1. Introduction

Piping methods of linepipe steels used for transporting crude oil or natural gas can be mainly classified by steel shapes. Piping methods such as U-ing, O-ing, and expansion (UOE), roll bending (R/B), and J-ing, C-ing, and O-ing (JCO) are applied for plate-type steels, while spiral seam piping and electric resistance welding (ER W) are applied to coil-type steels. Seam welding is then conducted on the fabricated pipes in order to keep the shape up. Submerged arc welding (SAW) is used for plate-type steels, while the ER W is used for coil-type steels. The welded pipes are transported to the places for crude oil or natural gas, and then are connected by welding finally to complete pipelines. For the girth welding to connect pipes, conventional methods of shielded metal arc welding (SMAW) and gas metal arc welding (GMAW) and recently-developed methods including electron beam welding (EBW) and laser-arc hybrid welding are used. To improve the welding speed while maintaining original properties of steels, new innovative girth welding methods, in which the input heat is reduced and the welding can be done in a shorter period of time, have been developed.

While linepipes are made by piping and welding processes, heat affected zones (HAZs) are formed according to thermal cycles of rapid heating and cooling, and Charpy impact properties of the welded regions can be deteriorated because of the growth of prior austenite grains, the precipitation of secondary phases or particles, and the occurrence of residual stress. To address these problems, oxide metallurgy technology using various oxides have been studied lately. In this technology, thermally stable oxides are formed in order to prevent the deterioration of HAZ properties. Some effective non-metallic oxides frequently used include Ti, Al, and Mg oxides. These oxides are known as suitable sites for inhomogeneous nucleation of ferrite in plain low-carbon steels because they readily nucleate ferrite inside austenite grains during the austenite to ferrite transformation.

The present study aimed at improving Charpy impact properties of the HAZs in API X70 linepipe steels by applying the oxide metallurgy technology. Oxides were formed by adding alloying elements such as Ti, Al, and Mg to the API X70 steels, and effects of these oxides on HAZ microstructures and Charpy impact properties were investigated. Oxides present in the steels were sized about 2 μm, and had characteristics of complex oxides composed of various elements. The steel containing more Al, Ti, and Mg showed higher volume fraction of oxides than the conventional steel. After the HAZ simulation test, the oxide-containing steel HAZs were composed mainly of acicular ferrite, together with small amount of bainite, whereas the conventional steel HAZ were composed mainly of bainite with a small amount of acicular ferrite. This formation of acicular ferrite in the oxide-containing steel HAZs was associated with the active nucleation of acicular ferrite at oxides, thereby leading to the great (five times or more) improvement of Charpy impact properties of the HAZs over the conventional steel HAZs.

KEY WORDS: API X70 linepipe steel; heat affected zone; complex oxides.
2. Experimental

2.1. API X70 Linepipe Steels

The steels used in this study were API X70 grade steels having minimum yield strength levels of 483 MPa (70 ksi), and their chemical compositions and rolling conditions are shown in Tables 1 and 2, respectively. An overall grain refinement effect was expected by rolling with a high rolling reduction ratio in the non-recrystallized region of austenite after austenitization at 1050–1150°C.14–16) Rolling was finished at the temperature of the austenite region above Ar3. After the finish rolling, the steels were rapidly cooled down to finish cooling temperatures of 470–500°C. For convenience, the conventional steel and the steel, in which Ti, Al, and Mg were additionally contained to form more oxides, are referred to as ‘S0’ and ‘S1’, respectively (Table 1).

