THE EFFECT OF PACKET BOUNDARIES ON THE FRACTURE TOUGHNESS OF A BAINITIC MICROSTRUCTURE

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ABSTRACT

A Cr-Mo-V alloyed pressure vessel steel has been investigated in a quenched and tempered heat treatment condition. The microstructure is fully bainitic with high dislocation density stabilized with small vanadium carbides. The fracture facet is of the same size as the cross-section of the bainite packet. The characteristic distance and the limiting notch root radius are several times larger than the fracture facet or the bainite packet but, however, definitely smaller than the prior austenite grain size. Micrographs taken from the broken fracture toughness specimens showed that microcracks extend from packet boundary to boundary on {001} type planes. Therefore the microcracks are parallel in the packets which have the same orientation. Microcracks can initiate at discontinuities like carbides lying on the high angle packet boundaries. The propagating crack changes its direction at the packet boundary, and the microcracking and crack propagation are controlled by the packet boundaries. The mechanism of crack formation in bainite is discussed.

KEY WORDS

Bainite; packet boundaries; microcracking; crack propagation; fracture toughness; characteristic distance; fracture facet.

INTRODUCTION

There have been many attempts to investigate the morphology of the cleavage fracture in complex bainitic microstructures (Naylor and Krahe, 1975; Naylor and Blondeau, 1976; Naylor, 1979; Kotilainen and Torronen, 1977). It has been shown that the habit plane of a cleavage crack is of the {001} type in polycrystalline pure iron and iron single crystals. There is evidence that the fracture planes also in bainite are {001} planes (Terasaki and Ohtani, 1972; Matsuda and co-workers, 1972; Lonsdale and Flewitt, 1978). Indirectly, it has been also indicated that some other low index planes can act as cleavage planes (Lindborg and Averbach, 1966; Naylor and Krahe, 1975).

The significant role of the sub-boundaries inside the previous austenite grains, i.e. the bainitic packet boundaries, on the fracture process has been stressed (Roberts, 1970; Naylor and Krahe, 1975; Kamada and co-workers, 1976; Kotilainen and Torronen, 1977; Roman and co-workers, 1979). The small angle boundaries, i.e. lath boundaries do not influence the crack propagation because of the small change of the direction of the cleavage planes at such a boundary (Naylor and Krahe, 1975; Naylor, 1979). A contradictory view of the role of the prior austenite grain boundaries has also been presented (Ritchie and co-workers, 1976).

As is evident from the contradictory views of the roles of different boundaries in bainite structures on the fracture processes, these complex microstructures raise the question of the effective grain size. In a ferritic structure the effective grain size is evident, but in bainite different opinions has been presented. In most cases the bainite packet is found to be effective (Naylor and Krahe, 1974; Kotilainen and Torronen, 1977; Lonsdale and Flewitt, 1978; Naylor, 1979), but also the austenite grain size is found to control the fracture (Ritchie and co-workers, 1976; 1979). Other microstructural factors such as carbides, retained austenite, trace elements etc. may influence the fracture process. However, very often the investigations do not include a detailed analysis of the microstructure and only few quantitative results are available.

The aim of this paper is to evaluate the effect of bainite packet boundaries on the fracture toughness through detailed microstructural, fractographic and fracture properties examination.

EXPERIMENTAL AND RESULTS

The steel investigated was a quenched and tempered Cr-Mo-V alloyed pressure vessel steel. The austenitizing was carried out at 980°C, the quenching simulating the cooling of a thick plate, and the tempering at 670°C for 20 hours. The quenching simulation resulted in a fully bainitic structure (Torronen, 1979).

The microstructural evaluation showed that the prior austenite grain size (~ 110 µm) consists of bundles of parallel and elongated bainite packets. The bundles have no definite boundaries, whereas the packet boundaries are definitely high angle boundaries. The packets are further divided into laths and/or cells between which low angle boundaries are observed. The average bainite packet size (width) was measured to be 3.0 µm.

Tensile tests were carried out between ambient and cryogenic temperatures, the results being given in Fig. 1. The scatter of the yield strength values is very small and even at the lowest test temperature no sign of twinning has been observed. This is probably due to the high dislocation density (Torronen, 1979). The true fracture strength shows a clear minimum at the temperature of 65 K and increases with the yield strength with decreasing temperature.

