THE DUCTILE FRACTURE OF HIGH-STRENGTH STEELS

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ABSTRACT

Air-melted and vacuum-remelted versions of En25 steel were tempered to give yield stresses in the range 950-1650 MPa, and initiation crack opening displacements were measured using a number of different techniques. These are compared and evaluated. It was found that vacuum-remelting En25 did not significantly improve its fracture toughness. It is suggested that, for the strength-range investigated, the tempered carbides play a more significant role than the non-metallic inclusions in determining the fracture toughness.

KEYWORDS

High-strength steels; vacuum-remelting; ductile fracture; "zig-zag" fracture; fracture toughness; crack opening displacement; crack replication; stretch zone.

INTRODUCTION

Non-metallic inclusions have been shown to have a major influence on the fracture toughness of structural steels (Green and Knott, 1976), but their effect on the ductile fracture of high-strength steels is less well established. It has, indeed, been suggested that the fracture toughness of rather clean, high-strength materials does not depend to any great extent on the distribution of inclusions (Smith, Cook and Rau, 1977). Since the modification of the inclusion content of steels by vacuum-remelting or electro-slag-refining is expensive, it is of considerable commercial interest to quantify the effect of inclusions on fracture in high-strength steels.

MATERIALS AND SPECIMEN PREPARATION

The steel chosen for study was En25. Two versions were examined: an air-melted condition (En25A) and a vacuum-remelted condition (En25V), whose chemical compositions are given in Table 1. Both versions were supplied as 29mm diameter bar, and were forged down to 25mm square bar, from which single-edge-notch bend specimens, measuring 20 x 20 x 100mm, were machined. The En25 was then austenitized for 4 hours at 950^{o}C, oil-quenched, and tempered for 2 hours at a temperature in the range 150-600^{o}C, followed by air-cooling.

To enable comparisons to be made between high- and low-strength steels, single-edge-notch bend specimens, measuring 24 x 24 x 101.6mm, were machined out of 25.4 x 101.6mm mild steel plate. The chemical composition of the mild steel is given in Table 1. The mild steel was then austenitized for 2½ hours at 900°C, water-quenched, and then spheroidized by holding at 700°C for 24 hours.

TABLE 1 Chemical Analysis of Steels

	Weight % of Alloying Element							
	C	Mn	P	S	Si	Ni	Cr	Mo
En25A	0.34	0.72	0.012	0.012	0.19	2.72	0.62	0.55
En25V	0.34	0.62	0.008	0.009	0.32	2.76	0.75	0.59
Mild Steel	0.18	0.81	0.009	0.019	0.28	-	-	-

TESTING METHODS

The initiation crack opening displacement (δ_i) was used as the fracture criterion, and the specimens were tested in general accordance with BS 5762 (British Standards Institution, 1979), with the exception that the mild steel specimens were tested in 4-point-bend, as permitted in DD 19 (British Standards Institution, 1972). It was thought sensible to validate the formulae used to calculate δ from the clip-gauge displacement (V).

From DD 19:
$$\delta = \frac{V}{1 + \dfrac{(a+z)}{r_1(W-a)}} \qquad (1)$$

From BS 5762:
$$\delta = \frac{K^2(1-\nu^2)}{m\sigma_y E} + \frac{V_p}{1 + \dfrac{(a+z)}{r_2(W-a)}} \qquad (2)$$

where a is the crack length, W the specimen width, z the knife-edge height, K the stress intensity factor, ν Poisson's ratio, m is a constant, σ_y the yield stress, E Young's modulus and V_p the clip-gauge displacement associated with the plastic deformation of the specimen. The first term on the right hand side of equation (2) gives the component of δ arising from the elastic deformation of the specimen, and the second term gives that arising from the plastic deformation. It has been shown that m usually takes a value in the range 1½-3 (Rice and Sorensen, 1978). BS 5762 (British Standards Institution, 1979) recommends that m be taken as 2. The plastic deformation is modelled as a rigid-body rotation around a "plastic-hinge" centred at a distance r(W-a) ahead of the crack-tip (Ingham and co-workers, 1971). Equation (1) assumes that all the specimen's deformation may be approximated by the "plastic-hinge" model. DD 19 and BS 5762 (British Standards Institution, 1972 and 1979) recommend that r_1 and r_2 be taken as 0.33 and 0.4 respectively. In order to validate these formulae and to determine m and r values experimentally, a more direct method of measuring δ is required.

