EFFECT OF THE TEST CONDITIONS ON THE HOT DUCTILITY AND FRACTURE MECHANIMS OF A C-Mn STEEL

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RESUMEN

El agrietamiento transversal en la superficie de algunos productos de colada continua es un problema que sigue teniendo una elevada incidencia y que depende en gran medida de la composición del acero. Para aceros procedentes del reciclaje de chatarra este problema se relaciona con la presencia de elementos residuales e impurezas tales como el Cu, Sn o S. Teniendo en cuenta que es previsible un aumento en el contenido de estos elementos a medida que el acero entre sucesivamente en el ciclo del reciclaje, es importante conocer los mecanismos de fragilización que éstos pueden inducir y determinar en qué condiciones pueden actuar. Para ello se escogió una calidad de acero industrial C-Mn con aproximadamente 0.5%Cu, 0.05%Sn y 0.02%S y que presenta problemas de agrietamiento superficial durante su colada continua en forma de palanquilla. Las condiciones termo-mecánicas a las que se pudo ver sometida la superficie de la palanquilla durante el proceso industrial se simularon mediante ensayos de tracción en caliente. La reducción de área (%RA) de las probetas una vez ensayadas se utilizó como parámetro para determinar las condiciones a las que el acero sería susceptible de presentar agrietamiento transversal. Los ensayos se llevaron a cabo para dos condiciones iniciales del material: en estado de colada, mediante probetas extraídas directamente de una palanquilla y en estado laminado, mediante probetas extraídas de varilla corrugada. Además, las probetas fueron sometidas a dos ciclos de austenización diferentes previos al ensayo, el primero comprendía un recalentamiento a 1100°C y el segundo a 1330°C. En términos generales se observa que el ciclo térmico no afecta significativamente a las curvas de ductilidad, a pesar de que las superficies de fractura muestran diferentes mecanismos de daño. Sin embargo, la condición inicial del material tiene una influencia más marcada en las curvas de ductilidad, dándose pozos de ductilidad más estrechos cuando el acero ha sido previamente laminado. El comportamiento del acero en función del ciclo térmico se relaciona con el efecto que éste puede tener en las segregaciones de las diferentes especies, principalmente el S y el Sn, mientras que las diferencias en las curvas de ductilidad que supone la variación en la condición inicial del material estarían más relacionadas con la morfología y distribución de los sulfuros dentro del material.

ABSTRACT

Transverse cracking on the surface of some strands is a problem related to the straightening operation of continuous casting that widely depends on the steel chemical composition. For those steels produced by scrap recycling, transverse cracking is related to the presence of residual elements and impurities such as Cu, Sn or S in its composition. These elements, which can not be easily removed, increase their content as steel re-enters the recycling cycle. Therefore, it is important to know the embrittlment mechanisms, related to the presence of residual elements and impurities, which can deteriorate the ductility of steel. In this work, a structural C-Mn steel with 0.23%C, 0.9%Mn, 0.5%Cu, 0.05%Sn and 0.02%S was studied. This steel grade is susceptible to transverse cracking on its surface during billet continuous casting. The hot tensile test was used to simulate the thermo-mechanical conditions experienced on the surface of the strand during the straightening operation. The reduction in area (%RA) of the samples tested to fracture was taken as a measure of the ductility and thus, of the susceptibility to present transverse cracking. Tests were carried out under two different initial microstructures, i.e. as-cast and as-rolled condition. Moreover, the samples underwent two different reheating temperatures of 1100°C and 1330°C previous to the test. The results showed similar ductility curves for samples tested after different reheating temperatures, even though the fracture features changed. On the other hand, the initial microstructure of the material had a strong effect on the ductility curve, the as-rolled structure promoting a narrower ductility trough. The hot ductility behaviour of the steel as a function of the thermal cycle was found to be related to its effect on the segregation patterns of the S and Sn. The differences introduced by the initial microstructures can be explained in terms of the morphology and distribution of the sulphides.

THEMATIC AREA: Metallic Materials Fracture.

KEYWORDS: C-Mn steel, continuous casting, hot ductility, residual elements, fractography.

1. INTRODUCTION

Transverse cracking in the surface is a problem related to the continuous casting steelmaking process. During the straightening operation, the top surface of the slab is tensile tensioned at temperatures and strain rates at which most steels present poor ductility [1]. The susceptibility to transverse cracking depends on operational conditions as well as on the chemical composition of the steel being produced.

