Experiment and Numerical Modeling of Boron Effect on Weld HAZ Hardenability of Steels

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\textbf{Abstract.} Boron is an alloying element used for increasing the strength of steels, but on the other hand, it sometimes deteriorates heat affected zone (HAZ) weldability and toughness. Although it is widely known that the hardenability of steel is improved due to grain boundary segregation of boron, its details and quantification have not yet been clarified. In this study, boron-free HT590 and boron-added HT780 steels were welded and microstructure and hardness distribution in HAZ were measured, in detail. Furthermore, we calculated and investigated the grain boundary segregation and diffusion of boron during weld thermal cycle by numerical analysis while comparing it with the experiment result. As a result, the present model made it clear that the effect of boron on HAZ hardenability can be expressed by non-equilibrium segregated boron at grain boundary, quantitatively.

1. Introduction

Boron (B) is a micro alloy element which increases hardenability of steels and it has been widely utilized in steel industry since many years ago [1]. Only a small addition of B, as low as 10ppm, has a tremendous effect on the hardenability of steels [2, 3]. It is widely accepted that B atoms are segregated at austenite grain boundary (GB) and reduce GB energy, thereby retarding austenite to ferrite, bainite and martensite transformation [4-11]. Relationship between concentration of B at G.B. and hardenability has been examined in a wide variety of steel chemical compositions and thermal histories, including isothermal and continuous cooling, intensively. During cooling process Fe-(C,B) precipitation may take place, in which case hardenability is reduced. It is also known that the addition of Molybdenum retards the precipitation, thereby maintaining the B hardenability effect [5, 10, 11]. The boron effect on steels’ hardenability is not limited to heat treatment of steels but also has an influence on phase transformation of weld heat-affected zone (HAZ) [12, 13]. In the latter case, HAZ hardness is elevated in B-added steels and sometimes promotes HAZ cold cracking and reduction in HAZ toughness. Generally, B has a deleterious effect on HAZ weldability.

As mentioned above, B increases HAZ hardenability but in somewhat unusual manner. In B-free steels, maximum hardness is realized at coarse-grained (CG) HAZ near fusion line (FL) and the
hardness decreases with increasing distance from the FL, along with lowered peak temperature of weld thermal cycle and reduced austenite grain size. However in B-added steels, maximum hardness often appears in the HAZ well apart from the FL, where the peak temperature is lower than that at CG HAZ [12, 13]. It is argued that the B segregation at GB is enhanced at lower peak temperature due to non-equilibrium segregation [13]. But concrete mechanism has not been established, yet.

In the present study, weld HAZ hardenability was investigated in B-free and B-added steel. Furthermore, numerical simulation on B diffusion and segregation was conducted and the influence of peak temperature and cooling rate on B hardenability are discussed.

2. Weld experiment
Welding was conducted to examine the B hardenability effect in HAZ.

2.1. Steels tested
Table 1 shows chemical compositions of the steel plates tested in the welding. They are as-rolled HT590 and quenched-tempered HT780 commercial steels. The HT590 steel contains practically no B and the HT780 steel contains 12ppm B. Plate thickness was 19mm in the both steels.

| Steel | Chemical compositions(mass%) | Plate thickness(mm) |
|-------|-----------------------------|---------------------|
|       | | C | Si | Mn | P | S | Cr | Ti | Al(sol) | B[ppm] | N[ppm] | |
| H590  | | 0.10 | 0.24 | 1.70 | 0.017 | 0.005 | 0.02 | 0.091 | 0.033 | 1 | 30 | 19 |
| H780  | | 0.15 | 0.23 | 1.28 | 0.012 | 0.006 | 0.25 | 0.012 | 0.025 | 12 | 37 | 19 |

2.2. Welding conditions
Submerged-arc welding was conducted. Table 2 shows the welding conditions, Heat input was 40kJ/cm. Weld bead was placed on the steel plate, i.e., bead-on-plate welding. A section sample was cut from the weld, polished and etched by 2% nital solution to reveal microstructure. Vickers hardness was measured traversing weld metal (WM), FL, CG HAZ, fine-grained (FG) HAZ, inter critical (IC) HAZ and base metal (BM) with indentation force of 0.5N.

| Wire | Flux | Wire diameter (mm) | Current(A) | Voltage(V) | Speed(cm/min) | Heat input(KJ/cm) |
|------|------|--------------------|------------|------------|---------------|-------------------|
| Linconweld | Lincolnweld | 2.4 | 500 | 50 | 37 | 40 |

2.3. Result
Figure 1 shows microstructure distributions with Vickers hardness indentations for the HT590 and HT780 welds. Figure 2 shows hardness distributions of the both welds. The HT590 weld shows typical hardness profile; maximum hardness appeared at CG HAZ near the FL and decreased with increasing distance from the FL. Contrary to the HT590 weld, the HT780 weld showed nearly flat hardness profile from the FL to FG HAZ but with considerable hardness drop at inter critical HAZ. It should be noted that the HT590 steel contains 0.091% Ti, which might have increased overall HAZ hardness by Ti carbide precipitation after phase transformation. Figure 3 shows microstructures at WM, CG HAZ, FG HAZ, IC HAZ and BM. The both welds showed columnar structure at WM, and predominantly bainitic microstructure in HAZ and ferrite-pearlite (HT590) and martensitic microstructures (HT780) at BM.
Figure 1. Microstructure distributions in weld.

Figure 2. Hardness distributions of welds.

Figure 3. Microstructures in weld.
Figure 4. Calculated thermal history.

