Communications

Effect of Boron on the Hot Ductility of Nb-Containing Steel

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The continuous casting process is widely used in the steel industry because it is a simple and cost-effective method to produce steels in large quantities. The steels are cooled rapidly since heat is extracted at a comparatively fast rate by heat transfer to the cold mold. However, such rapid cooling produces a steep thermal gradient between the interior and the exterior of the slabs, resulting in large thermal stress in the slabs. There is also a buildup of mechanically induced stress caused by friction in the mold, roll pressure, *etc*. The presence of these stresses can result in various types of cracking in the slabs. In the case of vertical continuous casters in particular, there is also an occurrence of tensile stress at the upper region and compressive stress at the lower region of the slabs when they pass through the straightening zone at around 700 °C to 900 °C. In the case of steels with poor hot ductility, the combination of the stresses mentioned previously quite often results in the formation of cracks. Among various types of steels, it is known that steels containing strong carbide formers such as Nb are particularly susceptible to cracking during continuous casting. Hence, the hot ductility of Nb-containing steels has been the subject of numerous investigations during the last decade.[1–5] It has been shown that the main causes for the large decrease in ductility are the presence of thin films of ferrite surrounding the austenite grain boundaries and/or the precipitation of fine Nb(C,N) particles within the ferrite and at austenite grain boundaries. A considerable amount of research has been conducted to improve the hot ductility of Nb-containing steels, mostly by the addition of alloying elements such as Ti,[6,7,8] Ca,[9] Zr,[6] and Y.[6] Another possible alloying element that can improve the hot ductility of the Nb-containing steels is B. It is well known that the addition of B to low C steels retards the formation of ferrite along austenite grain boundaries by segregating to these boundaries. It is therefore expected that the addition of B will change the microstructure at around 700 °C to 900 °C and, accordingly, the hot ductility. The objective of the present study is to examine the effect of B on the hot ductility of the Nb-containing steel.

A hot ductility test was conducted using the commonly practiced testing method.[1–9] Chemical compositions of the two steels used in the present study are shown in Table I. The steels were supplied as laboratory 50 kg vacuum melts, which had been hot rolled to 13-mm-thick plate. The specimens of 6-mm diameter and 36-mm length were machined with their longitudinal axes parallel to the rolling direction. Gleeble tests were carried out in order to simulate the thermal-strain history of the continuous casting process. The specimens were heated to 1350 °C and cooled at 1 °C/s to the test temperature and subjected to tensile testing at a strain rate of 5×10^{-3}/s. The test temperatures were from 700 °C to 900 °C with 50 °C intervals. After tensile testing, specimens were quenched by spraying with compressed air to preserve the proeutectoid ferrite formed at the test temperature. The reduction of area (RA) was measured as an evaluation of hot ductility. The cooling rate of 1 °C/s used in the Gleeble tests is the same as that used in the continuous caster at the Pohang Iron and Steel Co. Transmission electron microscopy (TEM) was employed to investigate the details of microstructural features. Specimens for TEM were made by electropolishing in a solution of 8 pct perchloric acid and 92 pct acetic acid.

Figure 1 shows the microstructures of the steels quenched after fracture at 750 °C. It shows that both steels have similar prior austenite grain sizes of 1 mm. Proeutectoid ferrite is present along the prior austenite grain boundaries in all steels. However, there are differences in the microstructures between the base steel (A steel) and the boron-containing steel (B steel). The B steel has a smaller amount of ferrite films along the prior austenite grain boundaries than the A steel (12.9 vol pct in the B steel *vs* 23.3 vol pct in the A steel). Also, the B steel contains proeutectoid ferrite within the prior austenite grains (10.9 vol pct). In addition to such differences in the morphology of ferrite between the A and B steels, the boron-containing B steel has 5-μm-sized coarse precipitates along the prior austenite grain boundaries as well as within the prior austenite grains (arrows in Figure 1). The TEM analysis was conducted to identify the nature of these precipitates, as shown in Figure 2. Selected area diffraction patterns (SADPs) of the precipitates show that they are $Fe_{23}(B,C)_6$. Scanning Auger microscopy (SAM) analysis was also conducted for the detailed identification of the composition of precipitates and it shows that they contain N besides Fe, C, and B (Figure 2(d)). The presence of intragranular ferrite in the B steel is worth noting. It has been previously shown that $Fe_{23}(B,C)_6$ particles have varying effects on the nucleation of proeutectoid ferrite, depending on their sizes;[10] fine particles retard the nucleation of ferrite and coarse ones promote the nucleation of ferrite, although the critical size is not known. As shown in Figure 1(b), there are many coarse $Fe_{23}(B,C)_6$ particles within the prior austenite grains and some of these particles are associated with intragranular ferrite. This result suggests that the matrix $Fe_{23}(B,C)_6$ particles act as the preferential sites for the intragranular nucleation of ferrite. This is in agreement with the result of Ueda *et al.*,[11] who reported the enhanced nucleation of ferrite within austenite grains by boron addition.

