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Effect of Copper Content on the Hot Ductility Loss of Low Carbon Steels

Fujda MARTIN

Department of Materials Science, Faculty of Metallurgy, Technical University, Košice, Slovakia

ABSTRACT

The hot ductility of low carbon steels with various Cu-content (0.04 %, 0.38 % and 0.63 %) was investigated at elevated temperatures ranging from 700°C to 900°C (temperature range of the $\gamma \rightarrow \alpha$ transformation), at a low strain rate (~$10^{-3}.s^{-1}$) by torsion testing in an Argon protective atmosphere. The highest hot ductility loss of analysed steels in the temperature range below the $\gamma \rightarrow \alpha$ transformation temperature Ar_3 is, predominately, caused by formation of proeutectoid ferrite along the austenite grain boundaries. The strain is localised in the proeutectoid ferrite film, which results in crack formation and the micro-void coalescence in proeutectoid ferrite at the γ - α interface. The higher ductility loss in the wider temperature region for the low Cu - content steel, if compared to the other analysed steels is a result of the embrittlement effect of combination of the ferrite film formation along the austenite grain boundaries and the AlN precipitation in the steel matrix and on the austenite grain boundaries.

Keywords : hot ductility, $\gamma \rightarrow \alpha$ transformation, film of proeutectoid ferrite, austenite grain boundaries, AlN precipitation, embrittlement

INTRODUCTION

Hot ductility of low carbon steel is determined by a number of factors, especially temperature, chemical composition, strain rate and thermal history. From the viewpoint of a large reduction of ductility there are three critical temperature regions: the region close to the melting point, the region of austenite, the austenite-ferrite two phase region (Harada, *et al.* 1990).

An important factor, which affects the steel slab surface cracking in the continuously casting process, is hot ductility of steel in the temperature range of the $\gamma \rightarrow \alpha$ transformation at a low strain rate (Harada, *et al.* 1990; Suzuki, *et al.* 1982; and Bohm, *et al.* 1992). The embrittlement of low carbon steels in this two phase region is principally caused by thin film formation of proeutectoid ferrite along the austenite grain boundaries. The localised strain in the proeutectoid ferrite film results in crack formation and microvoid coalescence in proeutectoid ferrite and at the γ - α interface (Harada, *et al.* 1990; Suzuki, *et al.* 1982; and Reveaux, *et al.* 1994). The hot ductility loss in this temperature region can be increased by MnS or AlN inclusion precipitation and precipitation of microalloying elements (Nb, V, Ti) particles. These precipitates on austenite grain boundaries are initial points for microvoid cracking and decrease grain boundary migration rate in the process of the austenite dynamic recrystallization (Reveaux, *et al.* 1994; Ouchi, *et al.* 1982; and Crowther, *et al.* 1987). AlN, (Nb, V, Ti) CN precipitation in the interior of austenite grains strengthens the steel matrix. Therefore, these precipitates can increase strength in the austenite grain boundary region in the deformation process (Ouchi, *et al.* 1982; and Crowther, *et al.* 1987).

A very important factor that affects the hot ductility of low carbon steels in this temperature region is the presence of detrimental impurities (Sn, As, Sb, Bi, Cu) in the steel. Their segregation on austenite grain boundaries decreases the hot ductility of steel through the decreasing of austenite grain boundaries cohesion (Nachtrab, *et al.* 1988; and Matsuoka, *et al.* 1997). The hot ductility reduction of low carbon steel can be also increased by fine copper sulphide and oxysulphide precipitation on austenite grain boundaries (Mintz, *et al.* 1995; and Llewellyn, 1995). The copper in steel can influence steel hot ductility by decreasing the nucleation rate and growth rate of ferrite on austenite grain boundaries in the $\gamma \rightarrow \alpha$ transformation temperature range (Ohtsuka, *et al.* 1997).

The aim of this work is to characterise a value and a reason for hot ductility loss of low carbon steel with various copper contents (0.04, 0.38 and 0.63 %) in the $\gamma \rightarrow \alpha$ transformation temperature region (700 – 900°C).

