Microstructure, mechanical properties and tear toughness of laser-welded DP980 dual phase steel

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Abstract

In this study, DP980 steel sheets were laser welded with a laser power of 4.5 kW and a welding speed of 4.5 m min⁻¹. The microstructural evolution and mechanical properties of the welded joint and the effect of notch location on tear toughness were evaluated. Results show that the fusion zone (FZ) was composed of lath martensite, the hardness (276 HV) of the heat-affected zone (HAZ) was lower than that of the base metal (305 HV) resulting from martensite tempering. The welded specimens failed at the soft HAZ with a 98.4% joint efficiency during the tensile test, and the ultimate tensile strength of the as-received steel and welded joint was 1026 MPa and 1010 MPa, respectively. The tear energy of the FZ and HAZ was lower than DP980 base metal (BM). Thus, it is considered that the fracture toughness of the joint decreased after welding. The crack growth path of the FZ gradually deviated toward the HAZ during tearing due to the asymmetrical plastic zone at the crack tip. Compared with the ductile fracture of the base metal, the significant decrease in the fracture toughness of the welded joint is due to the weak deformation resistance of tempered martensite.

1. Introduction

Dual phase (DP) steel is widely used in automobiles due to its low yield strength ratio, high initial work hardening rate, and excellent ductility [1, 2]. The Future Steel Vehicle (FSV) plan issued by the World Iron and Steel Association shows that the consumption of DP steel in the future car body will reach 31.3%. On the one hand, the large use of high strength steels such as DP steel can improve the vehicles’ safety performance. On the other hand, thinner high strength steels can be used to meet the requirements for auto parts, thereby reducing vehicle weight, fuel consumption and carbon emissions [3]. The connection between different parts is usually achieved by welding. Laser welding technology has incomparable advantages in many aspects, such as weld quality, welding efficiency, and process flexibility, and it is the most promising welding technology for automobile steel [4]. Cold forming of DP steel usually involves deep drawing, hydroforming and other large deformation operations. Due to the increasingly higher strength and inherent welding defects of laser tailor welded DP steel blanks, several cracking problems during cold forming have been reported [5, 6]. Thus, the crack resistance has become a vital factor limiting the applicability of high strength laser tailor welded DP steel blanks.

The mechanical properties and fracture behavior of dual-phase steel have been studied by numerous researchers. Frómeta et al [7] found that DP steel with a larger volume fraction of ferrite and homogeneously distributed martensite islands exhibited significantly higher crack propagation resistance. Han et al [8] investigated the mechanical properties, microstructural characteristics and the strain rate sensitivity of friction stir welded DP780 steel with different welding speed. Wang et al [9] studied the effect of heat inputs on microstructure and fracture behavior of laser welded joints of DP steels, the results showed that the microstructure of welded joints was similar, while the martensite decomposition and carbide precipitation gradually became significant with the increase of heat input, leading to the hardness decrease of the welded joint. Hossein Mostaan et al [10] investigated the microstructural observations, mechanical properties, and fracture behavior of laser welding of dual-phase steels with different silicon contents, they found that the fracture
mechanism of DP steel with low martensite volume fraction and small ferrite grain size was microcavity formation. While the fracture mechanism of DP steel with high martensite volume fraction was separation and decohesion. Wang et al. [11] studied the effect of martensite morphology and volume fraction on the low-temperature impact toughness of dual-phase steels, and found that refining the lath martensite substructure can improve the impact toughness while the large-size ferrite grains in DP steels preferentially occur cleavage fracture. However, the microstructure evolution of laser welded DP steel joints, namely, the morphology and distribution of martensite and ferrite are far more complex. Besides, the effect of mechanical mismatch on the crack propagation path of welded joint is also not fully understood, and cracks may originate from different regions of the welded joint due to the complexity of vehicle stress.

