THE RELATION OF TOUGHNESS TO HARDNESS AND RETAINED AUSTENITE IN SINTERED HIGH SPEED STEELS.

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An increase in toughness with decreasing hardness is reported for sintered high speed steels. Toughness is related to the properties of the matrix, but more specifically to retained austenite content. Austenite retention is caused by the stabilizing effect of carbon, freeded from MC carbides when transformed in MX carbonitrides. Fractographic examination has shown that although mainly failure proceeds by quasicleavage, the low austenite containing materials show flatter fracture surfaces than the materials containing high amounts of austenite.

INTRODUCTION.-

High speed steels, which have been traditionally used only for cutting tools, are increasingly being employed in wear and fatigue resistant applications (1). Improved fracture toughness is a requirement in these applications (2,3). Important improvements in fracture toughness using powder metallurgy processing are already reported in the literature (4). In the present work the fracture toughness of three high speed steels T15, T42 and T6 obtained by sintering in vacuum and in an atmosphere 90% N_2, 9% H_2 and 1% CH_4 has been measured. In the atmosphere sintering process, due to the presence of a high amount of nitrogen in it, MC carbides are transformed to MX carbonitrides and consequently the important amount of carbon available stabilizes the austenite. The specimens with high amount of retained austenite present a good toughness, and an important amount of austenite is transformed mechanically to martensite in a region close to the running crack. This mechanism seems to be responsible of the high toughness values observed in these materials.

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EXPERIMENTAL PROCEDURE.-

The powders were mixed for 2 h with an addition of 0.2% high purity graphite. Cylindrical specimens were uniaxially pressed in a 16mm diameter die using a pressure of 500 MPa. Sintering was carried out in a three zone Lindberg furnace provided with a refractory steel chamber. A 90% N_2, 9% H_2 and 1% CH_4 atmosphere and vacuum were used for sintering. Additional sintering details have been described elsewhere (5,6). After annealing, cylinders 12 mm in diameter and of 18 mm height were machined to "short rod" specimens for fracture toughness testing. The specimens were quenched and tempered (triple 1 h tempering at temperatures between 425º and 585ºC). The fracture toughness was measured using the "short rod" technique. This kind of tests, as reported by Barker (7) does not need previous precracking and, as described elsewhere(6) the fracture toughness was measured from the maximum load and the specimen geometrical factors. Fractographic examination was performed on one of fracture surfaces of the short rod specimens in a Philips 501B scanning electron microscope. The other surface after a deep polishing was used for metallographic observations, hardness measurements and for the determination of retained austenite. This was measured using X-ray diffraction techniques, using the method proposed by Miller (8).

RESULTS AND DISCUSSION.-

Fig. 1 shows the variation of fracture toughness with hardness for the different steels and heat treatments. A clear decrease of toughness with increasing hardness is observed. Within the experimental scatter, all the data can be reasonably fitted to a single straight line, indicating that fracture toughness is related to hardness and is independent of the amount of primary carbides present in the microstructure, the type of atmosphere used for sintering and of the steel's chemical composition. No influence of HIPping is observed. The straight line has the equation:

$$K_{Ic}(MNm^{-3/2}) = 64.5 - 0.054\ (HV10)$$

0.054 being the slope within the range 0.045 and 0.063 for the 95% confifence limits. Toughness in high speed steels has been inversely correlated with hardness(6), and interpreted as mainly related to the matrix properties and not to the amount, size and morphology of the primary carbides.

Fig. 2 shows the relationship between fracture toughness and amount of retained austenite. It is clearly apparent that increasing the amount of retained austenite results in an increase in fracture toughness. Additionally it can also be observed that variables as composition,sintering atmosphere or thermomechanical treatments, as HIP, do not have a large influence in fracture toughness. When Fig.2 is closely examined, it is evident that above

40- 50% retained austenite there is not a detectable increase of fracture toughness. The experimental results up to 40- 50% retained austenite can be fitted to a straight line of equation:

$$K_{Ic}(MN\ m^{-3/2}) = 9.3 + 0.48\ (\%\ \gamma)$$

The 95% confidence limits for the slope of the ajusted straight line are within 0.35 and 0.62. This kind of cut-off point, near 50% retained austenite has been also observed in an austempered ductile iron(9).

