Microstructure features and formation mechanism in a newly developed electroslog welding

Tomonori Kakizaki¹ · Shodai Koga² · Hajime Yamamoto² · Yoshiki Mikami² · Kazuhiro Ito² · Kei Yamazaki¹ · Shuji Sasakura¹ · Hirohisa Watanabe¹

Received: 24 August 2021 / Accepted: 15 November 2021 / Published online: 27 November 2021
© The Author(s) 2021

Abstract

Electroslog welding (ESW) is known to show higher heat input than electrogas welding (EGW), resulting in poor low-temperature toughness. However, a newly developed ESW (dev. ESW) method using low-resistivity slag bath exhibited excellent low-temperature toughness as a result of lower effective heat input than conventional EGW, as demonstrated by the faster cooling rates measured in weld metals and estimated using finite element method analyses. This led to much shallower molten pool in the dev. ESW, resulting in much finer columnar grains and thinner centerline axial grains. High cooling speed in the dev. ESW method appeared to contribute to increased acicular ferrite proportion. The uniform microstructure with large acicular ferrite proportion and small number of inclusions in the weld metal permitted the dev. ESW weld metal to possess little variation in Charpy impact energy across the center of weld metal.

Keywords Electroslog welding · Weld metal toughness · Acicular ferrite · Cooling speed · Effective heat input · Finite element method

1 Introduction

Electroarc welding (EGW) is a well-known method for upward butt-welding of steel plates in the vertical position with high heat input, leading to high-efficiency welding, e.g., for shipbuilding [1]. Employment of high heat input in the welding process results in increased deposition rate. Referring to the relationship among deposition rate, welding current, and welding speed in vertical position welding, a higher deposition rate can be obtained in EGW and electroslog welding (ESW) compared to other welding processes such as shield metal arc welding and gas metal arc welding [2]. EGW is a butt-welding of steel plates with a groove

surrounded by the steel plates, a water-cooled sliding copper plate, and a backing material [2]. Although it enables to work with high efficiency, there are several disadvantages: much spatter, poor low-temperature toughness in the weld metal and heat-affected zone (HAZ) due to high heat input. On the other hand, ESW is well known as an alternative vertical-position upward butt-welding method but with little spatter [3]. A butt-joint groove for the ESW is surrounded by the same components being used for EGW [4]. The major difference between these two methods is the heat source: EGW has an electric arc for its heat source, but ESW has a molten slag bath heated to elevated temperature by Joule heat. ESW may have a limitation in a long-distance welding relating to continuous slag leak-out through the gap between the sliding copper plate and base plate, in addition to the poor low-temperature toughness of the weld metal and HAZ due to high heat input as in EGW. As will be seen, however, our newly developed ESW (dev. ESW) was found to be able to supply slag automatically and provide superior and uniform toughness in the weld metals.

The weld metal toughness of low-carbon steels is mainly controlled by its microstructure, i.e., microstructural constituents formed after the transformation. The prior austenite grain boundary is usually a ferrite nucleation site for grain

Recommended for publication by Commission IX—Behaviour of Metals Subjected to Welding

✉ Kazuhiro Ito
ito@jwri.osaka-u.ac.jp

¹ Welding Process Dept. Technical Center, Welding Business, Kobe Steel, Ltd, 100-1 Miyamae, Fujisawa, Kanagawa 251-8551, Japan

² Joining and Welding Research Institute, Osaka University, 11-1 Mihogaoka, Ibaraki, Osaka 567-0047, Japan

boundary ferrite, ferrite side plate, bainite, etc., in weld metals. Formation of the ferrite side plate is followed by formation of the grain boundary ferrite in the large prior austenite grain, low carbon contents, and increase of degree of undercooling [5]. The large heat input increases the prior austenite grain size, leading to formation of coarse grain boundary ferrite and ferrite side plate. The upper bainite usually forms in the EGW and ESW welds with high heat input. The martensite-austenite constituent forms between upper bainite laths and causes brittle fracture [6]. These characteristics result in significant reduction of low-temperature toughness. An approach to suppress the grain boundary ferrite and upper bainite is the addition of boron (B). The B atoms tend to segregate at the prior austenite grain boundaries, and thus to suppress the formation of grain boundary ferrite and upper bainite [7]. However, the remained B can react with N in molten steel, and thus there is an optimum range of B content to make a high acicular ferrite content in the weld metals [8].

The acicular ferrite grows in radial directions and the growths impinge on each other, resulting in formation of fine microstructure [9]. Sympathetic nucleation also takes place, which leads eventually to a fine, interlocking microstructure [9, 10]. The acicular ferrite formation mechanism makes crystal orientation in the adjacent grains significantly different, leading to formation of high-angle grain boundary [11]. The feature of acicular ferrite is believed to enhance low-temperature toughness.