Therefore, the present work aims to investigate the microstructure evolution and tensile properties of DP980 joint after laser welding, and focuses on the effect of notch location on the crack initiation and propagation resistance of laser welded DP980 steel joints and to correlate the results with their microstructure characteristics. Fracture toughness was evaluated by the Kahn tear test (KTT) method [12], which can easily evaluate the plane stress fracture toughness of ductile thin plates. The main advantage of KTT is that it is very simple and provides an estimation of crack propagation resistance of the material. In addition, the fracture mechanism of the torn specimen was analyzed. This research is helpful to better understand the relationship between the toughness and strength of DP980 welded joints and improve the welding process.

2. Experimental procedure

Cold-rolled DP980 steel sheets with a thickness of 1.40 mm were used in this study. Table 1 lists the chemical compositions of the as-received DP980 steel. Blanks with size of 200 mm × 200 mm × 1.40 mm were laser welded by an IPG Photonics YLS-6000 fiber laser system with a defocusing distance of 0 mm and a laser spot diameter of 0.3 mm. Pure argon gas was selected for shielding with a gas flow of 15 l min⁻¹. The laser beam was perpendicular to the welded steel sheets, the welding speed was 4.5 m min⁻¹, and the laser power was 4.5 kW.

The laser welded joint was cut into cross sections and then mechanically grounded and polished. The metallographic specimen was prepared with 4% nitric acid solution. The welding microstructure was captured by Zeiss optical microscopy (OM) and Zeiss scanning electron microscopy (SEM). The microhardness of the polished specimen across the welded joint was measured using a Vickers microhardness tester with a 200 g load and 15 s dwell time. The distance between adjacent continuous indentations was 200 μm to avoid the potential impact of adjacent indentations.

The specimens for tensile tests were machined according to the GB/T 2651–2008 standard with a gauge length of 50 mm and a 12.50 mm parallel section width. The length direction of the specimens was perpendicular to the rolling direction, and the FZ was positioned at the center of the gauge length. Tensile tests of both the base metal (DP980-BM) and welded material were conducted at room temperature by a UTM 5000 electronic universal testing machine with a loading rate of 2 mm min⁻¹. To ensure the reliability of the results, each specimen was tested three times.

Three groups of Kahn tear test (KT) specimens were machined from the welded sheets in accordance with the ASTM B871–01 standard, and the specimens were cut in a transverse–longitudinal (T–L) direction. Notches were located in the base metal (KT-BM), heat-affected zone (KT-HAZ) and fusion zone (KT-FZ), the tear test involved a single edge notched specimen that was statically loaded through pin loading holes, as shown in figure 1 (the shaded part indicates the FZ). The Kahn tear test was carried out by an electrohydraulic servo test system capable of quasi-static loading at a crosshead speed of 1 mm min⁻¹. The force and displacement required to fracture the specimen were recorded for analysis, and three samples were tested for each notch condition. This test method provides a comparative measure of resistance of products to unstable fracture due to crack-like stress concentrators.

3. Results and discussion

3.1. Microstructural and hardness evolution

The macromorphology of the weld joint is shown in figure 2(a). The joint surface quality is decent under the selected laser tailor welding parameters. The joint surface is free of obvious defects such as spatter, cracks and

