1. Introduction

Advanced High Strength Steels (AHSS) are most popular car body material in the current automotive industry, since they can reduce the automobile body weight without compromising safety.1,2) Moreover, the current tendency of AHSS have been developed into low density, high strength, and high ductility. In recent years, a new generation of high strength and high ductility Medium-Mn steel has been successfully developed due to excellent properties with regard to the strength-ductility balance.3,4) Multi-phase, meta-stability, and multi-scale microstructure is obtained by austenite reverted transformation (ART-annealing) or Quenching and Partitioning (Q & P).5–7) High strength of the steel is obtained through phase transformation strengthening and grain refinement strengthening.4,8) High ductility of Medium-Mn steel is achieved by the effects of transformation induced plasticity (TRIP) of metastable austenite during tensile deformation.9,10)

Laser welding is a critical technology in the automobile manufacturing industry due to its flexibility and high energy density.1,11) Laser tailor-welded blanks (TWBs) are more and more used widely in the automotive industry due to high-efficiency welding.12,13) The applications of a new steel in the automobile body have been influenced by the welding quality and efficiency.

At present, little research focuses on the laser welding of Medium-Mn steel, especially on the microstructure and mechanical properties of HAZ. In general, the combination of high strength and high toughness can be deteriorated by welding thermal cycles due to the formation of local brittle zones in the welded joint.14) Lun et al.15) obtained high quality welded joint according to performing the medium-Mn steel (0.15C-10Mn-Fe) laser welding, but had not been studied the property of HAZ. Di et al.16) reported that the size and morphology of cementite and VC precipitates were influenced by the peak temperature (PT) at the simulated arc welded HAZ of the medium-Mn micro-alloy steel, and...
the microstructure of the simulated HAZ was composed of equiaxed austenite with some twins. Wei C et al.\textsuperscript{17} reported that the mechanical property of laser welded HAZ of dual phase steel was controlled with the packet size of lath martensite and the volume fractions of martensite. Thus, the change of PT exerts a significant influence on the microstructure, which results in the difference of mechanical properties in HAZ. Previous studies\textsuperscript{18,19} have shown that HAZ of the Medium-Mn steel laser welded joint can be divided into four distinct regions according to different peak temperature: (1) CGHAZ, where the PT is much higher than the $A_c^1$; (2) FGHAZ, where the PT is slightly higher than the $A_c^1$; (3) ICHAZ, where the PT is between $A_c^1$ and $A_c^3$; (4) SCHAZ, where the PT is below $A_c^1$.

Microstructure and mechanical properties of HAZ was studied by using thermal simulation experiment. A regional thermal cycle simulation experiment was conducted on a thermo-mechanical simulator. In this study, the effect of PT on the microstructure and mechanical properties of HAZ was investigated. Furthermore, the microstructure results of simulated HAZ was compared with that of actual HAZ.

2. Experimental Procedures

2.1. Experimental Materials

The Medium-Mn steel was melted in a vacuum induction furnace and cast into ingot, then forged into a square slab with a thickness of 30 mm. Thereafter the slabs were hot rolled to 4 mm thin sheets in the temperature range of 1 200–900°C. After 6 h of soft annealing at 650°C, the Medium-Mn steel sheets were cold rolled to thickness of 2 mm. Finally, the cold rolled sheets were processed with annealing heat treatment at 650°C for 15 min. The chemical composition and data sheets for the mechanical properties of medium-Mn steel is listed in Table 1 and 2, respectively.

2.2. Laser Welding and Welding Simulation

Cold rolled Medium-Mn steel sheets were butt-welded by IPG YLS-2000 fiber laser. Laser welding parameters is shown in Table 3. High purity argon gas was used as shielding gas.

Thermal simulated samples of medium-Mn steel were cut into the dimension of 60 mm × 10 mm × 2 mm, which were simulated by using a thermo-mechanical simulator Gleeble-1500D. Thermal simulated samples were heated to different PT at a heating rate of 1 000°C/s with the over shoot of ~50°C and held for 0.5 s, then quenched with water. K-type thermocouple was welded at the middle of the sample surface, and the heating and cooling curves were recorded during the whole thermal cycle. The sample was coated by lubricant to avoid oxidation at high temperatures at the samples surface. Different values of the PT were set based on the $A_c^1$ (630°C) and $A_c^3$ (780°C) temperature of medium-Mn steel.\textsuperscript{18,20} Besides, the $A_c^1$ and $A_c^3$ temperature were 670°C and 813°C respectively when the heating rate reached 1 000°C/s, which was calculated by the JMat-Pro software. Therefore, the PT of simulation HAZ selected 1 350°C, 900°C, 700°C, and 500°C to simulate the CGHAZ, FGHAZ, ICHAZ, and SCHAZ, respectively. The welding thermal cycle curve of HAZ is shown in Fig. 1.

