Microstructural evolution and mechanical properties of heat affected zones for 9Cr2WVTa steels with different carbon contents

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Abstract

The microstructures and mechanical properties of heat affected zones (HAZs) by Gas Tungsten Arc Welding (GTAW) were studied for 9Cr2WVTa steels with carbon content varying from 0.07 wt.% to 0.25 wt.% Enlarged HAZs samples with 8 mm to 10 mm wide uniform temperature zone were prepared by the thermal–mechanical physical simulator Gleeble 1500 based on the Finite Element Method (FEM) numerical simulation and experimental measurement for the welding thermal cycle process and weld profile. The microstructures were observed by optical microscope (OM), scanning electron microscope (SEM) and transmission electron microscope (TEM). In addition, the mechanical properties tests including micro-hardness test, tensile test and impact test were carried out to investigate the effects of the carbon content and the welding thermal cycle. The results show that the big blocky delta ferrite in 9Cr2WVTa steel with lower carbon content deteriorates the impact property. On the other hand, the quenched martensite, especially for the twin martensite in 9Cr2WVTa steel with higher carbon content, deteriorates the impact toughness as well. The weldability of 9Cr2WVTa steel can be improved by adjusting the carbon content between 0.14 wt.% and 0.17 wt.%.

1. Introduction

In 1999, the United States Department of Energy was the first to put forward the concepts of Generation-IV nuclear power. Three years later, the Generation-IV International Forum issued six reactor systems that offered the potential for meeting the Generation-IV goals, which included very high-temperature gas-cooled reactor (VHTR), gas-cooled fast reactor (GFR), sodium-cooled fast reactor (SFR), lead-cooled fast reactor (LFR), molten salt reactor (MSR) and super-critical water-cooled reactor (SCWR) [1]. The Generation-IV nuclear power is aimed at solving the problems of economy, safety, reliability, sustainability and nuclear non-proliferation. In the meantime, researches on the transmutation of long-lived radioactive nuclides are carried out with an emphasis on dedicated Accelerator Driven Systems (ADS). The reference design of ADS plant in Japan atomic energy research institute is the 800 MW, Pb–Bi eutectic (LBE) cooled, tank-type subcritical reactor loaded with (MA+Pu) nitride fuel [2]. Pb–Bi ADS reactor is one of the LFR with LBE as cooling medium for Generation-IV nuclear power station [3], and requires the materials to be used have higher corrosion resistant to LBE at high temperature [4]. Candidate materials have been considered, such as austenitic stainless steel, vanadium-base alloys [5], SiC/SiC composite materials [5] and high chromium ferritic/martensitic (F/M) steel [6] including reduced activation ferritic/martensitic (RAFM) steel [7,8] and oxide dispersion strengthen (ODS) steel [5]. According to the proposed demand for nuclear power companies, these materials are developed to increase nuclear power efficiency, which means burn less fossil fuel and coal at higher temperatures. Thereby, these materials can reduce costs and meet the increasingly stringent environmental requirements [9]. F/M steels with 9–12 wt.% Cr such as T91, EP823, F82H, HT9, Eurofer97 and 9Cr2WVTa have been considered for wide application in view of their lower corrosion rate and aging embrittlement [10], RAFM steel, which substitutes the elements of tungsten, vanadium and tantalum for molybdenum, niobium and nickel [7], has excellent irradiation resistance [11,12], good Pb–Bi corrosion resistance [13] as well as good mechanical strength such as high temperature strength [14] and high temperature creep property [15,16]. In brief, RAFM steel is the candidate material for Generation-IV nuclear power and needs to be thoroughly researched [17].