| C   | Si | Mn | Al | Cr | P  | S  | Ni | Fe |
|-----|----|----|----|----|----|----|----|----|
| 0.12| 0.4| 2  | 0.04| 0.03| <0.02| <0.02| 0.04| Bal. |
cavities, the weld width is uniform as a whole, and the transition between the joint and the base metal is smooth. Figure 2(b) shows the microstructure on the cross section of the DP980 welded joint. The weld joint achieved complete penetration, which is composed of a weld zone (FZ), heat-affected zone (HAZ) and base metal (BM). The microstructure of the DP980 base metal is composed of ferrite matrix and island-shape martensite which distributed in the matrix, as shown in figure 2(c). The martensite content is approximately 56% calculated by Image-Pro software. Ferrite provides a certain plasticity for dual-phase steel, while martensite ensures the strength [8].
The DP980 HAZ contained three different regions: the subcritical HAZ (figure 2(d)), intercritical HAZ (figure 2(f)) and upper-critical HAZ (figure 2(g)) [13]. The width of the heat-affected zone is approximately 0.50 mm. As the subcritical HAZ is close to the base metal, the microstructure is very similar to that of the base metal. Generally, martensite would be tempered below the Ac1 line with segregation of carbon. Martensite would directly decompose into cementite and fine ferrite [14, 15]. Therefore, the subcritical HAZ is mainly composed of cementite particles, ferrite and some untempered martensite, as shown in figures 2(d), (e). The microstructure of the intercritical HAZ is shown in figure 2(f). The temperature of this region was between the Ac1 line and Ac3 line during welding, resulting in partial austenitization of the base metal. Thus, martensite and undissolved ferrite were observed in this region after cooling. The grains in the intercritical HAZ are very fine, which results in an increase of grain boundaries in this region, and it is beneficial to hinder austenite growth and refine the grain to a certain extent. The microstructure of the upper-critical HAZ (above the Ac3 line) is coarse lath martensite with a small amount of ferrite, as shown in figure 2(g). Figure 2(h) exhibit the microstructure of FZ.

In the process of laser welding, the peak temperature of the weld zone exceeded the melting point of the DP980 base metal. Due to the small weld pool and fast cooling speed, the grain growth always follows the direction of the maximum temperature gradient, leading to the full coarse lath martensite being formed after welding. The microhardness profile across the joint is shown in figure 3. The average hardness of BM is 305 HV. A softened zone is observed in the subcritical HAZ attributed to the occurrence of tempered martensite, where the local hardness decreased significantly by 10% compared to the BM hardness. A significantly higher hardness value of 469 HV was observed in the FZ due to the existence of a full martensitic microstructure.

3.2. Tensile properties

Figure 4 shows the engineering stress–strain curves of the DP980 base metal and welded specimens. The tensile test results are shown in table 2. All the stress–strain curves exhibit smooth and continuous characteristics without an obvious yield point. The joint efficiency of the welded joint (defined as the ratio of the UTS (ultimate tensile strength) of the welded joint to the UTS of the corresponding base metal [16]) is more than 98.4%. Compared with the base metal, the elongation of the welded joint decreases by approximately 33%, while the yield ratio increases from 0.61 to approximately 0.66, which indicates that the formability of the welded joint deteriorated. The main reason for the low plasticity of the DP980 welded joint is the strain concentration in the softening zone during the tensile process, which leads to premature fracture of the joint. Energy absorption can be calculated using the area below the stress–strain curves [17], considerable energy absorption is helpful to reduce the damage caused by vehicle collisions.

As shown in figure 5, DP980 BM is uniformly deformed, and the fracture shape is shear after fracture. The fracture position of the welded tensile specimen is 1.72 mm away from the weld centerline, that is, the softening zone. During the tensile process, plastic deformation accumulates in the HAZ until the final failure. Stress concentration occurs in the softening zone, and the whole crack initiates and expands along the softening zone. Therefore, the crack direction of DP980 WJ after fracture is perpendicular to the tensile direction.

Due to the distinction of microstructure and properties between martensite and ferrite, dual phase steels have high work hardening properties. The process of plastic deformation shows the characteristics of multistage...
strain hardening, which is mainly related to the different effects of martensite and ferrite on stress transfer \[18\].

The strain hardening rate \( K \), which is defined by equation (1), could quantitatively reflect the strain hardening ability of materials and describe the deformation characteristics of the uniform strain stage.

\[
K = \frac{d\sigma}{d\varepsilon}
\]  

(1)

\( \sigma \) and \( \varepsilon \) represent the true stress and effective plastic strain, respectively. The relationship between the strain hardening rate and effective plastic strain of the DP980 base metal and welded joint was obtained by the data of tensile test, as shown in figure 6. As can be seen, the curve can be divided into two stages. In the first stage, the initial work hardening rate of DP980-BM and the welded joint decrease significantly, and the difference of the two specimens is not obvious, indicating the joint has a similar plasticity to the BM in the low strain stage. In addition, the high initial work hardening rate might be attributed to the deformation incompatibility between
the two phases and the movement of dislocations [19]. With increasing strain, the dislocations at the interface of the two phases interact with each other and rearrange, resulting in a significant decrease of work hardening rate [20].