2.3. Microstructural Characterization and Mechanical Properties Test

Thermal simulated samples were mechanical ground and polished, then etched with 4% nital. The microstructure of different HAZ were examined by FEI Quanta 650 electron back scattered diffraction (EBSD) using a scanning electron microscope (SEM) with a step of 0.01 μm and the area of 50 μm × 50 μm. The precipitated phase was examined by X-rays diffraction (XRD) based on the integrated intensities of the (200)$_m$, (211)$_m$, (200)$_s$, (220)$_s$, and (311)$_s$ diffraction peaks by the following Eq. (1).\textsuperscript{21,22} The microhardness and tensile test were performed on the FM-300 microhardness tester with a load of 200 g and the mechanical testing system

![Fig. 1. Thermal cycle curves of simulated HAZ. (Online version in color).](image)

| Table 1. Chemical composition of the Medium-Mn steel (wt.%). |
|-----------------|---|---|---|---|---|---|
| C | Mn | Si | S | P | N | Fe |
| 0.1 | 4.86 | 0.01 | 0.002 | 0.008 | 0.003 | Balance |

| Table 2. Mechanical properties of the Medium-Mn steel. |
|-----------------|---|---|---|---|
| YS (MPa) | UTS (MPa) | Elongation (%) | Microhardness (HV 200 g) |
| 605 | 756 | 38.6 | 225 |

| Table 3. Laser welding parameters. |
|-----------------|---|---|---|---|---|---|
| Laser power (kW) | Welding speed (m/min) | Focal length (mm) | Focus distance (mm) | Shielding gas flow (L/min) | Heat input (J/mm) | Fiber core diameter (mm) | Collimation focal Length (mm) |
| 2 | 4.8 | 300 | 0 | 15 | 25 | 0.1 | 150 |
(MTS) with strain rate of $10^{-3}$ s$^{-1}$ at room temperature, respectively. The tensile test sample has the width of 3 mm and the gauge length of 12 mm.

$$V_i = \frac{1}{1 + G(I_a / I_f)} \quad \text{........................... (1)}$$

3. Results and Discussion

3.1. Microstructure of the Base Metal

SEM and TEM images of the BM are shown in Figs. 2(a) and 2(b), respectively. Base metal is composed of ultrafine-grained ferrite and austenite, showing equiaxed grain. The austenite reverted transformation during intercritical annealing resulted in equiaxed ferrite and austenite grains.\(^7\) Research\(^2\) indicated that the solubility of C/Mn element in austenite is higher than in ferrite grains, and thus the image contrast of austenite is higher than that of ferrite grains, as shown in Fig. 2(b). Figure 3 shows the EBSD maps of BM. In the GB misorientation map, the black solid lines represent the high angle grain boundaries ($>15^\circ$) and the red solid lines represent the low angle grain boundaries ($2^\circ$–$15^\circ$). The average effective grain diameter is 1.56 $\mu$m. The proportion of high angle grain boundaries (~67%) is higher than that...
of low grain boundaries (~32%). The volume fraction of austenite in the BM obtained by XRD is about 18.26%, as is shown in Fig. 4.

The composition of steel exerts remarkable influence on the tensile strength and yield strength. The ultimate tensile strength (UTS) of traditional high strength low alloy steels (HSLA) for the automotive industry is less than 600 MPa. However, the tensile strength (756 MPa) and yield strength (605 MPa) of this steel is higher than that of traditional HSLA steels, as shown in Table 4. High tensile strength of the BM has been attributed to multiple strengthening mechanisms, such as solid solution strengthening and fine-grained strengthening. Moreover, the martensite that was transformed from metastable austenite during tensile test is positive for strength and elongation through the TRIP effect. Medium-Mn steel exhibits excellent elongation of 38.6%, which is related to fine grain size and high fraction of high angle grain boundaries (67%).

