The toughness of a cast hot-work tool steel

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Abstract

The mechanical and technological properties of cast hot-work tool steels need to be fully exploited, since cast-to-form tools are increasingly utilized in many sectors of the forming industry.
In the present research work, the fracture toughness of a cast H 11 hot-work tool steel is evaluated, using fracture mechanics and impact tests, on specimens machined from cast-to-form dies selected to represent different microstructural conditions.
The results of the mechanical and toughness tests are compared with corresponding properties of wrought H 11 steel, and interpreted in terms of the microstructural characteristics of the cast steel.

Riassunto

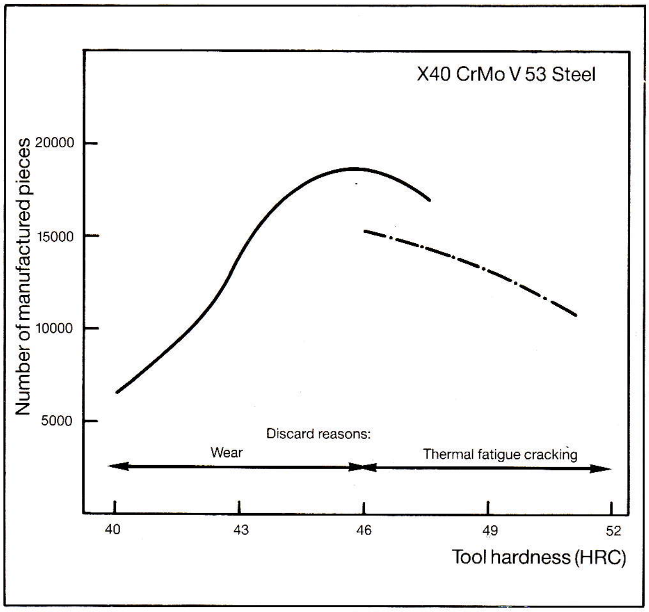
Tenacità di un acciaio per utensili a caldo allo stato di getto

L'impiego di utensili ottenuti mediante colata di precisione è in aumento in molteplici settori dell'industria di deformazione plastica a caldo e di conseguenza si avverte l'esigenza di valutare compiutamente le proprietà meccaniche e tecnologiche degli acciai per utensili per lavorazioni a caldo allo stato di getto.
Nel presente lavoro sperimentale si è valutata la tenacità alla frattura di un acciaio fuso del tipo H 11 mediante prove di meccanica della frattura e di resilienza su provini ricavati da matrici per lo stampaggio a caldo di acciaio, opportunamente selezionate in maniera da rappresentare differenti condizioni microstrutturali dell'acciaio H 11 allo stato di getto.
I risultati delle prove meccaniche e di tenacità sono stati confrontati con quelli relativi all'acciaio H 11 fucinato e sono stati interpretati in termini delle caratteristiche microstrutturali dell'acciaio fuso.

Introduction

The requirements that hot-work tools have to meet are predominantly connected with their performance; particularly where high production series are involved, manufacturers are generally compelled to seek improved tool perfomance so as to achieve overall savings by increased production reliability and extended service life.
The need for better tool performance is especially felt when the competition from alternative manufacturing processes is on the increase, as has been the case for some years in certain sectors of the metal forming industry, and particularly in the forging sector.
Furthermore, besides the requirement for single-tool manufacturing of a great number of pieces in a specific manufacturing process, other considerations must be taken into account from the standpoint of the overall economic importance of the tool itself. In this connection, the outstanding consideration is that the cost of the tool material is almost always a relatively small part of the total cost of manufacturing the tool; for instance, in the forging process, die costs are made up mainly of the cost of the tool steel and the cost of machinig, and the latter is usually five to ten times the former.
The handicap of high machining costs can be largely overcome in the case of cast-to-form tools; problems with this type of tools may arise from the solidification structure, whose mechanical and technological properties, and consequently the tool's service performance, may be quite different from those of wrought steel.
It must be emphasized, however, that casting has been put forward as a possible means of obtaining improved performances from existing tool materials - especially hot-working tool steels - it being unlikely that radical changes will occur in such alloys, considering the working conditions and the consequent demands on the tool steel (1).
Therefore, there is a special need for a comprehensive assessment of the characteristics of hot-work tool steels in the as-cast conditions, and the interrelationships between their properties and the microstructural features of the cast structure.
Reverting to the service performance of a tool in a hot-work manufacturing process, and taking as an example a forging operation in the automotive industry, Fig. 1 (2), die life is limited chiefly by wear or by thermal

