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Effect of niobium and titanium addition on the hot ductility of boron containing steel

Kyung Chul Cho[a], Dong Jun Mun[b], Yang Mo Koo[b], Jae Sang Lee[b,*]

[a] Department of Materials Science and Engineering, Pohang University of Science and Technology, Pohang 790-784, Republic of Korea
[b] Graduate Institute of Ferrous Technology, Pohang University of Science and Technology, Pohang 790-784, Republic of Korea

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ABSTRACT

Hot ductility of boron containing steel (B steel) with adding Nb (0.03 wt.%) (B–Nb steel) and B–Nb steel with adding Ti (0.0079 wt.%) (B–Nb–Ti steel) was quantified using hot tensile tests. The specimens were solution-treated at 1350 °C and cooled at 20 °C s^{-1} to tensile test temperature (T) in the range of $750 \le T \le 1050$ °C. After that, they were strained to failure at a strain rate of 2.5×10^{-3} s^{-1}. For the B–Nb steel, severe hot ductility loss was observed at $850 \le T \le 950$ °C, which covered the low temperature in which austenite (γ) single-phase exists, and the high temperature at which γ and ferrite (α) coexist. Ductility loss in the B–Nb steel was caused by the presence of a network of BN precipitates, rather than by Nb(C, N) precipitates at the γ grain boundaries. In contrast, hot ductility of the B–Nb–Ti steel was remarkably improved at $850 \le T \le 950$ °C. In the B–Nb–Ti steel, BN precipitates preferentially on TiN particles, resulting in increased BN precipitation in the γ grain interior and a decrease in the network of BN precipitates at the γ grain boundaries. These changes reduce strain localization at the γ grain boundaries and therefore increase the hot ductility of the steel.

1. Introduction

Loss of hot ductility during the continuous casting (CC) process has been a serious problem in microalloyed steels, because it is associated with surface cracking of slabs [1,2]. Recently, CC-direct rolling techniques have been adopted to save energy and cost in steel processing. In these processes, cracking of the slab surface must be prevented [2–4].

The hot ductility of continuously cast microalloyed steels appears to be strongly associated with a series of microstructural changes such as precipitation [5–7], austenite (γ) grain coarsening [8] and γ to ferrite (α) phase transformation [5,6,8]. These events can occur during the straightening operation of the CC process, which usually occurs in temperature (T) range of $700 \le T \le 1000$ °C [2,5], so the hot ductility of microalloyed steels in this temperature range is important and must to be controlled by adjusting the steel composition and the processing conditions.

In general, microalloyed steels with strong carbonitride-forming elements are particularly susceptible to slab surface cracking [6,7,9]. For instance, corner cracking on slabs of steels containing boron (B steel) easily occurs during CC because of poor hot ductility at $700 \le T \le 1000$ °C, possibly due to formation of boron nitride (BN) at γ grain boundaries [10–13]. According to this mechanism, if the morphology and distribution of the precipitates in the microalloyed steel can be controlled by adjusting its chemical composition, surface cracking may be efficiently suppressed in practical operation.

Thus, in this study, regarding the relation between the hot ductility of B containing steel and the morphology and distribution of B precipitate, the effect of small amount of Nb or of Nb plus Ti addition on the hot ductility of the steel was investigated in detail.

2. Experimental

This study was conducted using three types of B steel which contained almost the same level of elements except for Ti and Nb (Table 1). To quantify the effect of Nb addition on the hot ductility of B steel, 0.03 wt.% Nb was added to the B steel (B–Nb steel). Also, to quantify the effect of Ti addition on the hot ductility of the B–Nb steel, 0.0079 wt.% Ti was added to some B–Nb steels (B–Nb–Ti steel).

Steel ingots were cast using vacuum induction melting furnace then hot rolled into 20-mm thick plates with a final rolling temperature of 1000 °C. The tensile specimens were machined from each plate with their longitudinal axis parallel to the rolling direction; each such specimen had a diameter of 7 mm and a gauge length of 10 mm. Hot tensile tests were conducted using a hot deformation simulator with a dilatometer in an inert atmosphere of argon gas.

* Corresponding author. Tel.: +82 54 279 9025; fax: +82 54 279 9299.
E-mail address: js.lee@postech.ac.kr (J.S. Lee).

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Table 1
Chemical composition (wt.%) and equilibrium temperature of the γ/α phase boundary (Ae_3 (°C)) of steels used in this study.

Steel	B	Nb	Ti	C	Si	Mn	P	S	Al	N	Fe	Ae_3
B steel	0.0023	Trace	Trace	0.040	0.062	0.48	0.015	0.011	0.014	0.0051	Bal.	876.4
B–Nb steel	0.0024	0.030	Trace	0.042	0.048	0.46	0.013	0.010	0.010	0.0052	Bal.	876.1
B–Nb–Ti steel	0.0025	0.029	0.0079	0.041	0.050	0.51	0.014	0.012	0.012	0.0049	Bal.	874.6

The reduction of area (RA) was measured to evaluate hot ductility. Measurement of RA curves as function of deformation temperature is very useful in assessing the likelihood of slab surface cracking during CC [14].

The thermomechanical cycle (Fig. 1) used in this study was as follows: the specimens were heated from room temperature to 1350 °C at 10 °C s^{-1}, held for 10 min, then cooled to the test temperature range of 750–1050 °C at a cooling rate (CR) of 20 °C s^{-1}. Specimens were held at the test temperature for 5 min, and then deformed to failure at a strain rate of 2.5×10^{-3} s^{-1}. These testing conditions were chosen to simulate as closely as possible the conditions that occur during straightening operation of CC process. The intention of solution treatment at 1350 °C was to dissolve all precipitates and to create a coarse grain size reminiscent of the as-cast grain size. The CR of 20 °C s^{-1} was similar to that at the surface and corner of the slab during conventional CC. The test temperature range of 750–1050 °C and the strain rate of 2.5×10^{-3} s^{-1} were nearly the same as those in the straightening operation of CC process.

