RETRACTION

The following article was withdrawn due to the request of the authors.

Retraction: Microstructure and Properties of Fiber Laser Welded Joints of Ultrahigh-strength Steel 22MnB5 and its Dissimilar Combination with Q235 Steel

Jin JIA,1) Shang-lei YANG,1,2) Wei-yuan NI,1) Jian-ying BAI1) and Yang-shenglan LIN1)

1) College of Materials Engineering, Shanghai University of Engineering Science, Shanghai, 201620 China. 2) Shanghai Key Technology Development Center of High Intelligent Laser Processing and Equipment Production, Shanghai, 201620 China.

This article was published in ISIJ Int., 54 (2014), No. 12, 2881–2889.
Microstructure and Properties of Fiber Laser Welded Joints of Ultrahigh-strength Steel 22MnB5 and its Dissimilar Combination with Q235 Steel

Jin JIA,1) Shang-lei YANG,1,2)* Wei-yuan NI,1) Jian-ying BAI1) and Yang-shenglan LIN1)

1) College of Materials Engineering, Shanghai University of Engineering Science, Shanghai, 201620 China.  2) Shanghai Key Technology Development Center of High Intelligent Laser Processing and Equipment Production, Shanghai, 201620 China.

This paper is to analyze the microstructure, microhardness, tensile and fatigue properties of the welded joints performed by a fiber laser on the 22MnB5 and Q235 steels in similar and dissimilar combinations. The result shows that the weld zone (WZ) consisted of lathy martensite, and the heat affected zone (HAZ) of 22MnB5 could be divided into 3 parts: quenched zone, incomplete quenched zone and the tempered zone which didn’t appear in the Q235 HAZ. The WZ had the highest hardness, while the HAZ was very narrow with a very fast drop in hardness. A soft zone existed in the HAZ of 22MnB5, and it was absent on the Q235 side. With the welding speed increasing, the martensite in the WZ became thinner and shorter, and the width of HAZ decreased. All the tensile and fatigue specimens of the similar 22MnB5 welded joints were broken in the HAZ. The base metal (BM) had a higher fatigue life than the similar 22MnB5 welded joints. The fatigue fracture contained 3 parts: crack initiation region which occurred from the specimen surface, crack propagation region and the final fast fracture region.

KEY WORDS: laser welding; 22MnB5; Q235; microstructure; hardness; tensile; fatigue.

1. Introduction

With the increasing of the automobile pollution and the energy depletion, it has led to more requirements for lightweight vehicles.1,2) Meanwhile, driving safety is always the most important thing in automobile manufacture. This phenomenon has improved the development of the advanced high strength steel (AHSS) which is being increasingly implemented for automotive body-in-white applications. The AHSS like boron alloy steel (B), dual-phase steel (DP), transformation-induced plasticity steel (TRIP), complex-phase steel (CP), martensitic steel (MART), owns an excellent combination of high strength with good formability. Especially, the hot stamping B steel such as 22MnB5 with a fully martensitic microstructure and the highest strength among present AHSS has played an important role in the automobile safe parts, such as A, B-pillars and bumpers. This steel owns an ultrahigh strength far beyond the cold stamping steels, which makes it be the focus in the field of automobile manufacture and research.3–5)

Welding is a primary manufacturing process used in automobile industry, it is thus important to understand the mechanical and metallurgical phenomena concerned to the welding of AHSS. Furthermore, laser beam welding (LBW) has been very popular in auto manufacture, amounts of studies6–10) have been carried out to investigate the laser welding of AHSS used in light weight vehicles. This technology possesses many advantages such as high welding speed, deep penetration, low heat input, narrow HAZ and effective integration with welding robots.11–14) LBW has made great contributions to improve the size accuracy, lightweight and joints strength. Among the present laser welding processes, fiber laser welding (FLW) is the newest one with several advantages over other types, including higher efficiency, smaller beam diameter, compact size, low maintenance costs, higher precision and reliability, and more extensive work environment.15–17)

To provide a basic theory and experience of fiber laser welding blanks and promote the quality of laser welded joints in vehicles, in this paper, microstructural, microhardness, tensile and fatigue properties of the fiber laser welded joints were studied. Few papers have been reported so far on the similar and dissimilar combination between 22MnB5 and Q235 steels using the FLW process. This study, therefore, was a previous work which aimed at studying the microstructure and the mechanical properties of the similar 22MnB5 laser welded joints as well as the dissimilar combinations between 22MnB5 and Q235 steels.