The acicular ferrite formation is dependent on weld metal composition and welding condition that relates to the prior austenite grain size and cooling speed. The inclusion composition also varied with flux composition [12]. Some of the effective inclusions for acicular ferrite formation are reported to be TiO [13], oxides with spinel type [14], Ti_2O_3 [15], and MnS [16]. The former two inclusions are explained to have a good lattice matching with the acicular ferrite based on Bramfitt equation [17]. Accordingly, the acicular ferrite had Baker-Nutting orientation relationship with TiO and spinel-type oxides [13]. Although the latter two inclusions do not have good lattice matching with the acicular ferrite, a Mn-depleted layer around the inclusions is explained to enhance acicular ferrite formation [18]. The volume fraction of acicular ferrite decreased with decreasing number density of inclusions with Mn depletion layers, resulting in reduction of low-temperature toughness [19].

Regarding the welding conditions, the cooling speed of weld metals affects the formation of acicular ferrite. The weld metal microstructure consisted mainly of acicular ferrite at 30°C/s and volume fraction of the coarse grain boundary ferrite increased with decreasing cooling speeds below 15°C/s in low carbon steels [20]. The average prior austenite grain size increased with increasing heat input [21]. Such increase in prior austenite grain size was reported to tend

to increase Charpy impact energy [22]. This was explained to be associated with preferred formation of acicular ferrite from the inclusions instead of grain boundary ferrite.

The purpose of this study was to clarify the reason why the weld metal microstructure and low-temperature toughness are different between the newly dev. ESW and conventional EGW (conv. EGW) even under the same welding condition and heat input.

2 Low-temperature toughness of the butt-welded weld metals using developed electroslog welding and usual electrogas arc welding

Figure 1 shows cross-sectional macrostructures of welded joints made by dev. ESW and conv. EGW processes. Test steel plates and welding conditions are shown in Table 1, and the chemistry and mechanical properties of the weld metals in Tables 2 and 3, respectively. Ti, Al, and O contents were obviously different in Table 2. The oxygen was dissolved during the process, and its content could not be controlled. The Ti and Al contents were also controlled to result in the highest Charpy impact energy. The effects of difference in those contents and Al/O ratio can be seen in Chapter 4. The dilution ratios in both methods were almost the same, as shown in Fig. 1. To compare microstructure feature and related properties between the methods, test steel plates and welding conditions of groove angle, root gap, current, voltage, and welding speed were fixed to be almost the same for both methods. Composition of welding wires for both methods was optimized to reach a maximum in low-temperature Charpy impact energy.

Optical images indicate that columnar grains grew from the interface between a weld metal and a base metal toward

Fig. 1 Cross-sectional optical images of weld metal microstructure in the vertical-position upward butt-welded joints of steel plates using the dev. ESW and conv. EGW methods

Table 1 Test steel plates and welding conditions

	Steel type	Thick- ness (mm)	Groove angle (°)	Root gap (mm)	Current (A)	Voltage (V)	Welding speed (mm/ min)	Heat input (kJ/ mm)
Dev. ESW	SM490A	60	20	10	380	43	22	44.6
Conv. EGW	SM490A	60	20	10	390	43	23	43.8

Table 2 Contents of main solid solution elements of the obtained weld metals

	C	Si	Mn	P	S	Ti	Al	B	O
Dev. ESW	0.07	0.19	1.23	0.008	0.004	0.005	0.015	0.0024	0.022
Conv. EGW	0.08	0.19	1.38	0.008	0.007	0.026	0.008	0.0034	0.032

Table 3 Mechanical properties of the obtained weld metals

	YP (MPa)	TS (MPa)	Elonga- tion (%)	vE _{-20°C} (J)
Dev. ESW	460	594	23	170
Conv. EGW	458	618	23	95

the final solidification section at the center of the weld metal. The columnar grains in weld metals of the dev. ESW were much finer and the centerline axial grains in the final solidification section were much thinner than those of the conv. EGW specimens. The average grain sizes of prior austenite in coarse grain (CG) HAZ near the fusion lines were 249 and 305 μm , respectively, for the dev. ESW and conv. EGW specimens. The amount of heat input needed to make molten layers at the weld of steel joints is defined as the “effective heat input.” The microstructure features suggest that effective heat input in the dev. ESW was smaller than that in the conv. EGW method. This is contrary to the general understanding that ESW is the higher heat input welding. On the other hand, the HAZ widths at the mid-thickness were a

little thinner in the dev. ESW specimen: CGHAZ and HAZ widths were measured to be 2.95 and 8.93 mm for the dev. ESW specimen, and 3.05 and 10.18 mm for the conv. EGW, respectively.

The three points A, B, and C shown in Figs. 1(a) and 1(b) indicate the V-notch positions at the mid-thickness of weld metals for the Charpy impact tests conducted at -20°C using full-size specimens. The average of three Charpy impact specimens at each position is displayed in Fig. 2. The impact energy at each point was higher in the dev. ESW than the conv. EGW specimen. The values at points B and C in the dev. ESW were almost the same, while the value decreased as position furthered from the interface in the conv. EGW specimen resulting in the minimum at point A, which is the final solidification section. The extent of decrease was much greater in the conv. EGW specimens. The larger Charpy impact energies and their smaller variation (positions B and C) in the weld metal in the dev. ESW specimens could be related to the finer microstructure feature over almost the whole area of the weld metal. This indicates that the present dev. ESW has an advantage in comparison to the conv. EGW, and it is necessary to clarify the cause of this.