2.2. Microstructural Analysis

The steels were polished and etched in a 2% nital solution, and microstructures of longitudinal-transverse (L-T) planes were observed by an optical microscope and a scanning electron microscope (SEM, model; S-4300E, Hitachi, Tokyo, Japan, resolution: 0.2 μm). Chemical compositions of fine oxides were analyzed by energy dispersive spectroscopy (EDS) installed in the SEM. Oxides were electrolyzed in a 10% AA solution (methyl alcohol 89% / H11001 acetylacetone 10% / H11001 tetramethylammonium chloride 1%) under ultrasonic waves, extracted, separated, and then quantitatively analyzed.17) Since microstructures of the steels were quite complicated and its grains are very fine, the prior austenite grain size was measured by electron backscatter diffraction (EBSD) analysis (resolution; 0.2 μm).18) The EBSD analysis was conducted on the steels by using a field emission scanning electron microscope (FE-SEM, Model: S-4300SE, Hitachi, Tokyo, Japan). The data were then interpreted by an orientation imaging microscopy analysis software provided by TexSEM Laboratories (Provo, UT, USA), Inc.

2.3. Tensile and Charpy Impact Tests

Round tensile specimens having a gage diameter of 6 mm and a gage length of 30 mm were prepared in the transverse direction, and were tested at room temperature at a crosshead speed of 5 mm/min by an Instron machine of 100 kN capacity. Charpy impact tests were performed on standard Charpy V-notch specimens (size; 10×10×55 mm, orientation; transverse-longitudinal (T-L)) in the temperature range from −196°C to 20°C by a Tinius Olsen impact tester of 500 J capacity (model: FAHC-J-500-01, JT Toshi, Tokyo, Japan).19) In order to reduce errors in the data interpretation, a regression analysis for absorbed impact energy vs. test temperature was conducted with a hyperbolic tangent curve fitting method.20,21) Based on the regression analysis data, the energy transition temperature (ETT) which corresponds to the average value of upper shelf energy (USE) and lower shelf energy (LSE) was determined.

2.4. Weld Thermal Cycle Simulation Test

It is very difficult to systematically analyze actual welds because HAZ microstructures of multi-pass welds are complicated and inhomogeneous. Thus, weld thermal cycle simulation tests were conducted on the two API X70 steels for the systematic investigation of the HAZ microstructure and its effect on Charpy impact properties by a metal thermal cycle simulator (MTCS, model; Thermorestor-W, Fuji Electronic Industrial Co., Ltd, Tokyo, Japan). The thermal cycle of the weld simulation is characterized by the peak temperature (Tp) and the cooling time from 800°C to 500°C (Dt800–500). The thermal cycle used is shown schematically in Fig. 1 and Table 3. The coarse-grained HAZ near the welded fusion line, which is generally known as the most brittle one among HAZs, was considered as a local brittle zone. After reaching the peak temperature (Tp) of 1400°C and a resting time of 5 s, the specimens were cooled down with Dt800–500 of 18 s, 30 s, and 59 s. The cooling times of 18 s, 30 s, and 59 s are approximately equivalent to those of a submerged arc welding or electric resistance welding with heat inputs of 30 kJ/cm, 50 kJ/cm, and 100 kJ/cm, respec-

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Table 1. Chemical compositions of the two API X70 linepipe steels (wt%).

| Steel | C  | Si  | Mn  | Ni + Mo + Nb | Ti + Al | Mg  | P  | S  | O  | Cu* |
|-------|----|-----|-----|-------------|--------|-----|----|----|----|-----|
| S0    | 0.063 | 0.2 | 1.79 | 0.4 | 0.02 | -   | <0.008 | <0.002 | <0.005 | 0.41 |
| S1    | 0.057 | 0.2 | 1.77 | 0.4 | 0.03 | 0.001 | <0.002 | <0.002 | <0.005 | 0.41 |

* C eq (%) = C + Mn/6 + Si/24 + Ni/40 + Cr/5 + Mo/4 + V/14

Table 2. Rolling and cooling conditions of the two API X70 linepipe steels.

| Steel | Reheating Temp. (°C) | Cooling Temp. (°C) | Rolling Temperature (°C) | Cooling Rate (°C/s) | Thickness (mm) |
|-------|----------------------|--------------------|--------------------------|---------------------|----------------|
| S0    | Above Ar3            | 470–500            | 1400                     | 12                  | 14            |
| S1    | Above Ar3            | 470–500            | 1400                     | 12                  | 14            |

* SCT, FCT, and CR are start cooling temperature, finish cooling temperature, and cooling rate, respectively.

