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Hot ductility study on high P and Cu containing low alloy steel

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Abstract : Six numbers of laboratory heats of 25 kg each were made using a high frequency induction melting furnace. Various elements, viz, Ni, Cu, P and Si in the range of 0.01-2.0%, 0.22-0.47%, 0.12-0.19% and 0.40-0.86%, respectively were added during melting of the steels. Carbon and sulphur contents were kept at low levels, i.e., 0.05-0.087% and 0.005-0.01%, respectively. Hot tensile tests were carried out using Gleeble-350^{0}C simulator, which revealed reduction in area in excess of 70% at all deformation temperatures and at slow strain rate (0.005/s) testing conditions for all the experimental steels excluding low Si (0.40%) containing steels. The lower hot ductility (RA value <30%) of these steels resulted in brittle failure at 850^{0}C, which corresponds to the temperature experienced during straightening operation of continuous casting. On the other hand, hot ductility of low Si containing steels was significantly increased (RA values greater than 70%) at higher strain rate (6.5/s).

Keywords : Hot ductility, Slab cracking, Brittle failure, Weathering steel, Continuous casting.

INTRODUCTION

Copper, which is mainly contained in the scrap of cars and electrical appliances as well, is problematic in mechanical workings at elevated temperatures and limits the usage of scrap iron[1]. It is enriched at steel/scale interface by preferential oxidation of Fe, which leads to liquid embrittlement or surface cracking during hot working. This type of defect is well known as surface hot shortness[1]. Cu has been also reported to give surface cracking problem including transverse cracking in continuously cast products and has been found to be detrimental to surface quality[2, 3]. Broadly, two forms of hot shortness are categorised by Melford[4], (i) liquation cracking, i.e., hot shortness in the bulk of the steel and (ii) surface hot shortness. Liquation cracking is a problem associated with enriched liquid phase at grain boundaries, which occurs when relatively insoluble residual elements (e.g. sulphur) are present in steel resulting in cracking. Surface hot shortness, on the other hand is caused by the enrichment of residual elements on the subsurface during oxidation which is liquefied and subsequently, penetrated into austenite grain boundaries.

Other than surface hot shortness generally found in Cu bearing steels, different types of cracks are also appeared in continuously cast products in low carbon, Cu bearing and mirco-alloyed steel as shown schematically in Fig. 1. The most common defect found is transverse cracks, which may be formed on the broad face, narrow face, or corner of continuously cast slab. However, these are not always apparent to visual inspection unless the slab surface is scrapped or prepared for microscopic observation. These are usually associated with the depression of oscillation marks, and are predominantly found on the slab surface. The length of the cracks can be several 10s of mm, and generally follow austenite grain boundaries. The cracks are partially oxidised with some decarburization[5].

Transverse cracking occurs most frequently in steels micro-alloyed with Nb, which greatly promotes the formation of transverse cracks while it has no effect at low V and N levels.

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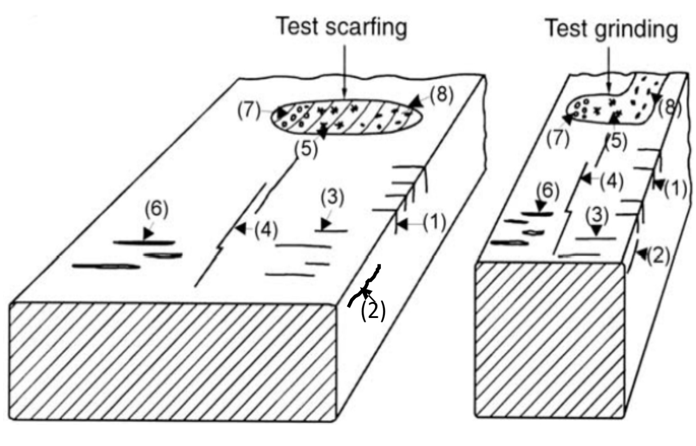
However, the combinations of 0.15% V and 0.02% N have been reported to lead to transverse cracking. In copper bearing steel, Burden et al. [6] have reported that surface cracks in continuously cast blooms are associated with enrichment of Cu, Ni and Sn. The work by Woollen [4] at British steel indicates that plate steels containing Nb with additions of 0.25% Cu and 0.25% Ni have a greater incidence of transverse cracking compared with similar grades without Cu and Ni additions.

Lankford[7] has reviewed the source of stresses during continuous casting which can arise from a large number of different causes. These include transformation effects, thermal effects (variable heat transfer within the mould, temperature gradients within slabs, effects of cooling water sprays, contact with rollers, etc.), friction between strand and mould, bulging of the strand caused by ferrostatic pressure, mechanical effects due to misalignment of the casting machine, and straightening strains. Out of these, straightening strains is most important. This is because large transverse cracks are observed in the final straightened slab, together with the fact that these are often most numerous on the surface of the slab (i.e. the surface which is in tension during straightening). This suggests that there is much crack propagation induced by the stresses experienced during the straightening process. When these stresses occur in the temperature range over which ductility is poor, the appearance of transverse cracking becomes severe.