Figure 3 shows the variation of values of RA with temperature. It shows that the A steel has a trough of RA between 700 °C and 850 °C. The temperature showing the smallest RA in the A steel is around 750 °C. The ductility trough is virtually nonexistent in the B steel. At 750 °C, which shows the smallest RA, the B steel has 95 % RA which is about 2 times larger than that of the A steel. This result indicates that the addition of B significantly increases the high-temperature ductility of Nb-containing steels. The SEM examination of the fracture surface shows that the fracture mode of

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Table I. Chemical Compositions of the Steels (Weight Percent)

Steels	Fe	C	Si	Mn	P	S	Al	Nb	B	N
A	bal	0.081	0.149	0.974	0.010	0.003	0.020	0.02	—	0.005 to 0.006
B	bal	0.081	0.139	0.983	0.011	0.003	0.031	0.02	0.002	

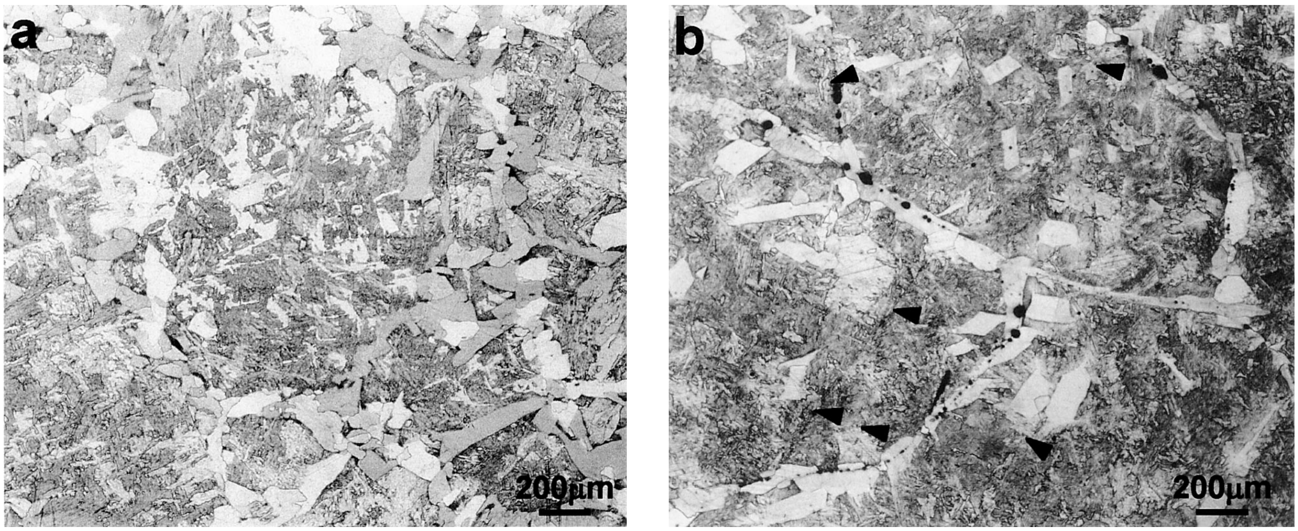


Fig. 1—Optical micrographs of the steels quenched after fracture at 750 °C: (*a*) A steel and (*b*) B steel.

