A COMPARISON OF BRITTLE FRACTURE BEHAVIOUR OF VARIOUSLY TEMPERED MARTENSITIC AND BAINITIC STRUCTURES OF SECONDARY HARDENING Cr-Mo-V PRESSURE VESSEL STEEL

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ABSTRACT

Brittle fracture behaviour of a Cr-Mo-V alloyed pressure vessel steel has been investigated after a variety of quenching and tempering treatments. Martensitic and bainitic microstructures were obtained after different quenching operations. Samples were subsequently tempered to several strength levels and the fracture behaviour, i.e. brittle crack initiation and propagation was evaluated as a function of the test temperature. A thorough microstructural and fractographical analysis was also performed. The effective grain size was found to be martensitic or bainitic lath packet size; the fracture facets are of the same size as the packets. The martensitic structure having smaller packet size is tougher than the bainitic structure at equal strength levels when considering crack initiation and propagation transition.

KEYWORDS

Brittle fracture; martensite; bainite; crack initiation; crack propagation; microstructure, fractography.

INTRODUCTION

Several microstructural factors are known to affect the brittle fracture behaviour of quenched and tempered steels. Grain size is normally the most apparent microstructural unit affecting toughness. The influence is not, however, well characterized expecially in complicated martensitic and bainitic structures due to other factors like dislocation density and carbides. In addition to this a controversy is evident concerning the effective grain size. In most cases the lath packet is proposed to be effective (Naylor and Krahe, 1974; Kotilainen and Törrönen, 1977; Naylor, 1979), but also the austenite grain size is found to control the fracture (Ritchie and co-workers, 1976; 1979). The laths themselves have only a minor effect on the fracture resistance due to their small mutual orientation differences (Der-Hung Huang and Thomas, 1971; Naylor, 1979).

The cleavage crack propagates in ferrite along the {001} planes. There is evidence that the fracture planes also in martensite and in bainite are of {001} type (Terasaki and Ohtani, 1972; Matsuda and co-workers, 1972). However, also planes of type {011}, {112} and {123} have been reported (Lindborg and Averbach, 1966; Naylor and Krahe, 1975; Naylor, 1979).

The correlations between cleavage crack initiation, propagation and microstructure in bainitic structures of secondary hardening Cr-Mo-V pressure vessel steels have been reported previously (Törrönen and Kotilainen, 1977; Kotilainen and Törrönen, 1977; Kotilainen and co-workers, 1980, 1981; Törrönen and co-workers, 1980; Kotilainen, 1979, 1980). The basic findings indicated that the changes in the dislocation density as governed by the coarsening of fine vanadium carbides, determine the changes in the fracture behaviour. A strong influence of lath packet size was also postulated but at that time experimental evidence was not sufficient. The present work was undertaken to clarify the role of the lath packet size in the brittle fracture behaviour of the previously studied Cr-Mo-V steel.

EXPERIMENTS AND RESULTS

The material studied was a Cr-Mo-V alloyed steel with 0.16 pct C, 2.8 pct Cr, 0.6 pct Mo and 0.3 pct V. The steel was quenched at two rates producing martensitic and bainitic structure, respectively. Subsequent temperings were carried out at temperatures between 600° and 760°C for 20 h.

The microstructural evaluation revealed that the prior austenite grain is divided into martensite or bainite lath packets. These packets are the smallest units surrounded by high angle boundaries. Parallel and elongated packets form bundles. Only two ferrite orientations are seen within a single bundle. The packets are further divided into parallel laths. The lath boundaries inside a packet are always low-angle boundaries. Further details of the microstructural studies are reported elsewhere (Törrönen, 1979). The average packet size (width), which remained constant during temperings, was measured to be 2.4 and 3.0 μm for martensitic and bainitic structures, respectively.

The fracture behaviour of the variously tempered martensitic and bainitic structures was evaluated using tensile tests down to cryogenic temperatures, instrumented impact tests and Pellini drop-weight tests.

The temperature dependence of the yield strength was found to be equal for both martensitic and bainitic structure tempered to equal room temperature strength level (Törrönen, 1979). The instrumented impact tests revealed, however, the superiority of the martensitic structure. An example is shown in Fig. 1 which gives the total absorbed energy as well as the crack initiation energy as a function of test temperature for a martensitic (MP) and bainitic (T) structure of equal strength level. The energy before maximum force was used as an estimate for the initiation energy (Koppenaal, 1974). The transition curves based on fracture surface appearance are given for all test materials in Fig. 2. The room temperature yield strength values varied between 380-1020 Nmm^{-2} and 480-930 Nmm^{-2} for bainitic and martensitic structures, respectively. The transition temperatures based on the transition in initiation behaviour and 15 pct shear area are shown as a function of the room temperature yield strength in Fig. 4. Reasons for these transition criteria will be discussed later. The nil-ductility transition temperatures representing the propagation behaviour are also given in Fig. 4.

