FRACTURE MECHANISMS AND DUCTILITY AT HIGH TEMPERATURES OF A CARBON STEEL

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ABSTRACT

The hot ductility and fracture behaviour of a C-Mn steel susceptible to transversal cracking in the corners is studied. In order to evaluate the influence of residual elements and inclusions, this steel was compared with another one with the same main composition obtained by electro-slag from almost pure iron. Tensile tests at temperatures ranging from 1000ºC to 650ºC and an initial strain rate of $1 \cdot 10^{-3} s^{-1}$ were carried out for reheated samples of both steels. The reduction in area of the samples tested to fracture was used as a measure of the hot ductility. For each temperature, the ductility behaviour was related to the phases present and to the fracture mechanisms acting. For this purpose the fracture surfaces of the tensile specimens were examined by scanning electron microscopy .

The two steels studied showed a very different ductile behaviour. The ductility trough for the steel with a high residual content was found to be very wide and deep. In the low ductility zone, the fracture appearance for temperatures above A_3 was intergranular decohesion, meanwhile for temperatures below A_3 the fracture was also intergranular with the existence of dimples in the grains surfaces. For this two-phase ($\alpha+\gamma$) temperature region even interdendritic fracture was observed. The electro-slag cast steel showed a narrow ductility trough located in the two-phase region and there fracture was intergranular with shallow dimples in the surface of the grains. The fracture mechanism that could be acting in the two-phase region for both steels could be the formation of a thin film of ferrite in the austenite grain boundaries, which would lead to a strain concentration as the ferrite is softer than the austenite and can recover dynamically at these temperatures. For temperatures in the single-phase austenite region, only the steel with a high residual content exhibited a poor ductility, so this brittle behaviour could be related to the existence of a high volume fraction of MnS inclusions and/or to a possible segregation of elements such as Cu or Sn.

1 INTRODUCTION

The hot ductility of steel has been the object of several studies as it is related to the presence of transverse cracks in the surface of some continuously cast products. These cracks propagate during the unbending operation, when the top surface of the strand is placed in tension at temperatures and strains rates for which most steels present a poor ductility, Mintz [1]. Hot tensile test is conventionally used to simulate the conditions of the unbending operation and the reduction in area of the samples tested to fracture is taken as a measure of the ductility.

Even though microalloyed grades give most of the troubles in terms of hot ductility, plain C-Mn steels also present deep and wide troughs of ductility, Crowther [2]. In the present case, a low carbon steel susceptible to the presence of transverse cracks in the corners is studied to determine the origin of such cracks and relate them to the hot ductility and the fracture mechanisms acting during high temperature conditions. This steel had a high residual content in elements such as Cu and Sn, and its Mn/S relation was about 42, thus to determine the influence of the cleanliness in the hot ductility of the steel, a second one, residual-free steel was also tested.

2 EXPERIMENTAL METHOD

As already mentioned, two C-Mn steels have been examined. Steel A, a commercial one, has a 0.23%C and a high content in residual elements such as copper, which is present in 0.59 %weight. It also has a relatively high S content. Steel D, a laboratory one, has a similar composition in C, Mn and Si, but contains no residual elements as it was obtained by electro-slag casting from almost pure iron by adding ferro-manganese and ferro-silicon in the right proportion. Composition for both steels is given in table 1.

Table 1 Chemical composition of steels A and D

Steel	%C	%Si	%Mn	%P	%S	%Cr	%Ni	%Cu	%Sn	Mn/S
A	0.23	0.125	0.9	0.011	0.021	0.075	0.121	0.59	0.053	42
D	0.23	0.13	0.82	0.025	<0.01	-	-	-	-	>82

Tensile specimens of 6 mm in diameter were machined for both steels with their tensile axis parallel to the casting direction. Samples were reheated in a radiation furnace to 1100ºC for 10 minutes, then cooled down to the test temperature, which ranged from 650ºC to 1000ºC, and held at this temperature for 5 minutes before starting the tensile test at a strain rate of $1{\cdot}10^{-3}s^{-1}$. The reduction in area of the samples tested to fracture was taken as a measure of the hot ductility. In order to relate the ductility behaviour with the phases present at each temperature, the transformation temperatures where calculated using the formula of Andrews, Andrews [3]. Both steels had the same A_1 which was 716ºC, and A_3 was 785º and 814ºC for steel A and steel D respectively. The fracture surfaces were examined using a 6400 Jeol Scanning Microscope.