Table 3. Weld thermal cycle simulation conditions.

| Peak Temp. (°C) | Holding Time at Peak Temp. (sec) | Cooling Time to 800°C (sec) | Δt800–500 (sec) | Heat Input (kJ/cm) |
|-----------------|---------------------------------|-----------------------------|-----------------|-------------------|
| 1400            | 5                               | 25                          | 18              | 30                |
| 1400            | 30                              | 50                          | 59              | 100               |

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Fig. 1. Schematic diagram of weld thermal cycles of the steels.
3. Results

3.1. Microstructure of API X70 Linepipe Steels

Figures 2(a) through 2(d) are optical and SEM micrographs of the L-T plane of the S0 and S1 steels, from which volume fractions of microstructures such as acicular ferrite (AF), martensite-austenite constituent (MA), and cementite were measured as listed in Table 4. The S0 and S1 steels consist mostly of AF, while about 3–4 vol% of MA or cementite are found. The effective grain sizes of the two steels are small at about 5–6 μm.

Figure 3(a) is an SEM micrograph of an oxide present in the S1 steel, which was analyzed by EDS as shown in Fig. 3(b). The EDS result clearly indicates that this oxide has characters of complex oxides as it is composed of many elements including Ti, Al, and Mg. The quantitative analysis result of complex oxides using the extraction and separation method as described in Sec. 2.2 is presented in Fig. 3(c).

Table 4. Volume fractions of acicular ferrite (AF) and secondary phases present in the steels and effective grain size.

| Steel | Volume Fraction (%) | Effective Grain Size (μm) |
|-------|---------------------|--------------------------|
| S0    | Bal. 2.9 ± 1.1       | 5.0 ± 1.5                |
| S1    | Bal. 3.7 ± 1.0       | 6.3 ± 1.9                |

*Secondary phases include cementite, martensite, and martensite-austenite constituent.

Fig. 2. Optical and SEM micrographs (L-T plane) of the (a) and (b) S0, and (c) and (d) S1 steels. Nital etched.

Fig. 3. (a) SEM micrograph and (b) EDS spectrum of a complex oxide present in the S1 steel, and (c) frequency of number of oxides in 5 g steel measure by the extraction and separation method.
in 1 cm³, average oxide size, and volume fraction of oxides were measured, and are listed in Table 5. The average oxide size is about the same in the two steels, ranging 2.1–2.3 μm. In the S1 steel, few oxides larger than 5 μm are found. Ti, Al, and Mg form oxides well, 7–13) so the S1 steel has the larger total number of oxides in 5 g steel than the S0 steel. Since oxides are sized fine, the overall volume fraction of oxides is extremely low. However, the oxide volume fraction of the S1 steel is twice as high as that of the S0 steel.

### 3.2. Tensile and Charpy Impact Properties of API X70 Linepipe Steels

Room-temperature tensile stress–strain curves of the two steels are presented in Fig. 4. From the curves, yield strength, tensile strength, elongation, and yield ratio were measured, and the results are summarized in Table 6. The two steels show a continuous yield behavior. The S0 steel shows higher yield strength but lower tensile strength than the S1 steel, and has high yield ratio of 85% and lower elongation of 18%. The S1 steel has lower yield ratio of 70% and higher elongation of 24%. The uniform elongation of the S1 steel is higher than that of the S0 steel because of the difference in hardenability. The difference of uniform elongation is due to the difference in hardenability, the lower hardenability, the higher uniform elongation.6,12) The two steels are mainly composed of acicular ferrite, but their cooling rates are different as shown in Table 2. Since the cooling rate of the S1 steel is slower than that of the S0 steel, the dislocation density present in the AF of the S1 steel might be lower than that of the S0 steel. Thus, under a tensile loading condition, the S1 steel shows the slower strain hardening and the higher uniform elongation.