Phosphorus may have some beneficial effect on hot ductility and transverse cracking.Owing to the reported catalytic effect of phosphorus on the formation of amorphous type FeOOH compact rust layer, its concentration in weathering steels has been remained at about 0.1 wt. %[8-10]. Therefore, effect of phosphorus on hot ductility need to be investigated.

The primary objective of this study is to evaluate the effect of Si, P and Ni on crack formation in a Cu-Cr-P low alloy steel. The technique involved to evaluate cracking problem is mainly hot tensile test using a Gleeble simulator which will simulate a processing environment that leads to surface hot shortness and hot cracking during continuous casting of slab/billet, and during reheating and hot rolling of copper bearing steels.



1. Transverse corner cracks; 2. Longitudinal corner cracks; 3. Transverse cracks;
4. Longitudinal cracks (broad face); 5. Star cracks; 6. Deep oscillation marks;
7. Pinholes; 8. Macro inclusions.

Fig. 1 : Schematic of surface defects in continuously cast products.

EXPERIMENTAL

Six numbers of laboratory heats of low alloy steels of 25 kg each were made using a high frequency induction melting furnace. Steel scrap of C-Mn rail steel was added to soft iron (0.001 wt. % C) to balance carbon content. In order to obtain Cr, P, Cu and Ni in the desired ranges, mother alloys of Fe-P, Fe-Cr along with Cu blocks and Ni lumps were added to the liquid steel in the furnace. Aluminium in the form of shots and wire were used for deoxidation purpose. The melts were cast into 100mm × 100mm square ingots, which were subsequently hot rolled to 16 mm thick plates through five passes after soaking at 11500C for 2.0 hours. The chemical composition of each steel was determined using an optical emission spectrometer. Chemical compositions of experimental steels after hot rolling are given in Table 1.

Table 1 : Chemical composition of different steels (in wt. %) obtained using OES after hot rolling

Steels	C	Si	Mn	S	P	Cr	Cu	Ni	Al
S1	0.059	0.82	0.31	0.005	0.12	0.36	0.41	0.13	0.017
S2	0.06	0.86	0.42	0.005	0.19	0.37	0.41	0.13	0.019
S3	0.065	0.80	0.35	0.008	0.13	0.36	0.47	trace	0.012
S4	0.068	0.81	0.38	0.005	0.20	0.37	0.47	trace	0.005
S5	0.085	0.40	0.39	0.01	0.19	0.02	0.40	trace	0.005
S6	0.087	0.40	0.42	0.01	0.19	0.34	0.40	trace	0.003

To understand the surface hot cracking behaviour of experimental steels, hot tensile test was carried out using a Dynamic Thermo-Mechanical Simulator (DTMS) Gleeble-350^{0}C. The specimens of 10 mm diameter and 120 mm length with threads at both ends were initially heated to the soaking temperatures 1300^{0}C at a heating rate of 20^{0}C/s and held for 1 min. This was followed by cooling at a rate of 5^{0}C/s to deformation temperature, viz., 1200, 1100, 1000, 900 and 850^{0}C. The deformation was allowed up to fracture with a strain rate of 0.005/s (i.e., the stroke rate is 0.125 mm/s with a gauge length of 25 mm) for S1 to S6 steels. The slow strain rate applied is similar to that occurring during straightening operation (~$10^{-3}s^{-1}$) of 240 mm thick slab on continuous casting[11]. In addition to this, deformation was also carried out at strain rate of 6.5/s (i.e., the stroke rate of 162.5mm/s) for S4 and S6 steels. The strain rate was selected to see the effect of hot ductility during rolling. Hot ductility was realized from the % of reduction in area (RA) at failure after tensile test.

RESULTS AND DISCUSSION

Hot Tensile Test

Fig. 2 shows variation in the percentage of reduction in area (RA) value with deformation temperatures after tensile test in Gleeble at a strain rate of 0.005/s in the range of 850^{0}C to 1200^{0}C. The different temperature zones of reduced hot ductility of steel related to embrittling mechanism can be co-related to that observed in experimental steels in Fig. 2. No major ductility trough is observed in any steel in any of the three expected reduced ductility temperature zones except S5 and S6 steels at 850^{0}C. In both cases of these S5 and S6 steels, the percentage of reduction in area (RA) value decreased to 20% at 8500 on decreasing silicon content from 0.8 wt.% to 0.4 wt.% while it was greater than 75 % at all other temperatures. This temperature is related to Type III in Figure 3 where phase transformation occurs[12]. Previously reported RA values above which transverse cracks don't occur are 75% [13], 60% [14] and