Fig. 2 Charpy impact energies measured at -20°C for weld metals in the butt-welded joints of steel plates using the dev. ESW and conv. EGW methods at the three points A, B, and C as the V-notch position in Figs. 1(a) and 1(b). The average values are represented in the bar chart. Optical images of fracture surface of both **b** dev. ESW and **c** conv. EGW specimens at position A, together with values of the brittle fracture surface ratio

3 Heat input in the developed electrosag welding compared with the usual electrogas arc welding

The effective heat input of the dev. ESW is expected to be smaller than that of the conv. EGW method, based on the microstructure of the weld metals. The lower effective heat input of the dev. ESW method was first clarified by the visualization of the molten pool depth. To make the weld pool exposed, the thickness was reduced to 20 mm, and the groove angle of the butt-welded joints and root gap were widened to 40° and 18 mm, respectively. Figure 3 shows optical images of weld pools taken through quartz glass in the dev. ESW and conv. EGW, as illustrated in Fig. 3(c). Welding conditions were also arranged for the visualization, and are shown in Table 4.

Significant difference between the dev. ESW and conv. EGW specimens was the molten pool volume. Its depth in the dev. ESW specimen was found to be much shallower than that in the conv. EGW specimen under the same heat input. The total molten-pool and slag-bath areas in the dev. ESW specimen (Fig. 3(a)) were estimated to be about 30% smaller than the molten pool area in the conv. EGW specimen (Fig. 3(b)). The molten-pool area itself in the dev. ESW specimen was measured to be only 17.5% of that in the conv. EGW specimen, demonstrating the molten pool volume in the dev. ESW was significantly smaller than that in the conv. EGW specimen. This fact implies the major factor controlling the columnar grain size would be a molten pool volume. Separate study on the welds made with fluxes with different resistivity showed that with increasing slag resistivity, the ratio of molten-pool area increased along with a concurrent change in the molten pool shape from a shallow-plate shape to a V shape. Nevertheless, columnar grains in its weld metal were similarly much finer.

To quantitatively evaluate difference in the effective heat input between the dev. ESW and conv. EGW methods under the same heat inputs, a finite element method (FEM) simulation of heat conduction was performed using the Abaqus 2018 software (Dassault Systemes Corp., France). Schematic images of the FEM model are presented in

Table 4 Welding conditions for visualization of weld pools

	Current (A)	Voltage (V)	Welding speed (mm/min)	Heat input (kJ/mm)
Dev. ESW	300	32	46	12.5
Conv. EGW	310	32	46	12.9

Fig. 4. Steel plate thickness, groove angle, and root gap in the butt-welded joint of steel plates (Fig. 4(a)) were the same as the values shown in Table 1. The range to be modeled for the FEM analysis was half of the butt-welded joints of steel plates with groove structure. A moving heat source with distance d (mm) was assumed to move upward at moving speed v (mm/min) in the groove (Fig. 4(b)). Element dividing over the range was set to be fine in and near the groove, consistent with the weld, becoming increasingly coarse further away from the groove (Fig. 4(c)). The physical properties of the groove material were assumed to be the same as those of the steel base metal. The FEM simulation was conducted for a material with temperature-independent density of 7.8×10^{-6} kg/mm³ and schematic temperature-dependent thermal conductivity and specific heat as shown in Fig. 5. The moving heat source was defined to have heat input of 44.6 kJ/mm and heat efficiency η . Heat transfer and radiation were assumed to occur at the boundary between the FEM model and atmosphere at 20°C under emissivity of 0.3, heat transfer coefficient of 2×10^{-6} W/(mm² K), and Stefan-Boltzmann constant of 5.669×10^{-8} W/(m² K⁴).

Figures. 6(a)–6(d) show typical temperature distributions over the partial range to be modeled using a moving heat source with η and d combinations. The combinations were selected for a temperature zone above the melting point to fall almost into the groove (Figs. 6(a)–6(c)), while the zone exceeded the groove for the combination of $\eta=0.9$ and $d=12$ mm (Fig. 6(d)). The boundary between base metal and groove is indicated as a broken line. The d increased with increasing η in keeping with the zone within the boundary, suggesting that the zone had a tendency to extend longitudinally parallel to the welding direction with increasing d for the same heat input. The figure shows one side opposite

Fig. 3 Optical images of weld pools in **a** the dev. ESW and **b** conv. EGW seen from the back bead through quartz glass, and **c** a schematic image of a butt-welded joint and observation direction

Fig. 4 Schematic images of **a** butt-welded joints of steel plates together with groove structure, **b** range to be modeled for FEM analysis with a moving heat source in the groove consistent with a zone surrounded by red lines in **a**, with a position for temperature measurement, and **c** element dividing in the range modeled for FEM analysis

Fig. 5 Schematic temperature-dependent **a** thermal conductivity and **b** specific heat of the materials modeled for FEM analysis

to the root. All combinations resulted in full penetration. Figures 6(e) and 6(f) show cooling curves, which varied with η at fixed $d=3$ mm and with d at fixed $\eta=0.75$, respectively. These curves were estimated at the specific position 120 mm away from the bottom at the center of groove after 240 mm run of the moving heat source (Fig. 4(b)). Temperature monotonically decreased with increasing time, and cooling curves varied significantly with η rather than d .