Steel grades produced from scrap recycling have high residual elements and impurities contents [2]. It has been shown that residuals, which can not be easily removed from the steel, impair the hot ductility of steels. By increasing the recycling tendency of steel products in the future, the problems related to residual elements will become more serious and therefore it will necessary to determine the conditions under which these elements participate in embrittlement mechanisms.

The hot tensile test is a useful method to simulate the thermo-mechanical conditions on the surface of the strand during the straightening operation in continuous casting. Optimal testing conditions have been set to evaluate the hot ductility of microalloyed steels, but these testing conditions need to be reviewed to guarantee a correct evaluation of the hot ductility of steels containing residual elements since they promote different embrittling mechanisms. Their embrittling effect is mostly related to their tendency to segregate both to interdendritic spaces during the solidification of the steel or to grain boundaries under certain thermal conditions.

In this work, the effect of different reheating temperatures, 1100°C and 1330°C, as well as different initial conditions of the material, as-cast and as-rolled, on hot ductility have been compared for a C-Mn steel with high residual and impurities contents.

2. EXPERIMENTAL METHOD

The study has been carried out on an industrial C-Mn steel with high residual elements level, especially Cu and Sn. Its hot ductility has been evaluated for two initial conditions of the material, the as-cast and the as-rolled. In the as-cast condition the samples were machined directly from the billet with their longitudinal axe parallel to the casting direction. This steel will be referred in the text as steel A1. In the as-rolled condition the samples were machined from the final product that was a corrugated bar, referred as steel A2. The chemical composition of the steels has been shown in table 1; as can be seen their composition is very similar. They have high amount of Cu and no enough Ni has been added to alleviate its negative effect. The amounts of Sn and S are also relatively high.

The tensile specimens were cylindrical with 9.5 mm in diameter and 125 mm in length. Previous to the hot tensile tests the samples followed an austenitizing thermal cycle that consisted on a reheating at 1100°C or at 1330°C for 5 minutes. Both thermal schedules are outlined in Figure 1. The reheating at 1100°C induced a grain size of 67μm for steel A1 and 55μm for steel A2, whereas at 1330°C being 344μm for steel A1 and 324μm for steel A2. Thermal cycles were applied to the specimens using an induction heating system. Specimens were fixed between the anvils of a fully computerized MTS machine and placed in a quartz tube under atmosphere of argon with 1% hydrogen to minimise oxidation. The temperature was measured by an infrared pyrometer focused on the middle section of the specimen and adjusted by a controller in accordance to the programmed thermal schedules.

Steel	C	Si	Mn	P	S	Ni	Sn	Cu
A1	0.23	0.12	0.9	0.01	0.02	0.12	0.05	0.59
A2	0.21	0.13	0.87	0.01	0.03	0.11	0.03	0.51

Table 1. Chemical composition (% in mass)

a)

$2°C/s$ 1100°C - 5' $10°C/s$ $\dot{\varepsilon} = 1\cdot10^{-3}\,s^{-1}$
$T = 650 - 1100°C$

b)

$2°C/s$ 1330°C - 5' $10°C/s$ $\dot{\varepsilon} = 1\cdot10^{-3}\,s^{-1}$
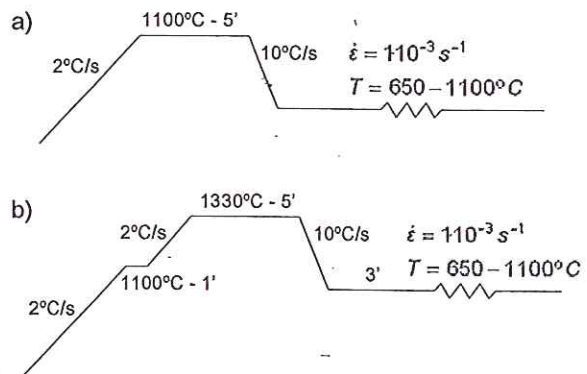$2°C/s$ 1100°C - 1' $3'$ $T = 650 - 1100°C$
$2°C/s$

Figure 1. Thermal cycle applied to the samples previous to the hot tensile test: a) reheating at 1100°C, b) reheating at 1330°C

After the reheating treatment the samples were cooled down to the test temperature at a constant cooling rate of 10 °C/s. Samples were then maintained at that temperature for 3 minutes to homogenise the microstructure in the whole section of the samples. Then, the tests were carried out at an initial strain rate of $5\cdot10^{-3}s^{-1}$. Both cooling rate and strain rate were close to the ones experienced by the billet surface during the straightening operation. Hot ductility was quantified by the reduction in area (%RA) of the samples tested to fracture.