It should be noted that average cooling rate between 800 °C and 500 °C was calculated as about 5 °C/s using in-house software of weld heat conduction analysis, see figure 4. The hardness profile shown in figure 2 indicates that HAZ hardenability is maintained in the HT780 weld even if the peak temperature of weld HAZ thermal cycle is low, as compared with the hardenability drop from CG to FG HAZ in the HT590 steel.

3. Numerical analysis of boron diffusion and GB segregation

For the purpose of understanding the possible change of the boron hardenability effect with thermal cycle conditions, segregation of B at GB was calculated in conjunction with B diffusion in austenite, numerically

3.1. Procedure of numerical analysis

The theory of equilibrium segregation of impurity elements was established by Mclean [14]. The present analysis follows his theory. Eq.(1) gives concentration of B at GB, $C_{B(GB)}$.

$$C_{B(GB)} = \frac{C_B \exp(-Q_B/RT)}{1 - C_B + C_B \exp(-Q_B/RT)}$$

where $C_B$ is concentration of B in austenite, $Q_B$ is activation energy, $R$ is universal gas constant and $T$ is temperature. This equation may be applied if temperature is kept for sufficiently long period. However, the sample is continuously cooled from peak temperature down to phase transformation temperature, in which case non-equilibrium segregation must be considered in conjunction with B diffusion in austenite grain, see figure 5. Coupled analysis of the non-equilibrium segregation and diffusion of B in austenite was conducted numerically in the present study. It is assumed that equilibrium distribution of B is achieved at peak temperature because peak temperature is relatively high and diffusion is fast. Upon cooling, local equilibrium between GB and austenite just adjacent to the GB can be assumed, in which case $C_B$ in Eq.(1) is regarded as B concentration at $x = 0$ (austenite just adjacent to the GB). Assuming one-dimensional diffusion, B diffusion in austenite grain is expressed as,

$$\frac{\partial C_B}{\partial t} = D_B \frac{\partial^2 C_B}{\partial x^2}$$

where $D_B$ is diffusion constant of B in austenite grain and a function of temperature. Flux of B atoms from austenite to GB is calculated as,

$$J_{B(GB)} = -D_B \frac{\partial C_B}{\partial x} \bigg|_{x=0}$$
Increase in $C_{B(GB)}$ during time increment can be calculated as,

$$\frac{dC_{B(GB)}}{dt} = \frac{J_{B(GB)}}{w_{GB}/2}$$

where $w_{GB}$ is GB width. $C_B$ and $C_{B(GB)}$ can be calculated numerically along the thermal cycle.

1: Set initial value of $C_{B(x)}$ (uniform in austenite grain) and $C_{B(GB)}$ at initial (peak) temperature.
2: Determine temperature at next time step, following the assumed weld thermal cycle.
3: Determine $C_{B(x=0)}$ in local equilibrium with $C_{B(GB)}$.
4: Solve diffusion equation using finite-difference algorithm and obtain $C_{B(x)}$.
5: Calculate $C_{B(GB)}$ using $dC_{B(GB)}$.
6: Repeat 2 to 5.

![Schematic diagram of boron diffusion](image)

Figure 5. Schematic diagram of boron diffusion.

3.2. Results of numerical calculations

Figure 6 shows change of boron concentration with time for peak temperature: 1000°C, 1400°C, where the cooling curves of figure 4 were applied. It can be seen that at lower peak temperature B diffuses faster to the GB in spite of the lower diffusion constant. This is due to the local equilibrium between GB B and that in grain adjacent to the GB. Figure 7 shows change of GB B concentration with time for different peak temperatures, but plotted against temperature during cooling. The GB B concentration increases with decreasing temperature during cooling but its value became higher for the lower peak temperature. Figure 8 shows GB B concentration at 650°C as a function of peak temperature. It is clearly seen that GB B concentration was higher with lower peak temperature, especially at 900°C.
Figure 6. Calculated Boron concentration in crystal grains.

Figure 7. Calculated Boron grain boundary segregation concentration.

Figure 8. Calculated peak heating temperature dependence of GB segregated boron concentration at 650°C.
3.3. Discussions

Although the hardness distribution of weld HAZ HT780 tended to be flattened than HT 590, the change in the hardness distribution by B was not necessarily clear in the present experiment. It is known that GB B is sometimes precipitated as Fe₂₃(CB)₆ in the course of cooling. In this case, the effect of B on hardenability is reduced [2, 7, 11]. On the other hand, Mo is known to suppress the B precipitation [10, 11]. The HT780 steel in the present experiment did not contain Mo. Therefore, results of HAZ hardness measurement of B added steel containing Mo by Kasuya et al [12] are shown in figure 9 and figure 10, together with B-free steel. Figure 9 shows schematic illustration of hardness distribution. While the B-free steel had hardness distribution similar to the present experiment, see figure 10(a), highest hardness is realized at a position away from FL in the Mo-B steel, see figure 10(b). This result corresponds to the dependence of GB B concentration on peak temperature, see figure 8. That is, the GB B concentration is high when the peak temperature is 1000°C or lower. This result suggests that the effect of B on HAZ hardenability can be explained by GB B concentration. However, it is necessary to consider the influence of B precipitation at the same time for Mo free B added steels.

![Figure 9](image_url)

**Figure 9.** Schematic illustration of hardness distribution [12].

![Figure 10](image_url)

**Figure 10.** Hardness distribution of the steel [12].

4. Conclusions

(1) It has been made clear from the present experiment together with the published data that weld HAZ hardness distribution changes with the addition of B.

(2) The present numerical simulation on the diffusion and grain boundary segregation of B showed a change of GB B concentration on peak temperature of weld HAZ thermal cycle.

(3) It is suggested that HAZ hardenability of B added steels can be explained by grain boundary B concentration.

In the future, it is necessary to analyze the grain boundary segregation of B considering B precipitation.
5. References
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