To compare the effective heat input between the dev. ESW and conv. EGW methods under the same heat inputs and dilution ratios (the dilution ratios in both methods did not significantly depend on heat input), a W-Re thermocouple was inserted into the center of a molten pool through a hole in the backing material at the root of butt-welded joints of steel plates as shown in Fig. 7(b). The welding conditions were the same as those in the FEM analyses described above. Cooling temperature was recorded down to 500°C. As shown in Fig. 7(a), the cooling rate of dev. ESW was about

two times faster than that of conv. EGW method even under the same heat inputs. To identify the effective heat input, the measured cooling curves were matched with those estimated by the FEM analyses. Figure 7(c) shows comparison of the cooling curve measured in the dev. ESW specimens with those estimated using several combinations of η and d with similar dilution ratios. The coefficient of determination, denoted R^2 , is the proportion of the variance in the dependent variable that is predictable from the independent variables. The R^2 exhibited the highest value of 0.99 in the combinations of $\eta=0.75$ and $d=15$, and $\eta=0.70$ and $d=12$. Similarly, the highest R^2 was obtained with the combination of $\eta=1.00$ and $d=33$ for the conv. EGW out of other combinations of η and d (Fig. 7(d)). The R^2 decreased with decreasing η . Note that the measured temperature–time curve can be obviously divided into two regions around 750°C. The curves estimated by the FEM analyses did not correspond completely, but the curve estimated in the case

Figure. 6 a–d Typical temperature distributions over the partial range to be modeled using a moving heat source with η and d combinations and cooling curves (e) varied with η at fixed $d=3$ mm and f with d at fixed $\eta=0.75$ estimated by FEM analysis

Figure. 7 a Cooling curves measured in the dev. ESW and conv. EGW specimens. b Schematic image of temperature measurement. Comparison of the cooling curve measured in c the dev. ESW and d conv. EGW specimens with those estimated by FEM analysis using several combinations of η and d with similar dilution ratios

of $\eta = 1.00$ and $d = 33$ exhibited a small deviation especially below 750°C at which phase transformation occurs. This is probably caused by difference of heat sources in the conv. EGW and FEM analyses: the former is a point heat source, and the latter is a surface heat source. The cooling curves did not vary with d , but varied markedly with η . These results suggest that difference of heat efficiency η between the dev. ESW and conv. EGW methods could be caused by difference of the effective heat input. Consequently, the heat efficiency estimated in the FEM analyses based on heat conduction in the butt-welded joints in the dev. ESW was found to be 0.70 and 0.75, which is lower than that of 1.00 in the conv. EGW method. This indicates that the effective heat input in the dev. ESW was reduced by about 25–30% in comparison with that in the conv. EGW method. This is good agreement with 30% reduction of liquid (molten-pool and slag-bath) area in the dev. ESW specimen (Fig. 3(a)) in comparison with the molten pool area in the conv. EGW specimen (Fig. 3(b)). In the dev. ESW specimens using lower and higher resistivity flux, the liquid area was estimated to be +4% and +2%, respectively, although heat input was set to 28% increase and 15% decrease. This indicates that the molten-pool and slag-bath volume increased with increasing slag resistivity under the same heat input. Thus, the 25–30% reduction in the effective heat input in the dev. ESW specimen is attributing to low-resistivity slag. In addition, molten slag should be kept during process in the dev. ESW method, and thus, latent heat of the slag needs to be subtracted from the heat input when accessing the difference of the effective heat input between the dev. ESW and conv. EGW specimens.

4 Effects of low effective heat input on weld metal microstructure

This section describes examination of how the low effective heat input in the dev. ESW affected the weld metal microstructure based on comparison of microstructure in specimens to which various thermal cycles were provided. The specimens were obtained from the position A in the as-welded weld metals (Fig. 1). For comparison, two welding wires with different B contents for the dev. ESW specimens and one wire for the conv. EGW were employed. Number density of inclusions and average grain size of prior austenite obtained at each position of the weld metals are shown in Fig. 8. To classify the inclusions, elemental mapping was conducted using an electron probe micro analyzer (EPMA). To visualize the prior austenite grain size, a commercially available etching set (A.G.S.; Yamamoto Scientific Tool Laboratory) was employed. The solution consists of mainly a picric acid saturated aqueous solution and surfactant. The total number densities of inclusions in the dev. ESW specimens were smaller than that in the conv. EGW at each