Figure 5 presents the variation of Charpy absorbed energy vs. test temperature. From these curves, USE, ETT, and absorbed energy at −20°C and 0°C were measured, and the results are summarized in Table 7. The USE of the S0 steel is higher (over 300 J) than that of the S1 steel. The main reason for the difference in USE of the base metal is the formation of different microstructures according to the alloying elements and rolling conditions. The volume fraction of secondary phases and prior austenite grain size of the S1 steel are higher or larger than those of the S0 steel. Thus, the USE of the S1 steel having a larger amount of secondary phases and larger grain size is lower than that of the S0 steel. The absorbed energies at −20°C and 0°C are about the same as the USE, and the absorbed energy at 0°C is slightly higher than the absorbed energy at −20°C. The ETTs of the two steels are very low below −100°C, while the ETT of the S1 steel is somewhat lower.

### 3.3. Microstructure of Simulated HAZ of API X70 Linepipe Steels

Figures 6(a) through 6(d) are optical and SEM micrographs of the S0 and S1 steels after the weld simulation test with the heat input of 30 kJ/cm, and Table 8 lists the volume fraction of bainite, AF, and secondary phases and the prior austenite grain size (AGS). The S0 steel HAZ is mostly composed of bainite according to rapid heating and cooling. It contains 32 vol% of AF and 5 vol% of secondary phases in the bainite matrix, while the S1 steel HAZ contains 3 vol% of secondary phases and 3 vol% of bainite in the AF matrix. The AGS is a little larger in the S1 steel HAZ and the S0 steel HAZ (Table 8). It is likely that the pinning effect of oxides does not occur because the AGS of the S1 steel having larger number of oxides is larger than that of the S0 steel. Optical and SEM micrographs of the steel HAZs with the heat input of 50 kJ/cm and 100 kJ/cm are presented in Figs. 7(a) through 7(d) and Figs. 8(a) through 8(d). As the heat input increases, the S0 steel
HAZs show the presence of bainite, secondary phases, and AF, like in the case with the heat input of 30 kJ/cm, although the volume fraction of AF and secondary phases tend to decrease. The S1 steel HAZs show the decreased volume fraction of AF with increasing heat input because of the reduced cooling rate.

3.4. Charpy Impact Properties of Simulated HAZ of API X70 Linepipe Steels

Table 9 lists the Charpy absorbed energy at $-20^\circ$C and $0^\circ$C of the two steels after the weld simulation test. Irrespective of heat input, the S0 steel HAZs show low absorbed energy at $-20^\circ$C (about 30 J). On the other hand, the absorbed energy at $-20^\circ$C of the S1 steel HAZ simulated with the heat input of 30 kJ/cm is 216 J, which is seven to eight times higher energy than those of the S0 steel HAZs. When the heat input increases to 50 kJ/cm, the absorbed energy at $-20^\circ$C of the S1 steel HAZ slightly decreases down to 185 J, and much further decreases down to 65 J under the larger heat input of 100 kJ/cm. The absorbed energies at $0^\circ$C are somewhat higher than those at $-20^\circ$C, but both of them show a similar tendency with respect to the steel HAZ and heat input variations.

Figures 9(a) through 9(f) and 10(a) through 10(h) are optical and SEM fractographs of Charpy impact specimens fractured at $-20^\circ$C for the S0 and S1 steel HAZs. The cleavage fracture is predominant on the fractured surfaces of the S0 steel HAZs (Figs. 9(a) through 9(f)). The S1-30 and S1-50 steel HAZs contain shear lips in the specimen edge area, and show cleavage fracture facets (dotted areas in Figs. 10(a) and 10(d)), together with ductile dimpled areas (Figs. 10(b) and 10(e)), whereas the S1-100 steel HAZ shows the cleavage fractured surface (Figs. 10(g) and 10(h)). The cleavage area fraction is measured to be 26% and 50% in the S1-30 and S1-50 steel HAZs, respectively. In the S1 steel HAZs, the cleavage area fraction increases as the heat input increases.