The beneficial effect of austenite in fracture toughness, at least for small amounts of austenite, has been associated to a blunting effect when the crack reaches the ductile austenitic phase(10). Other authors have also attributed this beneficial effect to crack branching and transformation-induced plasticity-TRIP-(11). In this material, containing an important amount of austenite, a substantial amount of it is transformed mechanically to martensite in a region close to running crack, as shown in Fig. 3. The energy spent during the transformation together with the compressive stresses created in front of the crack, due to the formation of martensite, could explain the improvement of toughness in this material. A similar effect has been observed in an austempered ductile iron for high amounts of austenite (9).

The fractographic analysis carried out by scanning electron microscopy on the failure surfaces has indicated that although all fracture surfaces can be described as quasi-cleavage, there are some differences between those corresponding to high and low hardnesses. The fractures observed on low hardness and high retained austenite containing materials are characterized by the presence of large facets which seem to correspond to M_6C primary carbides present at grain boundaries (Fig. 4).The majority of such facets are along (decorating) the grain boundaries. In addition the regions occupied by these carbides seem to form some kind of valleys. Indications are, therefore that cracks run mainly through these primary carbides present at the grain boundaries. For high hardness and low austenite containing steels the fracture surface can be differenciated from the previous by a flat aspect, showing no evidence of the "valleys". It can be observed that the size of the facets has decreased, but their amount has increased (Fig. 5). In this case the grain boundaries, decorated with M_6C, do not seem to be the preferred path for the advancing crack, its propagation being controlled by the matrix, not the carbides.

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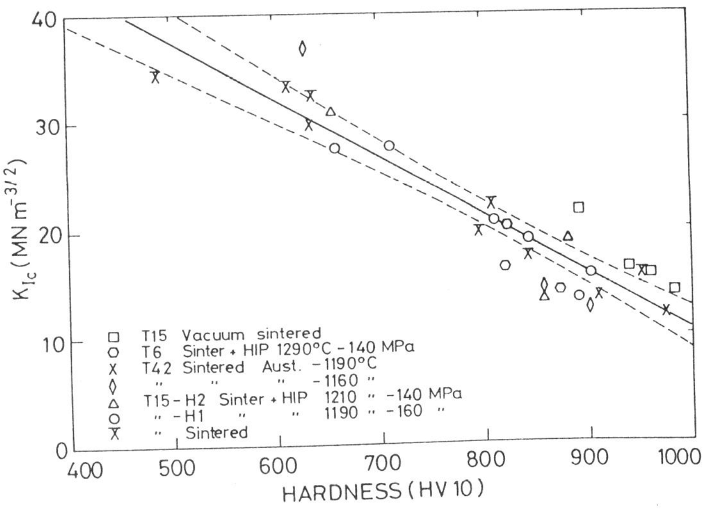

Figure 1 Dependence of fracture toughness on hardness.

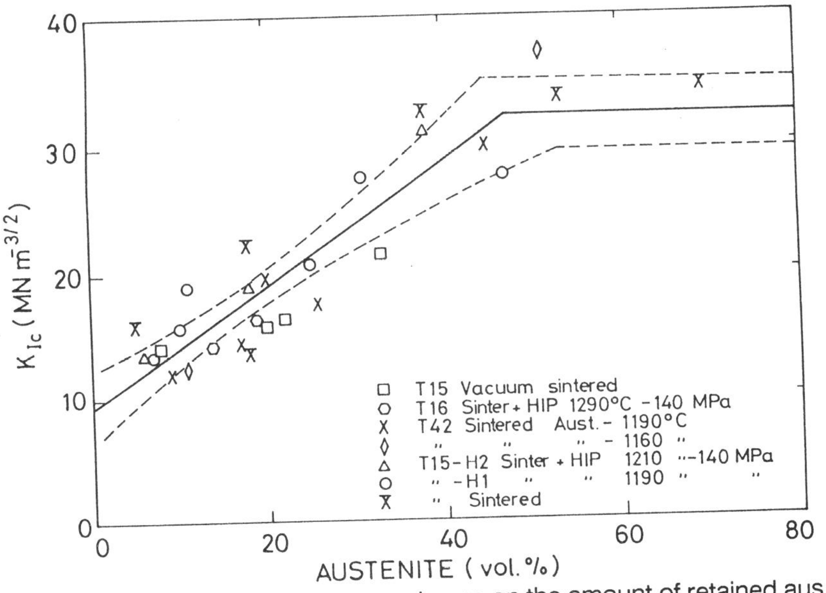

Figure 2 Dependence of fracture toughness on the amount of retained austenite.