Fig. 1 - Service performance of forging dies in an automotive industry as a function of tool hardness (Ref. 2).



fatigue cracking, tool hardness being, apparently, one of the more influencial parameters.
However, apart from the properties of resistance to wear and thermal fatigue cracking, it is primarily toughness that can affect tool life to a high degree, as pointed out in the literature (3).
Toughness is indeed an important characteristic when the material is subjected to stresses, as it gives a measure of the safety level against the risk of sudden failure.
The toughness of hot-work tool steels has generally been tested dynamically by means of notched or un-notched impact tests; fracture mechanics methods of toughness rating have rarely been used.
On the other hand, the fracture mechanics characterization of hot-work tool steels can provide a useful tool for investigating the microstructural parameters that control toughness, since a number of models (4) have been proposed and experimentally verified for the interrelationship between fracture toughness parameters and microstructural characteristics of steels, both in the case of ductile rupture and of brittle fracture.
The aim of the present research work is to make a contribution to fracture toughness assessment and to understanding the relationship between toughness and microstructure in cast hot-work tool steels.

Experimental procedure

The starting material for this investigation consisted of three forging dies, selected from those discarded in the forging division of an automotive industry, so as to represent three different microstructural conditions.
These dies had been made by the same manufacturer using a casting-to-form process, and had been discarded because of fatigue cracking, brittle fracture failure and excessive wear respectively.
The chemical composition of the die steels corresponding to AISI H 11 hot-work tool steels, is given in Table 1, where it can be seen that two dies, namely the first and third ones, had been made from the same heat.
Although it was impossible to go back to the specific details of the manufacturing process of three dies, macro-examination of them after acid etching revealed that, while steel 1 and steel 3 did not show segregation, steel 2 was strongly affected by dendritism and segregations, probably due to incorrect casting and/or homogenizing heat-treatment operations, and exacerbated by the effect of the higher carbon content (5).
The 10 × 10 × 55 mm specimens for the impact and three-point bending fracture-toughness tests were machined from the dies, in the same position within each die; in the impact specimens a Charpy-V notch was machined, with a root radius of 0.05 mm, in order to approximate the effect of a crack-like notch, and fracture mechanics tests were carried out according to the standard ASTM E399-81.
Tensile and fracture mechanics tests were made at room temperature by means of a 100KN screw-driven Instron Mod. 1195 instrument, at a crosshead speed of 0.1 mm/min; impact test properties on the other hand, were evaluated over the temperature range 20-400°C.
Metallographic and fracture surfaces were examined by means of an ISI Super II scanning electron microscope equipped with a PGT energy dispersive micro-analysis facility.

