Microstructure and mechanical properties of ultralow carbon high-strength steel weld metals with or without Cu-Nb addition

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Abstract

Two kinds of ultralow carbon steel weld metals with and without added Cu-Nb were prepared by using gas metal arc welding (GMAW) to investigate the correlation between microstructure and mechanical properties of weld metals. Results of microstructure characterization show that weld metal without Cu-Nb is mainly composed of acicular ferrite (AF), lath bainite (LB), and granular bainite (GB). In contrast, adding Cu-Nb to weld metal causes an evident transformation of martensite and causes grain coarsening. Both weld metals have high tensile strength (more than 950 Mpa) and more than 17% elongation; however, their values of toughness deviate greatly with a difference of approximately 40 J at -50°C. Analysis of morphologies of fracture surfaces and secondary cracks further revealed the correlation between microstructure and mechanical properties. The effects that adding Cu and Nb have on microstructure and mechanical properties of weld metal are discussed; the indication is that adding Cu-Nb increases both hardenability and grain size of weld metal and thus deteriorates the toughness.

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1. Introduction

Welding is the most essential joining technique that is used in industry, and it plays a significant role in determining the final performance of steel construction, especially when high-strength steel is used [1]. Researchers have conducted a lot of work on the microstructure and mechanical properties of high-strength steel welded joints in the heat affected zone (HAZ) [2-4]. However, the comprehensive properties of weld metal that is obtained by melting welding consumables also have a great influence on the overall high-strength steel welded structure. When conventional high-strength steels are welded, the mechanical properties of welded joints are obviously lower because of their high carbon equivalent (CE), and this leads to various defects during welding, such as welding cracks [5]. Because the amount of carbon accounts for a high proportion in CE for steels [6], reducing the carbon content is significant for improving the weldability of steel [7]. Therefore, because of their good weldability, high-strength low alloy (HSLA) steels that contain less than 0.10 wt. % carbon have been designed and widely used in many industrial fields, such as shipbuilding, pressure vessels, and pipelines [8-10].

The primary welding method for HSLA steel is arc welding techniques, including GMAW and gas tungsten arc welding (GTAW) [11-12]. In these techniques, design of the welding wire is crucial because it ensures a good match between the weld metal and base metal (BM) in terms of their mechanical properties. In previous studies [7,13-16], the carbon content of the welding wire used for high-strength steel is generally between 0.06-0.10 wt. %, and the tensile strength of the weld metal is in the range of 700-950Mpa. However, hardening and brittle phases, such as martensite and martensite-austenite (M-A), are easily generated in the weld metal when it contains more than 0.05 wt. % carbon, and this causes the impact toughness to deteriorate. In this case, post-welding heat treatment (PWHT) must be
implemented to improve the toughness of the weld metal although it usually increases costs [14-15].

For high-strength steel weld metal, using an ultralow carbon (<0.03 wt. %) welding wire with microalloying elements may be a feasible way to obtain applicable mechanical properties. In previous studies, it was suggested that adding Cu to a weld metal can improve the strength as a result of the solid solution hardening effect [17] and that adding a small amount of Nb produces undissolved precipitates, which thereby refine the austenite grain size [18]. In this context, it is worth further investigating if adding a certain amount of Cu-Nb into ultralow carbon welding wire can result in beneficial effects in a weld metal.

In this paper, two kinds of ultralow carbon welding wires (one with and one without added Cu-Nb) were designed. Acceptable mechanical properties for a weld metal without PWHT can be obtained using this kind of welding wire, and thus, the weldability of steels can also be improved. Meanwhile, with an ultralow carbon content level (<0.03 wt. %), the tensile strength of the weld metal exceeds 950 MPa in this work, and this value is significantly higher than that for weld metals that have been reported in previous studies [7,13-16,19-20]. The effects that adding Cu-Nb have on microstructure and mechanical properties of a weld metal are discussed.