2. Experimental

In this study Al–Si pre-coated boron steel 22MnB5 (Arcelor USIBOR 1500P, quenched) and galvanized Q235 steel sheets were used as the base metal. The chemistries and thickness of these steels are shown in Table 1.
Steel sheets were butt welded by an IPG YLS-5000 fiber laser system with a maximum output power of 5 KW attached to a KUKA robotic arm. The laser with a beam focal length of 150 mm, a spot size/diameter of 0.2 mm, a wavelength of 1070 nm, was used to weld the blanks. Welding was performed with a head angle of 4° and the shielding gas was Ar. The welding parameters used in this experiment are shown in Table 2.

The microstructure of the weldments was observed by an optical microscope (VHX-600) and the S-3400N scanning electron microscope (SEM) with the metallographic specimen etched with 4% Nital solution. The microhardness was tested by a HXD-1000 microhardness tester under a load of 100 g which was held for 15 s. Tensile and fatigue test specimens are shown in Fig. 1. Tensile tests were carried out via a universal material testing machine (AG-25TA) at room temperature and strain rates of $1 \times 10^{-3}$ s$^{-1}$. Fatigue tests were conducted on a computerized IBTC-300 testing system under load control. All the fatigue specimens were polished without any nick before test. The tension-tension cyclic loading at a stress ratio of $R = 0.1$ was applied at a frequency of 1 Hz and sinusoidal waveform. After the fatigue and tensile tests were completed, the fracture surfaces were observed by SEM.

3. Results and Discussions

3.1. Microstructure

With the welding process in Table 2 (at the welding speed of 3 m/min), the microstructure of the similar 22MnB5 welded joint is shown in Fig. 2. As shown in Fig. 2(a), a significant change in microstructure can be seen across the welded joint and the welded joint can be divided into WZ, HAZ and BM. The formation of the WZ was good without any cracks, pores or any other defects, but a small depression appeared due to no wire filling and high welding temperature. Figure 2(d) shows the 22MnB5 BM microstructure mainly contained fine lathy martensite with a uniform dispersion. As is known, it is hard to measure the real temperature of the weld to calculate the cooling rate. While the cooling time $t_{\text{cool}}$ expressed as the following equation (18) can be used in continuous-cooling transformation (CCT) diagram instead of the cooling rate to infer the microstructure in the WZ.

$$t_{\text{cool}} = \frac{\alpha}{4\pi \lambda^2} \left( \frac{q}{\rho c} \right)^{1/2} \left( \frac{1}{(T - t)^2} - \frac{1}{(T_0 - t)^2} \right)$$

where $\lambda$, $\alpha$, $\rho$, $c$, $q$, $h$, $v$, and $T_0$ were heat conductivity, heat diffusivity (heat conductivity/density $\times$ specific heat), heat efficiency, plate thickness, welding speed and pre-heating temperature.

After consulting present paper (19) (Thermal conductivity 30.0 W/m/K, Density 7860 kg/m$^3$, heat efficiency 50% and specific heat capacity 680 J/kgK) and the data in Tables 1 and 2, the $t_{\text{cool}}$ was calculated to be about 1.3 s, much faster than the critical speed of martensitic transformation of 22MnB5 steel (20). As a result, the microstructure of the WZ fully contained martensite with a lathy morphology, as can be seen in Fig. 2(c). When solidifying, the WZ transformed into martensite, and the martensitic phase firstly nucleated in the austenite grain boundary. The microstructure of primary martensite was fine and grew like a column towards the grain interior, due to a low content of carbon, so the final morphology was like a lath. The microstructure in the HAZ (Fig. 2(b)) changed significantly and both the color and appearance were different from the BM and WZ. The details of the HAZ microstructure will be discussed later in this section.