3.2. Microstructure Analysis of Simulated HAZ

3.2.1. Microstructural Evolution of Simulated HAZ

Macrostructure of the actual welded joint is shown in Fig. 5. The GB misorientation proportion of simulated HAZ by EBSD analysis is shown in Table 4.

![Fig. 4. Austenite phase intensity profiles measured by XRD of BM and HAZ.](Online version in color.)

![Table 4. The GB misorientation proportion of simulated HAZ by EBSD analysis.](GB misorientation proportion)

<table>
<thead>
<tr>
<th>GB misorientation proportion</th>
<th>GB misorientation proportion</th>
</tr>
</thead>
<tbody>
<tr>
<td>CGHAZ</td>
<td>FGHAZ</td>
</tr>
<tr>
<td>θ &lt; 15°</td>
<td>0.50</td>
</tr>
<tr>
<td>15° &lt; θ &lt; 50°</td>
<td>0.12</td>
</tr>
<tr>
<td>θ &gt; 50°</td>
<td>0.38</td>
</tr>
</tbody>
</table>

![Fig. 5. Macrostructure of the laser welded joint.](Macrostructure of laser welded joint.)

![Fig. 6. Microstructure of experimental HAZ and simulated HAZ at the PT of (a) 1350°C, (b) 900°C, (c) 700°C, (d) 500°C.](Microstructure of experimental HAZ and simulated HAZ)
The magnified images of thermal simulated and actual HAZ are shown in Fig. 6, and the experimental images are corresponding to the Figs. 5(a), 5(b), 5(c), 5(d). The microstructure of simulated HAZ exhibits similar characters compared to that of actual HAZ. The microstructure with PT of 1350°C (corresponding to CGHAZ) is martensite, as shown in Fig. 6(a) and the prior austenitic grain size of which is slightly larger than that of actual weld CGHAZ. The microstructure of simulated HAZ with PT of 900°C (corresponding to FGHAZ) has been changed into fine martensite and a small retained austenite, as shown in Fig. 6(b). Figure 6(c) shows the ICHAZ image subjected to PT of 700°C, which consists of UFG austenite, ferrite, and the precipitated phase. In this area, the difference of microstructure between thermal simulation and actual experiment is the content of the precipitated phase, and the former shows more precipitation. This is probably because actual ICHAZ is much narrower than thermally simulated ICHAZ.

The microstructure of simulated HAZ with PT of 500°C is similar to that of actual SCHAZ, as shown in Fig. 6(d). On the whole, the results of thermal simulation for HAZ have been proved reliable by comparison of microstructure characteristics.

The TEM micrographs of simulated HAZ for different PT are presented in Fig. 7. When the PT at 1350°C, the substructure of grain was martensite lath and austenite film, as shown in Fig. 7(a). When the PT reached 900°C, the substructure of grain had been changed into high density dislocations and martensite packet, as shown in Fig. 7(b). When the PT decreased to 700°C, however, the microstructure had been turned into ferrite and austenite. The spherical precipitated phase formed inside the austenite grain, as shown in Fig. 7(c). The microstructure of HAZ with PT of 500°C is similar to that of the BM containing low density dislocations, as shown in Fig. 7(d). The precipitated particles were further identified as \(M_2\text{C}\) carbides through TEM selected

![Fig. 7](image_url)
area electron diffraction (SAED) and energy dispersive X-ray analysis (EDAX) spectra, as shown in Figs. 7(e), 7(f).

Research\(^{25}\) indicated that the heating and cooling rates are very fast during laser welding, which can reach \(3.27 \times 10^3 \, ^\circ C/s\). Fast heating rate results in enhanced temperature of \(A_c_3\) and \(A_c_1\) of the steel. The simulated CGHAZ and FGHAZ exceeded the austenitizing temperature (735\(^\circ\)C) during the heating process. The microstructure has been changed into austenite at the heating process. On the one hand, the microstructure is promoted to transform into martensite during the rapid cooling process. On the other hand, the martensite-start temperature \(M_s\) is reduced due to the increasing content of Mn element,\(^{26}\) leading to the easy transformation of martensite. The martensitic lath structure is clear and small for the simulated CGHAZ. However, microtwins are formed in martensite, and the lath structure is fuzzy when the PT decreased to 900\(^\circ\)C. The austenitic grains have hardly grown at the simulated FGHAZ, thereafter, martensite transformation has occurred during the rapid cooling process. Many studies\(^{27,28}\) indicated that twins are formed in plate martensite at medium and high carbon steels. Substructure with microtwins and high-density dislocations formed in prior austenite could have resulted from extremely non-uniform diffusion behavior of elements during the heating process and strong quenching stress during the rapid cooling process. Strong \(M_3C\) precipitation formed in austenite grains, showing a diameter of \(~80\, nm\).