2. Experimental methods

2.1. Materials and welding procedures

10Ni5CrMoV HSLA steel plates with a thickness of 20mm were selected as the BM [10]. For the BM, steel plates were first cut into two pieces using flame cutting, and then the edges were processed to form each piece into a trapezoidal groove via carbon arc air gouging. The angle between the groove bevel and the vertical direction was 22.5°, as shown in Fig. 1a. Two types of ultralow carbon steel welding wires (each with a diameter of 1.2mm) were fabricated, and their chemical compositions are given in Table 1.
It is noted that the ultralow carbon steel welding wire had a relatively low CE. The impurity elements in the welding wire mainly included H, O, N, S, and P, and these must be precisely controlled within the following specified ranges: \([H]\leq3\text{ppm}, [O]\leq30\text{ppm}, [N]\leq30\text{ppm}, [S]\leq10\text{ppm}, [P]\leq10\text{ppm},\) and \([H]+[O]+[N]+[S]+[P]\leq60\text{ppm}.

Table 1. Chemical composition (wt. %) of welding wire 1 (W1) and welding wire 2 (W2)

<table>
<thead>
<tr>
<th>Welding wire</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>Al</th>
<th>Ti</th>
<th>Cr</th>
<th>Ni</th>
<th>Cu</th>
<th>Nb</th>
<th>Mo</th>
<th>Fe</th>
<th>CE</th>
</tr>
</thead>
<tbody>
<tr>
<td>W1</td>
<td>0.025</td>
<td>0.563</td>
<td>1.16</td>
<td>&lt;0.06</td>
<td>&lt;0.06</td>
<td>0.608</td>
<td>4.13</td>
<td>&lt;0.01</td>
<td>&lt;0.01</td>
<td>0.612</td>
<td>Bal.</td>
<td>0.37</td>
</tr>
<tr>
<td>W2</td>
<td>0.020</td>
<td>0.575</td>
<td>1.19</td>
<td>&lt;0.06</td>
<td>&lt;0.06</td>
<td>0.576</td>
<td>4.24</td>
<td>0.15</td>
<td>0.082</td>
<td>0.638</td>
<td>Bal.</td>
<td>0.38</td>
</tr>
</tbody>
</table>

Note [6]: \[CE = C + f(C) \cdot \left[\frac{Si}{24} + \frac{Mn}{6} + \frac{Cu}{15} + \frac{Ni}{20} + \left(\frac{Cr + Mo + V + Nb}{5} + 5B\right)\right] \text{ (wt. %)},\]

\[f(C) = 0.75 + 0.25\tanh \cdot \left[20(C - 0.12)\right] \text{ (wt.%)}\]

The weld metal in a butt joint was prepared using GMAW at room temperature without PWHT, and 5% CO\(_2\) + 95% Ar was selected to be the shielding gas. As shown in Fig. 1a, a multi-pass welding process was used to feed the molten welding wire into the groove. The weld metals W1 and W2 (with a heat input of 10 kJ/cm) were labeled as WM1 and WM2, respectively. Specific welding parameters of GMAW are presented at Table 2.

Table 2. GMAW butt welding parameters

<table>
<thead>
<tr>
<th>Welding current (A)</th>
<th>Arc voltage (V)</th>
<th>Welding speed (cm / min)</th>
<th>Gas flow rate (L / min)</th>
<th>Interpass temperature (°C)</th>
<th>Heat input (kJ / cm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>200±10</td>
<td>25±2</td>
<td>30±2</td>
<td>20±1</td>
<td>100–150</td>
<td>10</td>
</tr>
</tbody>
</table>
Fig. 1. (a) Weld groove size and multi-pass welding schematic diagram; (b) positions of specimens in the weld metal for microstructure analysis and mechanical tests.

2.2. Microstructure characterization and mechanical tests

Specimens for optical microscopy (OM) and scanning electron microscopy (SEM) had dimensions of 10×10×10 mm (Fig. 1b). The specimens were mechanical ground, polished, and etched with 4% nitric acid alcohol for 10-15s at room temperature. The microstructure was then observed using a 9XB-PC optical microscope and Zeiss Auriga scanning electron microscope equipped with an energy dispersive spectrometer (EDS). After mechanical lapping, metallographic specimens of the weld metal (with dimensions of 10×10×2 mm) were electropolished using a continuous-current plant with a solution of 10% HClO₄, and then electron backscattered diffraction (EBSD) was conducted. Transmission electron microscopy (TEM) samples (with dimensions of 10×10×0.3 mm) were extracted from the weld seam and mechanically ground to foils that had a thickness of 50-60 μm. The disks of 3 mm in diameter were
punched from the foils and then electropolished in a twin-jet apparatus with a solution of 10% HClO₄ at -20°C to further thin them to a thickness less than 100nm. TEM was conducted with FEI TecnaiF30 operated at voltage of 300 KV.