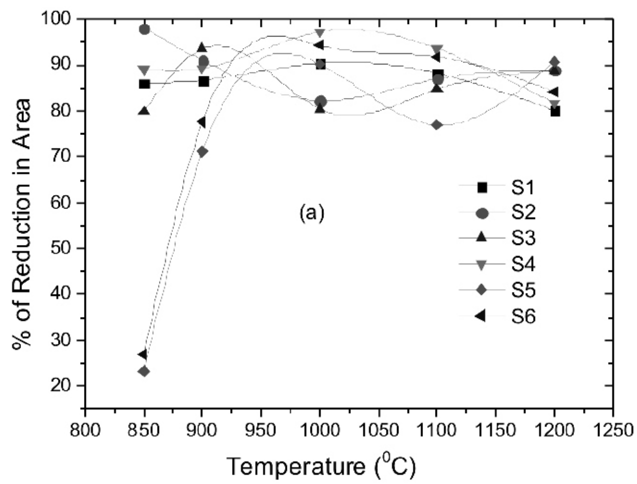


Fig. 2 : Experimental steels showing variation in percentage of reduction in area (a) and maximum strength (b) with deformation temperature after tensile test in Gleeble at a strain rate of 0.005/s.

30-40%[15]. Therefore, based on the present findings, it can be concluded that both S5 and S6 steels are not suitable for commercial production as there is good chance of the formation of transverse cracking during continuous casting.

The reason for Type III ductility trough in Figure 3 for low Si Steels (S5 and S6) can be attributed to the grain boundary cementite formation. Though this will be studied separately, Kozeschnik and Bhadesia [16] have recently reported that Si prevents cementite formation in austenite. Therefore, cementite formation in austenite may be the reason leading to low hot ductility in S5 and S6 steels unlike to that of S1 to S4 steels where Si content is in the range of 0.75-0.86%. In addition to this, as reported by many researchers that increasing sulphur content both deepens[17, 18] and widens[19] the hot-ductility trough. Therefore, as sulphur content

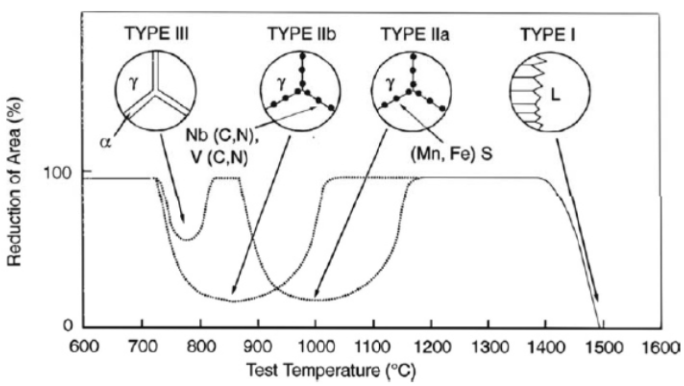


Fig. 3 : Schematic representation of temperature zones of reduced hot ductility of steel related to embrittling mechanism[12]

in S5 and S6 steel (0.01 %) is about double to that in S1 to S4 steel (~0.005%), this might have reduced the hot ductility at 850 and 900^{0}C. On the other hand, additions of Ni (0.13%) and Cr (0.34-0.37%) have little effect on hot ductility of experimental steels. Further, as shown in Fig. 4, the hot ductility of S6 steel has been improved in intermediate temperature range (900 and 1000^{0}C) at high strain rate, i.e., 6.5s^{-1}, which is in conformity with the findings of Suzuki et. al[20].

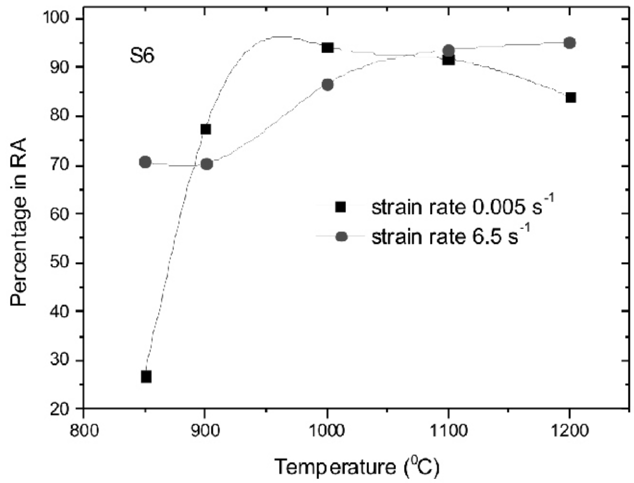


Fig. 4 : Effect of strain rate on the variation of percentage of reduction in area with deformation temperature after tensile test of experimental steel S6.

CONCLUSIONS

– Six different alloy chemistries were designed to study hot shortness and cracking effect during continuous casting. The chemistries were varied with respect to Ni, Cu, P and Si to make steel cost effective.

– All experimental steels exhibited good hot ductility at slow deformation rate (0.005/s) except S5 and S6 steel. However, S6 steel also exhibited good hot ductility at higher strain rate (6.5/s) and the similar result is also expected in the case of S5 steel.

– The failure of S5 and S6 steel in brittle manner at 850 0C is attributed to the compositional effect, especially Si and S content.

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