For the structural material, welding is an inevitable procedure for the fabrication of nuclear reactors [18]. However, few works are carried out on the weldability of the RAFM steel. Therefore, it is necessary to put forward the foundation of chemical composition design based on the perspective of its weldability. The HAZs
of 9Cr2WVTa steel include the coarse grained heat affected zone (CGHAZ), fine grained heat affected zone (FGHAZ) and intercritical heat affected zone (ICHAZ) [19,20]. Different microstructures of HAZs lead to different mechanical properties. The inhomogeneity of these micro-regions, which experience different welding thermal cycles, often cause the welding joint to be a weak area [21]. Therefore, it is valuable to investigate the microstructural evolution and mechanical properties of these micro-regions in HAZs exactly for analyzing the obdurability of the whole welding joint. The mechanical test samples of these micro-regions are not easy to obtain directly for its narrow size [17]. The micro-regions in HAZs can be enlarged by the thermal–mechanical physical simulator Gleeble 1500, which can provide convenience for the intensive research consequently [22,23].

In this study, the temperature field of Gas Tungsten Arc Welding (GTAW) on a 9Cr2WVTa steel plate was obtained by the thermal couples. Characteristic welding thermal cycle curves of the extracted HAZs were plotted with the aids of FEM and the measurements of phase transformation temperature. The enlarged HAZs samples with 8–10 mm wide uniform temperature zone were prepared by the thermal–mechanical physical simulator Gleeble 1500. Then, the microstructure observation and the mechanical properties of HAZs and base metal (BM) for 9Cr2WVTa steels with different carbon contents were conducted. Finally, the optimized scopes of the carbon content in 9Cr2WVTa steel were proposed to improve its weldability.

2. Materials and experimental methods

Five 9Cr2WVTa steels employed in this study were melted by a vacuum induction furnace, followed by hot-forging into a 45 mm square rod. The chemical compositions are given in Table 1. Specimens were austenitized at 1050 °C for 1.5 h followed by water quenching, then tempered at 750 °C for 2 h followed by subsequent air cooling.

The Young’s modulus, Shear modulus and Poisson’s ratio, thermal expansion, thermal diffusivity, specific heat and thermal conductivity at different temperatures were measured for the welding process simulation by the equipments RFDA HTVP 1750-C, Unitherm™-1252 Ultra High Temperature Dilatometer, and Flashline™-5000 Thermal Properties Analyzer according to the ASTM standards ASTM: E1876, ASTM: E228 and ASTM: E1461, respectively.

Enlarged HAZs samples were prepared by the thermal–mechanical physical simulator Gleeble 1500 based on the FEM numerical simulation and experimental measurement for the welding thermal cycle process and weld profile. The sizes of samples used in Gleeble 1500 machine were 11 mm × 11 mm × 80 mm and 3 mm × 11 mm × 80 mm. After thermal–mechanical physical simulation experiments, these samples were machined for standard impact samples and tensile samples. The tensile sample was designed in Fig. 1 in order to guarantee the fracture path in the test section. The Charpy V impact tests and tensile tests were carried out on SANS-ZBC2452-C instrument impact testing machine according to the ASTM standard ASTM: E23 and SANS-CMT 5205 universal testing machine at room temperature, respectively. The micro-hardness tests was carried out on LECO AMH43 automatic micro-indentation hardness testing system according to the ASTM standard ASTM: E384. The microstructures of HAZs and BM were analyzed by optical microscope (OM, ZEISS AXIOVERT 200 MAT), scanning electron microscope (SEM, Inspect F50) and transmission electron microscope (TEM, Tecnai G2 F20).