As shown in Fig. 1b, rod-like tensile samples (dimensions: Φ5×60 mm) that had a gauge length of 30mm as well as standard Charpy V-notch impact specimens (dimensions: 10×10×55 mm) were taken 1 mm below the upper surface of the BM along the welding direction, according to ISO 5178:2019 and ISO 9016:2012. Tensile tests were conducted in accordance with ISO 6892-1:2016 at room temperature using a WDW-200D electronic universal testing machine. For each weld metal, three Charpy V-notch impact specimens were tested on a ZBC2452-B pendulum impact test machine, according to ISO 148-1:2016. SEM was used to observe the morphologies of the fracture surface and secondary cracks beneath the impact fracture surface.

3. Results

3.1. Microstructure of weld metals

Fig. 2 shows microstructure of WM1 and WM2 observed by OM, and each has a prior austenite grain boundary (PAGB). As seen, the microstructure of WM1 mainly consists of acicular ferrite (AF), lath bainite (LB), or lath martensite (LM). By contrast, WM2 is predominantly lath structure. Moreover, WM2 has a relatively larger grain size than WM1 overall, which is embodied in both lath structure size and prior austenite grain size.
Fig. 2. OM images of weld metals: (a) WM1 and (b) WM2.

SEM images of WM1 and WM2 are shown in Fig. 3. As seen in Fig. 3a, WM1 is composed of fine AF and multiple forms of bainite, including granular bainite (GB) and LB. Also, there is fine-grained AF (approximately 1\(\mu\)m wide and 3\(\mu\)m long) that is mainly nucleated on the oxide inclusion and grows radially as indicated by the white circle in Fig. 3b [21]. The inset of Fig. 3b shows the EDS results of the inclusion; the results show that the inclusion is a complex oxide of Ti, Al, Mn, and Si. Meanwhile, Fig. 3c and d show that the microstructure of WM2 predominantly consists of LM with coarsened laths and that prior austenite grains are segmented by these lath blocks [2]. In addition, there are many slender M-A constituents that have high aspect ratio, and these constituents are distributed between the lath boundaries, as seen in Fig. 3d.
Fig. 3. SEM images of weld metals: (a) WM1 and (c) WM2; (b) and (d) magnified views of local zones (indicated by the rectangles marked in solid white lines) in (a) and (c), respectively.

The average prior austenite grain size (PAGS) and the area fraction of the main phases in the weld metals were measured using Image J software, and the results are given in Table 3. Both LB and GB are classified into bainitic ferrite (BF) because they have the same transformation mechanism.

<table>
<thead>
<tr>
<th>Items</th>
<th>WM1</th>
<th>WM2</th>
</tr>
</thead>
<tbody>
<tr>
<td>PAGS / µm</td>
<td>50.1±2.4</td>
<td>91.4±4.0</td>
</tr>
<tr>
<td>AF / %</td>
<td>20.5</td>
<td>4.2</td>
</tr>
<tr>
<td>BF / %</td>
<td>72.7</td>
<td>19.5</td>
</tr>
<tr>
<td>Martensite / %</td>
<td>6.8</td>
<td>76.3</td>
</tr>
</tbody>
</table>

TEM was used to further investigate the typical morphologies of the lath structure in WM1 and WM2, and the results are shown in Fig. 4. In Fig. 4a, LB in WM1 is composed of fine parallel ferrites that have an average length less than 2µm. Selected area electron diffraction (SAED) pattern in Fig. 4b reveals that there are retained austenite (RA) films distributed at the boundaries of LB. In WM2, lath structure of
WM2 is LM, and the lath is wider than that of WM1, as seen in Fig. 4c. Also, the SAED pattern in Fig. 4d for the slender M-A constituent shows that there are only martensite diffraction spots between laths, and this reveals that the austenite in M-A was almost transformed into martensite. This is more conducive to improving the toughness of the weld metal when there are RA films rather than slender M-A constituents between laths [22].