The equilibrium temperature of the γ/α phase boundary (Ae_3) calculated using Thermo-Cal (TCFE 6 database [15]) was ~875 °C (Table 1), which indicated that proeutectoid α was not formed at ≥900 °C in any of the three steels. Therefore, the temperature range of 900–1000 °C was designated here as the γ single-phase region and just below 850 °C as the $\gamma + \alpha$ two-phase region.

Undeformed and deformed specimens were examined both metallographically and fractographically using optical microscopy (OM), scanning electron microscopy (SEM), and transmission electron microscopy (TEM). The morphology and chemical composition of the precipitates which mounted on carbon replicas film were determined by using a TEM equipped with an energy-dispersive X-ray spectroscope (TEM-EDS). B distribution within undeformed specimens was determined using Particle Tracking Autoradiography (PTA).

3. Results

3.1. Hot ductility evaluation

All of the steels had good ductility at $T > 1000$ °C and $T < 800$ °C, and had low hot ductility at the high temperature range of the $\gamma + \alpha$ two-phase region ($T = 850$ °C) and the low temperature range of the γ single-phase region ($950 \leq T \leq 900$ °C) (Fig. 2). At $850 \leq T \leq 950$ °C, the ductility of the B steel was very similar to that of the B–Nb and both were lower than that of B–Nb–Ti steel. These results indicate that addition of only Nb with no Ti to the B steel does not increase hot ductility.

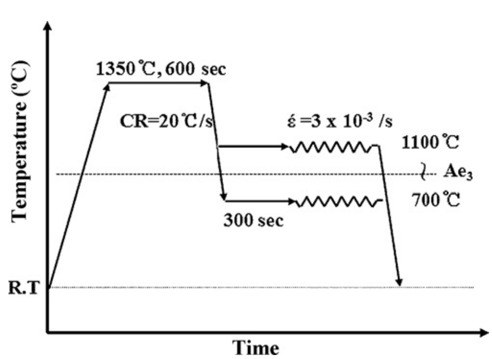

Fig. 1. Schematic diagram of thermomechanical cycle for hot ductility tests. Symbols and process are described are in the text.

3.2. Fractographs and microstructures

Scanning electron fractographs of the B–Nb steel and the B–Nb–Ti steel tested at 900 °C (>Ae_3: low temperature range of γ single-phase region) (Fig. 3) displayed fracture surfaces at the temperature at which ductility was minimum in both steels. In the B–Nb steel, the fracture surface exhibited typical intergranular decohesion, displaying flat grain surfaces that indicate failure due to grain boundary sliding (Fig. 3a) [2]. In the B–Nb–Ti steel, the fracture was primarily the result of intergranular separation, but it showed some ductile characteristics, i.e., the surfaces of grains were rough (Fig. 3b). The B–Nb–Ti steel appears to have experienced considerable plastic deformation before grain boundary separation. Comparison of Fig. 3a and b shows that the B–Nb–Ti steel is less susceptible to intergranular fracture than the B steel and the B–Nb steel in the low temperature region of γ single-phase.

In the cross-sectional microstructures of the tested steels quenched after fracture at 850 °C (just below Ae_3, high temperature range of $\gamma + \alpha$ two-phase region) (Fig. 4), microstructures were basically composed of prior γ and α. In the B–Nb steel, large prior γ grains were decorated with thin α layers, and voids nucleated at the grain boundary α (Fig. 4a). The microstructure of the B steel also had almost the same features. This similarity implies that fracture of the B steel and the B–Nb steel occurs by strain localization within the thin α layers along the γ grain boundaries. In contrast, the B–Nb–Ti steel having a relatively small γ grain showed extensive deformation of the γ and α phases near the fracture tip and a relatively thicker α layer at the prior γ grain boundaries than in the B–Nb steel (Fig. 4b). This indicates that the deformation of the B–Nb–Ti steel is not localized within the grain boundary α layers, as it is in the B steel and the B–Nb steel.

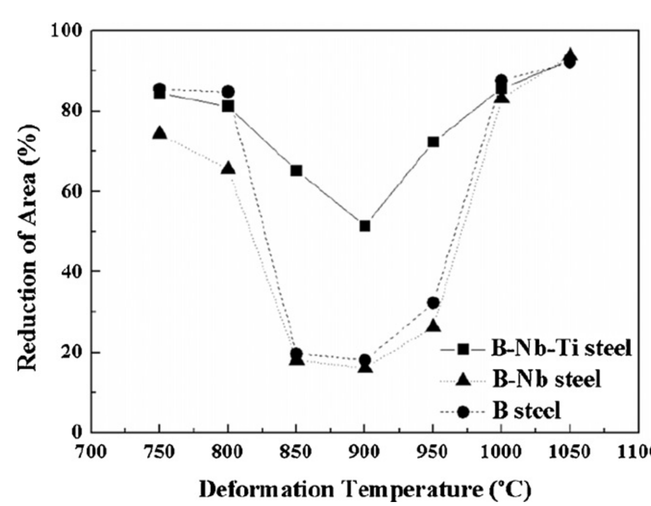

Fig. 2. Hot ductility curves of the B steel, B–Nb steel and B–Nb–Ti steel.