On the Characteristic Distance for Fracture in AISI 4140 Alloy Steel

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ABSTRACT

Using previously reported data on the variation with notch root radius of the toughness of AISI 4140 steel tested in different microstructural conditions, it is seen that the observed sharp crack toughness values can be rationalized in terms of the characteristic distance for the operating fracture mode. Critical fracture stress values, estimated from stress controlled fracture model and seen to be consistent with microstructural aspects pertinent to failure initiation, indicate that the argument proposed by Datta (1981) in that plane strain toughness improvement is to be attributed to an increase in the local (microscopic) stress level required for fracture initiation ahead of a sharp crak is not invariably valid. Finally, the concept of characteristic distance is used to explain why in some cases, as observed experimentally for AISI 4140 steel, a change from ductile to brittle fracture initiation mode can indeed be accompanied by an improvement in sharp crack toughness.

KEYWORDS

Characteristic distance, fracture initiation, ductile failure, brittle fracture, sharp crack toughness, notch root radius, critical fracture stress, critical fracture strain.

INTRODUCTION

Models introduced for the prediction of the fracture toughness, K_{Ic}, of a given material under different service conditions are generally dependent for their application on knowledge of the characteristic distance for the operating fracture mode. This concept is based on that a critical stress σ_f for stress controlled fracture, or a critical strain ε_f for strain controlled fracture, has to be exceeded over some distance ahead of the tip of a sharp crack before the conditions for fracture are fulfilled.

In the presence of a rounded notch, the fracture toughness (termed apparent toughness, K_A) would depend on the notch root radius, ρ. In fact critical fracture stress and critical fracture strain models (Malkin and Tetelman, 1971; Ritchie, Francis and Server, 1976; Ritchie and Horn, 1978) predict a linear relationship between K_A and $\rho^{1/2}$, as shown below by expressions (1) and (2) for stress controlled and strain controlled failure, respectively.

$$K_A \approx 2{,}9\ \sigma_y\ [\exp\ (\sigma_f/\sigma_y - 1)]^{1/2}\ \rho^{1/2} \qquad \rho \geq \rho_{eff} \qquad (1)$$

and

$$K_A \approx (\tfrac{3}{2}\ \sigma_y\ E\ \varepsilon_f)^{1/2}\ \rho^{1/2} \qquad \rho \geq \rho_{eff} \qquad (2)$$

where σ_y is the yield limit and E the modulus of elasticity. The parameter ρ_{eff}, termed the effective or limiting root radius, represents a measure of the characteristic distance and is defined experimentally as the notch root radius below which the apparent toughness, K_A, remains constant and equal to K_{Ic} (Malkin and Tetelman, 1971; Ritchie, Francis and Server, 1976; Ritchie and Horn, 1978). Accordingly K_{Ic} is obtained by replacing ρ in the above expressions by the value of ρ_{eff} corresponding to the operating fracture mode.

Simple models such as those represented by the aforementioned equations were applied by Malkin and Tetelman (1971) and by Ritchie, Server and Wullaert (1979) to obtain quantitative predictions of the variation with service conditions of the plane strain fracture toughness, K_{Ic}, in nuclear reactor pressure vessel steels. The application of equations (1) and (2) has also been successful in explaining why higher austenitizing temperatures result in higher K_{Ic} and lower K_A values in as-quenched and quenched and tempered high strength low alloy steels (Graça, Darwish and Pereira, 1984; Ritchie, Francis and Server, 1976; Ritchie and Horn, 1978). The increase in K_{Ic} with increasing austenitizing temperature was considered to be associated with an increase in the characteristic distance for fracture, whereas the decrease in rounded notch toughness, K_A, was interpreted in terms of a reduction in σ_f or ε_f depending on the operating fracture mode.