![Bright-field TEM micrographs for the weld metals WM1 and WM2: (a) WM1 and (b) WM2. (c) and (d) SAED patterns of areas indicated by dotted circles in (a) and (b) for WM1 and WM2, respectively.](image)

EBSD inverse pole figure (IPF) maps of WM1 and WM2 are presented in Fig. 5a and b. The color legend for the IPF maps is shown in the left bottom left corner of Fig. 5a. The boundaries, including high-angle grain boundaries (HAGBs, > 15°) and low-angle grain boundaries (LAGBs, 2 ~ 15°), are coded by black line and white lines, respectively. The low-angle boundaries consist of substructures such as dislocations or lath boundaries of LB/M [23]. For WM1 (Fig. 5a), the orientation colors of AF are
separated from each other, and this indicates that neighboring AF laths have different orientations, which makes the orientation distribution homogeneous. In contrast, grains in WM2 (Fig. 5b) exhibit less diverse crystallographic orientation, and large-sized LM (with the same orientation) are dominant in the IPF maps. The misorientation angles along the path indicated by the black lines in Fig. 5a and b are shown in Fig. 5c and d, respectively. Obviously, for WM1, the path crosses finer grains, whereas for WM2, the path crosses coarser grains. The red point-point misorientation values demonstrate that all of the HAGBs in the two weld metals have misorientation angles that are basically in the range of 50°–60°. Also, there are a few boundaries (10°–20°) and a myriad of low misorientation angles (<5°) distributed between HAGBs; these observations reveal that there is a high density of substructures and dislocations in the grains. From three different sets of EBSD data, quantified analyses of the grain size distributions of the weld metals were conducted. From the results presented in Fig. 5e and f, it is apparent that WM2 has much larger grains than WM1. The statistical results shown in Fig. 5c and d demonstrate that the amount of HAGB in WM1 is larger than that in WM2, and this is consistent with the grain size distribution.
Fig. 5. EBSD IPF maps of (a) WM1 and (b) WM2; misorientation distributions of (c) WM1 and (d) WM2 along the linear paths indicated in (a) and (b), respectively; effective grain size distributions of (e) WM1 and (f) WM2.

3.2. Mechanical properties of weld metals

Engineering stress-strain curves of the weld metals are presented in Fig. 6a. Tensile test data are summarized at Table 4. The weld metals for two kinds of welding wires all have high tensile strength (above 950Mpa) and favorable elongation (over 17%). These values are rare for a welding wire that has
less than 0.03 wt. % carbon. Also, sample WM2 has a slightly higher strength than WM1, and this is more evident in terms of yield strength. Fig. 6b shows histograms of the impact absorbed energy of the weld metals at -50°C. The average impact absorbed energy for WM1 can be as high as 50J, whereas that for WM2 is just over 10J. This implies that the high yield strength of WM2 is at the expense of its impact toughness.

Fig. 6. (a) Engineering stress-strain curves of the weld metals; □: tensile strength and ○: 0.2% off-set yield strength. (b) V-notch impact absorbed energies of the weld metals at -50°C.

Table 4. Tensile test results of the weld metals

<table>
<thead>
<tr>
<th>Weld metal</th>
<th>Tensile strength $R_m$ (MPa)</th>
<th>0.2% Off-set yield strength $R_{p0.2}$ (MPa)</th>
<th>Total elongation (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>WM1</td>
<td>957</td>
<td>842</td>
<td>18.8</td>
</tr>
<tr>
<td>WM2</td>
<td>974</td>
<td>930</td>
<td>17.2</td>
</tr>
</tbody>
</table>