MATERIALS AND EXPERIMENTAL PROCEDURES

Experimental Materials and Methods

The study was carried out on steel strips, hot rolled from parts of low carbon steel enriched by Cu continuously cast slabs, labelled A, B and C. Rolling in laboratory conditions was done between 950° to 1150°C. The chemical composition of strips are shown in Table 1.

Table 1 Chemical composition of analysed steels.

Steel	wt. [%]										
	C	Mn	Si	P	S	Al	N	Cu	Sn	Sb	As
A	0,12	0,36	0,21	0,010	0,014	0,016	0,005	0,63	0,003	0,001	0,004
B	0,17	0,35	0,22	0,012	0,015	0,009	0,006	0,38	0,005	0,002	0,011
C	0,15	0,48	0,13	0,009	0,014	0,067	0,010	0,04	< 0,001	0,001	0,005

The specimens for the hot ductility investigation had been cut out in the rolling direction and deformed by the torsion test. The test pieces were heated up to 1415°C close to the melting point of the analysed steel, held there for one and a half minutes, and cooled then to the torsion testing temperature, in the range from 700° to 900°C. The heating and deformation were carried out in an Ar atmosphere. A strain rate of 4.10^{-3} s^{-1} to 8.10^{-3} s^{-1} was used, and then slow cooling or fast one-quenching. The number of turns to failure (N_f) was the value characterising the ductility. The microstructures, carbon extraction replicas (on a nickel net) and fracture surfaces were analysed for the determination of hot ductility loss.

RESULTS AND DISCUSSION

The temperature dependence of the hot ductility values (N_f - number of torsion turns to failure) obtained by a torsion plastometer at given thermal-deformation conditions for three low carbon steels with different Cu contents (A-0.63 %, B-0.38 %, C-0.04 %) is shown in Figure 1. In this figure, the temperatures of the $\gamma \rightarrow \alpha$ transformation during cooling (Ar_1, Ar_3) are indicated for all three steels, too.

The minimum values of ductility for all three steels were achieved in the temperature range of $\gamma \rightarrow \alpha$ transformation. An apparent decrease and the lowest values of ductility were observed in the temperature range just below the Ar_3-temperature of the transformation onset. Only

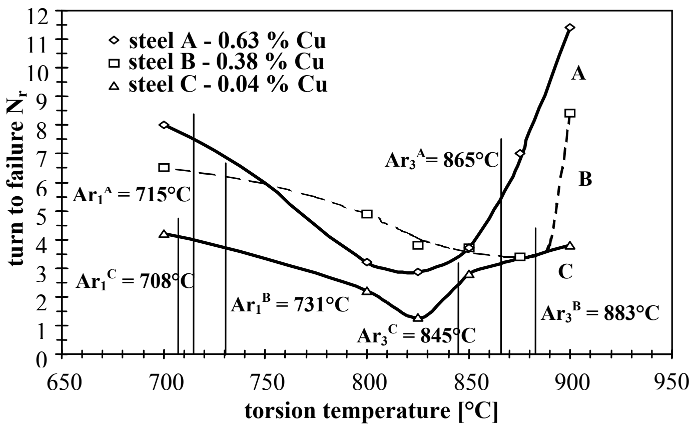

Figure 1 The temperature dependence of hot ductility for analysed steel A, B, C.

for steel A with the highest Cu-content (0.63 %), the minimum value of hot ductility was found at a lower temperature (825°C), not just below the Ar_3-temperature. The transformation temperature is highest for the steel B (883°C), lower for steel A (865°C) and lowest for steel C (845°C). The values of hot ductility in this temperature range are for steels A and B with high and middle Cu content respectively, almost the same. Ductility of steel C with the lowest Cu content is in this temperature range lower than that of steel A and B and the temperature range of the ductility loss is extended above the transformation temperature Ar_3.