In the second stage, the work hardening rate decrease slowly with increasing true strain. Martensite and ferrite undergo plastic deformation simultaneously during this stage, the incompatibility between the two phases is alleviated, and the dynamic recovery of ferrite reduces the internal stress concentration [21], which leads to a slow decrease in the work hardening rate. Compared with the base metal, the strain hardening rate of the joint decrease more faster in the second stage. The reason is that when the stress is transferred to the softening zone of the joint, the tempered martensite plastically deformed earlier in stage II compared with martensite [20], resulting in a decrease in the work hardening ability.

Dual phase steel shows multi-stage strain hardening characteristics, while the single value of $K$ in Hollomon equation is not enough to effectively reflect the strain hardening behavior in different stages. Collav points out that [22], the modified Crussard-Jaoul method based on swift equation can sensitively reflect the hardening mechanism in different deformation stages, and its expression is shown in equation (2).

$$\varepsilon = \varepsilon_0 + k\sigma^m$$  \hspace{1cm} (2)

where $\sigma$ is true stress; $\varepsilon$ is true strain; $\varepsilon_0$ is the maximum elastic strain; $m$ is the strain hardening index; $K$ is a constant. Transform equation (2) into equation (3):

$$\ln\frac{d\sigma}{d\varepsilon} = (1-m)\ln\sigma - km$$  \hspace{1cm} (3)

Figure 7 shows the $\ln\frac{d\sigma}{d\varepsilon}$~$\ln\sigma$ plots for DP980 BM and WJ. It observed that the strain hardening rate decrease with the increasing true stress. All the curves present two slopes $m_1$ and $m_2$, the lower the hardening index $m$, the stronger the work hardening ability. The corner point of curve is known as ‘transitional strain’, coded as $\varepsilon_0$. The strain hardening index and transitional strain of DP980 steel by the modified C-J analysis are listed in table 3. The $m_2$ of WJ increases compared to BM, indicating the strain hardening capacity of WJ impaired at second strain stage due to the lower dislocation density of the tempered martensite. Researchers [23] pointed out that the smaller the strength difference between hard phase and soft phase, the easier the hard phase is to participate in plastic deformation. The $\varepsilon_0$ of WJ is lower than BM, increasing that the laser welding sample at lower strain stage appeared plastic strain earlier.

### 3.3.3 Tear toughness

Generally, tear test is mainly used as an indicator of toughness. The fracture toughness and strength of the welded specimen are characterized by the unit propagation energy (UPE) and tear strength (TS). UPE provides a measure that combines the strength and ductility of a material and represents the toughness of a material against crack growth. Tear strength is the maximum nominal tensile and bending stress that a tear specimen could withstand. The expressions of the two parameters are as shown in equations (4) and (5):

$$UPE = \frac{PE}{bt} \text{ (N/mm)}$$  \hspace{1cm} (4)
Propagation energy (PE) is the energy required for crack propagation in tearing specimens, which is determined by integrating the area under the load–displacement curve (as shown in figure 8) from the point of maximum force to the point of complete fracture. \( P \) is the maximum applied load, \( b \) is the distance between the notch root and back edge of the specimen, \( M \) is the bending moment, \( t \) is the average specimen thickness, \( C \) is the distance from the neutral axis to the outermost fiber, and \( I \) is the moment of inertia. It is worth noting that UPE can only be used as a comparative value of crack growth resistance for a given material and is not intended to provide an absolute measurement of crack growth resistance for structural design.