Once the tests were carried out, one half of the sample was quenched in water in order to freeze the high-temperature microstructure. Before etching the samples, the distribution and morphology of the

inclusions was observed and then, the samples were etched in order to detect possible segregations. The other half of the sample was used to examine the fracture surfaces, using a scanning electron microscope JEOL T840 SEM. The grain boundary composition of some samples close to the fracture tips was measured by Auger electron spectroscopy (AES) using a Riber nanoprobe spectrometer equipped with a FEI field emission gun and a MAC 2 cylindrical analyzer.

3. RESULTS

3.1. Hot ductility curves

The hot ductility curves for steels A1 and A2 after different thermal cycles can be seen in Figure 2. The ductility loss for both steels took place in the austenite temperature region (Ae_3= 806°C, calculated using the Andrews equation [4]) but the trough extended to higher temperatures for steel A1 than for steel A2. This behaviour seems to be independent of the reheating temperature, even though the curves of both steels are closer to each other after reheating at 1330°C. Therefore, when the initial condition of the material is as-rolled, the ductility trough of the steel was narrower than when it is as-cast. On the other side, reheating temperature seemed to cause no significant difference on the above-mentioned behaviour by comparing Fig. 2a and 2b.

At the lower temperature range, a slight ductility recovery can be observed for samples tested after reheating at 1330°C. It was not possible to measure HD at the lower temperature range after reheating at 1100°C since they always fractured out of the samples gage length. The ductility recovers slightly at temperatures lower than Ar_3, which is around 723°C according to Ouchi [5].

It is also remarkable that the hot ductility curve of steel A1 passes below the one of steel A2 for both reheating cycles. The difference between both curves seems to decrease as the reheating temperature is increased.

3.2. Fractographies and metallographies

3.2.1. Reheating treatment at 1100°C

In Figure 3 the fracture features of steels A1 and A2 after reheating at 1100°C are shown. They correspond to samples tested at 900°C for steel A1 and at 850°C for steel A2, which are the highest temperatures end of the low ductility region. The fracture aspect seemed to be similar for both steels, showing some round features mixed with kind of intergranular component. These characteristics remained the same at all the brittle temperature range for both steels.

Fig. 4 illustrates microstructural differences between steel A1 and A2 after reheating at 1100°C. One of the
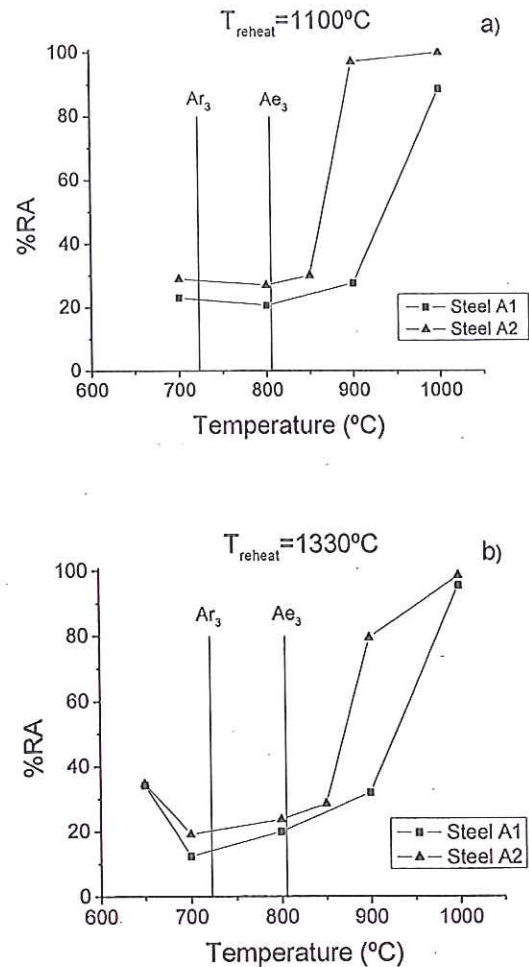
Figure 2 Hot ductility curves for steels A1 and A2 after reheating treatments at a) 1100°C and b) 1330°C

differences observed is related to the morphology and distribution of the inclusions. For steel A1 the MnS are located at the grain boundaries and they are round as can be seen in Figure 4a. On the other hand, for steel A2 the same are elongated and parallel to the deformation direction, probably as a consequence of the rolling process, Figure 4b.

