 7, previously listed, Eqs. and 10 can be converted to yield K_{Ic} by the usual relationship $K_{Ic} = \sqrt{(E\cdot J_{Ic}/(1-\nu^2))}$,

$$K_{Ic} = \left[\frac{\sigma_y^{(1-N)}\cdot\varepsilon^{*\,(1+N)}_{f,max}\cdot E^{(1+N)}}{(1-\nu^2)\quad F(\Gamma(N))}\right]^{1/2} \underline{s}^{1/2} \qquad (11$$

Eq. 11 employs all the five mechanical and microstructural quantities tha enter with various arrangements in the previously examined equations. Althoug only one microstructural parameter, $\underline{s}$, is directly indicated, the strong in verse dependance of $\varepsilon^*_{f,max}$ on v is evident from data previously reported f steels 1, 2, and 3 of Fig. 2. Therefore, it can be stated that the fractur toughness (K_{Ic} or J_{Ic}) in ductile ruptures is controlled both by the spacin between large inclusions and by the total second phase volume fraction, where as K_A or J_A values for blunt notch specimens depend mainly on v.

CONCLUSIONS

A mechanism of ductile fracture nucleation ahead of sharp cracks has been ra tionalized and a mathematical model derived. The proposed mechanism hypothe sizes that crack-tip blunts, up to achieving a finite radius, ρ_{eff}, which i of the same order of magnitude as the spacing between major non-metallic in clusions, $\underline{s}$. Then the onset of crack advance intervenes in the same way as i a specimen with a medium size blunt notch, i.e. when the maximum strain at th root of the notch reaches a limiting value, $\varepsilon^*_{f,max}$, which is inversely pr portional to the total inclusion volume fraction.

Consequently, a procedure to derive J_{Ic} data from the values of the applie J-integral at fracture nucleation in blunt notch specimens with notch-en radii greater than $\underline{s}$ has been devised and applied to compare calculated an experimental J_{Ic} pertaining to a number of low C, ferritic-pearlitic steels.

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