3.3. Fracture characteristics of the weld metals

Fig. 7 shows SEM images of the tensile fracture surfaces of the weld metals. Generally, the fracture surface is mainly divided into three regions: the fibrous zone (FZ), radical zone (RZ), and shear lip zone (SL) [24]. As seen in Fig. 7a and b, the macroscopic fracture surfaces of WM1 and WM2 have only a central FZ and a marginal SL with few characteristics of an RZ. These observations indicate that WM1 and WM2 all have excellent ductility at room temperature. In the FZ, there are some secondary cracks that extend below the fracture surface, and the number of cracks in WM2 is apparently higher than that...
in WM1. Enlarged FZ micrographs of WM1 and WM2 are shown in Fig. 7c and d, respectively. The FZ in WM1 has a few cleavage facets, but it is mainly covered with dimples. In contrast, the cleavage facet in the FZ of WM2 is obviously larger than that in WM1, and there are less dimples. Detailed morphologies of the SL in both WM1 and WM2 are presented in Fig. 7e and f, respectively. For both WM1 and WM2, the SL is dominated by dimples, and some cleavage facets (including unit cleavage facets) are also observed in the SL of WM2. It is noteworthy that the dimples in the SL are generally shallower than those in the FZ, and some dimples in the SL tend to transform into cleavage facets; these are marked as degenerated dimples.

![Fig. 7. SEM images of the tensile fracture surface for the weld metals WM1 and WM2:](image)

Macrofracture surfaces of (a) WM1 and (b) WM2; (c) and (d) enlarged micrographs of the FZ in (a) and (b), respectively; (e) and (f) enlarged micrographs of the SL in (a) and (b), respectively.

Macrofracture surfaces of V-notch impact specimens for WM1 and WM2 are presented in Fig. 8a and b, respectively. There are three zones: the plastic fracture zone (PFZ), cleavage fracture zone (CFZ), and shear lip zone (SL). PFZs are indicated by white dashed lines in Fig. 8a and b. Obviously, the PFZ of WM1 is wider than that of WM2, and this is attributed to the higher impact toughness. The PFZ of WM1
is predominantly fine dimples, whereas the PFZ of WM2 has hybrid microstructures of large-sized cleavage facets and dimples. Fig. 8e and f show enlarged images of the CFZs, and these consist of cleavage facets and other cleavage patterns, that is, cleavage steps and river patterns. In the CFZ of WM1 (Fig. 8e), there are cleavage facets, which are separated by dimple ridges. While in the CFZ of WM2, there are almost no dimples but cleavage facets. In Fig. 8f, it is observed that a cleavage fracture in WM2 initiates from the blocky M-A constituent.

Fig. 8. SEM images of the impact fracture surface for the weld metals WM1 and WM2: macrofracture surfaces of (a) WM1 and (b) WM2; enlarged images of PFZ in (c) WM1 and (d) WM2; enlarged images of the CFZ in (e) WM1 and (f) WM2. The inset in (f) shows EDS results of the M-A constituent enclosed by the dotted line.

Fig. 9 shows the SEM morphologies of secondary cracks beneath the impact fracture surfaces. As seen in Fig. 9a-c, cracks were preferentially initiated on M-A constituents, and this can be attributed to the debonding mechanism between M-A constituents and the matrix. In WM1, cracks were inclined to pass obliquely through the lath structure and to break the slender M-A constituents in the propagation path. Meanwhile, there are more crack deflections rather than crack arrests at the grain boundaries in Fig. 9d,
and this indicates that the microstructure of WM2 has poor resistance to crack propagation.

Fig. 9. SEM morphologies of secondary cracks in impact specimens of the weld metals: (a) and (b): WM1; (c) and (d): WM2.

4. Discussion

4.1 Effect that adding Cu-Nb has on the microstructure of the weld metals

Because the same welding procedure was used, the microstructure of the weld metal is primarily controlled by the chemical composition of the welding wire. At an ultralow carbon level, the weldability is improved, and the quantity of cementite and carbide in the weld metal is also reduced [25]. This contributes significantly to limiting the decrease in toughness. In a previous study [20], the ultralow amount of carbon coupled with microalloys contributes to obtaining fine bainitic structure in the weld metal, which has good comprehensive mechanical properties. By comparing the chemical compositions in Table 1, W2 (welding wire 2) contains additional amounts of Cu and Nb. Cu is mainly dissolved in austenite as a solid solution hardening element; this increases the stability of austenite and leads to easier
transformation of martensite and other low temperature phases [24, 26-27]. It has been found that the addition of 0.18 wt. % Cu to X120 steel welding wire increases the fraction of BF at the expense of GB in weld metal. Also, the prior austenite grain size and ductile-brittle transition temperature of the weld metal increase with the addition of Cu. This results in the deteriorated toughness [24]. From research on a low alloy steel weld metal [26], it was reported that an increase in Cu simultaneously increases the hardness and tensile strength as a result of the solid solution hardening effect. As indicated in Table 3, the amount of martensite in WM2 is as high as 76%, whereas the amount of AF decreases to 4.2%. This observation reveals that the phase of WM2 is inclined to transform at low temperature, and this reflects its strong hardenability. In contrast, WM1 is composed of phases, such as AF, GB, and LB, which transform at relatively high temperature. From the above discussion, it can be inferred that adding 0.15 wt. % Cu to W2 enhances the hardenability and induces more low-temperature transformation microstructures in WM2.