3. Results and discussion

3.1. GTAW heat source calibration and simulated HAZS samples fabrication

During welding process, the substrate material near the weld metal experiences different welding thermal cycle processes so that the microstructures and mechanical properties of these areas (HAZs) are complex and variable. Investigations on the microstructural evolution and mechanical properties of these areas (HAZs) are necessary to evaluate the weldability of the substrate material. GTAW self-melted welding is carried out on a 160 mm × 68 mm × 5 mm plate with welding speed of 2.33 mm/s and welding current of 160 A under 99.9% argon shielding. During this procedure, six points for temperature monitoring are gauged by K-thermocouples in the HAZs near the welding bead, as shown in Fig. 2. At the same time, the Sysweld software is used for simulating the welding process. Finally, the welding heat source model is calibrated by comparing the weld profiles between the simulation and the experiment. Fig. 3 exhibits the weld profiles consisting of experimental and simulated ones. The experimental measurements (Fig. 3(a)) show that the width and depth of the weld pool is 8.4 mm and 2.6 mm, respectively, while the width of HAZs is 2.5 mm. On the other hand, the results of numerical simulation show that the width and depth of weld pool is 8.6 mm and 2.6 mm, respectively, while the width of HAZs is 2.5 mm. So the simulated result is fitting well with the experimental ones, and the established welding heat source model is reasonable.

The simulated welding thermal cycle curves are extracted from the established welding heat source model. The positions of six monitoring points are shown in Fig. 2 (TC1–TC6). Fig. 4 shows the comparisons of these thermal cycle curves between experimental and simulated ones. The results show that these curves are fitting well, especially for the peak temperatures, which means that the established welding heat source model is accurate.

For rapid heating rate in the welding process, the phase transformation temperature of 9Cr2WVTa in the welding process is different from the equilibrium phase transformation point. In Mayr’s [24] result, the equilibrium phase transformation temperatures of 9Cr–1.5Mo steel were 835 °C (A1) and 887 °C (A3), while 903 °C (A1) and 1001 °C (A3) for rapid heating rate during welding. Therefore, faster heating rate contributes to a higher phase transformation temperature for steel. Therefore, the phase transformation temperature of 9Cr2WVTa steel under rapid heating rate is measured by the equipment L78 RITA Quenching and Deformation Dilatometer according to the ASTM Standard ASTM: A1033. Fig. 5 shows the effects of heating rate on the phase transformation

<table>
<thead>
<tr>
<th>Table 1</th>
<th>Chemical compositions of 9Cr2WVTa steels/wt.3%</th>
</tr>
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<tbody>
<tr>
<td>No.</td>
<td>C</td>
</tr>
<tr>
<td>---------</td>
<td>-------</td>
</tr>
<tr>
<td>1</td>
<td>0.07</td>
</tr>
<tr>
<td>2</td>
<td>0.11</td>
</tr>
<tr>
<td>3</td>
<td>0.14</td>
</tr>
<tr>
<td>4</td>
<td>0.17</td>
</tr>
<tr>
<td>5</td>
<td>0.25</td>
</tr>
</tbody>
</table>
temperatures ($A_1$, $A_2$) for 9Cr2WVTa steel. Considering the actual welding process, the heating rate is about 150–240 °C/s from 300 °C to peak temperature (Fig. 4). Here the average heating rate of the welding thermal cycle in HAZs is set at 180 °C/s. Therefore, the HAZs can be subdivided into CGHAZ (>1150 °C), FGHAZ (1035–1150 °C) and ICHAZ (908–1035 °C). Three representative thermal cycle curves with peak temperatures at 1315 °C, 1100 °C and 970 °C respectively are selected to represent the welding thermal cycle processes for the CGHAZ, FGHAZ and ICHAZ, as shown in Fig. 6. The heating rates are 234 °C/s, 189 °C/s and 163 °C/s and the $t_{8/5}$ are 21.7 s, 20.8 s and 20.5 s for the three welding thermal cycle curves. Finally, the enlarged HAZs samples with 8 mm to 10 mm wide uniform temperature zone are prepared by the thermal–mechanical physical simulator Gleeble 1500 for the following microstructure observation and mechanical property test.