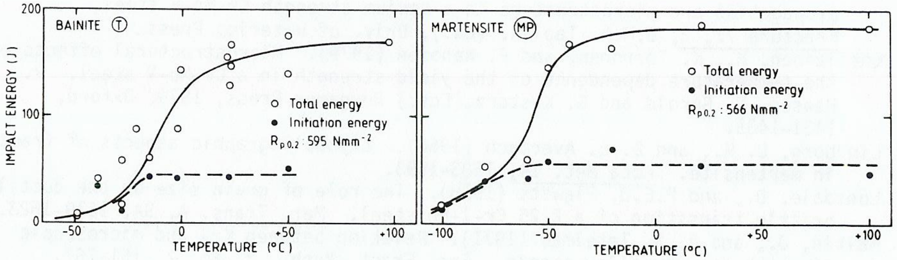

Fig. 1. Total and initiation energy for a martensitic and bainitic structure of equal strength level as determined in instrumented impact testing of Charpy V specimens.

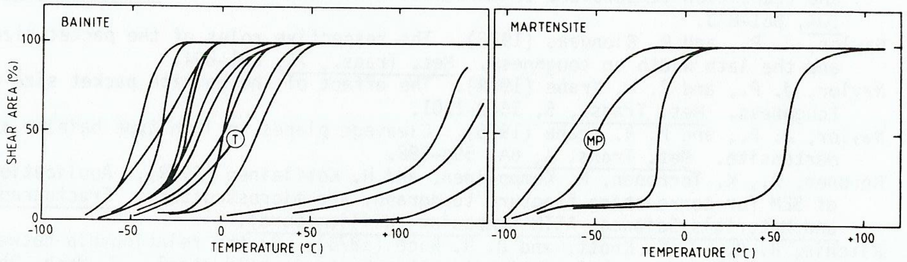

Fig. 2. Results of impact testing based on fracture surface appearance.

A fractographic investigation was performed to evaluate the connections between microstructure and crack propagation. Typical examples of the fracture surfaces and fracture profiles are shown in Fig. 3. The average cleavage fracture facet size was measured using the profile pictures to be 2.7 and 3.6 μm for the martensitic and bainitic structures, respectively. These mean values correspond well to the packet size considering that the cleavage fracture proceeds on the {001} planes, and the packets have <111> growth or long direction (Törrönen and co-workers, 1980).

DISCUSSION

The Charpy V test measures essentially the crack initiation energy to a temperature in the transition region which corresponds to roughly 1/3 of the upper shelf energy (Kotilainen and Sirkkola, 1979). This behaviour is evident also in Fig. 1. At this temperature the initiation energy reaches its maximum and stays constant with increasing temperature. It has also been found that at this temperature about 15 pct of the fracture surface is ductile the rest being quasi-cleavage. Thus a transition criteria of 15 pct shear area (T_{15}) is equal to the transition in initiation behaviour. This situation is schematically illustrated in Fig. 5. At this temperature the cleavage crack initiates in the plastic zone ahead of the notch, but the ligament R between the cleavage initiation region and the notch is broken by tear, i.e. by a ductile mechanism. Below this transition the microcracks are formed closer to the crack tip due to the higher triaxiality at the notch. Thus the ligament R is supposed to attain its maximum value at T_{15}. Decreasing temperatures result in lower values. At