To provide direct calibration, cracks in some specimens were infiltrated with a hardening silicone rubber (Robinson and Tetelman, 1973) whilst holding the specimens on-load at the desired clip-gauge displacement. After allowing the silicone rubber to harden and replicate the crack, the specimens were broken open at -196°C. It was not feasible to use this technique for steels having a yield stress greater than 1050 MPa because the cracks grew in an unstable fashion before the rubber could harden. Each replica was sectioned in several places to reveal the crack-tip

in profile, and the sections were photographed to enable the crack opening displacement to be measured (see Fig.1). The final, on-load, crack opening displacement of each specimen was thus obtained and the amount of fibrous growth was measured on the fracture surface. The initiation crack opening displacement was then determined by extrapolating a plot of δ against crack extension to zero fibrous crack growth (British Standards Institution, 1979).

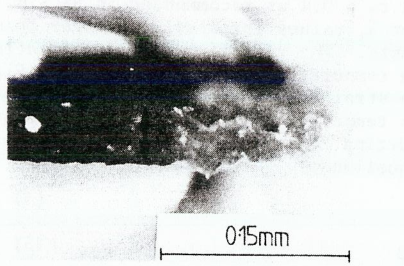

Fig.1. A section of a silicone rubber replica of the crack-tip region, showing how δ is measured at the fatigue pre-crack-tip.

For the mild steel, it was found that both equations (1) and (2) adequately described the relationship between V and δ. The rotational constants, r_1 and r_2, were determined to be 0.31 and 0.30 respectively and the best value for m was found to be 1.4. For the En25 with ∿ 1000 MPa yield stress, equation (1) described the relationship between V and δ poorly, but, with m = 2.9 and r_2 = 1.28, a good approximation to the replica crack opening displacement values is obtained using equation (2). Despite the very high value of r_2 (which places the centre of the "plastic hinge" outside the specimen), the assumption that r_2 = 0.4 and m = 2 (British Standards Institution, 1979) leads to only an 18% underestimation of δ_i. This is due to the fact that the errors in the assumed values of m and r_2 are, to a certain extent, self-compensating.

The steels with yield stresses greater than 1300 MPa gave valid plane strain fracture toughness (K_{IC}) values, as defined by BS 5447 (British Standards Institution, 1977). It was also possible to measure the stretch zone width (SZW) and use this to calculate the initiation crack opening displacement (Broek, 1974). Assuming a semicircular crack-tip profile (see fig.2), δ_i = 2x(SZW). Values of δ_i thus obtained were compared with the K_{IC} values:

$$\frac{K_{IC}^2(1-\nu^2)}{E} = m\,\sigma_y\,\delta_i \qquad (3)$$

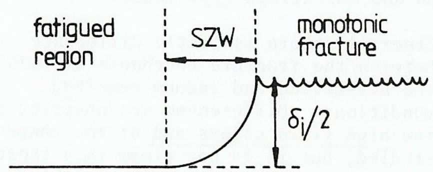

Fig.2. The fracture surface profile, showing the relationship between the stretch zone width (SZW) and the initiation crack opening displacement (δ_i).

The value of m was found to be 1.8. This is of a comparable magnitude to those calculated theoretically (Rice and Sorensen, 1978).