3.2. Microstructures of the simulated HAZs and BM

Fig. 7 exhibits the microstructures of BM and HAZs with different carbon contents. When the carbon content is 0.07 wt.%, the BM has dual phase microstructures with tempered martensite and thin strip δ-Fe. While in the areas of HAZs, the microstructural evolution is complex and influenced by both chemical compositions and cooling rates. In Zheng’s [23] result, the maximum cooling rate to form ferrite was 1 °C/s for CLAM steel. When the cooling rates ranged from 60 °C/s to 1 °C/s, fully martensite transformation was achieved and the cooling rates had little influence on the microstructure and micro-hardness of martensite. The pseudo-equilibrium phase diagram of 9Cr2WVTa steels with different carbon contents is calculated by Thermo-Calc software with database TCFE7, as shown in Fig. 8. When the carbon content is below 0.12 wt.%, the single delta ferrite phase (δ) zone appears. The δ-Fe retains in the BM with low carbon content of 0.07 wt.% because the transformation of δ-Fe to γ-Fe is a diffusion controlled process and the cooling rate during quenching process is high. The microstructure of CGHAZ is quenched martensite with big blocky δ-Fe. Since the peak temperature of the welding thermal cycle is high at 1315 °C, the original δ-Fe in the BM grows rapidly and aggregates at this high temperature range. In the cooling process, it does not provide sufficient time to complete the phase transformation process from δ-Fe to γ-Fe due to fast cooling rate in GTAW [25], so the big blocky δ-Fe retains in the CGHAZ for the carbon content of 0.07 wt.%. When it comes to the area of FGHAZ, the microstructure is the quenched martensite with thin strip δ-Fe. Besides, the microstructure of ICHAZ is mainly lath martensite, which includes tempered martensite, newly formed quenched martensite, and thin strip δ-Fe, as shown in Fig. 7. The volumes of δ-Fe in each typical area are listed in Table 2.

The transformation temperature of γ-Fe to δ-Fe becomes higher with increasing the carbon content, as shown in Fig. 8 from point A to E, which indicates that the complete austenitizing zone becomes larger. Therefore, the stability of austenitizing increases and the δ-Fe disappears in the BM gradually. As a result, the whole martensite forms with no δ-Fe existed in the BM when its carbon content exceeds 0.11 wt.% as shown in Fig. 7 and Table 2. During the welding process, as the peak temperature of welding thermal cycle is high at 1315 °C, which is over the phase transformation temperatures of γ-Fe to δ-Fe, 1170 °C (point B) and 1210 °C (point C) for the samples with 0.11 wt.% and 0.14 wt.% carbon contents, as shown in Fig. 8, the phase transformation of γ-Fe to δ-Fe occurs.
and newly $\delta$-Fe forms in the CGHAZ. Therefore, there are quenched martensite and blocky $\delta$-Fe in the CGHAZ for the samples with 0.11 wt.% and 0.14 wt.% carbon contents, as shown in Fig. 7. When the carbon content increases to 0.17 wt.%, the phase transformation temperature of $\gamma$-Fe to $\delta$-Fe is around 1235 °C (point D in Fig. 8), which is a little lower than 1315 °C. There is no enough time for the formation of $\delta$-Fe since the experience time over 1235 °C is quite short during the welding thermal cycle process. For the sample with carbon content of 0.25 wt.%, the phase transformation temperature of $\gamma$-Fe to $\delta$-Fe is around 1320 °C (point E in Fig. 8), which is higher than 1315 °C and there is no $\delta$-Fe in the HAZs either.

The peak temperatures of the representative welding thermal cycles for the FGHAZ and ICHAZ are 1100 °C and 970 °C respectively, which are lower than the $\gamma$-Fe to $\delta$-Fe phase transformation temperatures (point B: 1170 °C, point C: 1210 °C, point D: 1245 °C and point E: 1320 °C, as shown in Fig. 8) when the carbon content of the BM is over 0.11 wt.% so there is no $\delta$-Fe formed in FGHAZ.

Fig. 4. Comparison diagrams showing the thermal cycle curves between experiment and simulation (a) TC1 point, (b) TC2 point, (c) TC3 point, (d) TC4 point, (e) TC5 point and (f) TC6 point.

Fig. 5. $A_1$ and $A_3$ points for 9Cr2WVTa steel with different heating rates.
Fig. 6. Typical thermal cycle curves extracted from welding model which were shown in Fig. 3. (a) CGHAZ (1315 °C), (b) FGHAZ (1100 °C) and (c) ICHAZ (970 °C).