The fracture toughness has been measured using 25 mm thick three point bend and CT specimens and one 75 mm thick CT specimen. The valid K_{Ic} values are given with filled points in Fig. 2. The other non valid results are calculated using the estimation method described by Chell and Milne (1976). For the points measured at 77 K the fatigue precracking stress intensity level was a little too high according to ASTM E 399-74, but it is not thought to seriously affect the results. All specimens showed a flat cleavage type fracture.

In order to check the validity of the idea of the limiting notch root radius proposed by Tetelman and co-workers (1968) a set of Charpy size three point bend

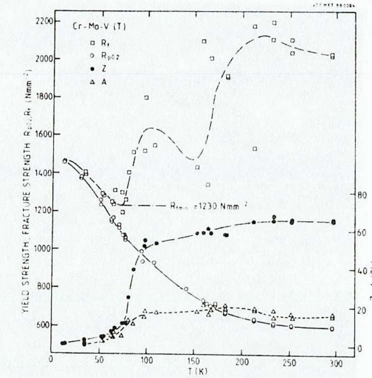

Fig. 1. Yield strength $R_{p0.2}$ and the true fracture strength R_f, the reduction of area Z and the fracture elongation A as a function of temperature.

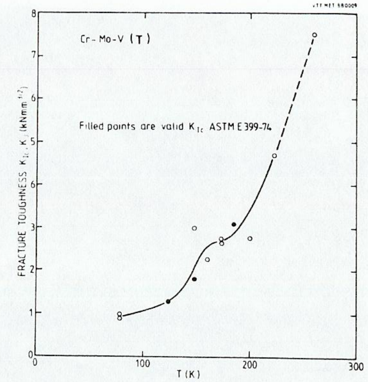

Fig. 2. Fracture toughness as a function of temperature.

specimens were tested at liquid nitrogen temperature. The results are given in Fig. 3 in which the straight line represents the best fit of all measured points. However, neglecting the highest values obtained by the round notches a limiting value of 14 µm can be found as shown by the dotted line.

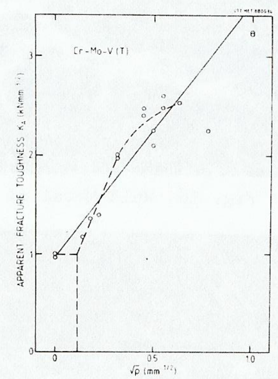

Fig. 3. Apparent fracture toughness K_A as a function of notch root radius ρ.

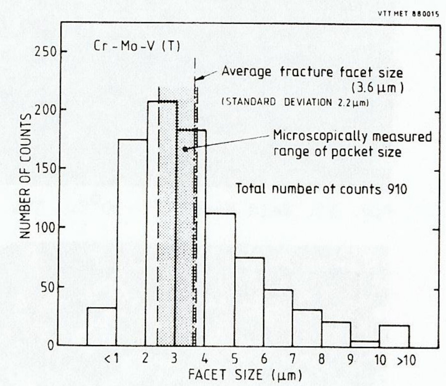

Fig. 4. Distribution of the sizes of the fracture facets.

The fracture facet size has been measured using a scanning electron microscope in which the profile of the fracture has been investigated. A typical example of the profile is shown in Fig. 6. Using the information obtained from similar figures a distribution of the fracture facet sizes has been obtained. The results are given in Fig. 4. The mean facet size is 3.6 µm, which corresponds very well to the bainite packet size considering that the cleavage fracture proceeds on the {001} planes, whereas the packets have a <111> growth direction (Nenonen and co-workers, 1979).

DISCUSSION

In order to be able to investigate the microscopical models for cleavage fracture the cleavage fracture strength has to be estimated. By using the idea of Wobst and Aurich (1977), the tensile cleavage fracture strength can be estimated by means of the tensile test. The minimum tensile cleavage fracture strength can be taken to be the minimum of the true fracture strength, which in this case is 1230 Nmm^{-2} (Fig. 1). The maximum tensile cleavage fracture strength σ_{fcmax} can be obtained by extrapolating the yield strength to 0 K temperature. By performing this calculation by means of a physically reasonable function a value of 1755 Nmm^{-2} is achieved (Kotilainen, 1979). These tensile cleavage fracture strength values deviate clearly from the microscopical cleavage fracture strength σ_f^* which can be estimated using the formula given by Malkin and Tetelman (1971)

$$K_{Ic} = 2.9\ \sigma_Y[\exp(\sigma_f^*/\sigma_Y - 1) - 1]^{1/2}\ \sqrt{\rho_0}, \qquad (1)$$

