Combined effects of welding heat input and peak temperature on precipitation and mechanical properties of the HAZ for modified austenitic medium manganese steels

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Abstract

We studied the microstructure and mechanical properties of the simulated welding heat-affected zone (HAZ) for modified austenitic medium manganese steels (MAMMS) in the welding peak temperature range of 750 °C to 1050 °C. The precipitation behavior of cementite and VC particle was sensitive to the welding thermal cycle in this temperature range. When the peak temperature increased from 750 °C to 850 °C, the intergranular cementite and VC particle became coarser, leading to deterioration of toughness. However, when the peak temperature raised to 1050 °C, the intergranular cementite disappeared. The sensibility of tensile strength to heat input at high peak temperature was higher than that at low temperature, which implied that it was affected by combined effects of the two factors of heat input and peak temperature, rather than a single factor. This variation tendency was highly related with the coarsening behavior of VC particle depending on the interaction effect of these two factors. The growth rate of VC particle with peak temperature of 850 °C was ~4 times as high as that with peak temperature of 750 °C. However, it only increased to ~1.3 times when the peak temperature increased from 850 °C to 1050 °C.

1. Introduction

Hadfield manganese steel, with a composition of Fe−1.2%C−13%Mn, displays high work hardening ability, remarkable toughness and excellent wear resistance under high-energy impact load [1, 2]. Therefore, it has been widely used as the wear resistant material in excavators, mineral crushing equipment and other severe mechanical conditions. However, the work hardening properties of Hadfield manganese steel is poor under low-impact conditions. Fortunately, the modified austenitic medium manganese wear resistant steel (MAMMS) with manganese content of 4.5−9 wt% and lower carbon content exhibits a high work hardening rate under low-impact energy condition, contributing to the remarkable wear resistance [3−5]. The high work hardening ability under low-impact energy for MAMMS stems from the strain-induced martensite transformations, which is related to the lower stacking fault energy (SFE) [6, 7].

One of the bottlenecks to the application of this MAMMSs is their poor weldability due to the high carbon and alloy contents. The weldments of austenite manganese steel are susceptible to hot cracking because of the high volumetric expansion coefficient of the austenitic phase and the volume variation after the precipitation of transformation products [8, 9]. The mechanical properties of heat-affected zone (HAZ) for austenite manganese steel are sensitive to the variation of austenite grain size, which is influenced by the welding thermal cycle. The coarsening of austenite grain is severe at the conditions of high welding heat input and peak temperature, and may deteriorate the strength and toughness [10]. When the heat input or peak temperature is low, conversely, the decrease in the grain size would diminish the propensity to ε-martensitic transformation and twinning.
activity, affecting the strain-hardening rate and ductility \[11\]. In addition, the strong carbide forming elements, such as Ti, V, Mo, Nb, are usually added to improve the strength and toughness of modified austenitic manganese steels \[11–15\]. The intergranular carbides are easy to form at reheating temperature range of 250 °C–850 °C and continuous cooling stage with slow cooling rate during welding \[16, 17\]. Such embrittling carbides precipitated in austenitic grain boundaries in the HAZ of the welded joint would significantly deteriorate the mechanical properties, especially toughness and strength \[18, 19\]. Our previous study showed that the morphologies of carbide was significantly affected by welding peak temperature when the peak temperature is between 750 °C and 1050 °C with heat input of 6 kJ cm\(^{-1}\) \[20\]. Unfortunately, the effect of welding heat input on precipitation behavior of HAZ for MAMMS was unclear, and few attempts have been carried out to investigate the influence of VC particles on mechanical properties in the HAZs of MAMMS after welding. Therefore, in the present work, the HAZs with different heat input for MAMMS were simulated with the peak temperature ranging from 750 °C to 1050 °C, where the precipitation was sensitive to the welding thermal cycle. The effects of peak temperature and heat input on the microstructural evolution and mechanical properties of HAZs were detailedly discussed.

2. Experimental procedures

2.1. Experimental materials and simulated HAZs procedures

The experimental material is the solution treated MAMMS, which was solution-annealed at approximately 1050 °C–1100 °C followed by water quenching. The chemical composition of this MAMMS is presented in table 1. Figure 1 shows the microstructure of as-received MAMMS that consisted of equiaxed austenite with twins. The average size of austenite was about 50 μm. Moreover, undissolved carbides are not observed in this condition.