3 RESULTS

The hot ductility curves for both steels show a great difference in their behaviour as shown in Figure 1. Steel A shows a wide and deep ductility trough. The ductility starts decreasing at temperatures above A_3 (785ºC), in the single-phase austenite region, the ductility remains lower than 20% as temperature decreases and only a slight recovery is observed at 650ºC where a major volume fraction of ferrite is supposed to exist. For steel D the trough is narrower and all the ductility values are larger than 50% indicating that the steel would not be susceptible to transverse cracking during the unbending operation. In this particular steel the ductility loss at high temperatures and its recovery at

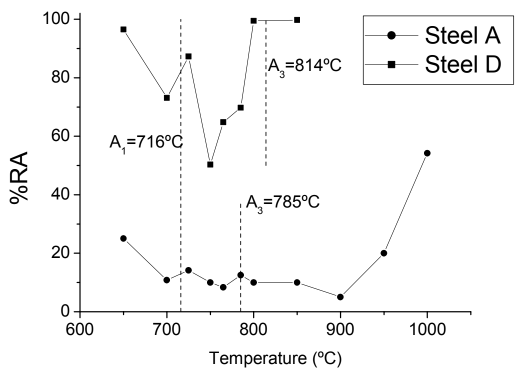
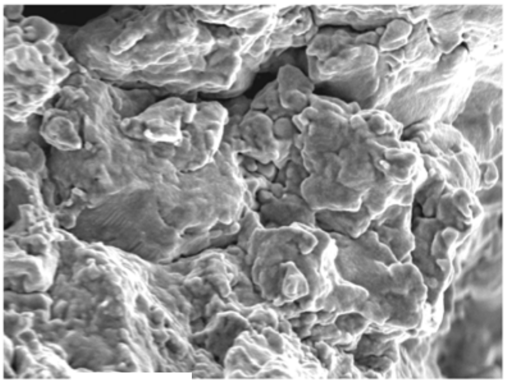


Figure 1 Ductility curve for steels A and D

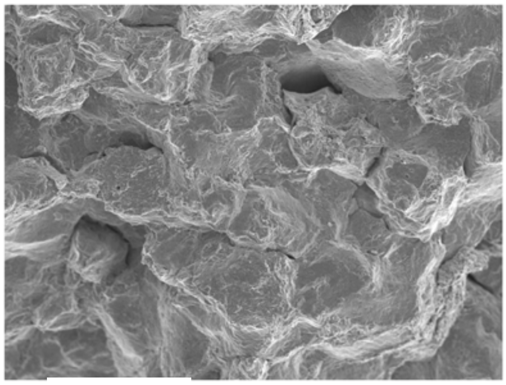
low temperatures seem to be related to the transformation temperatures A_3 and A_1 respectively. This is not the case with steel A.

The fracture appearances should be related to the dominant deformation mechanism and therefore to the test temperature and phases present. The best ductile behaviour for steel A was at 1000ºC and the fractography of the sample tested at this temperature showed an intergranular failure where the surfaces of the grains looked smooth and the existence of some little grains could be indicating that the crack propagated between new grains formed during dynamic recrystallisation (see Figure 2a). For temperatures in the single-phase austenite region where the ductility was very low, the fracture appearance was also intergranular with decohesion of the austenite grains as seen in Figure 2b for the sample tested at 850ºC. In the two-phase region ($\alpha+\gamma$), the fracture surface was also intergranular with the existence of some voids around MnS inclusions, Figure 2c. In this range of temperatures, there were samples that showed some interdendritic fracture. Figure 2d shows the surface of a dendrite with MnS inclusions that could be responsible for this kind of brittle fracture. This fracture surface reminds the surface of midway cracks which are smooth without any evidence of brittle or ductile fracture, Brimacombe [4].The corner cracks found in the billet of steel A had a similar fractography appearance.



