Investigation on microstructure and impact toughness of simulated heat affected zone of high strength low alloy steels by laser-arc hybrid welding

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Abstract

The microstructure and impact toughness of the simulated heat affected zone (HAZ) of a high strength low alloy steels by laser-arc hybrid welding (LAHW) were investigated in this paper. Homogeneous simulated HAZ specimens with different grain sizes were prepared using the welding thermal simulating technique. Instrumented impact test was conducted to investigate the toughness of the simulated HAZ. Multi-scale sub-structure characterization of the simulated HAZ specimens was implemented by optical microscopy (OM), scanning electron microscopy (SEM), transmission electron microscopy (TEM), electron back-scattered diffraction (EBSD). The instrumented impact test found that the peak temperature \( T_{Mf} \) mainly affected the crack propagation process, the crack propagation energy \( E_p \) decreased as the increase of \( T_{Mf} \), while the crack initiation energy \( E_i \) barely changed, the decrease of crack stable energy \( E_o \) led to the change of \( E_p \). The multi-scale sub-structure characterization showed that the prior austenite grain size (PAGS), packet width, block width both increased with the rise of \( T_{Mf} \). The block width was similar to the facet size in the crack unstable propagation zone of the simulated HAZ specimens. The block was the microstructure unit controlling the crack propagation process of the LAHW simulated HAZ specimens.

1. Introduction

LAHW is a high-quality and efficient welding process, which has broad application prospect in thick plate structure welding [1–4]. LAHW has been used in the welding of low strength steels, such as E36 and S355 steels, and has a remarkable effect on improving welding efficiency, guaranteeing joint quality, controlling structural deformation [5]. However, the toughness fluctuation and local embrittlement exist in the LAHW HAZ of high strength steels [6–8]. The toughness influencing factors and embrittlement mechanism are not clear yet.

Previous study [9] found that coarse grained HAZ (CGHAZ) was the local brittle zone of the LAHW joints of high strength low alloy steel. The CGHAZ was comprised of lath martensite (LM) thanks to the fast cooling rate of the LAHW process. For steels with a carbon content less than 0.4%, LM is composed of a variety of sub-structures. A prior austenite grain can be divided into several packets which have the same habit plane, but different crystal orientation [10]. Each packet is further divided into several blocks with the same or similar orientation. A block is composed of several martensite laths [10]. The schematic of the sub-structures of the lath martensite is shown in figure 1. The microstructure unit that controls the LM toughness is still controversial. Some scholars thought that the toughness of lath martensite is controlled by the packet [11], while others hold that the block was the microstructure unit that controls the LM toughness [12, 13].

Previous study [9] preliminarily deduced that block was the microstructure unit controlling the LM toughness of CGHAZ. While the relationship between the multi-scale sub-structure and the toughness was indirectly reflected by the high angle grain boundaries (HABs) ratio, as it was impossible to quantify the
toughness of each micro-zones in the LAHW HAZ. In this paper, welding thermal simulating technique was used to obtain homogeneous specimens of different LAHW CGHAZ micro-zones. Multi-scale sub-structure characterization and instrumented impact test were used to study the relationship between the sub-structure and the instrumented impact energies. The microstructure unit controlling the LAHW HAZ toughness was further investigated.

2. Materials and methods

The chemical composition of the investigated high strength low alloy steel is Fe-0.12C-0.25Si-1.09Mn-0.58Cr-0.53Mo-0.28Cu-0.04V (wt pct). The mechanical properties are listed in Table 1, the mechanical properties were tested according to GB/T 228.1–2010 and GB/T 229–2007.

LAHW combines the advantages of laser welding and arc welding. Normally, a LAHW system consists of laser device, welding source, welding robot system, hybrid welding head, water chiller and welding fixture. The schematic diagram of laser-arc hybrid welding is shown in Figure 2. In addition to the conventional process parameters such as laser power, defocus distance, arc voltage, current, welding speed, wire feeding speed and so on, the laser-arc angle, laser-arc distance and the relative position of laser and arc are also important parameters.

Figure 2 shows the laser-leading mode, which means the laser head in the front and arc torch in the back. Former studies on the welding thermal cycle of the LAHW process found that the average heating rate was about 400 °C/S, the temperature above 1100 °C dwelling time was 0.8 S ~ 1.3 S, and the cooling time from 800 °C to 500 °C was 4 S ~ 6 S [9].