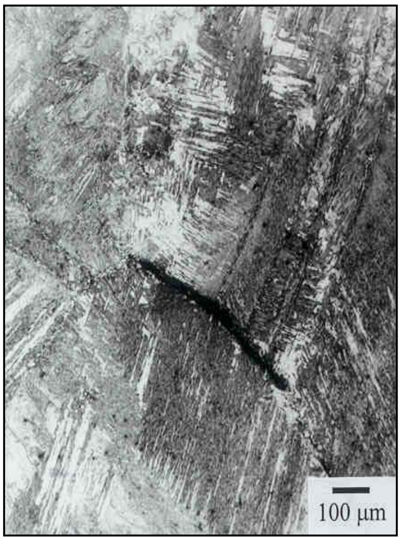

Figure 2 The film-like proeutectoid ferrite and the crack along the austenite grain boundaries (steel B, T = 87°C, quenching).

The loss of ductility in this temperature range is the first of all caused by the formation of film-like proeutectoid ferrite (Figure 2) along the austenite grain boundaries just bellow the Ar_3 temperature. Since the flow stress of the proeutectoid ferrite is significantly lower than that of the austenite grains at this temperature, plastic deformation occurs preferentially in this ferrite. Causing crack or void formation on the austenite grain boundaries, (Figure 3). The formation of such cracks during torsion deformation was manifested by smooth or slightly wrinkled intergranular facets on the fracture surface of the sample of steel C after the torsion test, (Figure 4).

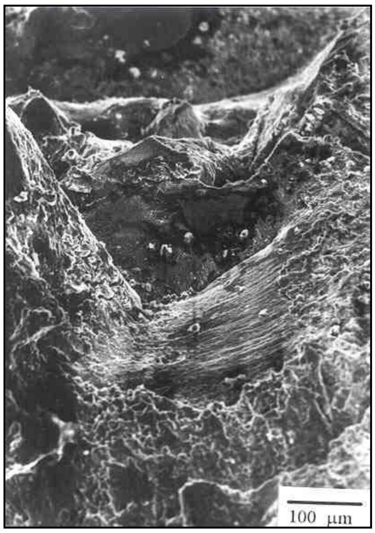

Figure 3 The intergranular facets on the fracture surface (steel A, SEM, T = 850°C).

Figure 4 The intergranular fracture on the fracture surface (steel C, SEM, T = 800°C).

A thin oxide film with a high Cu content, which is probably a result of copper enrichment and melting of grain boundaries in the processes of solution annealing, cooling to deformation temperature, and in torsion deformation observed at grain boundaries (Figure 5) of steel A with the highest Cu content, can increase the embrittlement of analysed steel A in this temperature range. The presence of increased copper content on austenite grain boundaries of steel A probably caused the decreasing of the nucleation rate and growth rate of ferrite on austenite grain boundaries and then it shifted the hot ductility minimal value from the temperature range just below the Ar_3-temperature to a lower temperature.

The embrittlement of steel C with the lowest (0.04 %) Cu content in this temperature range was a combination between thin ferrite film formation on austenite grain boundaries and AlN precipitation on the grain boundaries and inside the austenite grains (Figure 6) in this temperature range. These AlN precipitates strengthened the steel matrix and thus deformation was concentrated into the grain boundaries region and they also retarded the dynamic recrystallization of austenite by pinning up the grain boundaries in the γ region. These processes significantly the decreased hot ductility of steel C and widened hot ductility loss over the $\gamma \rightarrow \alpha$ transformation (Ar_3) temperature (to 900°C).