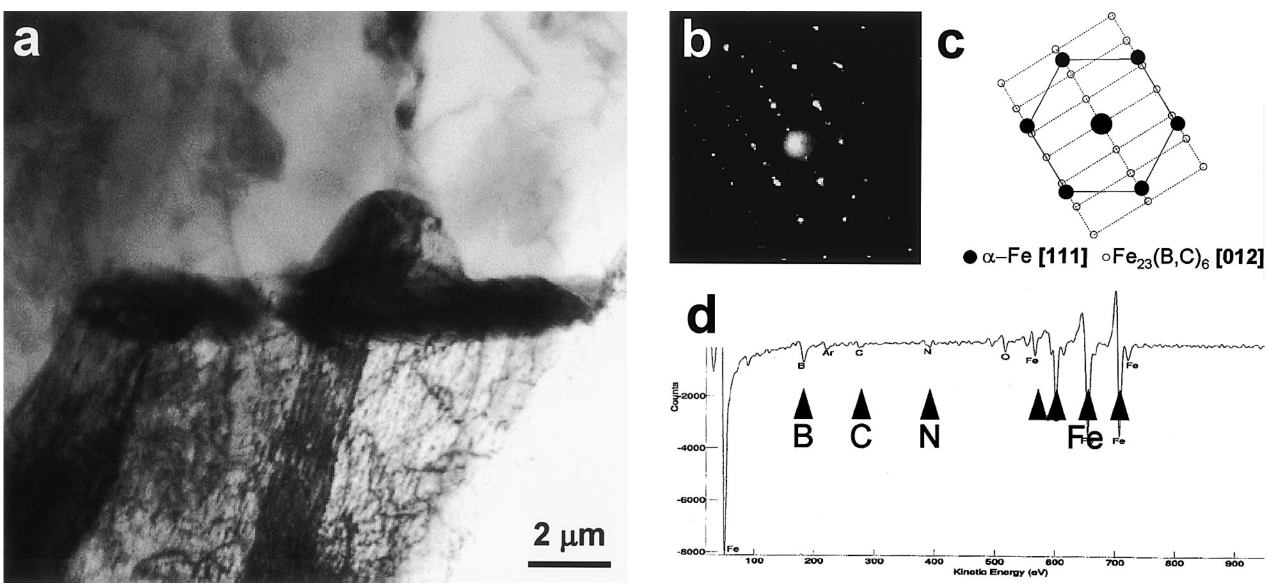


Fig. 2—(*a*) Bright-field TEM micrographs showing $Fe_{23}(B,C)_6$, (*b*) SADP, (*c*) schematic diagram of SADP, and (*d*) SAM analysis of $Fe_{23}(B,C)_6$.

the A steel is a typical intergranular fracture with fine dimples along the prior austenite grain boundaries (Figure 4(a)), while the B steel shows a completely ductile fracture (Figure 4(b)).

Several microstructural features have been suggested to be responsible for the ductility trough of Nb-containing steels at high temperatures. One is the presence of thin layers of ferrite along austenite grain boundaries.[1,2,12] During deformation at high temperatures, the strain is concentrated within the softer ferrite, leading to the formation of voids within the ferrite. Since ferrite exists in the form of thin layers along austenite grain boundaries, the localized deformation in ferrite results in intergranular fracture. Therefore, the temperature range for the ductility trough coincides with that of austenite-to-ferrite transformation, as shown in previous studies.[13–16] However, the ductility recovers as the thickness of ferrite layers increases with decreasing test temperature.[17] The other microstructural feature is the presence of fine Nb(C,N) particles within ferrite and at austenite grain boundaries.[7] The very fine Nb(C,N) precipitates increase

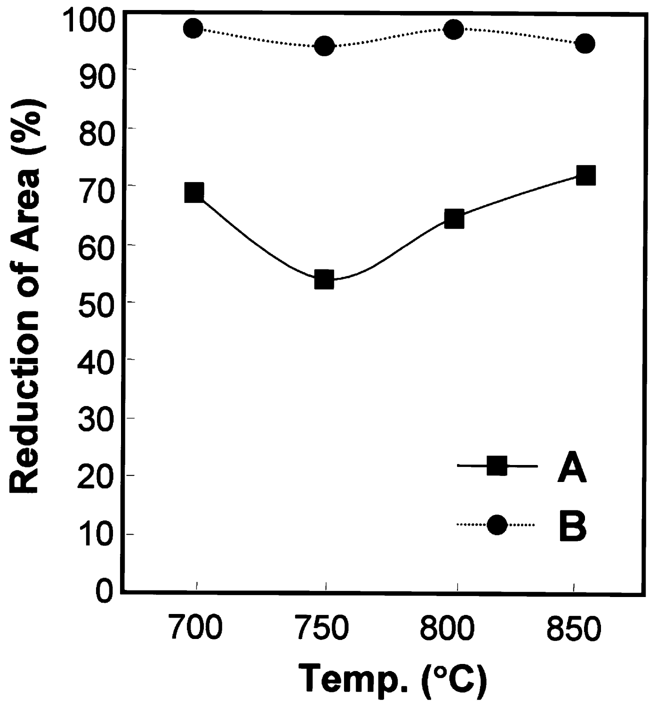


Fig. 3—Variation of RA with test temperature.