Simulated HAZ specimens were prepared by Gleeble-3800 thermal-mechanical simulating facility. The simulated thermal cycle parameters were set based on the above results. The heating rate was set as 400 °C/S, the peak temperature dwelling time was 1 S, t50/5 was 5 S. Different T_{Ae} (900 °C, 1100 °C, 1200 °C, 1300 °C) were selected to obtain specimens with different sub-structure grain sizes. Fine grained HAZ (FGHAZ) specimen (900 °C), which had the same microstructure with CGHAZ, was also studied as reference specimen. The welding

![Figure 1. Schematic of the sub-structures of the lath martensite.](image)

| Tensile strength (MPa) | Yield strength (MPa) | Reduction of area (%) | Absorbed energy (−40 °C) (J) |
|------------------------|----------------------|-----------------------|-----------------------------|
| 890                    | 833                  | 19                    | 188                         |
|                        |                      |                       | 177                         |
|                        |                      |                       | 153                         |

Table 1. The investigated steel mechanical properties.
thermal simulating cycle of 1300 °C specimen are shown in figure 3, the temperature setting curve basically coincides with the temperature measurement curve proves the accuracy of the thermal simulation technology.

The Vickers hardness of the simulated samples were measured with a force of 10 kgf (HV10) and 10 s dwell time according to the GB/T 2654–2008. Instrumented Charpy impact test (55 mm × 10 mm × 10 mm) was carried out at 233 K (−40 °C) with V-notches according to the GB/T 229–2007. The fracture surfaces of the Charpy impact specimens were observed by OM (Leica DM2500M), SEM (JEOL-JSM7200F).

Metallurgical specimens were etched with 4% nital solution, saturated picric acid to reveal the prior austenite grain boundaries, and observed by OM (Leica DM2500M), SEM (JEOL-JSM7200F) and TEM (FEI-Talos-F200). The EBSD characterization was carried out by a field emission SEM equipped with an EBSD detector (Oxford Symmetry). The EBSD samples were polished for 2 h by a VibroMetTM 2 vibratory polisher using the MasterMet silica polishing solution. The prior austenite grain size (PAGS), packet width, block width and lath width were measured by the linear intercept method according to GB/T 6394–2017.

Figure 2. Schematic diagram of laser-arc hybrid welding.

Figure 3. Welding thermal simulating cycles of 1300 °C specimen.
3. Results and discussions

3.1. Effects of peak temperature on the hardness and instrumented Charpy impact toughness

The hardness and instrumented impact test results of the LAHW simulated HAZ specimens are shown in table 2 and figure 4. The hardness increased as the $T_M$ increased. According to the load-deflection curve [9], the total impact energy $E_I$ is divided into crack initiation energy $E_i$ and crack propagation energy $E_p$. $E_p$ contains stable propagation energy $E_{a}$, unstable propagation energy $E_{b}$, and tearing energy $E_{c}$.

The total crack energy $E_I$ and crack propagation energy $E_p$ decreased as the increase of $T_M$, while the crack initiation energy $E_i$ barely changed. The change of $E_i$ was derived from the variation of $E_p$. $T_M$ mainly affected the crack propagation process. It was the decrease of crack stable energy $E_a$ led to the change of $E_p$, the change of the unstable energy $E_b$ and tearing energy $E_c$ were insignificant, especially for the CGHAZ specimens (1100 °C, 1200 °C, 1300 °C), as shown in table 2. The variation trend of hardness was in the opposite direction compared with the impact energies.

According to the morphology characteristics, the macro-fracture surface of the impact specimen can be divided into fibrous zone, radical zone and shear lip, which represent different fracture processes and impact energy. Under the impact load, the V-shaped notched specimen firstly elastic deformed until reaching the yield load, then plastic deformation occurred. When the load increases to the maximum, the plastic deformation had run throughout the whole notch section, the notch-root was in a three-way stress state, micro-cracks initiated inside the specimen at a certain distance from the notch. This process corresponded the crack initiation energy $E_i$. Then the micro-cracks propagated towards the width and thickness directions of the specimen, and the load began to decrease. When the crack had expanded to the whole width, a 'heel-shaped' fibrous zone formed. This process corresponded to the stable propagation energy $E_a$. As the crack increased to a critical size, the crack began to propagate rapidly, forming a radical zone, corresponding to the unstable propagation energy $E_b$. When the load further dropped to the termination load of the unstable propagation, the crack front entered the plane compressive stress stage, and the load decreased continuously to zero, forming the shear lip. This process corresponded to the tearing energy $E_c$.

Figure 5 are the micro-fracture surfaces at the crack stable propagation zones. As the $T_M$ increased from 900 °C to 1500 °C, the micro-fracture surfaces morphologies changed from large dimples, small dimples, small dimples with local cleavage facet, cleavage facet, which corresponded to the decrease of the stable propagation energy $E_a$ (18.4 J-9.0 J-6.8 J-0.9 J). The micro-fracture surface morphology characterization further proved the remarkable influence of $E_a$ on $E_p$.

![Figure 4. Effects of $T_M$ on the instrumented impact toughness and hardness.](image)

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| Specimen Number | $E_i$ (J) | $E_a$ (J) | $E_b$ (J) | $E_c$ (J) | $E_I$ (J) | $HV_{10}$ |
|----------------|----------|----------|----------|----------|----------|----------|
| 1300           | 27.9     | 0.9      | 2.5      | 6.7      | 10.1     | 38.0     | 436.5    |
| 1200           | 25.2     | 6.8      | 2.0      | 10.2     | 19.0     | 44.2     | 414.4    |
| 1100           | 25.0     | 9.0      | 2.3      | 10.5     | 21.8     | 46.8     | 405.5    |
| 900            | 29.0     | 18.4     | 7.5      | 12.3     | 38.2     | 67.2     | 396.0    |
3.2. Multi-scale sub-structure characterization of the simulated HAZ.