The segregation patterns show also differences depending on the initial condition of the steel. Figure 4c shows solidification dendritic structure of steel A1 after etching, showing segregation pattern. The segregation pattern for steel A2 was aligned in the rolling direction, as can be observed in Figure 4d.

3.2.2. Reheating treatment at 1330°C

Even though the hot ductility curves showed similar ductility troughs after different reheating treatment for each of the steels, the fracture surfaces of the samples

tested after reheating at 1330°C showed significant differences when compared to the ones reheated at 1100°C. For the higher reheating temperature, the samples with a brittle behaviour showed intergranular fractures as it is evident in Figure 5 for the steels A1 and A2 tested at 800°C. In this case, the facets of the grains looked smooth.
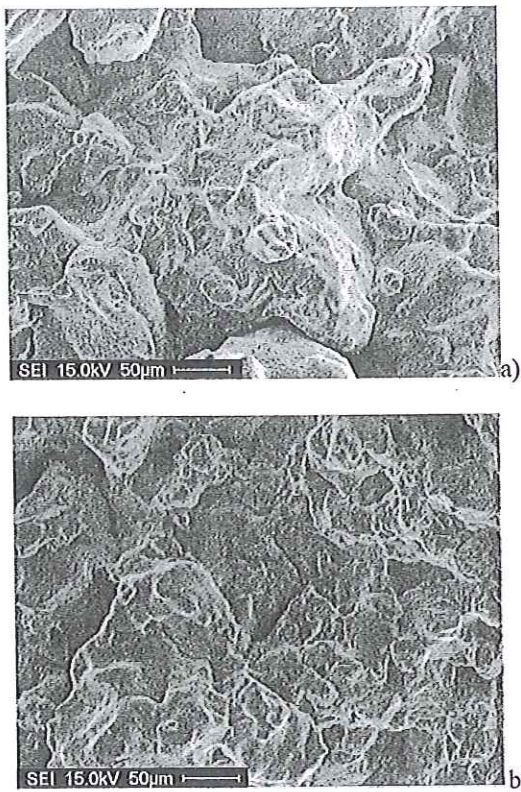

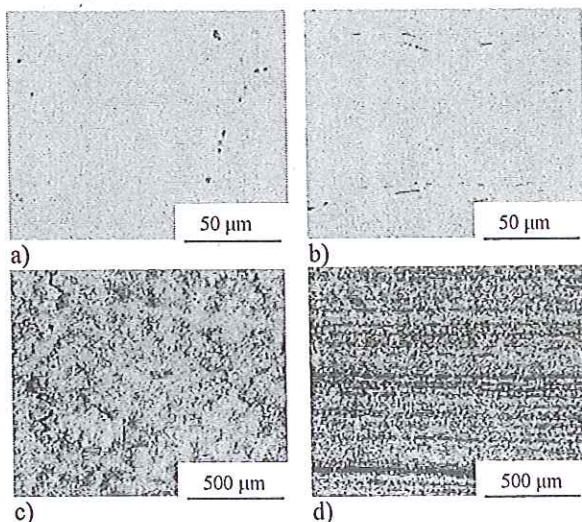Figure 3 Fractographies of a) steel A1 tested at 900°C and b) steel A2 tested at 850°C after a reheatment at 1100°C

Figure 4 Metallographies for samples of a) c) steel A1 and b) d) steel A2 after a reheating cycle at 1100°C
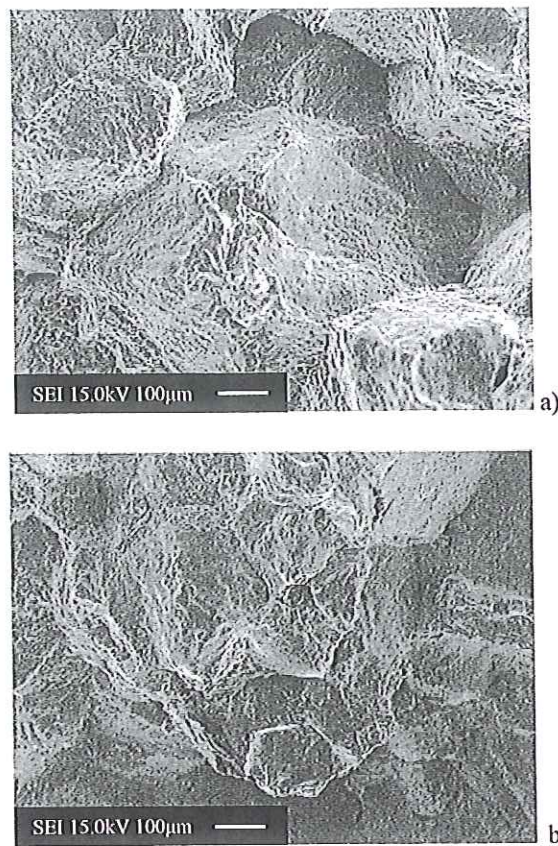

