EFFECT OF TITANIUM NITRIDE PARTICLES ON THE TOUGHNESS OF ENGINEERING STEELS

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Sharp cracks originated by cracking micron size TiN particles present in microalloyed ferrite-pearlite steels are shown to promote cleavage initiation. By developing appropriate microstructures which affect to the type of interfaces particle-matrix or increase the number of obstacles in form of grain boundaries, the propagation of the crack in brittle manner can be stopped and this improves the fracture toughness of these materials.

INTRODUCTION

The use of microalloying in medium C forging steels together with thermomechanical treatments developed for low C steels is becoming a route for avoiding the expensive heat treatments (quenching + tempering) and for obtaining required optimum mechanical properties directly after forging. Good strength levels have been obtained without difficulties but for the same strength level, toughness is better in quenched + tempered steels. In the last years the efforts have been concentrated in improving the toughness by refining the microstructure, and for that, Ti addition is becoming common method (inhibition of grain coarsening by small TiN precipitates). Several recommendations are made to obtain a fine TiN precipitation, but a certain proportion of TiN precipitates is usually in coarse form. These coarse particles can impair the toughness, being the origin of cracks which can propagate in a brittle manner. In this work it is shown that by changing the microstructure of these steels brittle failure at room temperature can be avoided.

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MATERIAL AND EXPERIMENTAL PROCEDURE

Two steels (chemical compositions in Table 1) were supplied by AFORASA in the form of 50mm square bars, obtained by hot rolling continuous cast 130 mm square billets in case of the Ti steel and by conventional casting to an square ingot and hot rolling to bar in the case of the Ti-V steel. Plane strain compression test specimens 50x50x10 mm were machined from the bars, and after reheating (30 min., 1100°C), were hot deformed to $\varepsilon = 0.22$ at 1000°C or 0.38 at 850°C at constant strain rate $\dot{\varepsilon} = 1s^{-1}$ and air cooled. Other specimens were hot deformed (1000°C, $\varepsilon = 0.22$, $\dot{\varepsilon} = 1s^{-1}$) and then accelerated cooled (4 ÷ 10°C/s). Tensile and J_{Ic} tests (three point bending, Bx2B, ASTM E813, B = 6 mm) were performed for hot deformed specimens at room temperature.

TABLE 1 - Chemical Composition of the steels in wt %

steel	C	Mn	Si	P	S	V	Al	Ti	N (ppm)	Ti/N
Ti - V	0.37	1.45	.56	.010	.043	.11	.024	.015	162	0.9
Ti	0.35	1.56	.33	.004	.007	-	.027	.028	89	3.1

RESULTS AND DISCUSSION

The addition of Ti to these steels results in coarse TiN precipitates formed in the interdendritic liquid as well as fine TiN precipitates (1, 2). The steels contain a significant proportion of particles with sizes between 1 and 3 µm. Figs. 1a and b show typical microstructures after slow cooling (ferrite and pearlite) and accelerated cooling (acicular ferrite). Tensile properties and relevant fracture toughness values are given in Table 2. Materials slow cooled after hot working at 850°C (indicated as: ferrite-pearlite, hw 850°C + sc) presenting a fine ferrite-pearlite microstructure and materials fast cooled (indicated as acicular ferrite in Table 2) presented stable crack propagation to yield valid J_{Ic} data. In the rest of the cases, cleavage fracture happened after a very small ductile crack propagation (<< 200µm). Fractography analysis showed that in all the cases of brittle fracture in the two steels the origin of the cleavage was associated with TiN particle fracture (particle/matrix debonding was never observed as a responsible of brittle fracture).

It is also observed in Table 2 that $\sigma_{0.2\%}$ and UTS values are higher in the case of the Ti-V steel due to V carbide precipitation strengthening. It is also apparent that precipitation hardening originates lower toughness values for equivalent process conditions.

The large scatter observed in the fracture toughness results for the as-rolled condition (more important in the case of the Ti-V steel) could be due to the fact that at room temperature the ferrite-pearlite coarse microstructures are in the ductile-brittle transition region. In the transition region a crack propagates initially in a ductile way by microvoid coalescence (Fig. 2). Latter on, if a TiN particle is found in the highly stressed region ahead of the crack front, a sharp crack would be generated by cracking the TiN particle, which could be the origin of the brittle cleavage fracture. For such event to be succesfull several steps are required (3).