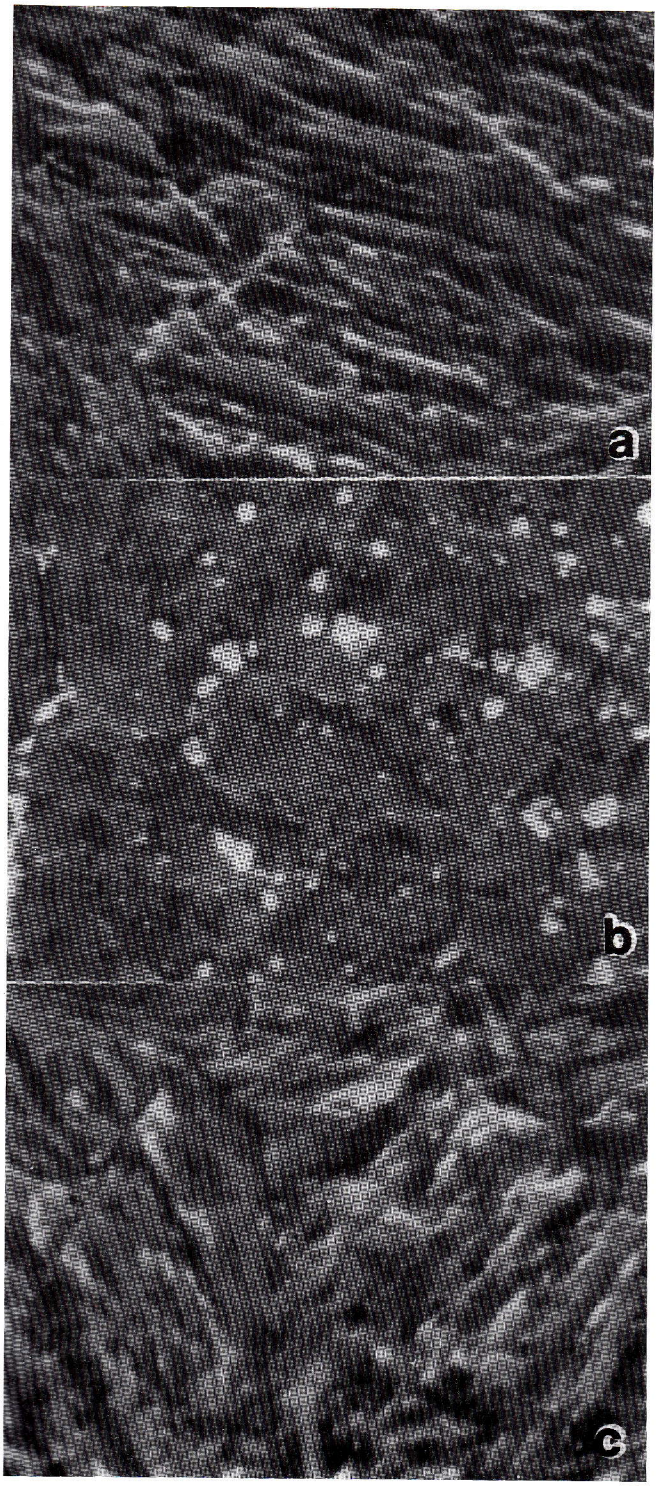
Results and discussion

The microstructure of the steel for all three dies consisted of tempered martensite, resulting from similar quenching and tempering heat treatment cycles.
These microstructures are shown in Fig. 2 at very high magnification; it can be observed that, unlike steels 1 and 3, steel 2 displayed a lot of relatively large carbides, identified as chromium-vanadium and molybdenum carbides, the occurrence of which agrees with the previously reported segregation of alloying elements.
Furthermore, from the microstructural viewpoint, there was a difference between steel 1 and steel 3, in that the latter presented an almost homogeneous austenitic grain size, of the order of 30 μm, while the former's austenitic grain size showed some scatter, with grain diameters up to about 80 μm and a mean austenitic grain size of order of 50-60 μm.
Mechanical properties of the three steels at room temperature are listed in Table 2. Steels 1 and 3 have

TABLE 1 - Chemical composition (wt. %) of the investigated hot-work tool steels.

	C	Si	Mn	Cr	Mo	V	S	P	Ni	Cu	Al
Steel 1	0.37	1.19	0.34	4.97	1.28	0.430	0.005	0.022	0.08	0.06	0.017
Steel 2	0.46	1.17	0.60	5.32	1.44	0.425	0.015	0.028	0.26	0.17	0.082
Steel 3	0.39	1.20	0.34	5.04	1.30	0.430	0.005	0.022	0.08	0.06	0.017

Fig. 2 - Microstructural appearance of the experimented steels; a) steel 1, b) steel 2, c) steel 3, (× 5000).



very similar yield strength and ultimate tensile strength. Moreover, these tensile properties are in accordance with those of wrought steel H 11 at the same hardness level (c.f. Ref. 6). However, the toughness related properties such as elongation and reduction of area are very low, and are comparable with the lowest values for transverse specimens of wrought H 11 steel at the same tensile strength level (7).