The Gleeble 3500 thermo-mechanical simulator was used for the welding HAZ simulation. The simulated specimens were machined into the size of 11.0 mm × 11.0 mm × 70.0 mm. It was reported that the steels were usually welded using low heat input less than 25.6 kJ cm\(^{-1}\) to avoid the precipitation of large-sized carbides and grain coarsening \[21\]. In this study, two heat inputs were performed, 17 kJ·cm\(^{-1}\) and 25 kJ·cm\(^{-1}\) respectively, to evaluate the effects of heat input. The specimens were heated to the peak temperature of 750 °C, 850 °C and 1050 °C, respectively, in which large amount of VC particles would precipitate and coarsen. The heating rate was 200 °C·s\(^{-1}\) to simulate the high heating rate during welding. After holding for 1 s, the specimens were continuously cooled to bellow 150 °C. Thermal histories of welding thermal cycle in different conditions were

![Figure 1. Microstructure of the MAMMS base metal.](image-url)
given by the 3-dimensional Rosenthal mathematical model embed in Gleeble 3500 thermo-mechanical simulator, presented in figure 2. The cooling time from 800 °C to 500 °C (t8/5) was calculated as 11 s and 17 s corresponding to input of 17 25 kJ cm⁻¹ and 25 kJ cm⁻¹, respectively.

2.2. Microstructure observation
The specimens used for microstructural observation were cut from the midpoint of thermal simulated specimens. The microstructure of the simulated HAZs was characterized using a scanning electron microscope (Hitachi s4800 SEM) and a transmission electron microscope (PHILIPS CM200 TEM), which is equipped with an ultra-thin window Oxford energy dispersive spectrometer (EDS). The SEM specimens were first polished using standard metallographic techniques and then etched with 4% alcohol nitric acid and 4% hydrochloric acid alcohol. For TEM observation, the specimens were mechanically polished to 100 μm in thickness, punched to disks with diameter of 3.0 mm, and electropolished using a solution of 95 vol% alcohol and 5 vol% perchloric acid (HClO₄) at temperatures ranging from −25 °C to −29 °C.

2.3. Mechanical properties test
Tensile tests were performed to determine the changes in the mechanical properties of the simulated HAZs. However, the tensile testing of normal size samples would not ensure failure at the region of interest and so might not represent the strength of the simulated HAZs. To avoid these problems, notch tensile specimen was used for tensile testing [22, 23]. Dimensions of notch tensile specimen of the simulated HAZs is illustrated in figure 3. The notch tensile tests were carried out at room temperature with a loading rate of 0.2 mm-min⁻¹ using high-precision micro-force testing machine with a 2.0 kN load capacity (Instron 5848 Microtester). Three specimens were tested for each single group simulated HAZ.

The Charpy V-notch specimens were prepared according to the ASTM E 23-07a standard, whose depth, notch depth, and width are 5.0, 1.0 and 5.0 mm, respectively. The Charpy V-notch impact test of the simulated HAZs was tested at 25 °C.
3. Results

3.1. Microstructure of simulated HAZs for MAMMS

Figure 4 presents the microstructure of simulated HAZs of MAMMS with heat input of 17 kJ cm\(^{-1}\) and 25 kJ cm\(^{-1}\). Figures 3(a)–(b) and 3(c)–(d) indicated that the microstructure of simulated HAZs consisted of austenite and precipitation at grain boundary when the peak temperature was 750\(^\circ\)C–850\(^\circ\)C. It was seen from figure 4 that the morphology of intergranular cementite was significantly influenced by welding peak temperature. When the peak temperature increased to 1050\(^\circ\)C, the intergranular precipitations were not observed. To determine the type and morphology of those precipitates, TEM was used to observed the HAZs with heat input of 17 kJ cm\(^{-1}\), shown in figure 5. The selected area electron diffraction (SAED) patterns obtained from the intergranular precipitates showed that they were cementite (figure 5(d)). EDS results of the precipitates showed that Fe atoms existed in the cementite was replaced in partly by Cr, Mn atoms (figure 5(e)).