TABLE 2 - Mechanical properties of the different structures studied

steel	struct.	process	$\sigma_{0.2\%}$ (MPa)	UTS (MPa)	J_c (KPa.m)	K_c (MPa√m)	J_{Ic} (KPa.m)	K_{Ic} (MPa√m)
Ti - V	acicular	hw 1000°C+fc to700°C+sc	666	875			76 102	133 154
		hw 1000°C +fc to rt	560	920			103 102	155 154
	ferrite-pearlite	hw 1000°C +sc	647	910	21.4 18	71 + 65 +		
		hw 900°C + sc	-	-	81.2 86.4	137.5 141.9		
		hw 850°C +sc	-	-			76.6 83	134 139
	ferrite-pearlite	as-rolled	590	875	21.7 43.5 12.8 64	67 +* 101 55 + 122		
Ti	acicular	hw 1000°C + fc to rt	519	819			133 123	176 169
	ferrite-pearlite	hw 1000°C + sc	-	-	107.1 62.3	158 120.5		
		hw 850°C + sc	-	-			113 123	162 169
		as-rolled	440	740	45 71.3	96.5 * 121.5 *		

hw, hot worked; **fc**, fast cooled; **sc**, slow cooled; **rt**, room temperature
* minimun thickness condition no fullfilled for K_{Ic} tests; + pop-in during test

First, it is necessary to break the particle and create a sharp crack (step 1). There is a second step, connected with the propagation of the crack created in the TiN particle to the surrounding matrix (step 2), in which the crack must surmount a certain energy barrier (G_{pm}) opposing propagation across the interface particle/matrix. The third step is the traversing of a high angle boundary in the matrix, which requires the crack to surmount an energy barrier, G_{mm}. Referring to the first step, it seems clear that larger particles will break at lower stresses than small particles (Weibull), but this does not mean that the crack will progress through the material. This will happen only if the second step can take place. The stress necessary to surmount the energy barrier G_{pm} around a circular sharp crack created in TiN particle of size "a" is (3):

$$\sigma_F^* = \left(\frac{\pi E G_{pm}}{(1-\nu^2)2a} \right)^{1/2} \qquad (1)$$

Values of G_{pm} have been measured experimentally (4) for the two alloys by measuring the fracture strength at liquid nitrogen temperature and identifying unambiguously the particle responsible for the brittle fracture initiation. From the position of cleavage origin and using the stress distribution as a function of the distance in front of a round notch (5), Fig. 3 can be plotted and the value of G_{pm} obtained from the slope results in the range 20 - 40 J/m^2. This means that stresses above 1900 - 2700 MPa are required to propagate a crack created in a particle of 2 μm in size. In Fig. 4 is shown a case of no brittle crack propagation through the boundary of a crack generated by a break of a TiN particle. This is a case commonly observed in the acicular microstructures of the two steels, in wich the stresses are not high enough for promoting the propagation of the crack through the interface. If the crack can trespass the interface, it can then find new obstacles in form of grain boundaries. In a recent work (4) the values of G_{mm} have been estimated, resulting in values lower than 200÷360 J/m^2. Stresses higher than 2700-3600 MPa are required to trespass grain boundaries corresponding to grain sizes of 10μm. If the stresses are lower than those or if for the same stresses the grain sizes are smaller (refined microstructure) the grain boundaries can stop the advance of the crack, as shown in Fig. 5, in which is clearly observed how the propagation of the crack has been stopped by the grain boundaries leaving brittle island totally surrounded by ductile fracture.

CONCLUSIONS

- In the two microalloyed medium C forging steels Ti addition produces coarse TiN particles which contribute to cleavage fracture. In coarse ferrite-pearlite microstructures TiN particles are the origin of the overall brittle fracture.

- By modifying the microstructure by thermomechanical treatments, although in the same coarse TiN particles cleavage fracture can be initiated, fine ferrite-pearlite or fine acicular ferrite microstructures stop the growing of cracks nucleated in TiN particles and ductile fractures occur.

ACKNOWLEDGMENTS

This work was carried out in collaboration with AFORASA under ECSC contract 7210-Ma/938. The authors wise to thank the Community for the economical aportation. M.A. Linaza and J.L. Romero acknowledge financial support from the Basque Government and Spanish Government respectively.

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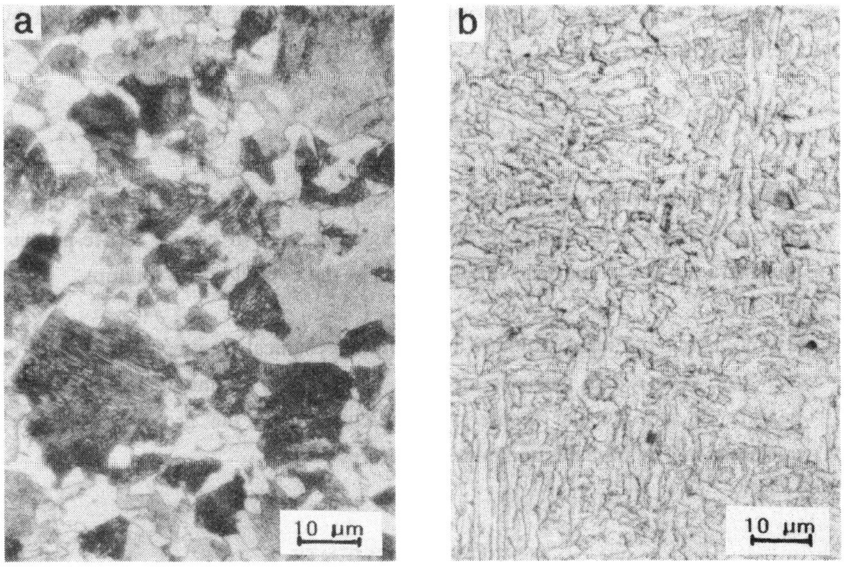

Figure 1 Microstructures of Ti-V steel: a) hot deformed at 850°C and slow cooled to room temperature and b) hot deformed at 1000°C and fast cooled.