Steel 2, notwithstanding a lower hardness and correspondingly lower yield and tensile strength than steels 1 and 3, has an even inferior elongation and reduction of area; both the low-hardness and the low-toughness related properties of steel 2 are probably connected with the above mentioned coarse carbide precipitation.

The experimental results of plane strain fracture toughness tests at room temperature are given in Fig. 3 as a function of hardness, together with the scatter band of fracture toughness data for wrought H 11 + H 13 steel (6).

Despite the difference in the hardness value, the

Fig. 3 - Plane strain fracture toughness of the tested steels as a function of hardness. The scatter band (Ref. 6) pertains to published data for wrought H 11 + H 13 hot work tool steel.

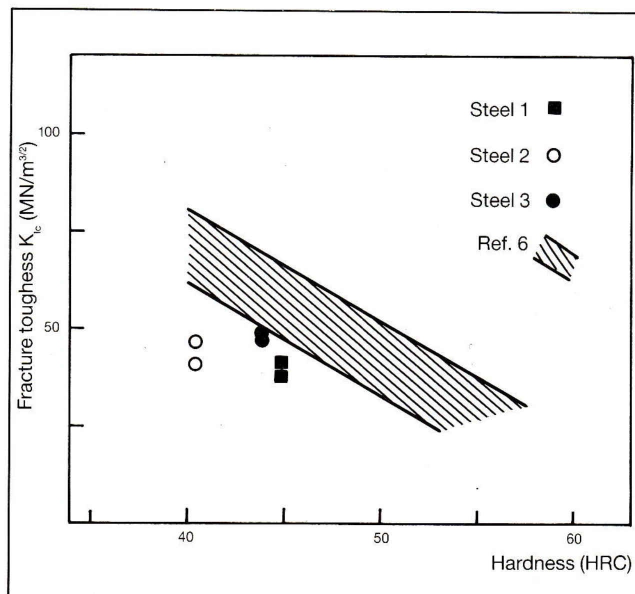
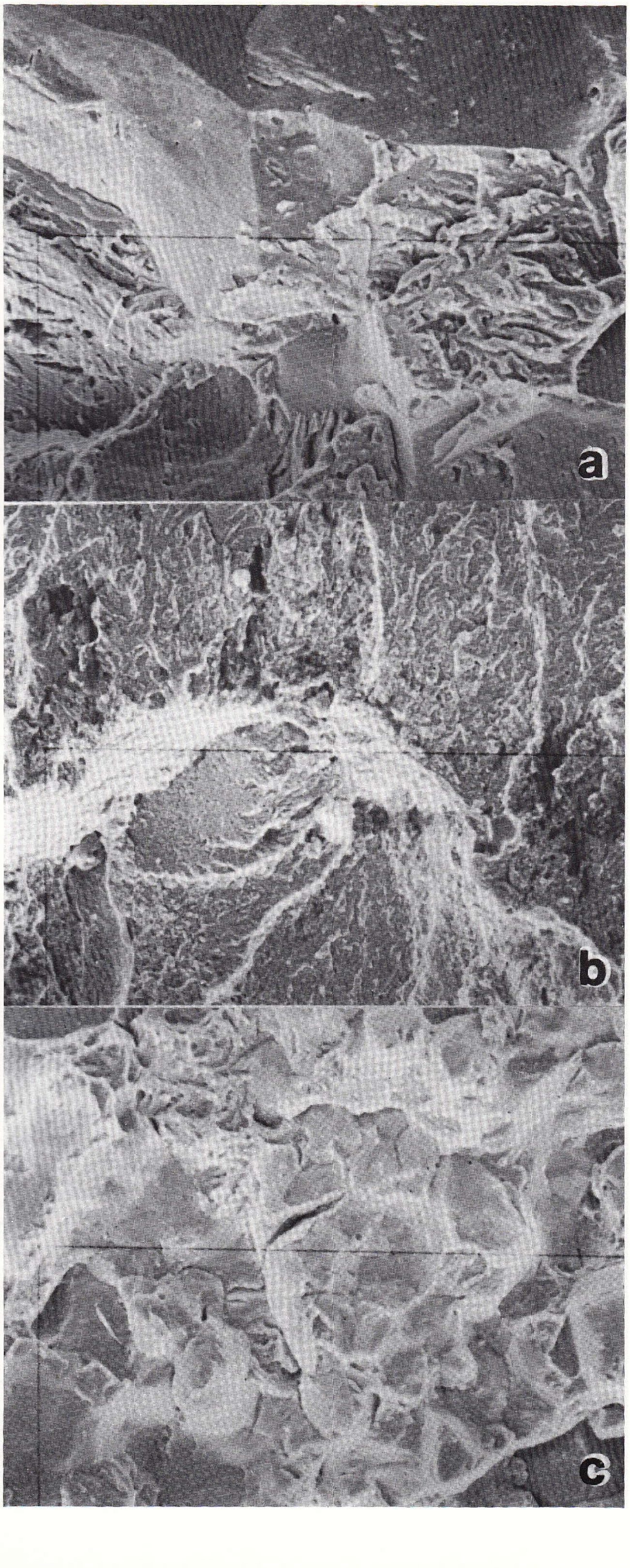