Fig. 8 Number density of inclusions and average grain size of prior austenite at each position (see the inset figure) of weld metals obtained using the dev. ESW and conv. EGW methods. The weld metals obtained using the dev. ESW contained low and high B content

position, probably due to lower content of dissolved oxygen. Position A was the largest among the three positions. The inclusion chemistries were little influenced by weld metal position, filler wires, or welding method. Average grain size of prior austenite was smaller in the dev. ESW specimens than that in the conv. EGW at positions B and C, suggesting the small effective heat input in the dev. ESW. In contrast, it became larger in the dev. ESW specimens than that in the conv. EGW specimen at position A, probably due to large agglomeration of inclusions at the final solidification section, especially in the conv. EGW. Contents of main solid solution elements in the three weld metals at position A are shown in Table 5.

Volume fractions of acicular ferrite estimated for each position of the weld metals are shown in Fig. 9. To estimate the values, coarse grains in the optical microscope (OM) images (probably polygonal ferrite or bainite) were painted manually in black in the images enlarged on a PC screen, and the images were binarized. The white area in the binarized images was considered to consist of acicular ferrite. Five OM images at 500 magnification were taken at each position. As the variation fell within several %, they are plotted in Fig. 9. The acicular ferrite proportion at positions B and C was large (more than 95%) for both methods, while it varied with weld metal and welding method at position A. The acicular ferrite proportion at position A exhibited almost no decrease, a small decrease, and significant decrease from positions B and C, respectively, in the weld metal with low and high B contents in the dev. ESW specimens that in the conv. EGW specimen. High-magnification OM images at position A for the three specimens are shown in Fig. 10. As little difference in microstructure was noted at this position, microstructural features and Charpy impact energy at position A are summarized in Table 6. Relatively large Charpy

Table 5 Contents of main solid solution elements in the three weld metals at position A

		Elemental contents (mass%)				
		Al	Ti	B	O	N
Dev. ESW	BL	0.012	0.008	0.0017	0.017	0.0030–
	BH	0.012	0.009	0.0026	0.015	0.0034
Conv. EGW		0.008	0.026	0.0034	0.032	0.0043

Fig. 9 Volume fraction of acicular ferrite estimated at each position of the weld metals obtained using the dev. ESW (low and high B content) and conv. EGW methods**Table 6** Features of microstructure and Charpy impact energy at position A

		Ave. grain size of prior austenite (μm)	Acicular ferrite proportion (%)	vE-20°C (J)
Dev. ESW	BL	188.9	96.5	181
	BH	224.1	92.9	160
Conv. EGW		101.1	87.4	95

impact energy was obtained in the dev. ESW method, and was proportional to the acicular ferrite proportion. On the other hand, average grain size of prior austenite at position A in the conv. EGW became almost half of those in the dev. ESW specimens, leading to reduction of acicular ferrite proportion and resulting in significant reduction in the Charpy impact energy.

Thermal cycles provided to the three types of specimens: low and high B dev. ESW and conv. EGW. All specimens were taken from the as-welded weld metals at position A. The thermal cycles employed are shown in Fig. 11. Average grain size of prior austenite was systematically changed by varying the holding times at 1400°C: 2.5, 10, and 40 min. The grain size of prior austenite increased in proportion to the square root of the holding time in all specimens, while the increase in high B specimen with the dev. ESW was about 1.5 times larger than that in the other two specimens. To investigate effect of cooling speed on microstructure, the five cooling speeds below 800°C were applied. Note that the cooling rates of 48 and 21°C/min represent the cooling speeds measured for the dev. ESW and conv. EGW, respectively.

Figure 12 shows OM images of thermal-cycled specimens of low B weld taken from as-welded dev. ESW. These results exhibited the effects of average grain size of prior austenite (together with holding times) and cooling speed below 800°C on microstructure. Both the acicular ferrite proportion and Vickers hardness are given on each microstructure. Vickers hardness tests were carried out on the

Fig. 10 High-magnification OM images obtained from position A of the weld metals obtained using the dev. ESW (a low and b high B content) and c conv. EGW methods

Fig. 11 Thermal cycles applied to specimens taken from the as-welded weld metals at position A

cross-section of specimens at room temperature with an applied load of 19.6 N and loading time of 15 s, to validate phase identification.