The driving force of austenite transformed into ferrite and carbides is higher than that of austenite transformed into ferrite and retained austenite through the calculation result, as shown in Fig. 8. Research\(^{29,30}\) has shown that the nanoscale

![Fig. 8. Driving force for austenite transformation of the BM as predicted by JMat-Pro calculation. (Online version in color.)](image)

![Fig. 9. Simulated HAZ GB maps obtained by EBSD analyses at the PT of (a) 1 350\(^\circ\)C, (b) 900\(^\circ\)C, (c) 700\(^\circ\)C, (d) 500\(^\circ\)C, (green color: retained austenite). (Online version in color.)](image)
Cementite could be precipitated in medium and low carbon steel via ultrafast cooling process. The prior austenite grain size slightly grew when the temperature reached 700°C and 500°C at extremely rapid heating rate. The size of austenite in 700°C and 500°C are small (~1.5 μm), which results in increasing thermostability for austenite, therefore, the martensite transformation has not been easily occurred during cooling process.

3.2.2. Grain Boundary Characterization of Simulated HAZ

The simulated HAZ GB maps examined by EBSD are presented in Fig. 9. The black lines represent the high angle grain boundaries (>15°) and the red lines represent the low angle grain boundaries (2°–15°). The green phase in the picture is austenite, and the white phase at ICHAZ (PT=700°C) and SCHAZ (PT=500°C) is ferrite, which is martensite and ferrite at CGHAZ (PT=1 350°C) and FGHAZ (PT=1 350°C). The average effective grain diameter of thermal simulated HAZ with different PT is shown in Fig. 10, and the average effective grain diameter is defined by the misorientation of the grain boundary (>15°). The grain boundary misorientation distribution results are shown in Table 4.

3.3. Mechanical Properties of Simulated HAZ

3.3.1. Microhardness of Simulated HAZ

The microhardness profile of different PT for thermal simulated HAZ is shown in Fig. 11(a), which is coincide well with the actual HAZ of welded joint, as shown in Fig. 11(b). The microhardness of simulated HAZ both PT=1 350°C and PT=900°C are well above 420 HV because of high content martensite (volume fraction~80%). Due to the reduction of martensite content, the microhardness of simulated HAZ both PT=700°C and PT=500°C decreased sharply.

3.3.2. Tensile Properties and Static Toughness

Figure 12(a) shows the engineering stress-strain curves of simulated HAZ, BM, and welded joint. The stress strain curves exhibit a big difference between each other. In the ICHAZ, SCHAZ, and BM, the curves display a distinct discontinuous yielding phenomenon, while the FGHAZ and CGHAZ exhibit continuous yielding. The UTS and YS increase with increasing PT for different HAZ, which is shown in Fig. 12(b). Static toughness is a capacity of material to absorb energy and plastically deform before fracturing, which is indicated by the area under the engineering stress strain curve. The static toughness formula is as Eq. (2).

\[
U = \int_0^\infty \sigma'd\varepsilon \quad \text{.................................(2)}
\]
where \(U\) represents the static toughness, \(\sigma'\) represents the flow stress, and \(\varepsilon_f\) represents the engineering strain at fracturing. The static toughness of HAZ for different PT and the BM is shown in Fig. 12(c). The static toughness of ICHAZ and SCHAZ is higher than that of BM, CGHAZ, and FGHAZ.