The microstructure changes of the dissimilar Q235-
22MnB5 welded joint are shown in Fig. 3. The weld formation was also very good. Figure 3(c) clearly shows that the microstructure of the WZ still consisted of lathy martensite, exhibiting a similar morphology with the WZ of the similar 22MnB5 welded joint (Fig. 2(c)). This phenomenon signified an insignificant effect of chemical contents on the WZ microstructure and formability. As shown in Fig. 3(d), The Q235 BM contained fine grained ferrite matrix with a uni-
form dispersion pearlite, much different from the 22MnB5 BM (Fig. 2(d)). As a result, the welded joint displayed two distinct HAZs (Fig. 3(a)) on each side where the original microstructure transformed differently during the FLW. The microstructure of the HAZ on the Q235 side (Fig. 3(b)) presented much light color compared with the WZ and the HAZ on the 22MnB5 side, and both the color and appearance resembled its BM. However, the HAZ on the 22MnB5 side resembled closely to the HAZ obtained in the 22MnB5 similar welded joints.

Figure 4 shows the SEM images detailing the HAZ microstructure obtained in 22MnB5 welded joint. Different places in the welded joints had a different distance from the welding heat source as well as the distinct phase transformation during the welding thermal cycle, so that it would result in forming different microstructure in the HAZ. Figure 4(a) shows the HAZ of 22MnB5 can be divided into 3 parts: quenched zone, incomplete quenched zone and the tempered zone. In the incomplete quenched zone, the temperature here was between Ac1 and Ac3 lines during FLW. When the temperature was above Ac1 line, the austenite started to form and subsequently transformed into martensite which was different from the original martensite in the BM. Ac3 was the ending line of austenitic transformation. However, the temperature in this area was always below Ac3, leading to amounts of original martensite transforming into ferrite rather than austenite, as seen in Fig. 4(c). This was the primary cause of the incomplete martensitic transformation in this area. In the quenched zone (Fig. 4(b)), the steel experienced a temperature above Ac1 line during the FLW with a complete austenitic transformation, exhibiting a much higher content of transformed martensite compared with the incomplete quenched zone. The tempered HAZ experienced a temperature below Ac1 line, which made amounts of pre-existing martensite in BM tempered. In this region, the microstructure was consisted of martensite, tempered martensite and ferrite as shown in Fig. 4(d).

SEM images of the HAZ in the Q235 are presented in Fig. 5. It should be noted that the microstructure of the Q235 BM was normalized without any martensite, the Q235 steel would never undergo any phase transformation below Ac1 line, so the tempered zone was not observed in its HAZ (Fig. 5(a)). The Q235 HAZ just can be divided into 2 parts: upper-critical zone and the two phase zone. Figure 5(b) presents the upper-critical HAZ microstructure containing ferrite along with amounts of acicular bainite. In the two-phase HAZ, the temperature during FLW was between Ac1 and Ac3. Under the joint actions of the welding thermal cycle and its poor hardenability, the microstructure in this region underwent an incomplete austenitic transformation and recrystallization without much martensitic transformation after welding. The final microstructure in the two-phase HAZ was similar with the BM, as can be seen in Fig. 5(c). Thus, during the laser welding, the different place experienced different welding thermal cycle, leading to forming a different microstructure in each region as well as the changes of mechanical properties.

Figure 6 shows the WZ microstructure of the similar 22MnB5 welded joints and the width of the incomplete quenched zones at different welding speeds. From Figs. 6(a) and 6(b), it can be indicated that the lathy martensite in WZ became thinner and shorter with the welding speed increas-
ing. As shown in Figs. 6(c) and 6(d), the width of incomplete quenched HAZ narrowed from 102.4 μm to 69.5 μm due to the decreasing of the welding heat input, when the welding speed increased. Apparently, with the welding speed changed, the microstructure and the width of HAZ would be different, which must have great effects on its properties in the realistic application.

Fig. 5. SEM micrographs showing the microstructure of HAZ in Q235 welded joint (a) overall view of the Q235 HAZ, (b) upper-critical HAZ, (c) two-phase zone.

Fig. 6. Microstructure of the 22MnB5 WZ and the width of incomplete quenched HAZ at different welding speeds.
3.2. Microhardness
The microhardness distribution across the similar 22MnB5 welded joints at different welding speeds is presented in Fig. 7. It is seen that the hardness profile exhibited a nonuniform characteristic with the highest hardness in WZ and the lowest hardness in HAZ. The hardness increased a little when transited from the WZ to the HAZ, as the cooling rate was higher in weld edge, increasing the volume faction of martensite and forming thinner, shorter lathy martensite morphology, as can be seen in Figs. 4(b) and 6(a). Both the WZ and the weld edge basically contained martensite, so that the increasing of microhardness was not large. Across the HAZ into BM, the microhardness dropped quickly and a soft zone formed in the HAZ with the lowest hardness just 63% of that in BM. Compared with BM, most parts in HAZ had the lower hardness, resulting in the most HAZ being the weak area.