Generally, adding a trace amount of Nb to steel produces undissolved precipitates (typically Nb (C, N)), and these significantly improve the refinement of the austenite grain size [18, 28]. Compared to the grain size of WM1, the grain size of WM2 increases remarkably when 0.08 wt.% Nb was added to W2, as indicated in Fig. 5e and f. These observations imply that adding Nb does not cause grain refinement of WM2. Moon [29] investigated the influence that adding Nb has on a Ti-containing steel weld HAZ. The results suggested that (Ti, Nb) (C, N) complex particles are produced instead of multiple particles of Ti (C, N) and Nb (C, N) when Nb was added to Ti-containing steel. However, (Ti, Nb) (C, N) particles grow more easily because of they have weaker bonds than either single Ti-N or Ti-C. Consequently, the comprehensive pinning effect of precipitated particles in steel is restrained, and this leads to an increase in the size of austenite grains. In this study, both W1 and W2 contain small amounts of Ti. Therefore,
the refinement effect of Nb in WM2 is likely restricted because of weaker pinning effect of (Ti, Nb) (C, N) complex particles. It is evident that the PAGS of WM2 is larger than that of WM1, and this also causes a more complete degree of martensite transformation as well as the greater hardenability of WM2 [31].

4.2 Correlation between microstructure and mechanical properties of the weld metals

Fig. 10 shows the true stress-strain curves of the weld metals and the corresponding strain hardening rate curves. According to the Hart criterion [32], under high uniaxial tensile stress, plastic elongation is easily affected by a local necking deformation, and this leads to early failure of the tensile specimen. This relationship can be described by the inequality of strain and strain hardening rate:

\[
d\sigma/d\varepsilon + m\sigma < \sigma
\]

where \(\sigma\) is the true stress, \(\varepsilon\) is the true strain, and \(m\) is the strain rate sensitivity, which can be negligible at room temperature [32]. Tensile specimen will enter unstable necking stage when inequality (1) is established. Therefore, to prevent the tensile specimen from fracturing, it is necessary for the strain hardening rate \((d\sigma/d\varepsilon)\) to be high enough to change the inequality (1). The intersection of the true stress-strain curve and the corresponding strain hardening rate curve \((d\sigma/d\varepsilon=\sigma)\) is the critical condition for the tensile specimen to enter the necking stage. When \(d\sigma/d\varepsilon<\sigma\), the strain hardening effect of the weld metal does not meet the requirement for a further increase in stress, and the fracture process for tensile specimen begins. As seen in Fig. 10, a higher work-hardening rate is observed for WM1 than for WM2, and this causes the WM1 sample to have a higher ductility. With a decrease in the strain hardening rate, the WM1 sample, which has higher \(d\sigma/d\varepsilon\), can easily keep up with the increase in \(\sigma\); this effectively delays the necking process, which improves the ductility. Meanwhile, the lower yield strength of WM1 is also beneficial for improving its impact absorbed energy during the crack initiation stage.
As seen from the tensile fracture surface shown in Fig. 7c and d, there are many cleavage facets that have parallel slip bands. The bands are formed via a mechanism of microvoid nucleation, growth, and polymerization [33]. Crack initiation of the cleavage facet usually results from a dislocation pile-up on slip bands. When shear stress is applied in a certain direction during the tensile process, small cleavage facets are likely to form via crack propagation. The larger size and higher quantity of cleavage facets in WM2 also demonstrate the poor ductility of WM2.