Volume fraction of the polygonal ferrite with white contrast was higher for sets of (82, 7), (82, 21), and (106, 7), respectively, for (average grain sizes of prior austenite (μm), cooling speed ($^{\circ}\text{C}/\text{min}$)). The microstructure exhibited acicular ferrite proportions below about 25% and Vickers hardness values below 200 HV. Acicular ferrite proportion increased to about 80% and Vickers hardness values were a little above 200 HV for sets of (82, 62), (106, 35), (106, 48), and (106, 62). In sets of (82, 35), (82, 48), (106, 21), acicular and polygonal ferrite coexisted. In contrast, relatively large Vickers hardness values more than 200 HV were obtained for sets of (141, 7) and (141, 35), although the acicular ferrite proportion was not so high. Especially, hardness value of the set of (141, 21) was significantly high even in the acicular ferrite proportion of about 80%. This is believed to

be caused by some martensite laths, which were observed using a scanning electron microscope (SEM). The measured cooling speed in the dev. ESW specimens was $48^{\circ}\text{C}/\text{min}$, and the as-welded weld metal did not include the martensite lath. The microstructure trend indicates that acicular ferrite proportion increased with increasing average grain size of prior austenite and cooling speed, and relatively high cooling speed measured in the dev. ESW specimens was beneficial to increase acicular ferrite proportion.

Figure 13 shows OM images of high B specimens made by the dev. ESW method. Some bainite laths and massive bainite with gray contrast were observed instead of polygonal ferrite at small average grain sizes of prior austenite such as 121 and $148\ \mu\text{m}$ and slow cooling speeds such as 7 and $21^{\circ}\text{C}/\text{min}$, leading to decreased volume fraction of acicular ferrite. Nevertheless, their hardness values exhibited more than 200 HV, probably due to the formation of bainite. Taking an overview of all the relations in Fig. 13, the hardness did not vary with parameters, while volume fraction of acicular ferrite increased with increasing average grain size of prior austenite and cooling speed. Note that the acicular ferrite content did not exceed 80%, suggesting the difference in mechanical properties between acicular ferrite and bainite was not great. In addition, the average grain sizes of prior austenite in the dev. ESW specimens with high B content shifted to large-size side, i.e., those arising with the holding times of 2.5 and 10 min at 1400°C were similar to those with 10 and 40 min in the other two specimens. The cause of this, which occurred when applying the thermal cycle to the specimen, is not clear. In fact, the as-welded weld metal with high B content in the dev. ESW specimens exhibited acicular ferrite proportion of 92.9% under the cooling speed of $48^{\circ}\text{C}/\text{min}$ (Table 6), which probably contributed to the increase of acicular ferrite proportion.

Figure 14 shows cross-sectional optical images of specimens taken from the as-welded weld metals at position A

Fig. 12 OM images of thermal-cycled specimens of weld metals of low B dev. ESW

		Cooling speeds below 800°C				
		$7^{\circ}\text{C}/\text{min}$	$21^{\circ}\text{C}/\text{min}$	$35^{\circ}\text{C}/\text{min}$	$48^{\circ}\text{C}/\text{min}$	$62^{\circ}\text{C}/\text{min}$
Average grain sizes of prior austenite	$141\ \mu\text{m}$ (40min)	59.7% 225HV	82.7% 262HV	62.7% 209HV		Dev. ESW-BL 100 μm
	$106\ \mu\text{m}$ (10min)	16.5% 178HV	49.9% 206HV	79.9% 201HV	94.1% 212HV	84.5% 213HV
	$82\ \mu\text{m}$ (2.5min)	14.8% 148HV	26.0% 165HV	40.5% 177HV	56.0% 185HV	77.3% 196HV

Fig. 13 OM images of thermal-cycled specimens of weld metals of high B dev. ESW

Cooling speeds below 800°C					
	7 °C / min	21 °C / min	35 °C / min	48 °C / min	62 °C / min
Average grain sizes of prior austenite 210 µm (40min)	55.7% 226HV	57.3% 235HV	75.5% 234HV		Dev. ESW-BH 100 µm
148 µm (10min)	55.5% 214HV	64.2% 220HV	69.7% 220HV	75.7% 216HV	71.5% 208HV
121 µm (2.5min)	54.2% 213HV	67.6% 215HV	70.7% 219HV	77.9% 214HV	75.6% 219HV

Fig. 14 OM images of thermal-cycled specimens of weld metals of the conv. EGW

Cooling speeds below 800°C					
	7 °C / min	21 °C / min	35 °C / min	48 °C / min	62 °C / min
Average grain sizes of prior austenite 144 µm (40min)	76.4% 227HV	88.4% 221HV	95.4% 223HV		Conv. EGW 100 µm
115 µm (10min)	59.0% 216HV	74.5% 220HV	86.3% 222HV	96.3% 211HV	95.6% 213HV
93 µm (2.5min)	40.1% 215HV	59.6% 208HV	66.8% 207HV	76.0% 206HV	83.3% 206HV

in the conv. EGW method, after providing several thermal cycles. Some bainite laths and massive bainite with gray contrast were observed and hardness values of more than 200 HV, suggesting no polygonal ferrite. Taking an overview of all the relations in Fig. 14, the hardness did not vary with parameters, while volume fraction of acicular ferrite increased with increasing average grain size of prior austenite and cooling speed. This is similar to the results for the dev. ESW specimens with high B content; however, the acicular ferrite proportion increased up to almost 100% at high cooling speeds of 48 and 62°C/min at the average grain size of prior austenite of 115 µm. Unfortunately, the cooling speed measured in the conv. EGW was 21°C/min, suggesting that the slow cooling speed led to a lower content of acicular ferrite in the conv. EGW specimen.