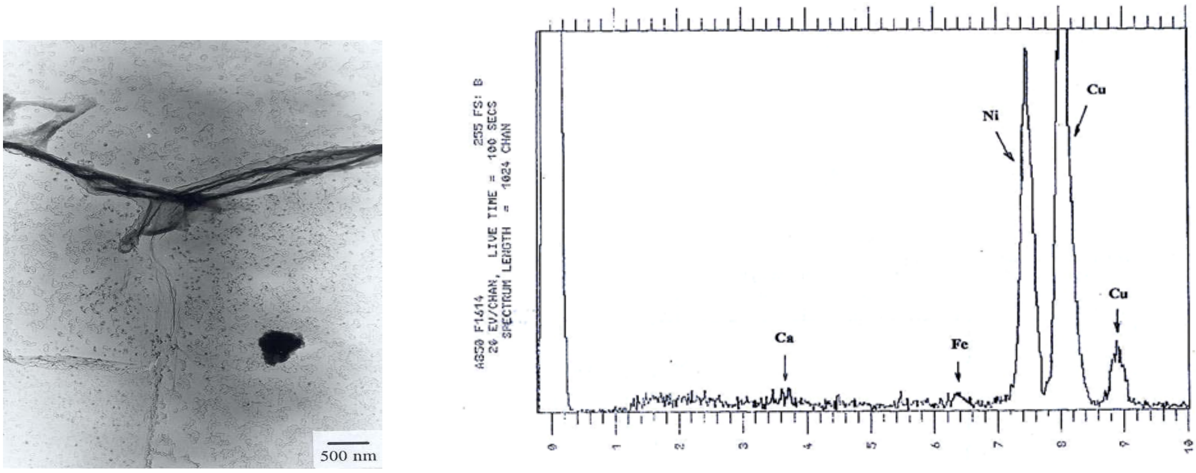

Figure 5 The oxide film with high Cu - content on grain boundaries (steel A, TEM, carbon replica, T = 850°C).

Figure 6 The AlN precipitates (steel C, TEM, carbon replica, T = 800°C).

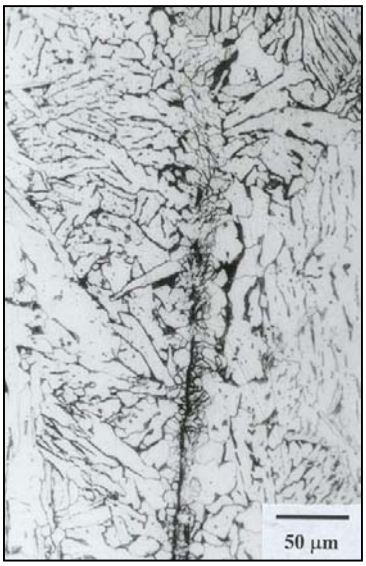

Figure 7 The microstructure of sample close to fracture surface (steel C, T = 800°C).

The matrix strengthening of steel C, the retardation of dynamic recrystallization of austenite and the deformation concentration to ferrite film in the austenite grain boundaries region are documented by acicular ferrite-pearlite microstructures after minimal austenite deformation and by cracks in the fine ferrite grains, which was formed on austenite grain boundaries near the fracture surface of the sample (Figure 7). On the contrary, the microstructure of the steels A and B near the fracture surface of the sample with the minimum value of hot ductility is fine ferrite-pearlite (Figure 8) after austenite deformation. It demonstrates a smaller embrittlement of these steels, than steel C. The better hot ductility of steels A and B is also indicated by a relatively small amount of intergranular facets (Figure 3) on sample fracture surface after torsion test in the critical temperature range in comparison with considerable amounts (near 100 %) of intergranular facets with a smooth and dimple morphology (Figure 4) on the sample fracture surface of steel C after the torsion test in critical the temperature range.

Figure 8 The microstructure of sample close to the fracture surface (steel A, T = 850 °C).

CONCLUSION

1. The ductility loss of analysed steels in this temperature range is first of all caused by the formation of film-like proeutectoid ferrite along the austenite grain boundaries. Since the flow stress of the proeutectoid ferrite is significantly lower than that of the austenite grains at this temperature, plastic deformation occurs preferentially in this ferrite, up to crack or void formation.

2. The combination of embrittling effect of thin ferrite film formation on austenite grain boundaries and AlN precipitation on the grain boundaries and inside the austenite grains was the reason for the lowest hot ductility of steel C (0.04 % Cu) in the $\gamma \rightarrow \alpha$ transformation temperature range. These precipitates strengthen the steel matrix and thus concentrate the deformation into weakened grain boundaries and in the γ region also retard the dynamic recrystallization of austenite by pinning up the grain boundaries. These embrittling effects of AlN precipitates significantly decrease hot ductility of steel C over the $\gamma \rightarrow \alpha$ transformation (Ar_3) temperature (to 900°C).

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