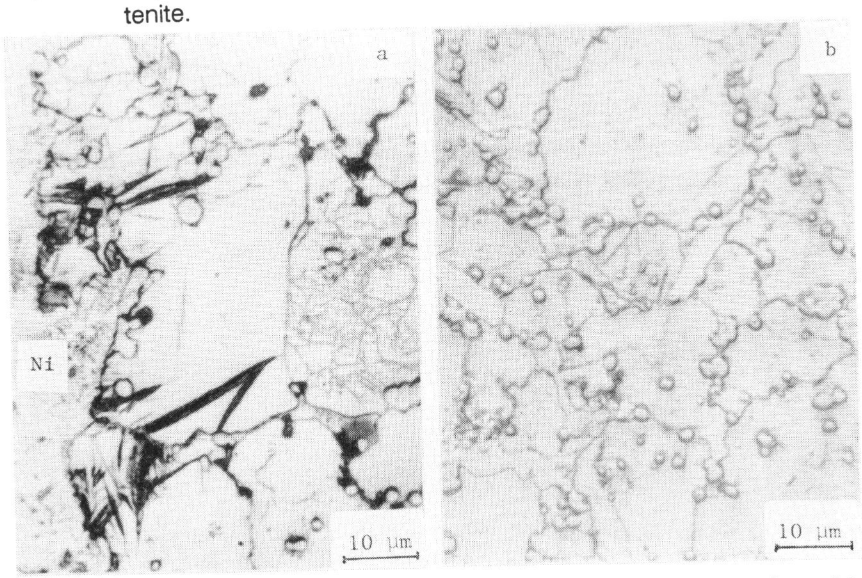

Figure 3 Microstructure of a "short rod" specimen. Near (a) and far from (b) the surface crack.

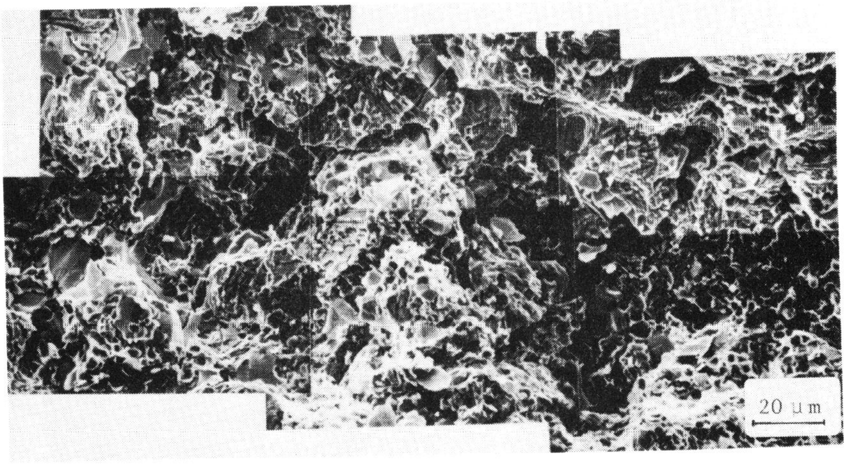

Figure 4 Fractograph corresponding to a steel with low hardness and high austenite content.

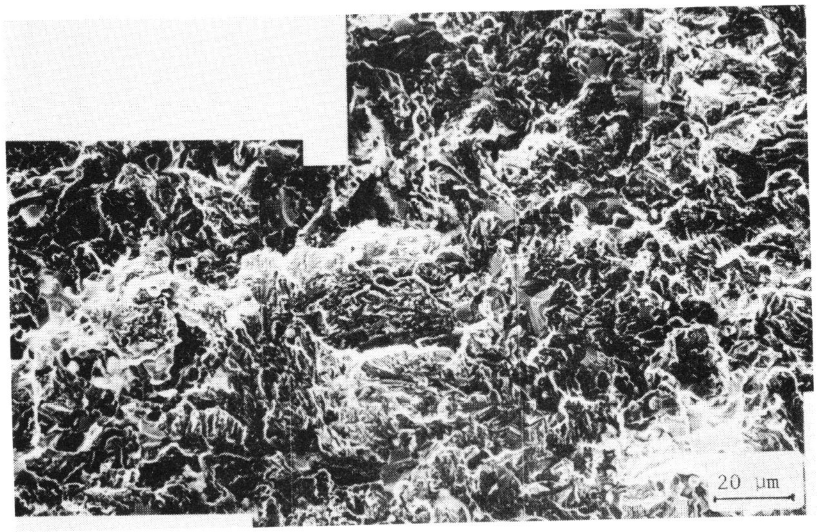

Figure 5 Fractograph corresponding to a steel with high hardness and low austenite content.