According to the previous studies [16, 17, 24], the carbon atoms segregated preferentially at the grain boundaries of the

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Figure 4. SEM micrographs of simulated HAZs with peak temperature of (a) 750\(^\circ\)C at 17 kJ cm\(^{-1}\), (b) 850\(^\circ\)C at 17 kJ cm\(^{-1}\), (c) 1050\(^\circ\)C at 17 kJ cm\(^{-1}\), (d) 750\(^\circ\)C at 25 kJ cm\(^{-1}\), (e) 850\(^\circ\)C at 25 kJ cm\(^{-1}\) and (f) 1050\(^\circ\)C at 25 kJ cm\(^{-1}\).

Figure 5. TEM micrographs of cementite at austenite grain boundaries in simulated HAZ with peak temperature of (a) 750\(^\circ\)C, (b) 850\(^\circ\)C and (c) 1050\(^\circ\)C for the heat input of 17 kJ cm\(^{-1}\), (d) selected area electron diffraction patterns of cementite and (e) EDS analysis of cementite.
austenite at temperatures ranging from 250 °C to 850 °C, which was prone to the formation of intergranular cementite. At the peak temperature of 750 °C, the cementite exhibited as vermicular with the size of ∼80 nm in width and ∼260 nm in length. When the peak temperature increased to 850 °C, it grew to the stringer particles distributed discontinuously along austenite grain boundaries. In this case, the cementite almost covered the whole austenite boundaries in coherent shapes (figure 5(b)). However, the cementite disappeared at the peak temperature of 1050 °C (figure 5(c)). As listed in table 2, the austenite grain size of HAZs increased slightly to ∼61 μm compared to that of base metal (49.6 μm). However, the heat input and peak temperature had no significant influence on austenite grain size of the simulated HAZs, which would be related to the pinning of the carbides.

Figure 6 shows the TEM High magnification micrographs of simulated HAZs at different heat input in the peak temperature ranging from 750 °C to 1050 °C. A considerable number of nanoscale precipitates were observed at austenitic matrix when the peak temperature was 750 °C and 850 °C. The particle size increased with the increase in peak temperature. When the peak temperature increased to 1050 °C from 750 °C, the precipitates coarsened to ∼30 nm from ∼20 nm. The density of the particles also decreased gradually at 1050 °C. Figure 7(a) shows the SAED analysis of the precipitate (particle A in figure 6(f)) and the austenite matrix. It was confirmed that these dispersed particles were VC with the face-centered cubic crystal structure. Moreover, figure 6(a) also indicated that the VC had an orientation relationship [(110)VC]//[111]austenite, (111)VC//[001]austenite with respect to the austenite matrix. Figure 7(b) shows the chemical compositions of particle A, which identified the VC particle. On the other hand, the effect of heat input on particles size was related to the peak temperature. When the heat

| Heat input, kJ cm⁻¹ | Peak temperature of thermal cycle, °C | Average diameter of austenite grain size, μm | Average diameter of VC particles, nm |
|----------------------|--------------------------------------|-----------------------------------------------|-------------------------------------|
| 17                   | 750                                  | 57.4                                          | 19.0                                 |
|                      | 850                                  | 58.1                                          | 23.8                                 |
|                      | 1050                                 | 61.3                                          | 27.0                                 |
| 25                   | 750                                  | 58.7                                          | 19.6                                 |
|                      | 850                                  | 58.9                                          | 28.4                                 |
|                      | 1050                                 | 61.8                                          | 31.6                                 |
| Base metal           | /                                    | 49.6                                          | Without                              |

Figure 6. TEM micrographs of simulated HAZs at heat input of 17 kJ cm⁻¹ with of (a) 750 °C, (b) 850 °C and (c) 1050 °C, and at heat input of 25 kJ cm⁻¹ with of (d) 750 °C, (e) 850 °C and (f) 1050 °C.
input increased from 17 kJ cm\(^{-1}\) to 25 kJ cm\(^{-1}\), the particle size did not change obviously at low peak temperature of 750 °C, but increased from 27.0 nm to 31.6 nm at high of 1050 °C.