The OM micrographs, SEM micrographs, IPF maps and TEM micrographs were used to measure the PAGS, packet width, block width and lath width of the simulated HAZ specimens, as shown in figure 6–9. More than 200 grains were measured for the PAGS of each specimen. Over 100 packets or blocks were used to statistic the packet width or block width of each specimen. The measure of lath width for each specimen counted at least 50 laths.

Figure 5. Micro-fracture surfaces at the crack stable propagation zones of: (a) 900 specimen, (b) 1100 specimen, (c) 1200 specimen, (d) 1300 specimen.

Figure 6. PAGS characterizations of: (a) 900 specimen, (b) 1100 specimen, (c) 1200 specimen, (d) 1300 specimen.
The statistical data of the multi-scale sub-structure characterization was listed in table 3. The PAGS, packet width, block width both increased with the rise of $T_M$, while the lath width almost remained unchanged.

3.3. The microstructure unit controlling the toughness

Figure 10 shows the effects of the multi-scale sub-structure grain sizes on the stable propagation energies $E_a$ of the simulated HAZ under different $T_M$. As the $T_M$ increased, the PAGS, packet width, block width both increased, while the crack stable propagation energy $E_a$ decreased. Which was the microstructure unit controlling the toughness of the simulated HAZ specimens cannot be determined.

The micro-fracture surface morphologies at the crack unstable propagation zones are shown in figure 11. The 900, 1100 and 1200 specimen showed mixed fracture morphologies with small dimples and cleavage facet. The 1300 specimen showed a cleavage facet. The facet sizes of the simulated HAZ specimens were measured, as shown in figure 9. The facet size showed good agreement with the block width, indicated that the block substructure was the dominant microstructure of the crack stable propagation process.

The distribution of HABs also proved the above view. EBSD All-Euler map of the specimens are shown in figure 12, HABs above 45° are outlined by the yellow lines. Different colors represent different crystal orientations in the All-Euler map, which means different blocks. The HABs above 45° mainly distributed along the block boundaries, which enhanced the ability of the block to inhibit crack propagation. The decrease of the crack stable propagation $E_a$ as the increase of $T_M$ can be explained as below: As $T_M$ increased, the block width enlarged, which led to the reduction of the HABs ratio. The decrease of the HABs ratio further reduced the ability to hinder crack propagation, finally led to the decrease of $E_a$.

4. Conclusions

1. The peak temperature $T_M$ mainly affected the crack propagation process of the simulated HAZ specimens. The crack propagation energy $E_p$ decreased as the increase of $T_M$, while the crack initiation energy $E_i$ barely changed. It was the decrease of crack stable energy $E_a$ led to the change of $E_p$.

2. The PAGS, packet width, block width both increased with the rise of $T_M$, while the lath width remained unchanged.

3. The block width was similar to the facet size in the crack unstable propagation zone of the simulated HAZ specimens. The block was the microstructure unit controlling the crack propagation process of the LAHW simulated HAZ specimens.
Figure 8. Block width characterizations of: (a) 900 specimen, (b) 1100 specimen, (c) 1200 specimen, (d) 1300 specimen.
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Data availability statement

The data that support the findings of this study are available upon reasonable request from the authors.

Table 3. Multi-scale sub-structure grain sizes of the simulated HAZ specimens under different $T_{32}$.

| Specimen | PAGS ($\mu$m) | packet width ($\mu$m) | block width ($\mu$m) | lath width ($\mu$m) |
|----------|---------------|----------------------|---------------------|---------------------|
| 900      | 18.7          | 7.7                  | 6.2                 | 0.17                |
| 1100     | 42.9          | 13.8                 | 11.5                | 0.18                |
| 1200     | 51.4          | 17.8                 | 12.8                | 0.21                |
| 1300     | 75.8          | 22.5                 | 16.9                | 0.23                |
Author contributions

Writing—original draft, Liangliang Bao; Writing—review & editing, Liangliang Bao; data curation, Yiping Huang; project administration, Fujian Liu, Tao Han; methodology, Yanhong Xu, Kai Ouyang.

Conflicts of interest

The authors declare no conflict of interest.

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Figure 10. Effects of the multi-scale sub-structure grain sizes on the stable propagation energies of the simulated HAZ under different $T_{M}$. 

Figure 11. Micro-fracture surfaces at the crack unstable propagation zones of: (a) 900 specimen, (b) 1100 specimen, (c) 1200 specimen, (d) 1300 specimen.
Figure 12. EBSD All-Euler map of (a) 900 specimen, (b) 1100 specimen, (c) 1200 specimen, (d) 1300 specimen.

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