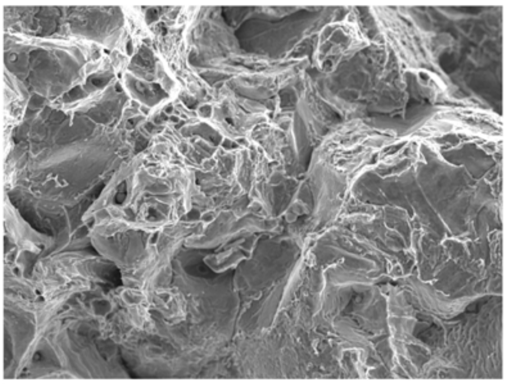
60 μm

Figure 2a Sample tested at 1000ºC



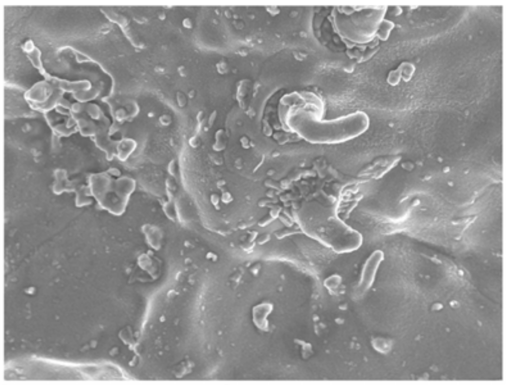
300 μm

Figure 2b Sample tested at 850ºC



100 μm

Figure 2c Sample tested at 725ºC



60 μm

Figure 2d Dendrite surface with MnS

In steel D the fracture modes are totally different. At high temperatures where the steel shows a good ductility, the fracture is due to the coalescence of voids as displayed in Figure 3a for the sample tested at 800ºC. In the two-phase region, there is a loss of ductility and the fracture is intergranular with the existence of shallow dimples that increase in volume fraction as the temperature decreases (Figure 3b). The ductility is recovered at temperatures below A_1 and for these temperatures the fracture occurs again by void coalescence.

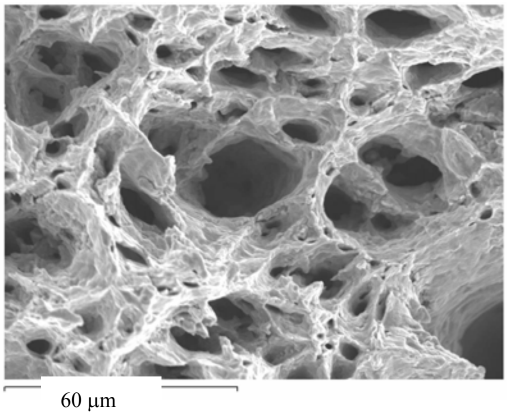


Fig. 3a Sample tested at 800ºC

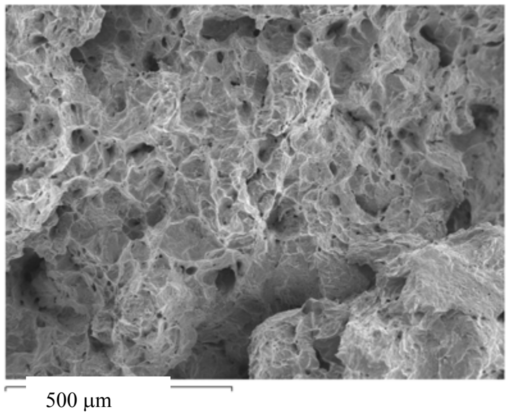


Fig. 3b Sample tested at 765ºC

4 DISCUSSION

Different ductility behaviours are observed for each one of the two steels studied. In the high temperature range ($T>A_3$), only steel A exhibited low ductility and the best ductility for this steel was only slightly better than the worst ductility for steel D, i.e. the ductility trough for steel A is deeper and wider in comparison with the one of steel D.

For both steels the ductility recovery at high temperatures can be associated with the apparition of the softening mechanism of dynamic recrystallisation. The stress-strain curves in Figure 4 show that the ductility improvement correspond to the temperatures for the onset of dynamic recrystallisation. For steel A this temperature corresponds to the highest test temperature of 1000ºC while for steel D this temperature was 800ºC. The later apparition of recrystallisation for steel A can be attributed the high sulphur content in this steel, as it raises the temperature for the onset of dynamic recrystallisation, Mintz [5]. This softening mechanims prevents cracks formed by grain boundary sliding from joining up since grain boundary migration is acting Mintz [5].