FRACTURE TOUGHNESS RESULTS

For comparison purposes it was decided, wherever possible, to use δ_i values calculated by equation (2) with m = 2 and r_2 = 0.4 as recommended in BS 5762 (British Standards Institution, 1979), but δ_i values for steels stronger than 1300 MPa were obtained from SZW measurements. The variation of the initiation crack opening displacement with tempering temperature is shown in Fig.3. Figure 4 shows how the yield stress (σ_y), the true strain to fracture (ε_f) and the work hardening exponents (n_1 and n_2) vary with tempering temperature. The mild steel results have been plotted at the spheroidizing temperature. The work hardening exponents were obtained by fitting the equations

$$\sigma = k_1 \, \varepsilon^{n_1} \tag{4}$$

and

$$(\sigma - \sigma_y) = k_2 \, (\varepsilon - \varepsilon_y)^{n_2} \tag{5}$$

to the true stress/true strain curves of cylindrical tensile specimens. σ and ε are the true stress and true strain respectively, ε_y is the true strain at yielding, and k_1 and k_2 are constants.

Hahn and Rosenfield (1968) proposed a model which predicts that

$$K_{IC} = (A \tfrac{2}{3} E \, \sigma_y \, n_1^{\,2} \, \varepsilon_f)^{\frac{1}{2}} \tag{6}$$

where A is 25.4mm. The predictions of this model have been shown, in several cases, to correspond with experimental fracture toughness data (Hahn and Rosenfield, 1968; Garrett and Knott, 1976). From equation (3) with m = 2 it follows that the predicted relationship between the initiation crack opening displacement and the tensile properties is

$$\delta_i = A \, \frac{(1-\nu^2)}{3} \, n_1^{\,2} \, \varepsilon_f \tag{7}$$

It can be seen in Fig.5 that the correct trend is predicted only for the 500-600°C range of tempering temperatures. The value of δ_i seems to relate more simply to n_2 than to n_1, but the dependence of the plastic zone width on n_2 needs to be established before n_2 may be used in a Hahn and Rosenfield-type model.

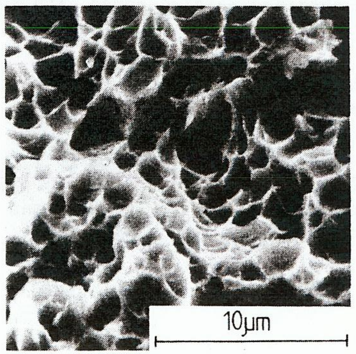

Fig.6. The dimpled fracture surface of the En25V tempered at 150°C.

Generally there is little difference between the fracture toughness of En25 in the air-melted and vacuum-remelted conditions. Differences are observed at the high yield stress end of the range studied, but it is not clear that these differences are attributable to the different inclusion contents. The dimples, which completely cover the fracture surface (see Fig.6), are separated by distances which are much smaller than the inclusion spacings, but could correspond to a "microstructural element size" (Schwalbe and Backfisch, 1977). The En25A is tougher, has larger dimples, and has a coarser microstructure than the En25V.

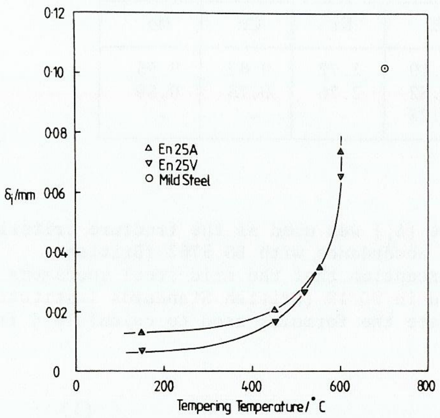

Fig.3. The variation of the initiation crack opening displacement (δ_i) with tempering temperature.

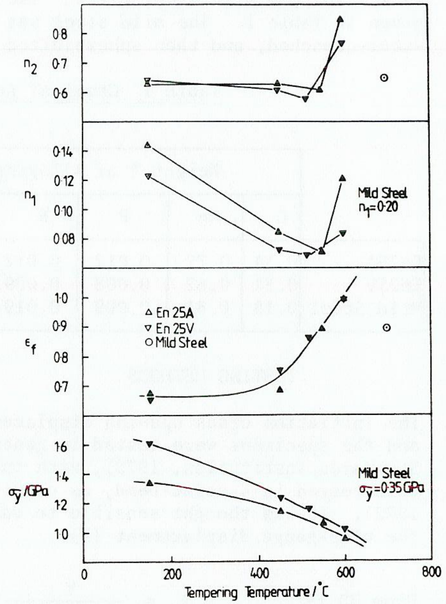

Fig.4. The variation of the tensile properties with tempering temperature.