Table 2
The volume of delta ferrite in HAZs and BM with different carbon contents.

<table>
<thead>
<tr>
<th>Carbon content/%</th>
<th>BM/%</th>
<th>CGHAZ/%</th>
<th>FGHAZ/%</th>
<th>ICHAZ/%</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.07</td>
<td>9.27</td>
<td>32</td>
<td>10.28</td>
<td>12.20</td>
</tr>
<tr>
<td>0.11</td>
<td>0</td>
<td>11.45</td>
<td>0</td>
<td>0</td>
</tr>
<tr>
<td>0.14</td>
<td>0</td>
<td>7.23</td>
<td>0</td>
<td>0</td>
</tr>
<tr>
<td>0.17</td>
<td>0</td>
<td>0</td>
<td>0</td>
<td>0</td>
</tr>
<tr>
<td>0.25</td>
<td>0</td>
<td>0</td>
<td>0</td>
<td>0</td>
</tr>
</tbody>
</table>

Fig. 7. The OM images of the samples with different carbon contents.
and ICHAZ for the samples with carbon content over 0.11 wt.%, as shown in Fig. 8.

### 3.3. Mechanical properties of the simulated HAZs and BM

According to the three representative thermal cycle curves, enlarged HAZs samples with 8–10 mm wide uniform temperature zone for the CGHAZ, FGHAZ and ICHAZ are fabricated by the thermal–mechanical physical simulator Gleeble 1500. Mechanical tests for these typical zones (CGHAZ, FGHAZ and ICHAZ) are carried out.

#### 3.3.1. Micro-hardness and tensile test

The results of Micro-hardness tests are shown in Fig. 9. The Vicker hardness of BM and HAZs increases with the carbon content. The hardness of the HAZs is much higher than that of the BM because the microstructure of HAZs is mainly quenched martensite and the microstructure of the BM is mainly tempered martensite. The microstructure of ICHAZ composes of quenched martensite and tempered martensite, and its hardness is between that of FGHAZ and BM. Comparing the hardness of CGHAZ with that of FGHAZ, the hardness of FGHAZ is higher than that of CGHAZ when the carbon content is below 0.17 wt.%, especially for the samples with carbon content of 0.07 wt.%. Micro-hardness test for the samples with 0.07 wt.% carbon content shows that the hardness of δ-Fe in HAZs is remarkably lower than that of martensite, as shown in Table 3. δ-Fe in the CGHAZ decreases its hardness when the carbon content is below 0.17 wt.%. When the carbon content reaches to 0.25 wt.% the highest hardness moves to the area of CGHAZ, which can be interpreted by the formation of coarsening quenched martensite and there is no δ-Fe formed.

Fig. 10 gives the tensile strength for the HAZs and BM with different carbon contents. The variation of the tensile strength is similar to that of the Vicker hardness. When the carbon content is below 0.17 wt.%, the strength of CGHAZ is lower than that of FGHAZ. When the carbon content reaches to 0.25 wt.%, the strength of CGHAZ is the highest and almost one point five times higher than that of BM.

#### 3.3.2. Impact test

The results of impact tests are shown in Fig. 11. The toughness of HAZs strongly depends on the microstructure. With the carbon content increasing, the impact toughness decreases in the areas of BM, FGHAZ and ICHAZ, respectively. However, the toughness of CGHAZ increases initially and reaches to the peak value when the carbon content is 0.17 wt.%, followed by a decrease as carbon content exceeding 0.17 wt.%. This tendency is the result of competition between quenched martensite and big blocky δ-Fe in the CGHAZ.