![True stress-true strain curves and strain hardening rate curves](image)

**Fig. 10.** True stress-true strain curves (dashed lines) and strain hardening rate curves (solid lines) of tensile specimens for the weld metals.

It is well known that AF, which has a fine grain size and cross-interlocking structure, is the desired microstructure of a weld metal in terms of improving toughness [21]. The radiating structure and HAGB of AF effectively restrain crack propagation. Moreover, grain size is one of the most important factors that affect impact toughness [30], and thus, grain coarsening of WM2 has an obvious deterioration effect on impact toughness. Lan [34] investigated the effect that refining prior austenite grains has on the coalescence of bainite; they reported that the combination of both strength-ductility and toughness were significantly increased when the grain size was refined. Meanwhile, as seen from the IPF map of WM2 (Fig. 5c and d), the lath boundaries (indicated by the white solid line) are low misorientation angle (<15°)
boundaries, which have a weak effect on impeding crack propagation [23]. Therefore, the HAGB of WM2 has a relatively low density compared to that of WM1, and this causes the CFZ to be completely covered with cleavage facets. As seen in Fig. 9, it is noted that HAGB can induce deflection or arrest during crack propagation. This enhances the consumption of crack extension energy and improves the toughness. Because the amount of AF and fine grains is moderate, crack propagation in WM1 is easily arrested by HAGB. In contrast, it is easy to propagate cracks in WM2 because of the poor crack resistance of the coarsened LM. Hence, it has been illustrated that refining grain size to increase the density of HAGB is an effective way to improve the toughness of the weld metal.

Both RA and M-A constituents are important factors that affect the toughness of weld metals. On one hand, there are several mechanisms that support RA to improve the toughness of steel, these are as follows [35,36]:

(i) The deformation-induced martensite transformation (DIMT) from RA dissipates energy, and this reduces the energy available for crack initiation and propagation.

(ii) DIMT from RA delays the nucleation and coalescence of microvoids.

(iii) RA without DIMT can be deformed, and thereby cracks can be blunted and arrested.

On the other hand, the martensite-austenite (M-A) constituent is most commonly a mixture of high carbon martensite plus retained austenite (RA) [37]. The hardness of the slender M–A constituent is higher than that of the ferrite matrix. This induces stress concentrations in the neighboring ferrite matrix, and this is a key factor that is related to the fracture initiation and deterioration of toughness [38]. Wang's study [15] showed that M-A constituents are the main nucleation sites for cracks in multi-pass weld metal. In our previous work [39], we investigated the effect that the morphology of M-A constituents has on impact toughness, and we found that MA constituents that have a slender shape contain almost complete
martensite, which leads to lower toughness. The morphologies of the secondary cracks (Fig. 9a and c) indicate that the crack tends to propagate through the slender M-A constituent, and this is also consistent with the findings of our previous work. Due to the enhanced hardenability in WM2, the martensite in M-A constituent will be more completely transformed by inhibiting the carbon from diffusing into RA [35]. This leaves a good deal of slender M-A without RA. Thus, the absence of RA between LM indicates that WM2 has a higher degree of complete transformation to the M-A constituent than WM1, and this cause the toughness loss of WM2 to be more serious.

5. Conclusions

(1) Two kinds of ultralow carbon weld metals were fabricated using GMAW. The weld metal without added Cu-Nb (WM1) has a high tensile strength of 957MPa, a total elongation of 18.8%, and an average impact energy of 50.2J at -50°C. The yield strength of the weld metal with added Cu-Nb (WM2) increased by approximately 100MPa, whereas the impact toughness and strain hardening rate obviously decreased.

(2) WM1 is mainly composed of AF, GB, and LB, and the fraction of AF is about 20%; this is beneficial for the ductility and toughness of WM1. Also, WM2 is predominantly coarsened LM (about 76%). The grain size (including PAGS) of WM2 is significantly larger than that of WM1, and this cause the HAGB to have a relatively small density.

(3) Slender M-A constituents are distributed between martensitic laths in WM2, and these are transformed into martensite, whereas some austenitic films are still retained between laths in WM1.

(4) In WM2, the strength increases, but the toughness decreases; these observations are attributed to the enhanced hardenability and grain coarsening in the weld metal, as induced by the addition of Cu-Nb.

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