To compare the relationship between microstructure, grain size of prior austenite, and cooling speed, contour maps were constructed as shown in Fig. 15 for specimens taken from the weld metals of all three specimens. The

acicular ferrite proportion increased with increasing average grain size of prior austenite and cooling speed in general, but it was little affected by the grain size in the high-B dev. ESW. The acicular ferrite proportions for the as-welded dev. ESW and conv. EGW specimens are also marked with their austenite grain size and cooling speed. The cooling speeds from 800 to 500°C were estimated to be 48°C/min and 21°C/min from Fig. 7(a) for the dev. ESW and conv. EGW specimens, respectively. The acicular ferrite proportions and grain sizes of prior austenite for the as-welded weld metals at position A are indicated in Table 6. Those in the dev. ESW specimens were located in the high acicular ferrite proportion area, while that in the conv. EGW was in the relatively low proportion area. This was caused by the lower effective heat input in the dev. ESW than that in the conv. EGW method. In addition, the number density of inclusions in the dev. ESW specimens was lower than that in the conv. EGW, suggesting that small number of fracture origins, as well as large acicular ferrite proportions, led to significantly

Fig. 15 Acicular ferrite proportions and their distribution as a contour map for specimens taken from the weld metals at position A with **a** low and **b** high B contents in the dev. ESW and **c** in the conv. EGW after providing several thermal cycles, together with those as-welded weld metals

larger Charpy impact energies in the dev. ESW than that in the conv. EGW specimen (Table 6). From a macroscopic viewpoint, fine columnar grains in the dev. ESW specimens caused by the lower effective heat input could contribute to increased Charpy impact energy as well.

5 Conclusions

In this study, microstructural features of the weld metals made by the newly dev. ESW method were investigated and compared with those of conv. EGW. The results can be summarized as follows:

(1) The weld metal produced using the dev. ESW method consisted of much finer columnar grains and thinner center-line axial grains than those by the conv. EGW. Consequently, the Charpy impact energy of the weld metal in the dev. ESW welds was higher than that in the conv. EGW. The degree of reduction in impact energy with the V-notch position approaching to the center of the weld metal was smaller in the dev. ESW than in the conv. EGW specimens.

(2) The molten pool observed in the dev. ESW specimens was much shallower than that in the conv. EGW. The cooling rate of the dev. ESW was about two times faster below 800°C than that of the conv. EGW.

(3) FEM analyses indicated that heat efficiency of the moving heat source was 0.70 and 0.75 in the dev. ESW,

while 1.00 in the conv. EGW, suggesting that the effective heat input in the dev. ESW is lowered by about 20–25% in comparison with that in the conv. EGW method.

(4) Large cooling speed in dev. ESW tended to contribute to increase of acicular ferrite proportion. In addition, small number of inclusions in the dev. ESW specimens also contributed uniform microstructure not varying with position in the weld metal, and small number of fracture origins. Thus, the thermal cycle and small number of inclusions in the dev. ESW specimens led to large acicular ferrite proportion and high Charpy impact energy across the entire area of weld metals.

Acknowledgements This investigation was a result of collaboration between research programs of the Welding Process Department Technical Center, Welding Business, Kobe Steel, Ltd. and Joining and Welding Research Institute (JWRI), Osaka University.

Author contribution Conceptualization: Tomonori Kakizaki, Kazuhiro Ito, Kei Yamazaki, Hirohisa Watanabe; investigation: Tomonori Kakizaki, Shodai Koga, Hajime Yamamoto, Shuji Sasakura; data curation: Tomonori Kakizaki, Shodai Koga, Shuji Sasakura; validation: Tomonori Kakizaki, Shodai Koga, Hajime Yamamoto, Kazuhiro Ito, Kei Yamazaki, Shuji Sasakura; FEM analysis, software: Yoshiki Mikami; supervision: Kazuhiro Ito, Kei Yamazaki, Hirohisa Watanabe; writing: Kazuhiro Ito; project administration: Hirohisa Watanabe.

Declarations

Competing interests The authors declare no competing interests.

Open Access This article is licensed under a Creative Commons Attribution 4.0 International License, which permits use, sharing, adaptation, distribution and reproduction in any medium or format, as long as you give appropriate credit to the original author(s) and the source, provide a link to the Creative Commons licence, and indicate if changes were made. The images or other third party material in this article are included in the article's Creative Commons licence, unless indicated otherwise in a credit line to the material. If material is not included in the article's Creative Commons licence and your intended use is not permitted by statutory regulation or exceeds the permitted use, you will need to obtain permission directly from the copyright holder. To view a copy of this licence, visit <http://creativecommons.org/licenses/by/4.0/>.