Figures 11(a) through 11(d) show SEM micrographs of the cross-sectional area beneath the cleavage fracture surface of the Charpy impact specimens fractured at $-20^\circ$C. The nickel was coated in order to protect the fracture surface as marked in micrographs. The cleavage crack path propagates linearly along effective grains having high-angled boundaries, and is deviated at boundaries between effective grains. Since bainite having small misorientations between subgrains has large effective grains, its unit crack path is long, whereas the unit crack path of AF having large misorientations between grains is short. In the S0-30 steel HAZ, cleavage cracks propagate linearly without microstructural deformation, and tend to be deviated frequently at bainite/AF interfaces. In some area, a crack initiated at a bainite/AF interface propagates into the interior of bainite. This crack linearly propagates long because of large effective grain size of bainite, and its unit path reaches 40 $\mu$m or longer (Fig. 11(a)). In ductile fractured regions in the S1-30 and S1-50 steel HAZs, deformed regions can be observed as indicated by arrows in Figs. 11(b) and 11(c). In the S1-100 steel HAZ, the crack propagation path is deviated at AF boundaries, while the crack linearly propagates at bainite regions without deviation of the prop-

Table 8. Volume fractions of bainite, acicular ferrite, and secondary phases present in the steel HAZs.

| Steel HAZ | Volume Fraction (%) | Prior Austenite Grain Size (µm) |
|-----------|---------------------|-------------------------------|
|           | Bainite | Acicular Ferrite | Secondary Phases*          |                      |
| S0-30     | Bal.    | 32.3 | 5.4 | 75 ± 6 |
| S0-50     | Bal.    | 14.9 | 4.5 | 91 ± 7 |
| S0-100    | Bal.    | 7.6  | 2.4 |
| S1-30     | Bal.    | 94.2 | 2.6 |
| S1-50     | Bal.    | 93.9 | 2.3 |
| S1-100    | Bal.    | 42.5 | 3.0 |

* Secondary phases include cementite, martensite, and martensite-austenite constituent.

Fig. 6. Optical and SEM micrographs of the (a) and (b) S0-30, and (c) and (d) S1-30 steel HAZs. Nital etched.
agation path (Fig. 11(d)). The crack propagated into the interior of AF is deviated at boundaries between AFs, and its unit path is quite short (about 15 μm). Cracks are also observed inside bainite regions.

4. Discussion

The HAZ microstructures of API X70 steels are mainly composed of bainite, AF, and secondary phases such as martensite or MA as mentioned in Sec. 3.1. Due to rapid cooling or quenching, martensite is formed by diffusionless shear transformation from austenite without changes in its chemical composition, is shaped in a lath or a plate, has high dislocation density inside, and shows low toughness.23) In bainite, aligned, elongated, and parallel-island-type retained austenite (RA) and MA are present together with well-developed laths. Since the orientation relations between nearby laths are almost the same, laths can be

Table 9. Charpy absorbed energy of the steel HAZs.

| Steel | Charpy Absorbed Energy at 0°C (J) | Charpy Absorbed Energy at -20°C (J) |
|-------|----------------------------------|-----------------------------------|
|       | 30 kJ/cm | 50 kJ/cm | 100 kJ/cm | 30 kJ/cm | 50 kJ/cm | 100 kJ/cm |
| S0    | 48       | 34       | 43        | 29       | 30       | 32        |
| S1    | 238      | 215      | 173       | 216      | 185      | 65        |

Fig. 7. Optical and SEM micrographs of the (a) and (b) S0-50, and (c) and (d) S1-50 steel HAZs. Nital etched.