The presence of martensite in CGHAZ and FGHAZ is the reason for high UTS (1,464 MPa, 1,406 MPa), YS (1,097 MPa, 1,002 MPa), and low total elongation (TE) (11.4% and 19.1%). High hardness and strength are the substantive characteristics of martensite. A large amount of martensite formed in CGHAZ and FGHAZ during the rapid cooling process, therefore, it has higher UTS and YS. The difference in yielding behavior between ICHAZ, SCHAZ and FGHAZ, CGHAZ, is considered to be caused by different strain distributions between the different microstructures. ICHAZ and SCHAZ consist of ferrite with lower density of dislocations and austenite with higher density of dislocations, as shown in Figs. 7(c), 7(d). Soft ferrite grains with lower density of dislocations are easily deformed, while the austenite grains are difficult to be deformed due to the high density of dislocations and the solid solution strengthening of C/Mn element. This difference results in inhomogeneous deformation. In the CGHAZ and FGHAZ the microstructure consists of a large amount of martensite and little austenite. Furthermore, the austenite is simultaneous strained with the martensite at the early stage of plastic deformation due to hardening with high density of dislocations, which results in continuous yielding. Moreover, a serrated fluctuation appeared in the stress-strain curve of ICHAZ and SCHAZ, which is attributed to dynamic strain aging (DSA), resulting from dynamic interactions of mobile dislocations and C/Mn solute atoms. ICHAZ exhibits the best static toughness property with value of 343.4 MJ/m², while values of both FGHAZ and CGHAZ are lower than that of the BM due to their worse ductility despite high UTS. The ductility of SCHAZ, especially ICHAZ exceeded 40%, which is attributed to the TRIP effect of austenite during tensile test.

### 3.3.3. Tensile Fractography Analysis

The tensile fracture morphologies of simulated HAZ with different PT are shown in Fig. 13. The central fibrous region of fractured samples with PT of 1,350°C and 900°C exhibits lamellar tearing. When the PT decreased to 700°C and 500°C, the central interior region presents irregular and fibrous. The central region magnified images of all samples are listed in Figs. 13(b)–13(e).

At the PT of 1,350°C, the simulated HAZ showed lower ductility and toughness. This is mainly due to the coarser prior austenite grain size obtained at that temperature. As shown in Fig. 13(b), the fracture surface of the HAZ with PT of 1,350°C is characterized by shallow dimples, and local regions show the characteristics similar to brittle fracture, which is consistent with the mechanical properties. When the PT decreased to 900°C, the fracture surface is characteristic of shallow and slightly big dimples showing in Fig. 13(c). However, many dimples are observed in HAZ with PT of 700°C and 500°C fracture surfaces, and each
Dimple is one half of a microvoid that formed and then separated during the fracture process. The dimple characteristic is deep and small, showing the ductile fracture mechanism, and the diameters of dimples are similar to the grain size as shown in Figs. 13(d), 13(e). From the microstructure analyses presented above, the ferrite and austenite are easily deformed due to good ductility during tensile test, which indicates ductile fracture at the PT of 700°C and 500°C.

4. Conclusions

The microstructure evolution and mechanical properties of simulated HAZ of a Medium-Mn steel subjected to different PT were investigated, and the following conclusions could be drawn:

1. The microstructure of medium-Mn steel consists of UFG ferrite and austenite, with an average effective grain size of 1.56 μm and 67% high angle grain boundary, leading to an excellent elongation and UTS with values of 38.6% and 756 MPa, respectively.

2. At the PT of 1350°C, the microstructure corresponding to that of CGHAZ consists of martensite and austenite film. At the PT of 900°C, the microstructure corresponding to that of FGHAZ consists of fine martensite packet with high density dislocations. At the PT of 700°C, the microstructure corresponding to that of ICHAZ consists of...
of austenite, ferrite, and $M_7C$ carbides. When the PT reached to 500°C, the microstructure corresponding to that SCHAZ consists of ferrite, austenite.

(3) The UTS and YS of HAZ increase with enhancing PT due to the increasing of martensite volume fraction. The CGHAZ exhibits the highest tensile strength and the worst ductility due to the martensite. Due to the good balance of strength and ductility, the ICHAZ possess the best static toughness with a value of 326.2 MJ/m$^3$.

(4) The fracture surface exhibits a typical ductile fracture mechanism due to the austenite and ferrite microstructure in ICHAZ, SCHAZ and ductile character with bigger dimples in FGHAZ. However, the fracture surface of CGHAZ mainly exhibits ductile fracture and local regions show brittle characteristics.

Acknowledgements

This research was supported by the National Key R&D Program of China (No. 2016YFB1101100) and the International Science and Technology Cooperation Program of China (No. 2015DFA51460).

REFERENCES