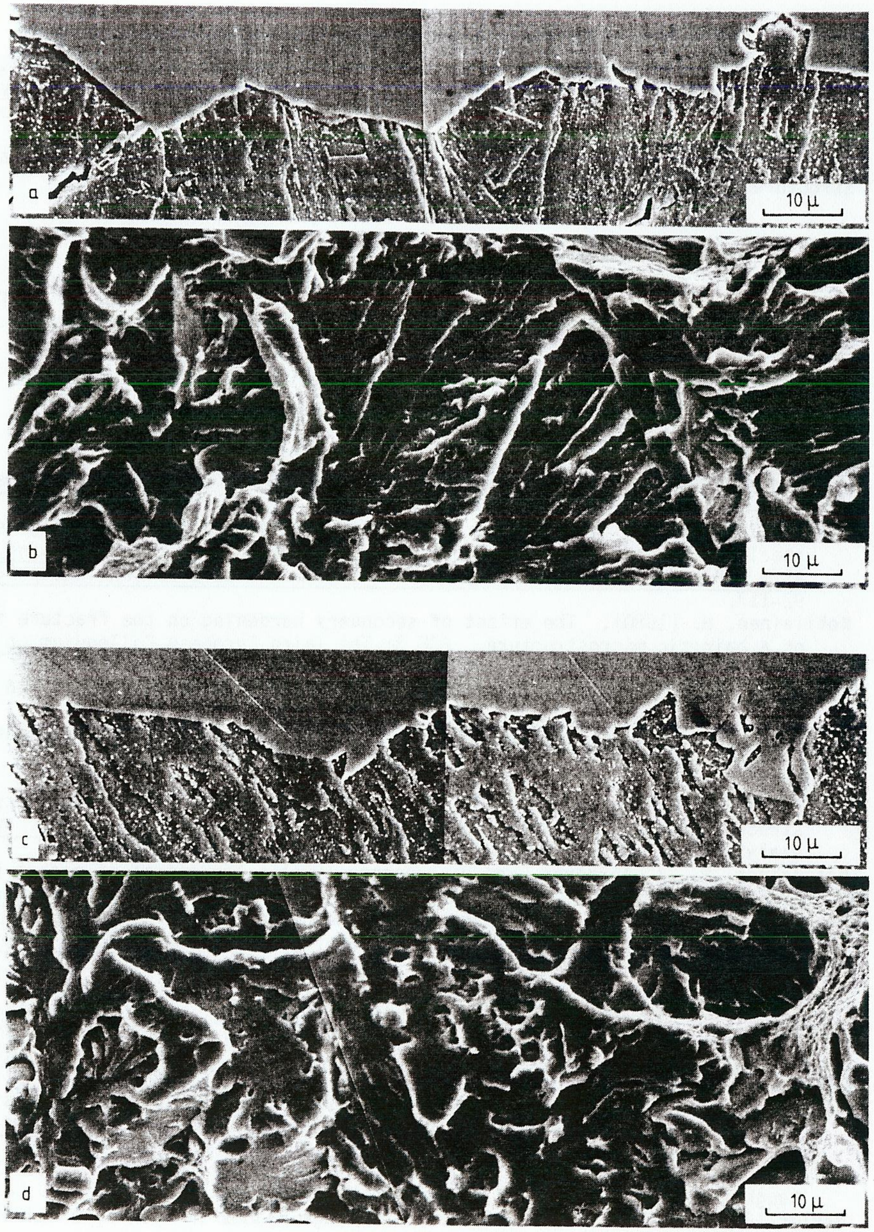

Fig. 3. Fracture profiles and the corresponding microstructures with the cleavage fracture surfaces. Martensitic structure a, b; bainitic structure c, d.

higher temperatures the fracture initiation is assumed to be caused by plastic tear with consequent increase in the absorbed energy, although after a certain amount of ductile tear the condition for cleavage fracture can be reached.

When this initiation transition is plotted as a function of the room temperature yield strength a linear relationship can be seen (Fig. 4). This indicates that the same microstructural factors that govern the yield strength may also control cleavage crack initiation. It has been shown earlier (Törrönen and co-workers, 1979; Törrönen, 1979) that dislocation density, which in turn is dependent on the coarsening of vanadium carbides, is solely responsible for the room temperature yield strength. As is evident in Fig. 4 the initiation transition temperature is, however, different for bainitic and martensitic structure of equal strength the difference remaining almost constant througout the studied strength levels. The only basic difference between the martensitic and bainitic structures is the lath packet size, which is smaller for martensitic material (2.4 µm versus 3.0 µm). Thus the transition temperature can also be correlated to the packet size, smaller size giving lower transition temperatures.

The initiation process can be considered to include both the formation of the plastic zone and the microcracks; it is thus controlled by slip dislocations by definition. Therefore a certain amount of plastic deformation for the initiation is required.

The same factors which increase the yield strength increase also the cleavage fracture strength (Kotilainen and co-workers, 1980). To obtain microcracks the cleavage fracture strength must be exceeded. If both the cleavage fracture strength and the yield strength are high, the cleavage microcracks are initiated easily because deformation processes cannot relax the stresses at the notch tip. Thus it is evident that the transition temperature depends on both the yield strength and the cleavage fracture strength.