When the carbon content is below 0.14 wt.%, the impact toughness of CGHAZ is lower remarkably than that of FGHAZ and ICHAZ due to the existence of the blocky δ-Fe. As mentioned above in Section 3.2, the δ-Fe existing in 9Cr2WVTa steels has different shapes and volumes, as shown in Fig. 7 and Table 2. Fig. 12 gives the fracture morphologies of the impact samples with 0.07 wt.% carbon content. The fractographies of BM, FGHAZ and ICHAZ (Fig. 12(a), (c), and (d)) exhibit typical ductile dimple fracture patterns. On the contrary, the fractography of CGHAZ (Fig. 12(b)) shows quasi-cleavage fracture with typical river-like patterns, step patterns and ligule patterns. Fig. 13 gives the cross-section samples perpendic-

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**Table 3**
The micro-hardness of delta ferrite and martensite with 0.07 wt.% carbon content

<table>
<thead>
<tr>
<th>Area</th>
<th>CGHAZ</th>
<th>FGHAZ</th>
<th>ICHAZ</th>
<th>BM</th>
</tr>
</thead>
<tbody>
<tr>
<td>δ</td>
<td>155</td>
<td>171</td>
<td>186</td>
<td>188</td>
</tr>
<tr>
<td>M</td>
<td>359</td>
<td>333</td>
<td>305</td>
<td>228</td>
</tr>
</tbody>
</table>

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**Fig. 9.** Micro-hardness tests of the BM and HAZs with different carbon contents.

**Fig. 10.** Tensile tests of the BM and HAZs with different carbon contents.

**Fig. 11.** Impact tests of the BM and HAZs with different carbon contents.
ular to the impact fracture surface. It is clearly shown from Fig. 13(a) that the second-cracks form on the big blocky ε-Fe in the CGHAZ. The hardness of ε-Fe in the CGHAZ is about 155 HV0.01, which is much lower than that of lath martensite (359 HV0.01). Relevant literature showed that the crack tends to propagate into the softer grain [22]. When deformation occurs during the impact process, the dislocation slips in ε-Fe easily and piles up near the interface between the ε-Fe and the martensite, which leads to the stress concentration at the interface between ε-Fe and martensite. Therefore, the cracks initiate easily at the interface between ε-Fe and martensite, then the cracks preferentially propagate in the soft phase (ε-Fe) and form pre-cracks (Fig. 13). According to Griffith strength theory, the critical fracture stress is inversely proportional to the length of pre-crack. So the critical
fracture stress of martensite matrix with big blocky δ-Fe (CGHAZ) is remarkably lower than that with thin strip δ-Fe (BM, FGHAZ and ICHAZ). Therefore, the impact toughness of CGHAZ is much lower than that of BM, FGHAZ and ICHAZ. For the samples with carbon contents of 0.11 wt.% and 0.14 wt.%, δ-Fe just forms in the CGHAZ only and makes its impact toughness lower than that of FGHAZ and ICHAZ.

When the carbon content exceeds 0.14 wt.%, the impact toughness of HAZs samples with 0.17 wt.% and 0.25 wt.% carbon contents decreases as a whole comparing with that of HAZs samples with relatively low carbon content (<0.14 wt.%), as shown in Fig. 11. There is no δ-Fe existed in the CGHAZ for the samples with high carbon contents of 0.17 wt.% and 0.25 wt.%, as shown in Fig. 7. Relevant literatures showed that the impact property for martensitic material depends on the matrix microstructures, including the martensite type, precipitated phase and the lath packet size. In Wang’s [26] result, the packet boundaries can strongly hinder fracture propagation, thus the martensitic packets can act as the effective microstructure unit for cleavage [27]. Fig. 14 shows that the matrix is the dislocation martensite for the 0.17 wt.% carbon content sample (Fig. 14(a)), while twin martensite for the 0.25 wt.% carbon content sample (Fig. 14(b)). The twin martensite has high quenching stress and leads to a significant decrease of toughness. The fractography of FGHAZ with 0.17 wt.% carbon content (Fig. 15(a)) exhibits ductile dimple fracture pattern. The dimples are relatively large with some ligaments around them, which hinder the cracks propagation and show good toughness. The fractography of FGHAZ with 0.25 wt.% carbon content (Fig. 15(b)) exhibits the quasi-cleavage fracture, with some cracks on the fracture surface. These cracks are almost caused by the second-phase particles and carbides. Fig. 15(c) and (d) show the fractographies of CGHAZ and FGHAZ with 0.25 wt.% carbon content. The dimples are small and light on them, which caused by the fracture of twin martensite. These small and light dimples hinder the growth of large dimples, that is to say, the plastic deformation around them induces the small dimples, which makes the matrix absorb less energy in the process of fracture. Therefore, the impact toughness of the HAZs with 0.25 wt.% carbon content is lower overall.