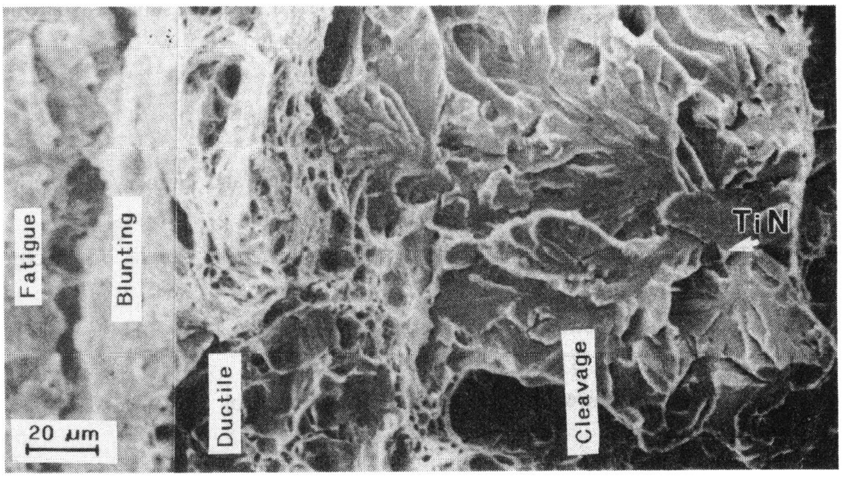

Figure 2 Fracture propagation sequence in the transition zone. Cleavage has been initiated in a TiN particle.

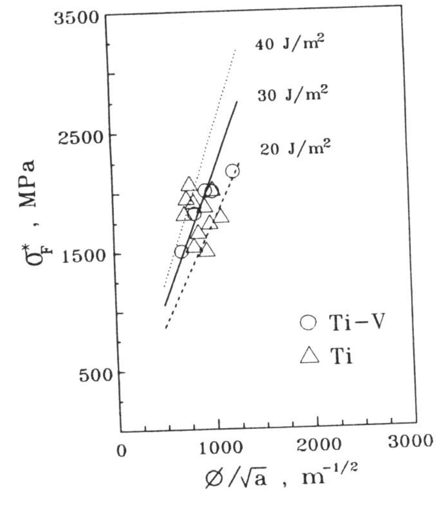

Figure 3 $\sigma_F{}^*$ vs reciprocal square root of the size of TiN particle which produced brittle fracture initiation.

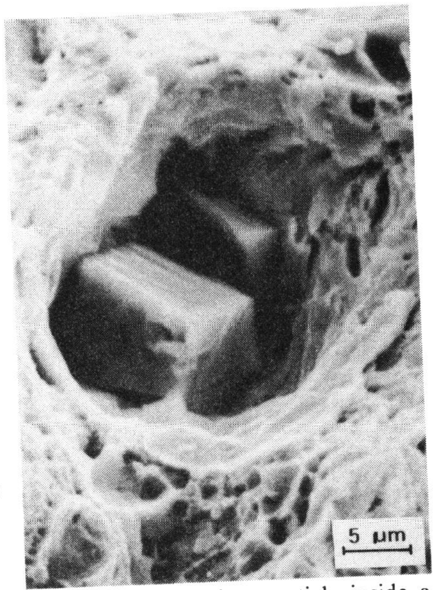

Figure 4 TiN broken particle inside a dimple (Ti-V steel).

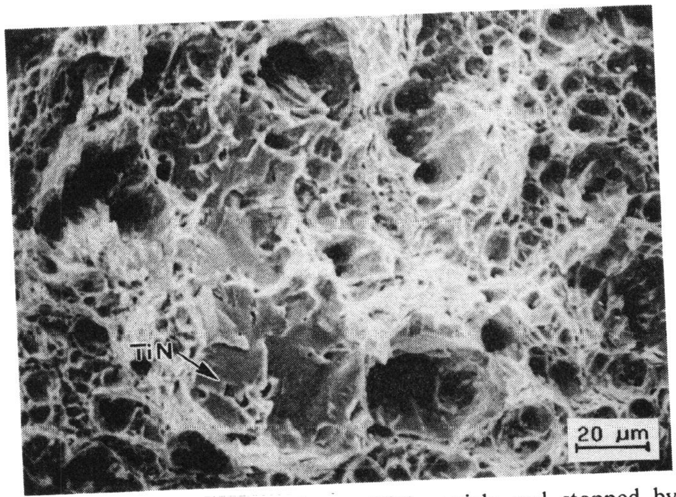

Figure 5 Cleavage fracture initiated at a TiN particle and stopped by ductile micromechanisms (Ti-V steel, hot worked at 900°C and slow cooled).