References

- Funatsu Y (2012) Structural steels for ships (Materials). J Japan Welding Society 81:257–263. <https://doi.org/10.2207/jjws.81.257>
- Sato M (1997) Electro gas arc welding process. J Japan Welding Society 66:274–278. <https://doi.org/10.2207/qjws1943.66.274>
- Yamaguchi M (1997) Welding consumables and procedures for electros slag welding. J Japan Welding Society 66:279–283. <https://doi.org/10.2207/qjws1943.66.279>
- Kiyohara M, Ohwa T (1978) Fusion welding equipment. Sanpo publications Inc, Tokyo
- Krahe PR, Kinsman KR, Aaronson HI (1972) Influence of austenite grain size upon the widmanstatten-start (W_s) temperature for the proeutectoid ferrite reaction. Acta Metall 20:1109–1121. [https://doi.org/10.1016/0001-6160\(72\)90175-7](https://doi.org/10.1016/0001-6160(72)90175-7)
- Koseki T (2004) Microstructure development and control in steel welds. Tetsu-to-Hagane 90:61–72. https://doi.org/10.2355/tetsu tohagane1995.90.2_61
- Thewlis G (1994) Transformation kinetics of ferrous weld metals. Mater Sci Technol 10:110–125. <https://doi.org/10.1179/mst.1994.10.2.110>
- Kitani Y, Ikeda R, Ono M, Ikeuchi K (2009) Improvement in weld metal toughness in high heat input electros slag welding of low carbon steel. Quarterly J Japan Welding Society 27:240–246. <https://doi.org/10.2207/qjws.27.240>
- Ricks RA, Howell PR, Barritte GS (1982) The nature of acicular ferrite in HSLA steel weld metals. J Mater Sci 17:732–740. <https://doi.org/10.1007/BF00540369>
- Terasaki H, Komizo Y (2006) In situ observation of morphological development for acicular ferrite in weld metal. Sci Technol Weld Joining 11:561–566. <https://doi.org/10.1179/174329306X149795>
- Gourgues A-F, Flower HM, Lindley TC (2000) Electron backscattering diffraction study of acicular ferrite, bainite, and martensite steel microstructures. Mater Sci Technol 16:26–40. <https://doi.org/10.1179/026708300773002636>
- Okazaki Y, Ishida H, Suenaga K, Hidaka T (2009) Influence of oxide inclusion compositions on microstructure and toughness of weld metal for high strength steel. Quarterly J Japan Welding Society 27:131–138. <https://doi.org/10.2207/qjws.27.131>
- Yamada T, Terasaki H, Komizo Y (2009) Relation between inclusion surface and acicular ferrite in low carbon low alloy steel weld. ISIJ Int 49:1059–1106. <https://doi.org/10.2355/isijinternational.49.1059>
- Grong Ø, Kluken AO, Nylund HK, Dons AL, Hjelen J (1995) Catalyst effect in heterogeneous nucleation of acicular ferrite. Metall and Mater Trans A 26:525–534. <https://doi.org/10.1007/BF02663903>
- Byun J-S, Shim J-H, Suh J-Y, Oh Y-J, Cho YW, Shim J-D, Lee DN (2001) Inoculated acicular ferrite microstructure and mechanical properties. Mater Sci Eng, A 319–321:326–331. [https://doi.org/10.1016/S0921-5093\(00\)02014-1](https://doi.org/10.1016/S0921-5093(00)02014-1)
- Tomita Y, Saito N, Tsuzuki T, Tokunaga Y, Okamoto K (1994) Improvement in HAZ toughness of steel by TiN–MnS addition. ISIJ Int 34:829–835. <https://doi.org/10.2355/isijinternational.34.829>
- Bramfitt BL (1970) The effect of carbide and nitride additions on the heterogeneous nucleation behavior of liquid iron. Metallurgical Transactions 1:1987–1995. <https://doi.org/10.1007/BF02642799>
- Byun J-S, Shim J-H, Cho YW, Lee DN (2003) Non-metallic inclusion and intragranular nucleation of ferrite in Ti-killed C-Mn steel. Acta Mater 51:1593–1606. [https://doi.org/10.1016/S1359-6454\(02\)00560-8](https://doi.org/10.1016/S1359-6454(02)00560-8)
- Song FY, Shi MH, Wang P, Zhu FX, Misra RDK (2017) Effect of Mn content on microstructure and mechanical properties of weld metal during high heat input welding processes. J Mater Eng Perform 26:2947–2953. <https://doi.org/10.1007/s11665-017-2597-7>
- Kitani Y (2010) Acicular ferrite formation in weld metals produced by laser welding and electro-slag welding of low carbon steel, Osaka University
- Suzuki S, Kamo T, Komizo Y (2009) Influence of martensitic islands on fracture behaviour of high heat input weld HAZ. Weld Int 23:397–402. <https://doi.org/10.1080/09507110802542817>
- Han P, Sato M (2003) Influence of austenite (γ) grain size on low temperature toughness of weld metal. R&D Kobe steel engineering reports 63:11–15

Publisher's note Springer Nature remains neutral with regard to jurisdictional claims in published maps and institutional affiliations.