### 3.2. Mechanical properties of simulated HAZs for MAMMS

The tensile properties of simulated HAZs were measured by notch tensile tests. The values of yield strength and elongation were unreliable due to the small effective testing zone of simulated HAZ (less than 2 mm in width) in notch tensile specimens (figure 3). However, the tensile strength could be obtained. Figure 8(a) presents the variation of tensile strength in the simulated HAZs. It was seen that the tensile strength increased when the peak temperature increased from 750 °C to 850 °C, but it sharply decreased at the high peak temperature of 1050 °C. Overall, the tensile strength of the simulated HAZ decreased with the increase in heat input. It was noted that the effect of heat input on tensile strength was highly related to peak temperature, where the decrement of strength with heat input at low peak temperature was more remarkable than that at high peak temperature. Therefore, the tensile properties of simulated HAZs were governed by the combined effects of welding peak temperature and heat input. Presented in figure 8(a), the tensile strength with peak temperature of 750 °C was higher than base metal when the heat input was lower than 17 kJ cm\(^{-1}\). However, the tensile strength of all HAZs was higher than base metal when the peak temperature was 850 °C, and it was lower than base metal when the heat input was 25 kJ cm\(^{-1}\) at peak temperature of 850 °C or lower than 17 kJ cm\(^{-1}\) at peak temperature of 1050 °C.

The impact absorbed energy of the base metal was 30.5 J at 25 °C in the present study. The effects of the heat temperature and peak temperature on the impact toughness of the simulated HAZs are shown in figure 8(b). It can be observed that the impact toughness of simulated HAZs was lower than base metal and obviously decreased with increase in heat input. Compared to peak temperature of 750 °C and 1050 °C, the toughness at 850 °C was lowest.
4. Discussion

The mechanical properties of MAMMS were attributed to chemical compositions, grain size and second phase precipitation. In this study, the HAZs of MAMMS had the same chemical compositions, and their grain size did not change distinctly in spite of the change of heat input and peak temperature. Therefore, the second phase precipitation, including cementite and VC particle, had significantly effects on mechanical properties.

4.1. Effect of cementite on mechanical properties in simulated HAZs

Figures 4 and 5 showed that there were a lot of cementite located along the austenite grain boundaries, which would seriously worsen the toughness. The brittle cementite can change the nature of the grain boundary or its adjacent region, which is weaker than the grain interior leading to the loss of ductility [25]. This will result in strain concentrations and, consequently, grain boundary decohesion during deformation. Therefore, the toughness loss will become more serious when cementite precipitated at the grain boundaries [18, 19]. In addition, the cementite at the austenite grain boundaries were initial points for microvoid, indicated by arrows (figure 3), which was consistent with previous research [26].

Furthermore, deformation twinning was the major deformation mechanism involved in most of the austenite manganese steels [26–28]. When the deformation twinning was activated during the hardening stage, the interactions between the twins and the intergranular cementite produced local shear stress concentrations at the grain boundaries. With further deformation, the twinning did not result in a further hardening regime but the activation of a damage mechanism triggered by twin-intergranular carbides interactions, which caused the formation of large amount of cracks [19]. Therefore, the HAZs toughness of MAMMS decreased due to the formation intergranular cementite, especially when there was stringer cementite distributed discontinuously along austenite grain boundary at peak temperature of 850 °C. Fortunately, the toughness increased when the grain boundary cementite disappeared at 1050 °C. Although the cementite deteriorated the toughness dramatically, it effected less on the strength, which agreed well with the results obtained by Lee et al [18]. This was because the size of cementite was in nanometer level and its amount was not large.

4.2. Coarsening behavior of VC particle and its effect on mechanical properties in simulated HAZs

Because 0.15% V and 0.98% C were primarily in solid solution state in base metal, VC particles would precipitate from austenite during reheating and cooling due to the supersaturation. For welding thermal cycle, the precipitation of VC particles during heating stage was ignored due to the high heating rate (200 °C·s⁻¹).

According to the solid solubility product of [V][C] in austenite [29]:

\[
\log \left( [V] \cdot [C] \right)_0 = 6.72 - \frac{9500}{T} \]

the fully solution temperature of experimental MAMMS was estimated to be 1225 K (952 °C). Thus, the VC particles would precipitate from austenite, accompanied by the coarsening, when cooled to below 950 °C during welding.