in which σ_Y is the yield strength, ρ_0 the limiting notch root radius and K_{Ic} the fracture toughness. By applying the value of 14 µm, σ_f^* is not constant but approaches at low temperatures a value of 3100 Nmm^{-2}. Assuming that at low temperatures premature microcracks exist, the Griffith-Orowan equation (Eq. 2) can be applied for the estimation of the microscopical cleavage fracture strength

$$\sigma_f^* = \left[\frac{2E\gamma_p}{\pi(1-\nu^2)a}\right]^{1/2} \qquad (2)$$

in which 2a is the crack length, γ_p the effective surface energy and E Young's modulus. By applying 120 Jm^{-2} as the surface energy (Brozzo and co-workers, 1977; Roman and co-workers, 1979) and the average size of the fracture facet as the crack length a value of 3100 Nmm^{-2} for the cleavage fracture strength can be obtained. This is of the same value as the estimate based on Eq. (1). If the limiting notch root radius is a microscopically significant material constant, it is about 5 times larger than the bainite packet.

According to Ritchie and co-workers (1973) a characteristic distance constant for the material in question can be found. If the stress exceeds the cleavage fracture strength over the characteristic distance, fracture occurs. In Fig. 5 the characteristic distances calculated by means of the Tracey's (1976) stress distribution are given. Both the maximum tensile cleavage fracture strength 1755 Nmm^{-2} and the microscopical cleavage fracture strength 3100 Nmm^{-2} are used for the calculation. The calculation with 1755 Nmm^{-2} does not give any definite minimum or constant value of the characteristic distance but approaches a value of 65 µm at low temperatures. The microscopical cleavage fracture strength gives a value of 8 µm. It should be pointed out that also the stress distribution also affects the results.

However, the characteristic distance is about 3 to 20 times larger than the

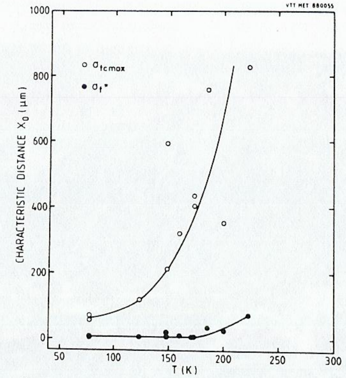

Fig. 5. Characteristic distance X_0 calculated with macroscopical (σ_{fcmax}) and microscopical (σ_f^*) cleavage fracture strengths.

bainitic packet, but always smaller than the prior austenite grain size. Therefore, this observation clearly contradicts the results of Ritchie and co-workers (1973, 1976) in which they found that the characteristic distance is about 2 to 4 times the austenite grain size. The difference need not, however, be very serious because in a bainitic microstructure, due to the high dislocation density, the yielding units are very fine and more units are assumed to be involved in the yielding than in ferrite. It is unreasonable to assume that in a structure with high dislocation density dislocation pile-ups can form and produce a crack nucleus (Kotilainen and co-workers, 1979). Therefore, one must distinguish between the yield triggered crack initiation and crack propagation, when considering the effective grain size.

The yielding and the cleavage fracture strength are controlled by the same factors (Kotilainen and co-workers, 1979). However, the formation of microcracks and crack propagation are controlled by the packet boundaries. This is clearly demonstrated in Fig. 6 in the upper half of which a section through the fracture profile is shown. In the lower part of Fig. 6 the fracture surface is shown with a line along which the section has been taken. Fig. 6 clearly shows that fracture changes its propagation direction at every high angle packet boundary. This is consistent with the observations of Naylor and Krahe (1974) and Naylor (1979) and others. The important role of the bainite packets is shown quantitatively in Fig. 4 according to which the fracture facet size corresponds very well to the size of the bainite packets.

A detailed mechanism for the microcrack formation cannot be given. It is, however, a special feature in this particular bainitic microstructure that only two orientations of bainite packets exist within a packet bundle (Torronen, 1979). Due to this fact it is assumed that only suitably oriented packets can yield and thus relax the stress concentration at the crack tip. The packets having another orientation do not yield but because of the compatibility of the material they must break. Locally this phenomena can be assisted by some discontinuities at packet boundaries e.g. by carbides. Figure 7 shows an example of the tip of an arrested crack. The discontinuous mode of propagation is evident. At the top of the Fig. 7 the microcrack has probably initiated between two small carbides. The tip of the microcrack is blunted. All microcracks are parallel because also in bainite the crack is formed on {001} type planes (Nenonen and co-workers, 1979). No cracked carbides are found in the