To illustrate the discussion above the initiation energy for cleavage crack is plotted as a function of the room temperature yield strength in Fig. 6. A linear relationship is evident for both bainitic and martensitic structures. Smaller lath packet size of the martensitic structures is shown to require a higher initiation energy value as compared to the bainitic structures of equal strength levels. This strongly suggests that the cleavage crack initiation is governed by the Griffith-Orowan type analysis, which relates the cleavage fracture strength to the initial crack length, which in this case is obviously related to the packet size. This is further discussed in a related paper in this conference (Kotilainen and co-workers, 1981).

A comparative behaviour is evident concerning the cleavage crack propagation. As seen in Fig. 4 the NDTT is again linearly dependent on the room temperature yield strength but there is a clear difference between the behaviour of the bainitic and martensitic structures. The transition in behaviour occurs at lower temperatures for the martensitic structure, i.e. when the packet size is smaller. An explanation for this behaviour can be seen in Fig. 3 showing how the brittle fracture path changes its direction always when it crosses the lath packet boundary. The crack never propagates along packet boundaries but cross the packets at various angles. This is consistent with the knowledge that mainly {001} type cleavage planes are formed and that the long direction of the packets is <111> (Törrönen, 1979; Törrönen and co-workers, 1980). If the appearance of the fracture surfaces of the martensitic and bainitic structure are compared two basic differences are evident. The fracture facets of the martensitic structure are often elongated whereas those of the bainitic are more irregular. This is due to the differences in the cross section of the

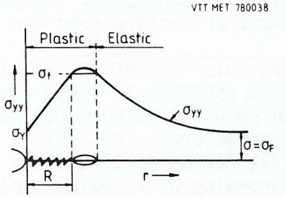

Fig. 5 Schematic representation of the fracture initiation at the tip of a round notch. R corresponds to the length of the shear zone between the microcrack and notch.

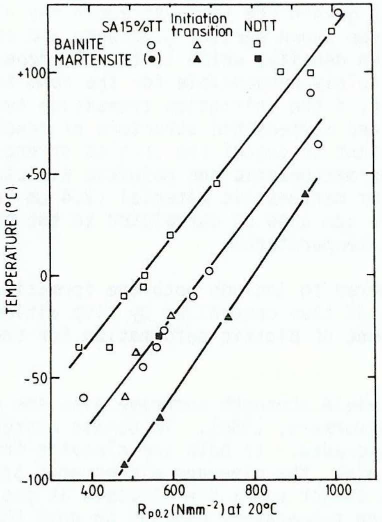

Fig. 4. Initiation and propagation transition temperatures as a function of the room temperature yield strength.

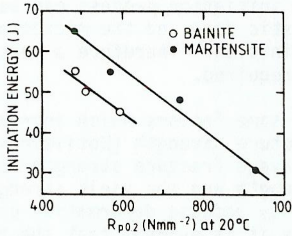

Fig. 6. Initiation energy for cleavage crack as a function of the room temperature yield strength.

packets in the two structures; the packets of the martensitic structure are shown to be elongated, plate like, sometimes crossing each other, whereas those of the bainitic structure are irregular and finger like (Törrönen, 1979). Another difference is that smaller angular changes of the propagating crack are seen is the martensitic structure. This indicates that also other planes than {001} planes may cleave in a martensitic structure, as suggested by Naylor and Krahe (1975) and Naylor (1979). What makes the difference in this case between martensite and bainite is not known.

The above discussion indicates that the packet size is the effective grain size. As the propagating crack crossing the packet boundary changes its direction this step is evidently the most energy consuming, i.e. the propagation in a structure of smaller packet size requires more energy. Thus it is obvious that a transition to ductile propagation occurs at lower temperatures in the martensitic structure as compared to the bainitic one.

A practical consequence of the observed and discussed behaviour is also evident. If the presented results are applied to the weldments in heavy section components it is obvious that the martensitic region in the heat affected zone shows better resistance to brittle fracture initiation and propagation than the bainitic base metal.

CONCLUSIONS

It has been shown that the effective grain size in a Cr-Mo-V alloyed secondary hardening steel is the martensitic or bainitic lath packet size. The cleavage fracture processes are influenced by this structural unit as well as by the same microstructural factors which control the yield strength. The martensitic structure having smaller lath packet size is found to be superior to the bainitic structure at equal strength levels.

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