Despite the proven applicability of these models to a variety of problems, some criticism has been raised against the concept of characteristic distance in that it cannot be directly correlated with microstructural aspects of the material, especially if the detailed local mechanisms of critical failure events are not well characterized. In a model proposed by Ritchie, Knott and Rice (1973), a distance of two ferritic grain diameters resulted in good agreement between predicted and experimentally determined cleavage fracture toughness values in mild steel. However, no fundamental physical significance could be attached to defining the characteristic distance for cleavage as two ferritic grain diameters. For ductile failure initiation, a lower bound on the dimension taken to represent the characteristic distance is obviously equivalent to the spacing between the void initiating particles. In certain instances, though, the characteristic distance is estimated as a multiple of that spacing, meaning that the coalescence of several voids represents the critical event responsible for ductile fracture initiation. In trying to best fit experimental toughness data for nuclear pressure vessel steels with theoretical predictions from models compatible with the operating fracture mechanism, Ritchie, Server and Wullaert (1979) used various estimates of the characteristic distance for a given initiation mode. It is thus concluded that the dimension taken to represent the characteristic distance must be regarded as essentially an empirical quantity presumably of relevance to microstructural aspects of fracture initiation.

More recently, various arguments against the concept of characteristic distance were raised by Datta (1981). He argues that the improvement in sharp crack toughness associated with high temperature austenitization of AISI 4340 steel tested in the as-quenched condition could not be attributed to an increase in the characteristic distance. Rather, the improvement in K_{Ic} can be explained by the presence of substantial amounts of autotempered ε carbide particles, thereby causing dispersion strengthening and hence raising the microscopic stress level required to create a microcrack (that is, to initiate fracture) ahead of the sharp crack tip. Both the improvement in toughness and the presence of fine ε carbide precipitates, brought about by high temperature austenitizing, were considered consistent with the predominance of a ductile failure initiation mode in as-quenched test specimens. For conventional austenitizing (870°C), on the other hand, a quasi-cleavage-intergranular cracking initiation mode was found to prevail in precracked specimens tested in the as-quenched condition. This finding was attributed by Datta (1981) to the lack of ε carbide particles in significant amounts, which, in turn, implies in a lower microscopic stress level for fracture initiation and hence in a lower K_{Ic}, compared with the case of high temperature austenitization. Based on the foregoing arguments Datta (1981) concludes that in sharp crack testing the grain size does not affect the toughness level, whereas in the presence of a blunt notch, the grain size does play an important role in defining apparent toughness values. Here the smaller grain size obtained in the conventional austenitizing treatment invariably results in a superior toughness (K_A) in virtue of the large number of grain boundaries a microcrack formed ahead of the notch, away from its tip, has to cross to join the notch root.

The purpose of this paper is to discuss the validity of the characteristic distance concept, using previously reported experimental data (Graça, Darwish and Pereira, 1984) on the variation of the apparent toughness with notch root radius in commercial AISI 4140 alloy steel austenitized at 870 and 1200°C and tested at ambient temperature in the as-quenched and quenched and tempered at 200 and 350°C conditions. Both sharp crack and rounded notch toughness values are presented and discussed in light of Datta's arguments.

EXPERIMENTAL DATA

Fractographic studies have indicated that fracture occurs primarily by intergranular cracking, in AISI 4140 steel specimens tested in the 870→350 and 1200→350 conditions (referring to 870 and 1200 austenitization followed by quenching and 350°C tempering). Specimens austenitized at 1200°C and tested in the as-quenched condition (1200→Q) have been found to fail by a mixture of intergranular cracking and quasicleavage. Based on these fractographic observations, it is concluded that fracture initiation in these three microstructural conditions can be considered to be stress controlled. On the other hand, fracture is considered to be essentially strain controlled, in conventionally austenitized specimens tested in the as-quenched (870→Q) and quenched and low temperature tempered (870→200) conditions, since failure in these two cases is shown to initiate predominantely by ductile rupture. Finally, high temperature austenitized-low temperature tempered specimens (1200→200) are seen to display a mixture of ductile rupture, quasicleavage and intergranular cracking. Fracture surfaces as observed by scanning electron microscopy are shown in Fig. 1, for the different microstructural conditions considered in this work.