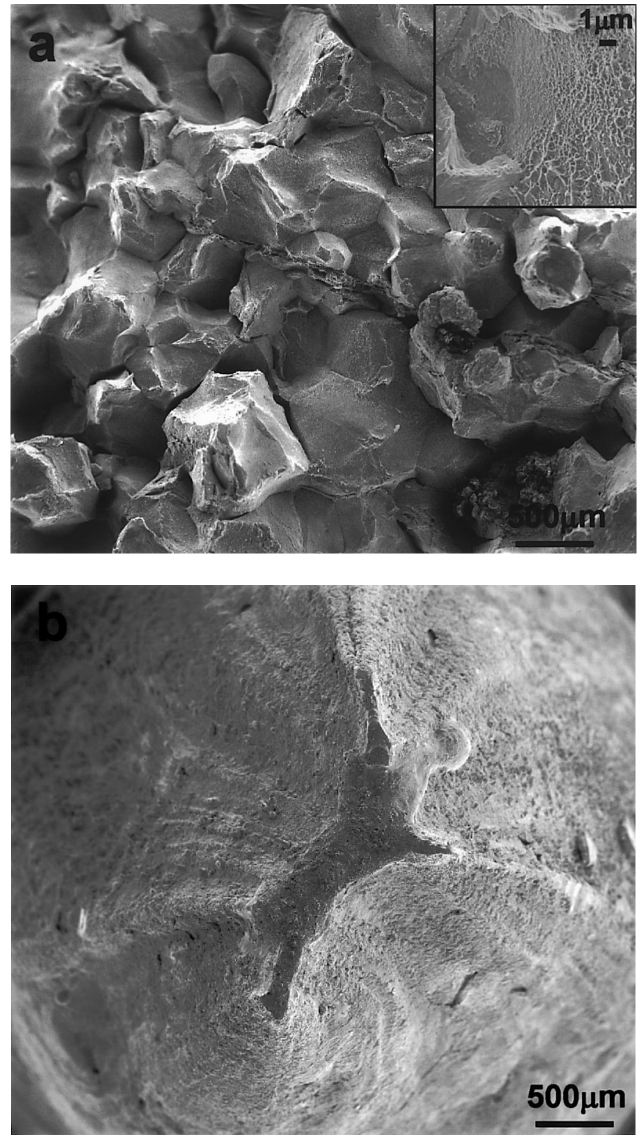
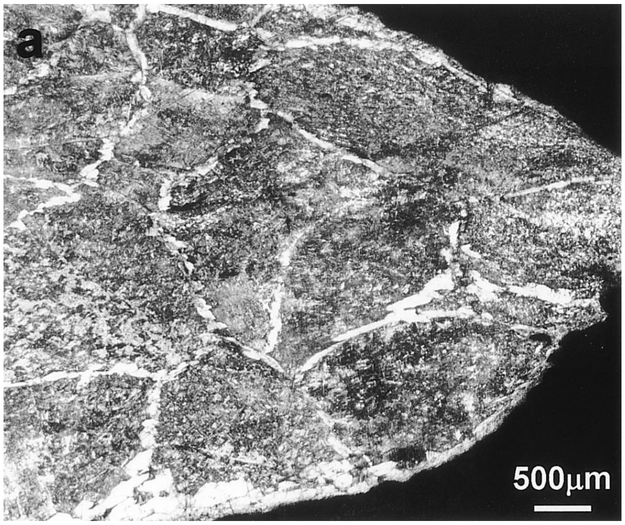


Fig. 4—SEM fractographs: (*a*) A steel (insert; magnified image) and (*b*) B steel.

the stress required for deformation and therefore increase the stress in the grain boundary region. Also, precipitate-free zones are often present at the boundaries, concentrating the strain into these regions. The result again is intergranular fracture. It should be noted that the ductility trough at high temperatures is associated with fine precipitation of Nb(C,N) particles. At low temperatures, on the other hand, poor ductility is usually associated with coarse precipitates.