Our previous study confirmed that the average size of VC particles was 14.4 nm, 15.6 nm and 18.2 nm when the peak temperature was 750 °C, 850 °C and 1050 °C with heat input of 9 kJ cm⁻¹, respectively [20]. In this
study, it was found that the VC particle size increased with the increase in heat input. When the heat input increased to 25 kJ cm\(^{-1}\), the average size of VC particles increased to 19.6 nm, 28.4 nm and 31.6 nm at peak temperature of 750 °C, 850 °C and 1050 °C, respectively, listed in table 2.

For MAMMS, the VC precipitation coarsening primarily depended on the volume diffusion of vanadium in the austenite matrix [8, 30]. Based on the mechanism of vanadium volume diffusion, the coarsening kinetics of the VC precipitates during welding cooling stage can be simplified as [20]:

\[
r^3 - r_p^3 = A(f(T_p), t_{8/5})
\]

(1a)

where \( r \) is the average particle radius after welding cooling, \( r_p \) is the average particle radius at welding peak temperature, \( m \); \( A \) is the particle growth rate factor, \( m^3 \cdot K \cdot s^{-1} \); \( f(T_p, t_{8/5}) \) is the function of welding peak temperature \( (T_p) \) and cooling time from 800 °C to 500 °C during welding cooling stage \( (t_{8/5}) \), \( s \cdot K^{-1} \), which can be expressed as:

\[
f(T_p, t_{8/5}) = \frac{R\theta t_{8/5}}{Q_v(T_p - T_0)} \exp\left(-\frac{Q_v}{RT_p}\right)
\]

(2)

where \( R \) is the gas constant, 8.314 J (mol·K)\(^{-1}\); \( T_0 \) is the initial temperature and considered to be 298 K in this study; \( Q_v \) is the volume diffusion activation energy of vanadium in austenite, 264000 J mol\(^{-1}\) [31]; \( \theta \) is given by [32]:

\[
\frac{1}{\theta} = \frac{1}{773 - T_0} - \frac{1}{1073 - T_0}
\]

(3)

Here, the precipitation and coarsening of VC particles during welding heating stage was ignored because the precipitation at base metal was not observed and the welding heating rate was very high. Under this condition, the particle radius at welding peak temperature \( r_p \) can be ignored. Therefore, when the peak temperature \( T_p \) was given, the dependence of the average radius of the VC particles on \( t_{8/5} \) follows an approximately linear relationship, where the VC particles radius was proportional to \( f(T_p, t_{8/5}) \) and \( A \) is the slope of \( r^3 - f(t_{8/5}) \) plot.

Figure 9 presents the plot of \( r^3 - f(t_{8/5}) \), which clarified the linear relationship between \( r^3 \) and \( f(t_{8/5}) \). The linear relationship between \( r^3 \) and \( f(t_{8/5}) \) indicated that the coarsening kinetics of VC precipitate was in accordance with equation (1) obtained by the mechanism of vanadium volume diffusion, which implied the growth of VC particles was controlled by the volume diffusion of vanadium in this study. The diffusion rate of interstitial carbon atoms in austenite is higher than that of vanadium atoms in austenite. Therefore, the growth of VC was mainly restricted by the diffusion of vanadium elements. It was considered that the coarsening of MC carbides was controlled by the volume diffusion of solute M elements [33, 34]. The A of peak temperature of 750 °C, 850 °C and 1050 °C was 2.15 \( \times \) \( 10^{-7} \) \( m^3 \cdot K \cdot s^{-1} \), 6.70 \( \times \) \( 10^{-8} \) \( m^3 \cdot K \cdot s^{-1} \) and 1.57 \( \times \) \( 10^{-9} \) \( m^3 \cdot K \cdot s^{-1} \), respectively.