Figure 5 Fracture surfaces of the samples tested at 800°C, reheated at 1330°C for steel a) A1 and b) A2

The distribution and shape of inclusions after a reheating at 1330°C can be seen in Figure 6 for both steels. In general, the inclusions in the samples of steel A1 reheated at 1330°C, Figure 6a, were bigger than the ones observed for the same steel reheated at 1100°C. Furthermore, some of the inclusions in steel A2 reheated at 1330°C, Figure 6b, seemed to be elongated but the rest were round and aligned, probably in the direction of the original inclusions that gave way to them. Little segregation could be detected on samples reheated at 1330°C, probably indicating that this reheating temperature could have been high enough to homogenise the composition and therefore reduce segregation features.

In addition, when the samples were tested in the ferrite plus austenite temperature region, ductile intergranular fractures were observed, as can be seen in Figure 7 for the sample of steel A1 tested at 650°C. This kind of fracture is related to the presence of a ferrite film surrounding the austenite grains, which concentrates the deformation at the grain boundaries, leading to the nucleation of voids around MnS.
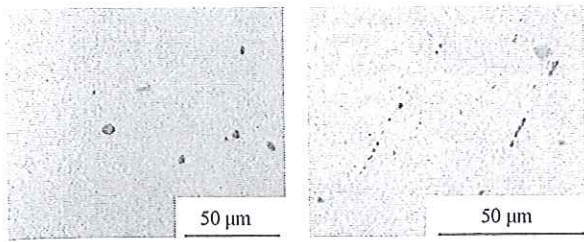
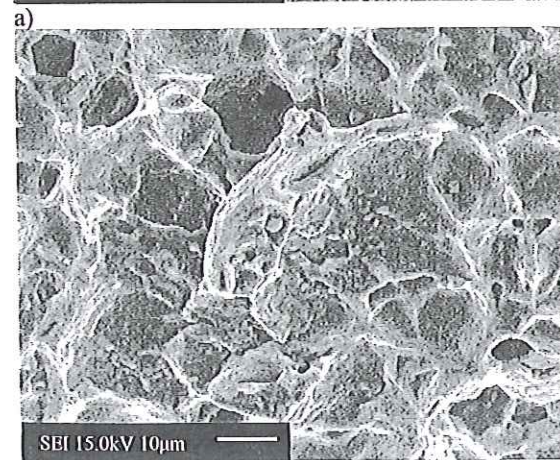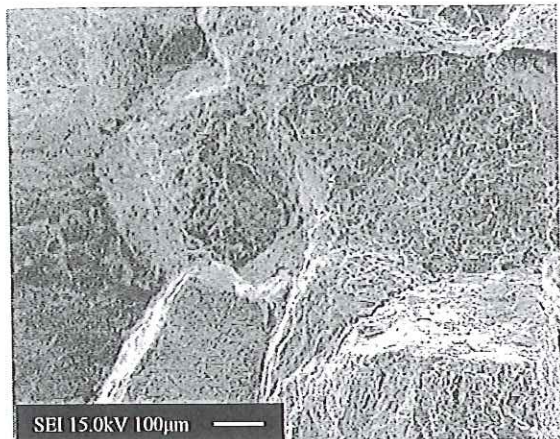Figure 6 Metallographies for samples of a) steel A1 and b) steel A2 after reheating at 1330°C

a)

b)

Figure 7 a) Fracture features of a sample of steel A1 tested at 650°C after reheating at 1330°C b) the same but in higher magnification

4. DISCUSSION

From the fracture features and microstructures shown in the previous section, two different embrittling mechanisms seem to be acting. The first one acts inter-dendritically as is apparent for samples tested after reheating at 1100°C. The second mechanism acts at the intergranular scale and embrittles grain boundaries owing to an intregranular decohesion fracture. This mechanism acts preferentially after reheating treatments at higher temperatures.

The lower reheating temperature (1100°C) would not be high enough to remove solidification microsegregations. Regarding the composition of both steels, there are several elements susceptible to enrich the interdendritic liquid during solidification, mainly being S and P that would strongly reduce the interfacial energies of the dendritic grains when segregated [6]. This embrittlement of the interdendritic areas would be apparent at lower temperatures during HD test, promoting an interdendritic decohesion as the one evident in Figure 3.