From Fig. 7, it is also seen that, with the welding speed increasing, the width of the WZ and HAZ became narrow and the microhardness became higher. The highest hardness in the WZ increased from 524.6 HV to 544.3 HV. Meanwhile, the lowest hardness in soft zone increased from 319.6 HV to 330.1 HV. This phenomenon above was a result of the change in welding speed which had great effects on cooling rate, heat input and weld thermal cycle in the welded joints. The increasing of the welding speed not only increased the content of martensite but also reduced the level of tempering, so that the microhardness increased and the HAZ and WZ narrowed.

To have a better comparison, the microhardness data across the similar 22MnB5, Q235 and the dissimilar 22MnB5-Q235 welded joints at the welding speed of 3 m/min are plotted in Fig. 8. The correspondence between the microstructural zones (quenched, incomplete quenched zone and tempered zone or upper-critical and two-phase zone) and the hardness is also marked in Fig. 8 according to the overview of the HAZ in Figs. 4(a) and 5(a). It is seen that the HAZ in the similar Q235 and 22MnB5 welded joints respectively had a similar trend of hardness distribution with the HAZ on the Q235 and 22MnB5 side in the dissimilar 22MnB5-Q235 welded joint. Similarly, the highest hardness of 22MnB5-Q235 welded joint was also appeared in the WZ. However, it is particularly observed that there were two distinct hardness regions inside the WZ of dissimilar 22MnB5-Q235 welded joint (~470 HV on the Q235 side and ~530 HV on the 22MnB5 side). This was a result of the difference in the BM and chemistry between the 22MnB5 and Q235 where the contents of the alloying elements and carbon were much lower than that of 22MnB5, resulting in a poor hardenability and a higher critical speed on the Q235 side. Moreover, since the WZ solidification time during the FLW was very short, there was little or insufficient diffusion in the WZ, which made the WZ still in the condition of segregation between the two steels. However, compared with the similar Q235 welded joint, the dissimilar welded joint had a much higher hardness in the WZ, indicating the alloying elements already had a good combination without any defects, in spite of not fully complete combination.

3.3. Tensile Tests
The details of the tensile test results are given in Table 3, and some of the fractured samples are shown in Fig. 9. From Fig. 9(a), it is seen that the sample 5 failed in the WZ due to its incomplete penetration at the welding speed of 6 m/min. While other fractured samples of the similar 22MnB5 welded joints all failed in the HAZ, suggesting that the welding speed didn’t deteriorate the tensile properties of the welded joints if complete penetration. This was because the plastic deformation of the welded joints concentrated in the soft zone due to its lower strength. The soft zone accounted for about 60% of the total welded joint, as can be seen in the Fig. 7. In addition, with the specimen shown in Fig. 1(b), the tensile strength (TS) of the WZ was tested to be 1412.2 MPa, higher than that of the HAZ. The results given above made a further explanation why all the tensile samples failed in the HAZ. The similar 22MnB5 welded joints had an elongation of about 2% without any visible occurrence of necking due to the occurrence of the coarse grains and the hard, brittle phase. However, with the welding speed increasing, the TS and elongation was improved by narrowing the soft zone, refining the grain and increasing the remaining austenite.21 Interestingly, the content of the ductile phase “ferrite” reduced in the soft zone, but the elongation still improved at a higher welding speed. The elongation was...
jointly decided by the WZ, HAZ and BM, while the soft zone was just a small part of welded joint, so the soft zone didn’t have the decisive effects on the elongation.

On the contrary, the failure of the similar Q235 welded joint was all in the BM with the TS of 391.7 Mpa and a much higher strain to failure (28%) than that of 22MnB5. Though the soft zone in HAZ had a side effect on the tensile properties of welded joints, the dissimilar 22MnB5-Q235 welded joint was broken in the Q235 BM (Fig. 9(b)) owning a similar TS with the similar Q235 welded joint, due to the hardness in 22MnB5 soft zone still higher than that of the Q235 BM. However, its elongation was just 20% which was lower than that of the similar Q235 welded joint. The elongation of 22MnB5 is much lower than Q235, leading to the most plastic deformation occurring on the Q235 side with little on the 22MnB5 side, so that the elongation of the whole welded joint reduced finally.