Once the microscopic fracture mechanism is known, the linear relationship between K_A and $\rho^{1/2}$, determined for a given microstructural condition, can be used to obtain σ_f or ε_f depending on the predominant initiation mode,

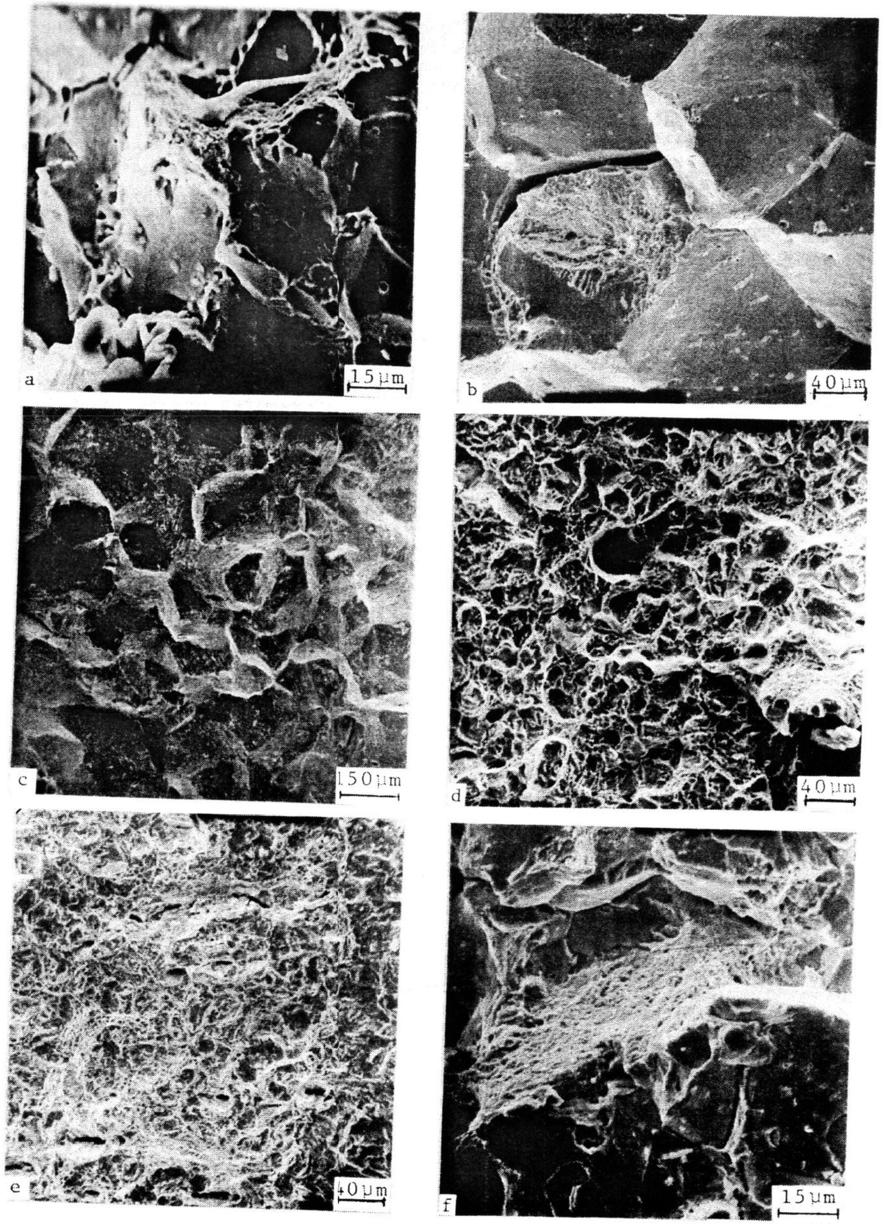

Fig. 1. SEM of fracture surfaces for, (a) 870→350, (b) 1200→350, (c) 1200→Q, (d) 870→Q, (e) 870→200, and (f) 1200→200 conditions.

and from knowledge of K_{Ic} one can estimate ρ_{eff}. The values of σ_f, ε_f and ρ_{eff} reported by Graça, Darwish and Pereira (1984) are presented in Table 1, together with K_{Ic} and K_A (for ρ = 0.25 mm) toughness levels.