TABLE 2 - Mechanical properties of the investigated hot-work tool steels

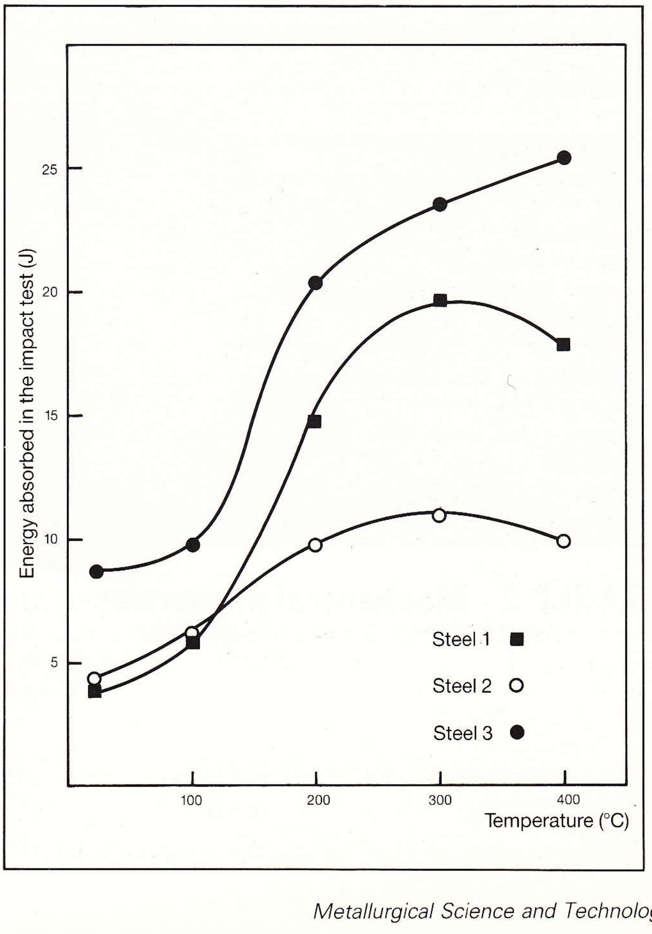
	HRC	σ_y (N/mm^2)	σ_{UTS} (N/mm^2)	Elongation (%)	Reduction of area (%)
Steel 1	44.8	1346	1535	5.88	9.69
Steel 2	40.6	1118	1162	0.38	0.67
Steel 3	43.7	1294	1480	3.51	8.68

Fig. 4 - Microfractographic appearance of room temperature toughness specimens; a) steel 1, b) steel 2, c) steel 3, (× 500).