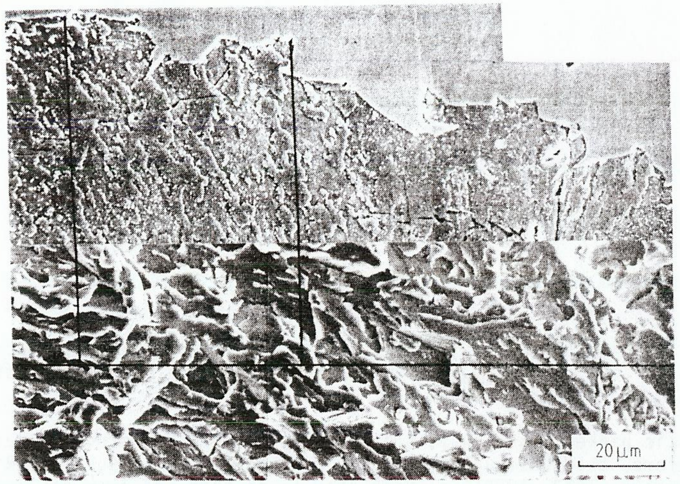

Fig. 6. Fracture profile and corresponding microstructure with the cleavage fracture surface.

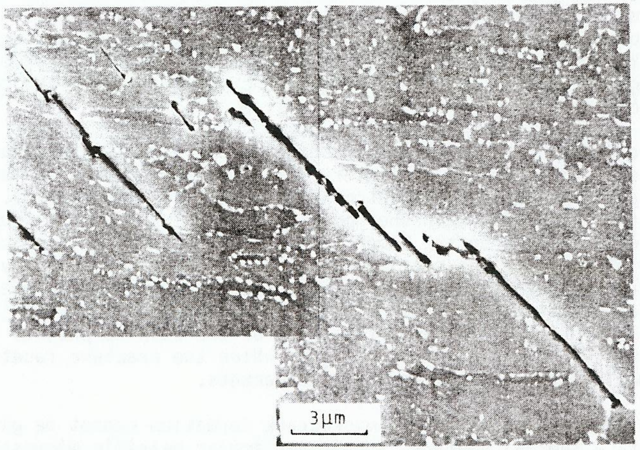

Fig. 7. Microcracks at the tip of an arrested crack.

micrographs. The same packet can be cracked at several locations leaving behind unbroken ligaments of the other packets, as shown in Fig. 8. This phenomena supports the idea of the influence of the two packet orientations on the microcracking process. Figure 8 clearly demonstrates the importance of the packet boundaries as a strong obstacle for cleavage crack growth. In Fig. 8 also the microcrack ends inside a packet. The arrest can be due to the loss of driving stress during the arrest period, or because the microcrack cannot follow the borders of the packets having very irregular shape.

At low temperatures in which the fracture is assumed to be controlled by the mechanism and not by slip dislocations, finding a suitably oriented packet could

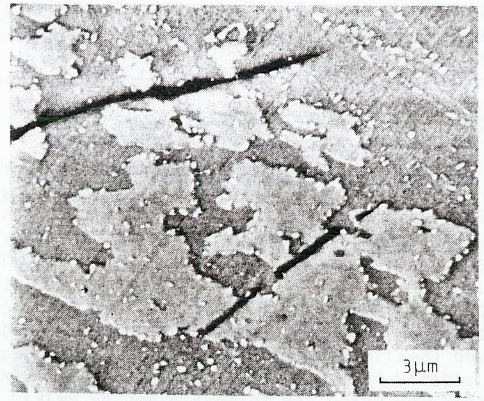

Fig. 8. Broken bainite packets.

be a probability function of the whole distribution of packets. Also small packets can act as sites of microcracks whereas at higher temperatures the fracture could be determined by the average size of the packets. Very large packets are rare according to Fig. 4, and thus the initiation of fracture at those sites is not very probable.

CONCLUSIONS

A bainitic Cr-Mo-V steel has been studied to elucidate the role of the bainite packet boundaries in cleavage fracture. The formation of microcracks and crack propagation are found to be controlled by the packet boundaries. The size of the fracture facet corresponds very well with the size of the bainite packet. The microcracks usually extend from boundary to boundary having a common habit plane inside a packet. The microcracks can initiate at packet boundary carbides, although these carbides do not break.

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