SEM images of the tensile fracture surface are shown in Fig. 10. The fracture edge of simple 1 displayed typical dimple pattern which were small and shallow along with some small cleavage planes (Fig. 10(a)). In the center of the fracture (Fig. 10(b)), the dimples became bigger and deeper. There were some big dimples surrounding by small dimples and a few ridges appeared, presenting a characteristic of ductile fracture. However, the fracture surface of sample 2 exhibits an obvious feature of quasi-cleavage fracture. In Fig. 10(d), it is seen that the fracture surface contained some small dimples along with amounts of cleavage planes. While on the edge of its fracture surface (Fig. 10(c)), the dimples reduced and the cleavage planes increased and grew up. This was because the cleavage planes initiated on the grain boundary and then connected with each other by the plastic deformation. Quasi-cleavage fracture is a typical form which mainly appears in the regions consisted of lathy martensite. Compared the fracture of sample 1, it can be easily summarized that the BM ductility and toughness was obviously better than that of the welded joints.

Table 3. Results of tensile test.

<table>
<thead>
<tr>
<th>Sample</th>
<th>Type</th>
<th>Tensile strength (MPa)</th>
<th>Elongation (%)</th>
<th>Fracture sites</th>
<th>Occurrence of necking</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>22MnB5(BM)</td>
<td>1584.2</td>
<td>8.9%</td>
<td>–</td>
<td>Not clear</td>
</tr>
<tr>
<td>2</td>
<td>3 m/min</td>
<td>1241.9</td>
<td>1.7%</td>
<td>HAZ</td>
<td>No</td>
</tr>
<tr>
<td>3</td>
<td>4 m/min</td>
<td>1259.5</td>
<td>1.8%</td>
<td>HAZ</td>
<td>No</td>
</tr>
<tr>
<td>4</td>
<td>5 m/min</td>
<td>1285.3</td>
<td>2.1%</td>
<td>HAZ</td>
<td>No</td>
</tr>
<tr>
<td>5</td>
<td>6 m/min</td>
<td>1299</td>
<td>2.2%</td>
<td>WZ</td>
<td>No</td>
</tr>
<tr>
<td>6</td>
<td>Q235(BM)</td>
<td>391.7</td>
<td>28%</td>
<td>–</td>
<td>Yes</td>
</tr>
<tr>
<td>7</td>
<td>22MnB5-Q235</td>
<td>397.1</td>
<td>20%</td>
<td>Q235(BM)</td>
<td>Yes</td>
</tr>
</tbody>
</table>

Fig. 9. Fractured locations in the specimens (a) 22MnB5-22MnB5 welded joints, (b) Q235 similar and dissimilar welded joints.

Fig. 10. SEM morphology of tensile fracture at different location. (a) Fracture edge of sample 1, (b) Fracture center of sample 1, (c) Fracture edge of sample 2, (d) Fracture center of sample 2.
3.4. Fatigue Tests

Uniaxial tensile cyclic loading fatigue tests of the similar 22MnB5 welded joints and its BM were experienced at a stress ratio of $R = 0.1$ and with the max stress $\sigma_{\text{max}}$ of 800 MPa. The result showed that the similar 22MnB5 welded joint exhibited a fatigue life of $2.62 \times 10^5$ cycles, lower than that of 22MnB5 BM ($1.76 \times 10^6$ cycles). As shown in Fig. 11, after etched by 4% Nital solution, the failure in 22MnB5 welded joints fatigue samples was seen in the HAZ, due to a sensitive effect of the soft zone in the HAZ. As known above, the grain was much coarser in the soft zone, increasing the degree of slippage under the alternating stress. So that the initiation of fatigue cracks in this region was accelerated. Furthermore, when the cracks propagated to the grain boundary, both the spread path and fatigue striation width were changed, which would avoid the cracks to keep growing. As the coarse grains had fewer boundaries to avoid the growth of cracks, it would result in a low fatigue life of the welded joint. The relationship between grain size and the fatigue life was just like the Hall Petch formula. In addition, some studies indicated that the ferrite and austenite was the centralization slippage zone under alternating stress, having great contributions to the fatigue cracks initiation. This is also the result why the soft zone had a lower fatigue life than the BM.