Table 1. K_{Ic}, K_A, σ_f, ε_f and ρ_{eff} for the different heat treatments

Heat treatment	K_{Ic} (MPa$\sqrt{m}$)	K_A (MPa$\sqrt{m}$)	σ_f (MPa)	ε_f (%)	ρ_{eff} (µm)
870→Q	57±1	113±4	-	15.8	51
870→200	70±2	129±1	-	27.7	43
870→350	52±1	102±3	3,677	-	39
1200→Q	71±4	85±2	2,565	-	188
1200→200	74±4	124±3	-	-	91
1200→350	56±2	68±3	2,183	-	196

DISCUSSION

The effect of heat treatment on the sharp crack toughness of the steel considered in this study is summarized in Fig. 2, where it is observed that an

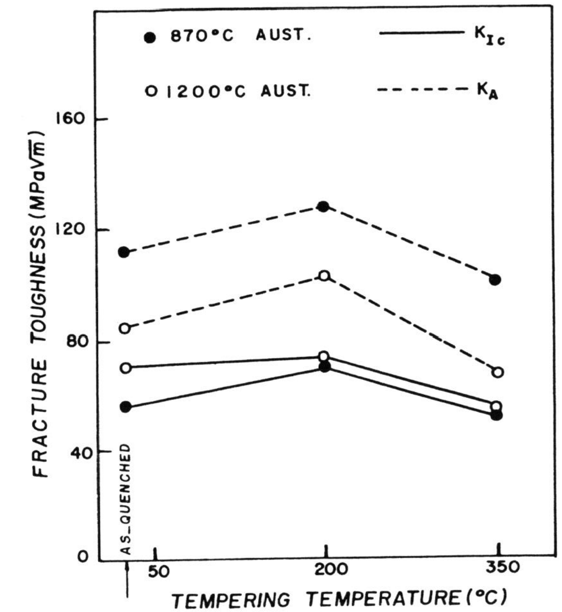

Fig. 2. Variation of K_{Ic} and K_A (measured for ρ = 0.25 mm) of AISI 4140 steel with heat treatment conditions.

increased K_{Ic} value for as-quenched specimens is indeed associated with high temperature austenitization. However, when the steel is quenched and tempered, the improvement in toughness, brought about by increasing the austenitizing temperature, becomes less significant. A loss in toughness is observed for the material tempered at 350°C, consistent with the finding by Wood (1975) that AISI 4130 and 4140 steels are prone to tempered martensite embrittlement (TME) when heated in the 280→350°C temperature range. Figure 2 also depicts the variation with heat treatment of rounded notch toughness, representing K_A values measured for ρ = 0.25 mm. Here one can observe that lower austenitizing temperature results in a higher apparent toughness and that the 350°C tempering treatment causes a considerable loss in K_A.

Effect of the 350°C Tempering Treatment

The loss in toughness observed for the steel tempered at 350°C indicates that TME has in fact occurred as a result of this type of heat treatment. However, based on the percentage loss, the effect is seen to be less pronounced for conventional austenitizing, consistent with the σ_f values estimated from the critical fracture stress model (Table 1). Based on the predominance of an intergranular cracking failure mode in specimens tempered at 350°C (Fig. 1), Graça, Darwish and Pereira (1984) attribuited the observed TME to embrittlement of prior austenite grain boundaries. This conclusion is borne out by the fractographic evidence (Fig. 1) that intergranular cleavage facets correlate well with the prior austenite grain size, estimated by Graça, Darwish and Pereira (1984) at 32 and 157 μm, for 870 and 1200°C austenitization, respectively.