As mentioned previously, the major difference between the base steel and the B-containing steels is the presence of intragranular ferrite in the latter. In the case of the base steel, which has only intergranular ferrite, the fracture mode is intergranular fracture with fine dimples (Figure 4(a)), indicating that fracture occurs by strain localization within the ferrite layers along the austenite grain boundaries. Observation of the cross section of the fractured specimen shows that the deformation of austenite grains is minimal in the A steel, even near the tip of the fractured surface, as shown in Figure 5(a). On the other hand, there is an extensive deformation of austenite and ferrite layers in the B steel (Figure 5(b)), indicating that the deformation of the B steel is not localized within the ferrite layers at austenite grain boundaries. This can be confirmed by the analysis of hot stress-strain curves. As shown in Figure 6, the B steel shows much higher strength and larger ductility than the A steel at 750 °C. Considering that the strength of austenite is much higher than that of ferrite at high temperatures,[2,18] the results shown in Figure 5 indicate that the deformation of the B steel is mostly controlled by austenite despite the presence of ferrite layers at austenite grain boundaries. It is believed that the presence of intraganular ferrite improves the homogeneity of deformation and accordingly the hot ductility of the B steel. As shown in previous investigations of multiphase steels, the morphology of the second phase is one of the most important factors controlling the deformation behavior and the resultant mechanical properties.[19,20] It has been shown that homogeneous deformation can be obtained in steels having a uniform distribution of a second phase. Although the majority of ferrite is at the austenite grain boundaries in the B steel, the presence of intragranular ferrite makes the distribution of ferrite more uniform in the B steel than in the A steel. Also, the presence of intragranular ferrite can have a beneficial effect on reducing the effective austenite grain size of the B steel. It has been shown that the refinement of austenite grain size decreases the depth and width of the ductility trough in plain C-Mn steels.[2,21] In microalloyed steels, the effect of grain size on the hot ductility has not received much attention because of the overriding effect of the precipitates. However, work considering the effect of grain size while keeping the precipitate distribution constant shows that the variation of the hot ductility with grain size is similar to those observed in plain C-Mn steels.[22] Therefore, the overall effect of the intragranular ferrite in the B steel is to improve the hot ductility.



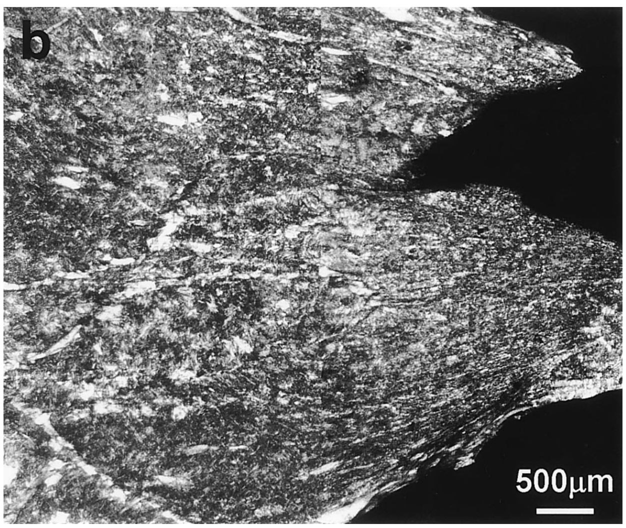


Fig. 5—Cross-sectional micrographs of specimens tested at 750 °C: (*a*) A steel and (*b*) B steel.

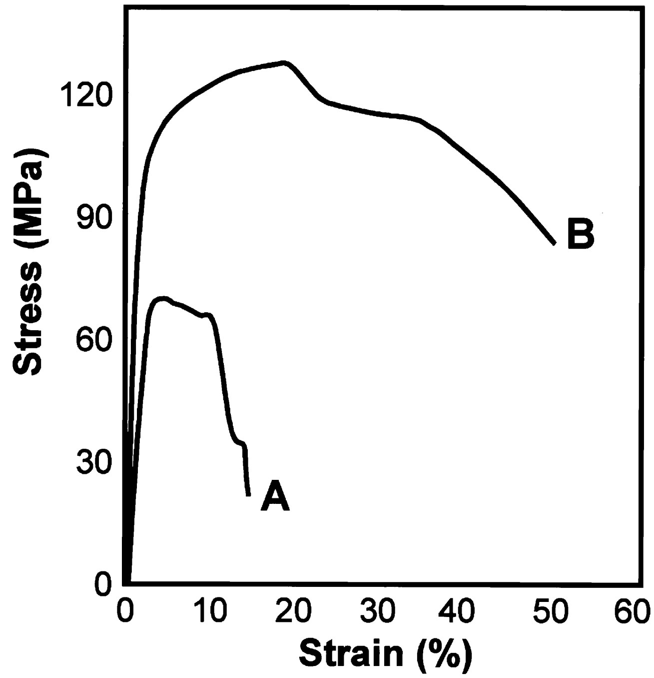


Fig. 6—Stress-strain curves obtained by Gleeble test at 750 °C: (*a*) A steel and (*b*) B steel.