Some typical SEM images of the fatigue fracture surface of the similar 22MnB5 welded joint are shown in Fig. 12. In Fig. 12(a), it is seen that the fatigue fracture can be divided into 3 parts: crack initiation region, crack propagation region and the final fast fracture region, which was marked as “b, c, d” in the picture. Magnified image of the crack initiation region is presented in Fig. 12(b), it can be seen that crack initiation mainly took place from the specimen surface due to more stress concentration and less constraint than that inside. There is no doubt that more slippage occurred in the surface so that the crack initiation appeared here more easily. In Fig. 12(c), fatigue striations and secondary cracks can be seen in the propagation region where fatigue damage gradually accumulated together till the propagation approached the final fracture. Fatigue striations were usually divided into two types: the toughness striations and the brittleness striations. As shown in the image, many crystal faces appeared in this region, perpendicular with the fatigue striations, suggesting a result of being brittleness fatigue striations. Figure 12(d) shows the final fast fracture zone contained the typical dimples along with the plastic ridges just like its tensile fracture surface (Fig. 10(d)).

Compared with the fracture surface of similar 22MnB5 welded joint, the fatigue fracture surface of its BM exhibited a very similar surface, indicating a similar fatigue fracture principle between the BM and the welded joints of 22MnB5.
4. Conclusions

Depending on the present study on the microstructure, microhardness, tensile and fatigue tests of the similar 22MnB5-22MnB5 and dissimilar 22MnB5-Q235 welded joints. Several conclusions can be drawn below:

(1) Both the WZ of the similar and dissimilar welded joints consisted of lathy martensite due to a fast cooling rate during FLW. The HAZ of 22MnB5 was divided into 3 parts: quenched HAZ, incomplete quenched HAZ and tempered HAZ. While the HAZ on the Q235 side just had two zones: upper-critical HAZ and two phase zone without any tempered martensite exhibited.

(2) A characteristic asymmetric microhardness profile across dissimilar 22MnB5-Q235 welded joints was observed. A soft zone appeared in the 22MnB5 HAZ, but it was absent on the Q235 side. Inside the FZ two hardness sub-regions were observed (~470 HV on the Q235 side and ~540 HV on the 22MnB5 side) due to the difference in the alloying elements between two steels along with the fast cooling during FLW.

(3) When the welding speed increased from 3 m/min to 5 m/min, the microhardness of the welded joints became higher (the highest hardness in WZ increased from 524.6 HV to 544.3 HV, the lowest hardness in soft zone increased from 319.6 HV to 330.1 HV), the width of the incomplete quenched HAZ became narrower (from 102.4 to 69.5 μm), and the martensite in WZ became shorter and thinner.

(4) The failure in tensile samples of the similar 22MnB5 welded joints all occurred in the HAZ. Both the tensile strength and the elongation were lower than that of the BM, due to the soft zone and the coarse grains in welded joints. Observation of the fractured tensile samples by SEM indicates that the ductility and toughness of the BM was obviously better than the welded joints.

(5) BM had a higher fatigue life than similar 22MnB5 welded joints. The fatigue failure still occurred in the HAZ of 22MnB5, due to a sensitive effect of the soft zone.

(6) The fatigue crack initiation was observed to occur from the specimen surface due to more stress concentration and a less constraint in this place. Fatigue striations and secondary cracks can be seen in crack propagation. The final fast fracture zone just looked like its tensile fracture surface.

Acknowledgements

The authors sincerely would like to thank Dr. Yan from the College of Materials Engineering, Shanghai University of Engineering Science, for his help in preparation of specimen and helpful discussions. This research was supported by Shanghai Key Technology Development Center of High Intelligent Laser Processing and Equipment Production (Shanghai University of Engineering Science). This work was financially supported by National Natural Science Foundation of China (Project No.51075256), and a state key project (No.11ZZ177).

REFERENCES

2) W. J. Joost: JOM, 64 (2012), 1032.
18) Z. W. Gu, S. B. Yu and L. J. Han: ISIJ Int., 52 (2012), 484.