Now if one considers the K_{Ic} values listed in Table 1, it is noticed that the sharp crack toughness resulting from the 350°C tempering is slightly higher (about 8%) for the 1200°C austenitizing treatment compared to low temperature austenitization. According to Datta (1981), this observation should imply in a (slightly) lower local stress level at the moment of fracture initiation ahead of the sharp crack tip for the conventionally austenitized specimens, which is seen to be inconsistent with the σ_f levels reported in Table 1. In fact, microstructural variations brought about by the heat treatment cycle (austenitizing, quenching and tempering) were shown to lead to more pronounced TME for the larger grain sized material (austenitized at 1200°C) due to more effective weakening of the prior austenite grain boundaries (Banerji, McMahon and Feng, 1978; Briant and Banerji, 1979; Graça, Darwish and Pereira, 1984; Horn and Ritchie, 1978; Materkowski and Krauss, 1979; Sastry and Wood, 1980; Ustinovshchikov, 1983), consistent with the σ_f levels estimated from the critical fracture stress model. The observed K_{Ic} values resulting from the 350°C tempering treatment are thus best rationalized on the basis of the characteristic distance approach. In this respect, one may mention the ρ_{eff} values of 39 and 196 μm which result from conventional and high temperature austenitizing, respectively and which correlate reasonably well with the respective prior austenite grain sizes.

As-Quenched Specimens

Specimens tested in the 1200→Q condition were characterized by quasicleavage and intergranular fracture appearance, indicating some form of embrittlement due to microstructural changes occurring during austenitizing and probably during mechanical testing (Graça, Darwish and Pereira, 1984). For conventional austenitizing, on the other hand, fracture initiation in as-quenched specimens was found to proceed essentially by ductile failure. The change from ductile to brittle initiation mode, brought about by increasing the austenitizing temperature from 870 to 1200°C, is seen to be accompanied by an improvement of about 25% in sharp crack toughness. That is, the predominance of a ductile failure mechanism does not necessarily imply in higher K_{Ic} level and hence cannot be used as an argument in itself to explain improvement in sharp crack toughness. It is thus concluded that despite the embrittling effect associated with high temperature austenitizing, the accompanying coarsening of the microstructure, leading to a considerably larger characteristic distance (ρ_{eff} = 188 μm) for the operating fracture mode, seems to be responsible for the significant improvement in toughness level over that observed for ductile failure in 870→Q specimens. Only in the 1200→350 heat treatment condition did the steel's sharp crack toughness, sufficiently degraded by TME, become as low as that determined for the 870→Q specimens (Table 1).

Effect of Low Temperature (200°C) Tempering

Low temperature tempering of quenched specimens results in the formation of fine carbide precipitates (essentially ε carbide), which according to Datta's arguments (Datta, 1981) should be accompanied by an increase in both sharp crack and rounded notch toughness levels. However, for high temperature austenitizing, it is seen from Fig. 2 that while K_A did in fact improve considerably as a result of low temperature tempering, K_{Ic}, on the other hand, remained essentially unaltered. Now if one considers the data reported in Table 1, it may be concluded that since fracture initiation in 1200→200 specimens proceeds by a mixture of ductile rupture, quasicleavage and intergranular cracking, the corresponding ρ_{eff} value probably reflects an average between the limiting root radii for a stress controlled and strain controlled failure mechanisms (Graça, Darwish and Pereira, 1984). Thus the reduction in ρ_{eff} resulting from low temperature tempering can be considered responsible for maintaining K_{Ic} essentially at the same level encountered for as-quenched specimens despite ε carbide precipitation during tempering.

For specimens tested in the 870→200 condition, both K_{Ic} and K_A levels were found to be higher than those obtained for as-quenched specimens (870→Q) as a result of the increase in ε_f that accompanies the transfer of carbon from the dislocations, during low temperature tempering, to form fine carbide precipitates. This microstructural change is seen to be accompanied by a decrease in ρ_{eff} for ductile failure (Table 1). However this reduction in characteristic distance is more than compensated by the significant increase in ε_f and K_{Ic} is found to exhibit higher level for 870→200 specimens compared to that observed for as-quenched specimens (870→Q).

CONCLUSIONS

The characteristic distance concept seems to be appropriate for rationalizing plane strain fracture toughness values determined for AISI 4140 steel in different microstructural conditions.

Higher sharp crack toughness is not necessarily associated with higher local stress level at the moment of fracture initiation ahead of the sharp crack tip.

A change in initiation mode from ductile failure to brittle fracture does not necessarily lead to an improvement in sharp crack toughness. Thus in defining K_{Ic} level, one should take into account the characteristic distance for the operating fracture mode.

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