As mentioned previously, the other microstructural feature responsible for the ductility trough of Nb-containing steels is the presence of fine Nb(C,N) precipitates. The effect of such precipitates is more dominant when the test temperature is at the low austenite region than at the austenite plus ferrite two-phase region since the ferrite layers have a overriding effect when austenite and ferrite coexist. It has been shown that the coarsening of precipitates is quite effective in improving the hot ductility of various steels.[2,6,8] The $Fe_{23}(B,C)_6$ particles present in the B steel are very coarse and therefore advantageous for the hot ductility, although most of them are located at austenite grain boundaries. Moreover, $Fe_{23}(B,C)_6$ particles contain N, as shown in Figure 2. This would reduce the amount of grain boundary N and C available for the precipitation of fine Nb(C,N) particles, which are detrimental to the hot ductility.

REFERENCES

1. Y. Maehara, K. Yasumoto, H. Tomono, T. Nagamichi, and Y. Ohmori: *Mater. Sci. Technol.*, 1990, vol. 6, pp. 793-806.
2. B. Mintz, S. Yue, and J.J. Jonas: *Int. Mater. Rev.*, 1991, vol. 36, pp. 187-215.
3. B.G. Thomas, J.K. Brimacombe, and I.V. Samarasekera: *Iron Steel Soc. Trans.*, 1986, vol. 7, pp. 7-20.
4. J.K. Brimacombe, F. Weinberg, and E.B. Hawbolt: *Metall. Trans. B*, 1979, vol. 10B, pp. 279-92.
5. J.K. Brimacombe and K. Sorimachi: *Metall. Trans. B*, 1977, vol. 8B, pp. 489-505.
6. K. Suzuki, S. Miyagawa, Y. Saito, and K. Shiotani: *Iron Steel Inst. Jpn. Int.*, 1995, vol. 35, pp. 34-41.
7. R. Abushosha, R. Vipond, and B. Mintz: *Mater. Sci. Technol.*, 1991, vol. 7, pp. 613-21.
8. R. Abushosha, O. Cominelli, and B. Mintz: *Mater. Sci. Technol.*, 1999, vol. 15, pp. 278-86.
9. B. Mintz, Z. Mohamed, and R. Abushosha: *Mater. Sci. Technol.*, 1989, vol. 5, pp. 682-88.
10. Ph. Maitrepierre, D. Thivellier, and R. Tricot: *Metall. Trans. A*, 1975, vol. 6A, pp. 287-301.
11. S. Ueda, M. Ishikawa, and N. Ohashi: in *Boron in Steel*, S.K. Banerji and J.E. Morral, eds., TMS-AIME, New York, NY, 1980, pp. 181-98.
12. X.P. Li, J.K. Park, J. Choi, and C.H. Yim: *Met. Mater.*, 1999, vol. 5, pp. 25-32.
13. G.I.S.L. Cardoso, B. Mintz, and S. Yue: *Ironmaking and Steelmaking*, 1995, vol. 22, pp. 365-72.
14. B. Mintz and J.J. Jonas: *Mater. Sci. Technol.*, 1994, vol. 10, pp. 721-27.
15. B. Mintz, R. Abushosha, and M. Shaker: *Mater. Sci. Technol.*, 1993, vol. 9, pp. 907-14.
16. D.N. Crowther and B. Mintz: *Mater. Sci. Technol.*, 1986, vol. 2, pp. 671-76.
17. D.N. Crowther, Z. Mohamed, and B. Mintz: *Metall. Trans. A*, 1987, vol. 18A, pp. 1929-39.
18. I. Tamura, H. Sekine, T. Tanaka, and C. Ouchi: in *Thermomechanical Processing of High-Strength Low-Alloy Steels*, Butterworth & Co. Ltd., London, 1988, pp. 124-25.
19. N.J. Kim and G. Thomas: *Metall. Trans. A*, 1981, vol. 12A, pp. 483-89.
20. H.J. Koh, S.K. Lee, S.H. Park, S.J. Choi, S.J. Kwon, and N.J. Kim: *Scripta Metall.*, 1998, vol. 38, pp. 763-68.
21. T. Maki, T. Nagamichi, N. Abe, and I. Tamura: *J. Iron Steel Inst.*, 1985, vol. 71, pp. 1367-74.
22. D.N. Crowther and B. Mintz: *Mater. Sci. Technol.*, 1986, vol. 2, pp. 1099-1105.