The key factor influencing the impact toughness is the type of microstructure for 9Cr2WVTa steel. The big blocky δ-Fe formed in the sample with low carbon content (0.07 wt.%), and the twin martensite formed in the sample with high carbon content (0.25 wt.%) deteriorate the impact toughness remarkably. The high strength matching with good toughness of the HAZs can be obtained when the carbon content of 9Cr2WVTa steel is controlled between 0.14 wt.% and 0.17 wt.%.

4. Conclusions

The weldability of 9Cr2WVTa steels with the carbon contents between 0.07 wt.% and 0.25 wt.% for Pb-Bi ADS reactor of Generation-IV nuclear power station is studied systematically. The following conclusions are drawn:

(1) The welding heat source model for GTAW is established based on the FEM numerical simulation and experimental measurement for the welding thermal cycle process and weld profile. Characteristic welding thermal cycle curves are selected to represent the CGHAZ, FGHAZ and ICHAZ considering the influence of the welding heating rate on the phase transformation temperature. Enlarged HAZs samples with 8–10 mm wide uniform temperature zone for the mechanical property test to the micro-regions of HAZs are prepared by the thermal–mechanical physical simulator Gleeble 1500.

(2) The chemical composition of the BM and the peak temperature of the welding thermal cycle affect the microstructure of HAZs comprehensively. The retained δ-Fe in the BM of 9Cr2WVTa steel with low carbon content (0.07 wt.%) grows rapidly and aggregates after experiencing the welding thermal cycle with peak temperature of 1315 °C. Though the BM of 9Cr2WVTa steel with middle carbon content (0.14 wt.%) is the whole martensite, δ-Fe is steel newly formed in CGHAZ because the peak temperature (1315 °C) of the welding thermal cycle is much higher than the γ-Fe to δ-Fe phase transformation temperature (1210 °C) for the sample with
0.14 wt.% carbon content. With increasing the carbon content to 0.25 wt.% in 9Cr2WVTa steel, the γ-Fe to δ-Fe phase transformation temperature increases to 1320 °C, which is higher than the peak temperature of the welding thermal cycle for CGHAZ (1315 °C), and there is no δ-Fe formed in the HAZs during the welding thermal cycle process.

(3) Mechanical property of the HAZs for 9Cr2WVTa steel depends to a large extent on the microstructure, including the volume of δ-Fe and the type of the martensite. After welding thermal cycle process, the tensile strength of the micro-regions (CGHAZ, FGHAZ and ICHAZ) is higher than that of the BM because the main microstructure is quenched martensite in the HAZs while tempered martensite in the BM. As carbon contents are in the range of 0.07 wt.% to 0.14 wt.% for 9Cr2WVTa steels, the tensile strength of CGHAZ is a little lower than that of FGHAZ, while the impact toughness of CGHAZ is much lower than that of FGHAZ because there is blocky δ-Fe formed in the CGHAZ. As carbon content in the 9Cr2WVTa steel increases to 0.25 wt.%, there is no δ-Fe existed in the CGHAZ, but forms twin martensite which also makes the impact toughness of the HAZs much lower than that of HAZs with 0.17 wt.% carbon content. Therefore, the optimized scopes of carbon contents for 9Cr2WVTa steels with relatively good tensile strength and impact property for HAZs are between 0.14 wt.% to 0.17 wt.%.

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