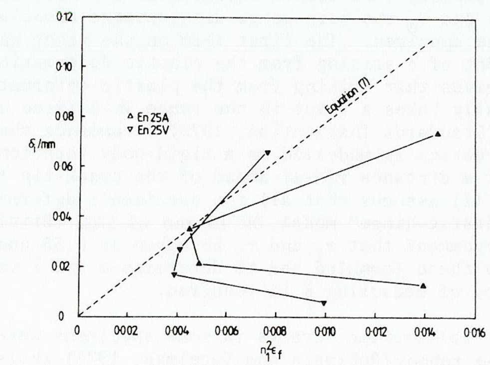

Fig.5. A comparison of the Hahn and Rosenfield (1968) prediction (dotted line) with the experimental data.

For lower yield stresses, the difference in δ_i between the two steels becomes insignificant. The fracture path "zig-zags" much more (see Fig.7) and, in addition to dimples, there are large, relatively featureless areas on the fracture surfaces (see Fig.8). The featureless areas probably correspond to the fine cracks seen just ahead of the main crack-tip in the two lowest strength En25 steels (see Fig.9). Fine (non-crystallographic) cracks were also seen in the spheroidized mild steel and are apparently characteristic of an important fracture mechanism. In the En25, there was some evidence that the carbides nucleate voids (see Fig.10) and that the fine cracks may advance by void coalescence along microstructurally pre-determined paths. Whatever the precise fracture mechanism, it is apparent that some form of interaction occurs between the fine cracks and the carbides.

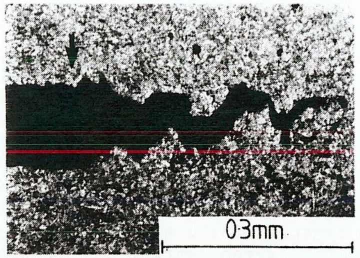

Fig.7. The "zig-zag" fracture path of the ductile crack in En25V tempered at 600°C. The position of the fatigue precrack-tip is arrowed.

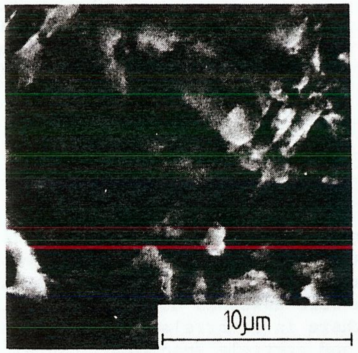

Fig.8. A relatively featureless area on the fracture surface of the En25A tempered at 600°C.

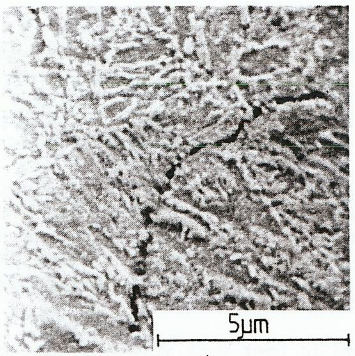

Fig.9. A fine crack, just ahead of the main crack in En25A tempered at 600°C.

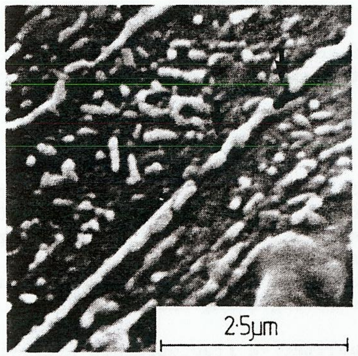

Fig.10. The arrows indicate voids nucleated at inter-lath carbides, seen just below the fracture surface of En25A tempered at 600°C.

CONCLUSIONS

The vacuum-remelting of En25 has little effect on its initiation crack opening displacement, and a more promising approach to attain tougher high-strength steels would seem to be the modification of the carbides. Studies of the effect of electro-slag-refining on the fracture toughness of a similar steel are continuing, extending the methodology to include the J-integral fracture criterion.

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