The grain size obtained after reheating treatments at 1100°C might favour the interdendritic fracture since it is about one half of the secondary arm spacing determined for steel A1 which is around 150μm. If any intregranular segregation takes place during the austenitizing treatment, grain boundaries will also be weakened and thus they could be an easy path for the linkage of the cracks formed at the microsegregated areas.

The higher reheating temperature (1330°C) seems high enough to redistribute the elements segregated at the interdendritic spaces during the solidification; however such thermal cycle would be promoting intergranular segregations. S and Sn are elements prone to segregate to intergranular spaces. The segregation of S was confirmed by Auger electron spectroscopy at grain boundaries of one of the specimens as can be seen in Figure 8. No evidence of Sn segregation was detected but according to other authors [7], this element would probably segregate in a non-equilibrium manner. This means that its segregation would depend on the soaking time at the testing temperature and also the cooling rate from the reheating temperature.
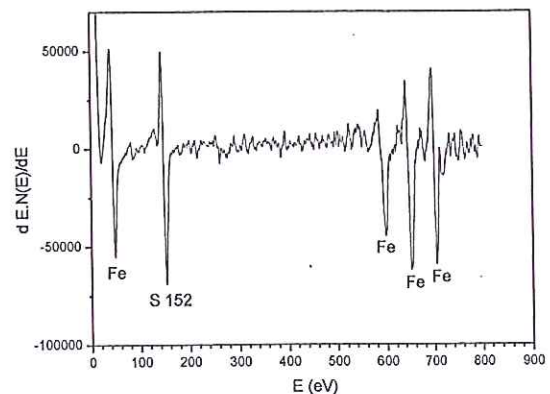

Figure 8 Typical Auger scan of the fracture surface of samples tested at brittle temperatures after a reheating at 1330°C

The role of S seems to be important in the overall steel behaviour. This element can appear in steel in form of

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inclusions, solid solution or segregated. During the solidification of the steel it microsegregates to the interdendritic areas as shown in Figure 4a since the inclusions are located at such regions. Due to the high cooling rates experienced during solidification, the inclusions formed were probably (Mn, Fe)S type [8]. This kind of inclusions has lower melting temperatures than pure MnS whose melting point is at 1530°C [9]. A reheating treatment at 1100°C probably had no effect on the size and distribution of the inclusions formed during the solidification of the steel, but 1330°C would probably favour the formation of MnS by increasing the diffusivity of Mn in the austenite. This temperature is also high enough to put some S in solid solution that would be able to segregate to the intergranular spaces leading to intergranular decohesion.

Grain size has usually an important effect on the hot ductility promoting wider troughs as the grain size increases [10]. In this case, this effect was not observed since the troughs obtained after different reheating treatments (different grain sizes) are very similar. The negative effect of the larger grain size for the samples reheated at 1330°C could be alleviated by the larger size of the inclusions that form mainly in the matrix after this thermal cycle, giving similar ductility troughs for each steel after different reheating temperatures. But, grain size differences could influence the negative effect of each embrittlement mechanism promoting deeper troughs for steel A1 (slightly larger grain size) than for steel A2 both after reheating at 1100°C and after reheating at 1330°C.

The difference of the ductility trough width between steel A1 and A2 can be explained in terms of the inclusions and segregations. As it was observed in Fig. 4b, 4d and 6b their orientation parallel to the tensile direction can maintain the good ductility until lower temperatures.

The ductile intergranular fracture observed for samples tested at 650°C can be related to an increase of the ferrite volume fraction. This phase, can appear induced by deformation at temperatures between the Ae_3 (equilibrium transformation temperature) and Ar_3 (continuous cooling transformation temperature). In this temperature range the ferrite forms surrounding the austenite grains and embrittles the steel by the concentration of the deformation at such softer phase.

5. CONCLUSIONS

Residual elements seemed to have strong effect on the hot ductility of steels studied in this work since they are prone to segregate. Microsegregations taking place during the solidification of the strand can cause the embrittlement of the steel under typical straightening conditions as can be detected in samples reheated at 1100°C. Higher reheating temperatures could induce intergranular segregations which resulted in similar ductility troughs to the ones obtained after lower

reheating. S is susceptible to be one of the elements having a major influence on the hot ductility of the steels. Concerning its effect it would be better to carry out *in-situ* melting tests in order to obtain more reliable results, since this kind of test put all S in solution.

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