According to the coarsening kinetics of precipitation:

\[
r^3 - r_0^3 = Kt
\]

(4)

where \( r_0 \) is the average particle radius at the onset of coarsening and \( K \) the temperature-dependent growth rate of VC particles, \( m^3 \cdot s^{-1} \), the equations (1) and (2) can be written as:

\[
r^3 = K(T_p)t_{8/5}
\]

(5)

\[
K(T_p) = A \cdot \frac{R\theta}{Q_v(T_p - T_0)} \exp\left(-\frac{Q_v}{RT_p}\right)
\]

(6)

Substituting the values of \( A, R, \theta, Q_v, T_p \) and \( T_0 \) into equation (6), the value of \( K(750 \, ^°\text{C}), K(850 \, ^°\text{C}) \) and \( K(1050 \, ^°\text{C}) \) was calculated as 4.02 \( \times \) \( 10^{-25} \) \( m^3 \cdot s^{-1} \), 1.73 \( \times \) \( 10^{-24} \) \( m^3 \cdot s^{-1} \) and 2.3 \( \times \) \( 10^{-24} \) \( m^3 \cdot s^{-1} \), respectively. Therefore, it was concluded that the peak temperature significantly influenced the coarsening rate during the welding cooling stage. The \( K(850 \, ^°\text{C}) \) was \( \sim 4 \) times as high as \( K(750 \, ^°\text{C}) \). This indicated that the growth of VC particles was sharply promoted when the peak temperature increased form 750 °C to 850 °C. However, the \( K(1050 \, ^°\text{C}) \) only increased to \( \sim 1.3 \) times as high as \( K(850 \, ^°\text{C}) \), implying the reduction in increment of growth rate at high peak temperature. Therefore, although the increase in heat input increased the size of VC particles, the peak temperature also significantly influenced the VC particle coarsening behavior. The VC particle coarsening depended on the interaction effect of heat input and peak temperature. Based on this point, the variation tendency of tensile strength with welding heat input and peak temperature (shown in figure 7(a)) would be explained. The tensile strength of HAZ was sensitive to peak temperature at low heat input of 9 kJ cm\(^{-1}\) that it decreased 76 MPa when the peak temperature increased from 850 °C to 1050 °C. However, the decrement of tensile strength was only 26 MPa when the peak temperature increased from 850 °C to 1050 °C under high heat input of 25 kJ cm\(^{-1}\) condition, indicating the low sensibility of tensile strength to peak temperature. Similarly, the sensibility of tensile strength to heat input at high peak temperature was higher than that at low temperature,
presented in figure 7(a). As mentioned above, the VC precipitation played an important role in the change of strength of HAZs, because there was no significant difference in grain size of HAZ and the strength was not distinctly influenced by cementite. Therefore, the reduction of tensile strength was mainly depended on the coarsening of VC particles, which was affected by combined effects of the two factors (heat input and peak temperature), rather than a single factor.

5. Conclusion

The effects of welding heat input and peak temperature on microstructural evolution and mechanical properties of the simulated HAZs for MAMMS were investigated. The following conclusions were attained through this study:

(1) The microstructure of simulated HAZs consisted of intergranular cementite and austenite with large number of VC particle when the peak temperature was $750 \sim 850 \, ^\circ \text{C}$ at heat input of $17 \sim 25 \, \text{kJ cm}^{-1}$. With increase in peak temperature and heat input, the austenite size did not change significantly, but the intergranular cementite and VC particle coarsened. When the peak temperature increased to $1050 \, ^\circ \text{C}$, the intergranular cementite disappeared.

(2) The tensile strength of simulated HAZs was highest at peak temperature of $850 \, ^\circ \text{C}$ and decreased with the increase in heat input, but the variation tendency of tensile strength depended on the combined effects of heat input and peak temperature, rather than a single factor. The toughness of the simulated HAZs was lower than base metal, which was related with the growth of austenite grain and precipitation of intergranular cementite and VC particle.

(3) The sensibility of tensile strength to heat input at high peak temperature was higher than that at low temperature. This variation tendency was highly related with the coarsening of VC particle, which depended on the interaction effect of heat input and peak temperature. The growth rate of VC particle was $4.02 \times 10^{-25} \, \text{m}^3 \, \text{s}^{-1}$ with peak temperature of $750 \, ^\circ \text{C}$ and sharply increased to $1.73 \times 10^{-24} \, \text{m}^3 \, \text{s}^{-1}$ with peak temperature of $850 \, ^\circ \text{C}$. However, when the peak temperature increased to $1050 \, ^\circ \text{C}$, it only slightly increased to $2.3 \times 10^{-24} \, \text{m}^3 \, \text{s}^{-1}$.

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Data availability statement

All data that support the findings of this study are included within the article (and any supplementary files).

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