fracture toughness is almost the same for all the three steels; however, while fracture toughness data of steels 1 and 3 are close to the lower bound of the scatter band, fracture toughness data of steel 2 are rather low compared with data for wrought H 11 + H 13 steels at the same hardness level.
Fig. 4 shows the appearance of the fracture surfaces of the three steels, close to fracture initiation from the fatigue precrack; in steels 1 and 3, fracture at room temperature is of an almost brittle intergranular type, with some quasi-cleavage areas, while in steel 2 the fracture propagated prevalently in a quasi-cleavage manner, mainly through the interdendritic regions.
The results of the impact tests are given in Fig. 5, which shows that the temperature of trasition from brittle to ductile behaviour is almost the same for all three steels; the transition temperature is very close to that reported for precracked Charpy-V specimens (8) but shifted slightly toward higher temperatures compared with the data for wrought H 11 steel Charpy-V samples (7), in consequence of the sharper notch-root radius. However, the level of energy absorbed in the impact tests is different for the three steels, and

Fig. 5 - Temperature transition curves of Charpy-V type specimens having the notch root radius equal to 0.05 mm.

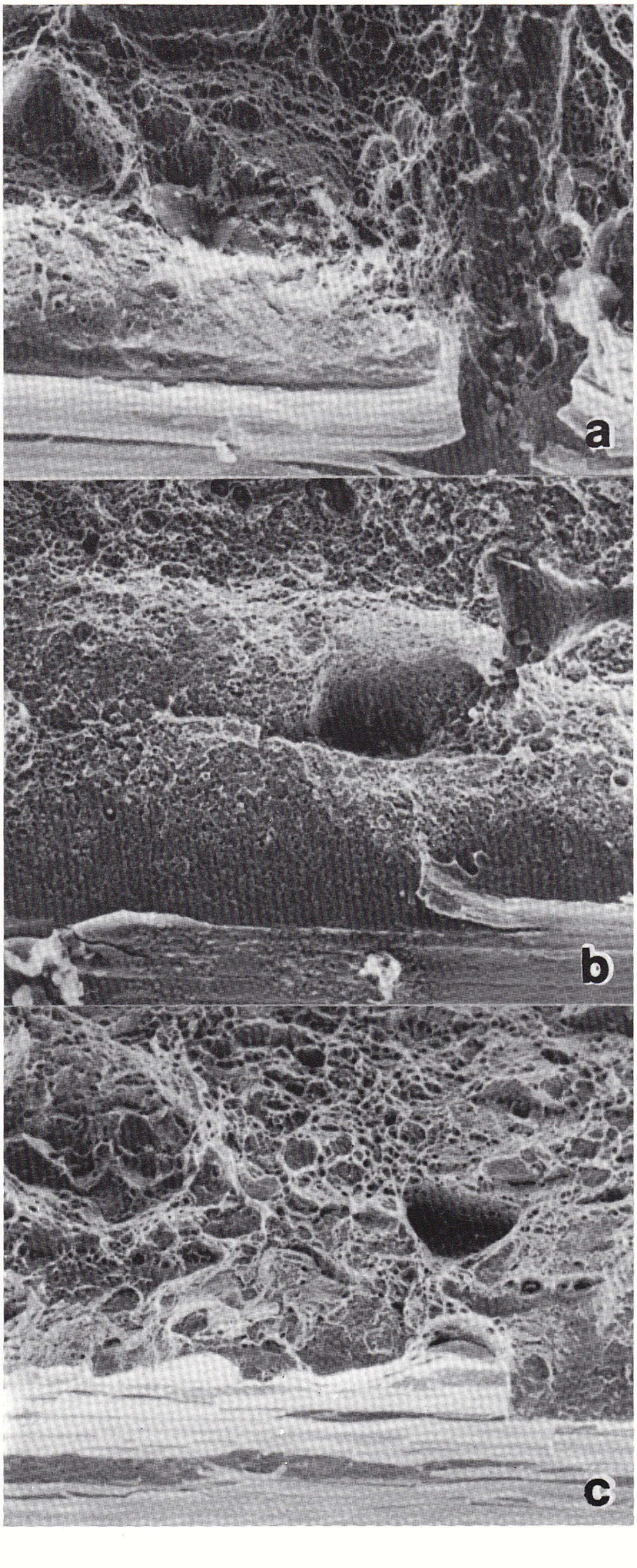


must be differently interpreted for the brittle and the ductile regions.
Infact, in the brittle region, steels 1 and 2 show the same absorption of impact energy, in contrast to steel 3, which absorbed almost double the energy; such behaviour can be understood by considering the model proposed by Ritchie et al. (9, 10) linking data for sharp-cracked samples with results for blunt-notch samples. Such a model is applicable to strees-controlled fractures, like the brittle intergranular or interdendritic fractures in the experimented steels, and is expressed by the following relationship:

$$K_{Ic} \simeq 2.9\,\sigma_y \left[\exp\,(\sigma_f/\sigma_y) - 1\right]^{1/2} \cdot \rho_{eff}^{\ 1/2}$$

Where σ_y is the yield strength, σ_f the critical fracture stress and ρ_{eff} the limiting value of the notch-root radius below which K_{Ic} is constant and independent of ρ. ρ_{eff} is a measure of the characteristic distance over which, according to the theories developed by Tetelman et al., (11, 12), the critical fracture stress must be exceeded in order to initiate crack propagation.
As pointed out in previous papers (13, 14), in brittle intergranular fracture of martensite structural steels, the controlling parameter is the austenitic grain size, and therefore the notch-root radius employed is much too great for steel 3 to have the same effect as a crack-like notch; consequently more energy is absorbed. On the other hand, toughness tests indicate that the plane strain fracture toughness is almost equal for all three steels, notwithstanding the increase in ρ_{eff} from steel 3 to steel 2 (for which the dendrite size can be supposed to represent a