Fig. 8. Optical and SEM micrographs of the (a) and (b) S0-100, and (c) and (d) S1-100 steel HAZs. Nital etched.
grouped in packets. It has high interior dislocation density, and prior austenite grain boundaries are well preserved.\textsuperscript{18,24} AF is nucleated and grown arbitrarily inside austenite grains, showing an acicular shape with irregular boundaries. It has a good combination of strength and toughness because of high interior dislocation density and fine grain size.\textsuperscript{18,24,25}

The S0 steel HAZ simulated with the heat input of 30 kJ/cm is mainly composed of bainite, with minor presence of AF and secondary phases (Table 8). However, in the S1 steel HAZ, 94 vol\% of AF is mainly formed besides bainite, and some AFs have a dense structure intersected each other (Figs. 6(a) through 6(d)). When the heat input increases, the S0 steel HAZs do not show much variation in their microstructures, mostly composed of bainite. In the S1 steel HAZs, AF is well developed, although the volume fraction of AF is significantly reduced according to slower cooling rate. Such variations in the HAZ microstructure vs. steel and heat input can be attributed to the presence of complex oxides working as nucleation sites for AF inside the HAZ. Because complex oxides inside austenite grains in the S1 steel HAZs provide nucleation sites, at which AF can be easily formed during the transformation from austenite to bainite, the formation of AF is observed around complex oxides, whereas other regions are mostly composed of bainite. Despite the presence of a considerable amount of complex oxides in the S0 steel HAZs, AFs are rarely formed. To explain this, effects of size and volume fraction of complex oxides on nucleation of AF need to be closely examined.

According to Barbaro et al.,\textsuperscript{26} AFs are well nucleated when non-metallic inclusions are sized over 0.5 $\mu$m. As their size increases, activation energy required for ferrite nucleation decreases,\textsuperscript{27} which effectively works for the nucleation of AF. Since complex oxides formed in the present two steels are sized about the same around 2 $\mu$m, most complex oxides can work as nucleation sites for AF. The volume fraction (or the number) of complex oxides in the S1 steel is about the double of the S0 steels. This higher volume fraction beneficially affects the formation of AF in the HAZ. However, the volume fraction of AF is much higher in the S1 steel HAZs than that in the S0 steel HAZs. In order to explain the large increase in AF fraction in the S1 steel HAZs, additional microstructural factors such as AGS in addition to complex oxide volume fraction should be taken into consideration.
During the austenite–ferrite transformation, the AGS affects the formation of AF in the HAZ because both austenite grain boundaries and complex oxides can work as ferrite nucleation sites. Bhadeshia et al. reported that AF is well formed when the AGS exceeds a certain size level, e.g., about 100 mm. Since austenite grain boundaries require lower activation energy for the transformation than complex oxides, they quickly provide nucleation sites at a low degree of supercooling, thereby leading to the ready formation of bainite. On the other hand, complex oxides require relatively higher activation energy, and thus AF is readily nucleated at a higher degree of supercooling. When the AGS is large enough, despite the fast transformation at austenite grain boundaries, ferrites can be nucleated at complex oxides since the space inside austenite grains is sufficient. When the AGS is too small, however, complex oxides cannot properly work as ferrite nucleation sites because microstructures such as bainite transformed from austenite grain boundaries take up the interior space of austenites.

According to Table 7 on the variation in Charpy impact properties vs. microstructure, both the two steels show high USE and absorbed energy at −20°C and 0°C over 200 J before the weld simulation test. However, in the S0 steel HAZ simulated with the heat input of 30 kJ/cm, Charpy absorbed energy at −20°C and 0°C is low below 50 J (Table 9). This is because the microstructure after the weld simulation test is mostly composed of bainite having low toughness. Unlike the S0 steel HAZ, the S1 steel HAZ consists of 94 vol% of AF, together with some bainite, because complex oxides work as AF nucleation sites during the austenite–bainite transformation. Consequently, the S1 steel HAZ shows much higher absorbed energy (over 200 J) at −20°C and 0°C than the S0 steel HAZ after the weld simulation test with the heat input of 30 kJ/cm. When the heat input increases to 50 kJ/cm and 100 kJ/cm, the absorbed energy at −20°C and 0°C is abruptly reduced. This is because the AF volume fraction decreases, while the bainite volume fraction increases, as the cooling rate becomes slow by the increased heat input. In the HAZ microstructures simulated with the high heat inputs, the dense and closely intersected microstructures of AF and bainite are not readily formed, and a number of bainite of large packets are formed, thereby leading to the reduced impact properties.