The loss in ductility for temperatures well in excess of A_3 in steel A can only be controlled by the presence of MnS inclusions in the steel and/or by a possible segregation of residual elements, mainly Cu and Sn which are the more detrimental. The sulphides present in the steel may embrittle the austenite grain boundaries, encouraging cracks formed to link up more readily. The result is an intergranular fracture with the appearance of the one shown in Figure 2b. Moreover, the MnS would raise the temperature at which dynamic recrystallisation occurs, Abushosha [6]. But a possible segregation at the grain boundaries of Cu and Sn must also be considered. It is reported by some authors, Nachtrab [7] that a segregation of these impurities has the effect of lowering the grain boundaries surface energy and increases the rate of grain boundary cavity formation and growth. Cu seems to deepen ductility troughs, Cooper [8] for steels with high Cu content (though lower than in the present case for steel A). Therefore in the austenite zone, the combination of inclusion with a possible impurity segregation can lead to a severe embrittlement and ductilities as low as 5%.

Poor ductility in the two-phase region can be caused by the apparition of the softer phase ferrite at the austenite grain boundaries. This phase can be deformation induced and can appear at temperatures just below the equilibrium transformation A_3. As shown in Figure 2c for steel A, and in a similar way for steel D, the ductile intergranular fracture is by the coalescence of voids that form and grow around the MnS inclusions in the ferrite as this phase can easily recover leading to a concentration of the deformation in it. Steel A with a great volume fraction of inclusions shows the worst ductility behaviour in this temperature range.

It is worth noting an additional mechanism that took place in some samples for steel A when tested to fracture, that is interdendritic fracture. The general aspect of a fracture surface showing interdendritic decohesion is shown in Figure 5 for the sample tested at 700ºC. The surface of the dendrites was like the one observed in Figure 2d. This kind of fracture has the same appearance as midway cracks of billets formed by "hot tearing" at temperatures 70ºC to 100ºC below the solidus temperature, associated to a zone of low ductility, Brimacombe [4] and Lankford [9]. In this region, sulfur and other solutes segregate between growing dendrites and lower the solidus temperature locally. In combination with iron, sulphur forms FeS, which solidifies at 1200ºC. The Mn:S ratio for steel A may not be high enough to avoid the formation of this phase, then severe hot tearing may be acting which forms internal cracks that remain even during the testing and lead to interdendritic decohesion.

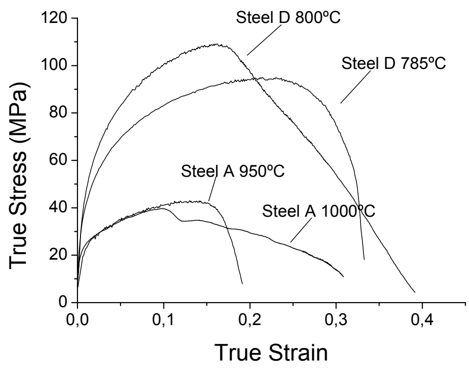


Figure 4 Stress-strain curves for both steels at different temperatures

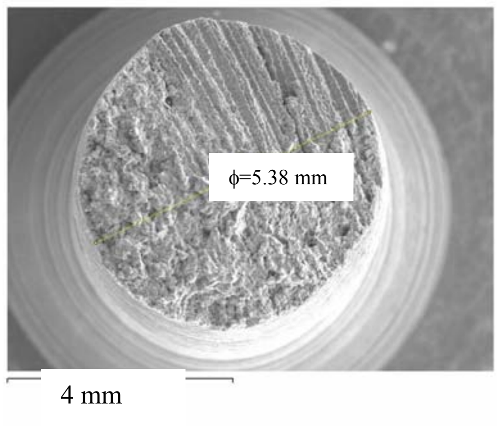


Figure 5 Fracture surface showing interdendritic decohesion

5 CONCLUSION

The combination of a high volume fraction of MnS with a possible Cu and/or Sn can lead to wide and deep ductility troughs. The width of the trough was controlled by MnS particles which raised the temperature for the onset of dynamic recrystallisation to 1200ºC. The depth of the trough could be attributed to the effect of a possible Cu segregation to the austenite grain boundaries.

Hot tearing during the production of steel is to be avoided in order to reduce the incidence of cracks formed by a lack of interdendritic cohesion. A higher Mn:S ratio could be helpful for steel A.

With no residual elements and a low %S the only embrittling mechanism for temperatures between A_1 and A_3 was the concentration of the deformation in the ferrite formed around the prior austenite grains as ferrite can recover easily for temperatures at which both phases are present.

Acknowledgements
This work was supported by grants for the DURSI and from the MCYT (DPI2002-04479-C02). We wish to thank the CENIM for the use of their facilities to produce the material.

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