measure of the characteristic distance); however, this can be explained considering that, for the test steels, with increasing ρ_{eff} there may be a drop in σ_f, which is easily related to segregations at the grain boundaries of larger austenitic grains in steel 1 and at the dendritic boundaries in steel 2. Consistently with the above explanation, in steel 3 impact-test specimens, the initial part of crack propagation from the notch-root followed a slip line, in accordance with previous results for martensitic-structure steel specimens with different notch-root radii (15).
Regarding the ductile rupture region, on the other hand, it must be considered that fracture is governed by a critical strain criterion and that the microstructure parameters playing a role in controlling fracture toughness are related to inclusions, around which microvoids nucleate (4); fracture surface of impact specimens tested at 400°C are illustrated in Fig. 6, where similar fracture appearances can be observed for steels 1 and 3, while steel 2 shows a large number of small microvoids among the larger microvoids around which the fracture nucleated.
The lower impact energy level of steel 2 may be related to its higher sulphur content and, in this respect, the present results agree with previously published data for

Fig. 6 - Microfractographic appearance of impact specimens tested at 400°C; a) steel 1, b) steel 2, c) steel 3; (× 250).



wrought H 13 steel (8).
As regards the impact-test results for steels 1 and 3, the difference in the absorbed energy can not be fully explained at present, as impact data include both initiation and propagation contributions. Initiation energy or ductile fracture toughness, for the same steel with the same inclusions content and distribution, should be the same, and therefore the difference in total absorbed impact energy should relate to weak areas along the crack-propagation path in steel 1.

Conclusions

The fracture toughness of cast H 11 hot-work tool steel has been evaluated as a function of different microstructure conditions, by means of fracture mechanics tests at room temperature and impact tests over the 20-400°C temperature range.
Results showed that fracture toughness at room temperature is only slightly lower than that for wrought steel at a comparable hardness level, provided that no heavy segregations and dendritism affect the microstructure of the steel. In brittle fracture, the microstructure parameter controlling fracture toughness is the austenite grain size; but segregations at austenite grain boundaries or in the interdendritic regions, also play an important role in controlling the critical fracture stress.
In ductile rupture, fracture toughness is controlled mainly by the amount and distribution of non-metallic inclusions - particularly sulphides - around which microvoids nucleate; therefore, ductile fracture toughness is controlled chiefly by the sulphur content of the steel.

Acknowledgements

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