As mentioned above, the HAZ toughness is greatly influenced by the HAZ microstructure, and improves with increasing AF fraction. Microstructural factors affecting the AF formation in the HAZ are the number and size of oxides. As the number and size of oxides increase, the AF fraction and HAZ toughness increase. If the number of oxides is same in the S0 and S1 steels, factors affecting the AF fraction are the oxide size. Because the oxide size is similar in both steels, the S1 steel HAZ having larger number of oxides (1339 cm−3) shows the higher HAZ toughness than the S0 steel HAZ (652 cm−3) because of the higher AF fraction. When the number of oxides is same, the AF fraction is influenced by the oxide size. In this case, the HAZ toughness of the S0 steel HAZ having slightly larger oxides would be somewhat higher than that of the S1 steel HAZ, although the toughness difference might be very small.

Figure 12 illustrates a schematic drawing describing the austenite–ferrite transformation behavior in the HAZ. Complex oxides present inside austenite grains provide nucleation sites for AF during the transformation from austenite, while bainite is formed at austenite grain boundaries. Here, major microstructural factors affecting the transformation include the size and volume fraction of complex oxides, and AGS. When complex oxides are coarse and widely dispersed, AF nucleation sites increase. When the AGS is too small, the volume fraction of AF decreases since the space for the AF nucleation inside austenites is not sufficient. This can lead to deteriorating the HAZ toughness. Thus, the AGS should be over a certain size so that complex oxides inside austenites can work as AF nucleation sites and promote the formation of a number of AFs, thereby effectively

![Fig. 11. SEM micrographs of the cross-sectional area beneath the cleavage fracture surface of the Charpy impact specimens fractured at −20°C for the (a) S0-30, (b) S1-30, (c) S1-50, and (d) S1-100 steel HAZs, showing the crack propagation path. Fractured surfaces were coated by Ni.](image-url)
enhancing the HAZ toughness.

By using the oxide metallurgy in which AFs are nucleated inhomogeneously inside austenite grains by forming fine complex oxides in linepipe steels as in the present study, linepipe steels with excellent Charpy impact properties can be fabricated. This oxide metallurgy technology can be applied to grain refinement by inhomogeneous nucleation when a high level of deformation cannot be imposed due to the limited rolling reduction ratio as in the case of thick high-strength steel plates. It can also be applied to the development of high-strength steel plates with excellent welding properties by controlling the HAZ microstructures.

5. Conclusions

In the present study, complex oxides were formed inside API X70 linepipe steels, and effects of these oxides on HAZ microstructures and Charpy impact properties were investigated.

(1) Oxides present in API X70 linepipe steels were sized about 2 μm, and had characteristics of complex oxides composed of various elements. The S1 steel containing more Al, Ti, and Mg showed higher volume fraction of oxides than the S0 steel.

(2) The HAZ microstructures of the S0 steel simulated with the heat input of 30 kJ/cm were composed of bainite, but the S1 steel HAZ contained 94 vol% of AF, together with a small amount of bainite. Thus, the absorbed energy of the S1 steel HAZ at −20°C and 0°C was five to eight times higher than that of the S0 steel HAZ.

(3) When the weld simulation test was conducted with increased heat input of 50 kJ/cm and 100 kJ/cm, the absorbed energies at −20°C and 0°C of the S1 steel HAZs decreased down because of the increased amount of bainite as the cooling rate became slower.

(4) In the S1 steel HAZs in which the AGS was relatively large (91 μm) and a number of complex oxides were contained, complex oxides worked as AF nucleation sites, thereby promoting the active formation of AF in the HAZ and improving Charpy impact properties.

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Fig. 12. Schematic drawing describing the austenite–ferrite transformation behavior in the oxide-containing steel HAZ.