Figure 8 shows the load–displacement curves of Kahn tear specimens with different notch positions. The calculated TS and UPE parameters are shown in table 4. The initiation energies (IE) are higher than the propagation energies for KT-BM and KT-HAZ specimens, while KT-FZ is the opposite. Compared with KT-BM, the load required to cause fracture in KT-FZ increased by approximately 20%, while the UPE decreases by 37%, and the TS and UPE in KT-HAZ decreased by approximately 18% and 56%, respectively. It can be inferred that the toughness of the welded joint is decreased after laser welding. This is mainly because the welding process inevitably brought some defects and deteriorated material properties. Moreover, the ferrite-martensite dual-phase structure of the base metal can give full play to the strength and plasticity of the steel, improving the toughness. However, the microstructure of FZ is mainly hard and brittle martensite, so the toughness decreased. Besides, KT-haz has the worst toughness due to its heterogeneous microstructure, which also included numerous hard and brittle martensite.

Figure 9 shows the macroscopic fracture diagram after the tear test. KT-BM and KT-HAZ fractured specimens are in the catalog of type A in ASTM B870–01 (the crack path did not deviate more than 10° from the test plane). The crack growth in KT-BM is stable because the microstructure of base metal is relatively uniform. For the KT-HAZ condition, the crack is confined to expand in the HAZ, so the crack growth path is relatively straight. This is because the HAZ is relatively narrow with a softening zone, and the strength of the BM and FZ near it is higher. For the KT-FZ specimen, it can be observed that the crack initiating from the FZ deviated from its original growth direction after a stage of steady growth, and gradually deviated from the FZ toward the HAZ side, and finally fractured at the soft subcritical HAZ, showing a catalog of type B (the angle between the crack path and test plane deviates from 10° to 20°).

### Table 3. Parameters of the modified C-J analysis.

| Samples     | \( m_1 \) | \( m_2 \) | \( \varepsilon_0/\% \) |
|-------------|-----------|-----------|------------------------|
| DP980 BM    | 2.03      | 7.10      | 1.93                   |
| DP980 WJ    | 2.09      | 8.03      | 1.82                   |

\[
TS = \frac{P}{bt} + \frac{MC}{I} = \frac{4P}{bt} \text{(MPa)}
\]  

(5)
The main reason for crack propagation deflection can be attributed to the inhomogeneity of the joint microstructure, resulting in the existence of asymmetric plastic zones. As mentioned above, the microstructure of the FZ is fully lath martensite with high strength, which is beneficial to resist crack initiation and growth. The deformation resistance of HAZ is weakened due to the martensite tempered in this region. Thus, the mismatch of strength between the FZ and HAZ caused the deformation gradient increased from the FZ to the softened HAZ. Figure 10 shows the asymmetric plastic zones near the crack of KT-FZ. The shape of the plastic zone is similar to the crack instability propagation deviation model by Laukkanen [24]. As can be seen, the plastic zone at the crack tip tends to shift to the softening zone due to local plastic deformation. This is the reason why the cracks in the FZ deviated toward HAZ.
3.4. Microfracture mechanism

The fracture morphology after tensile fracture and Kahn tear test is presented in figures 11 and 12, respectively. The fracture surface of the tensile specimens of DP980 base metal and welded joint exhibit equi-axed dimples with some micro holes between dimples, as shown in figures 11(a), (b). This is a typical feature of ductile fracture. The Kahn fracture surface of KT-BM is similar to the tensile fracture, as shown in figure 12(a), showing an obvious ductile fracture mode. However, the tear fracture surface of the KT-HAZ and KT-FZ specimens are mainly composed of relatively smooth surfaces and tiny dimples. The fracture surface of WJ is smoother compared with KT-BM, as shown in figures 12(b), (c).
To further confirm the microfracture mechanism of Kahn tear tests, the microstructure beneath the fracture surface of the torn specimen was observed, as shown in figure 13. The fracture plane of all the specimens is approximately 45° to the loading direction. For KT-BM, it can be seen that the microcracks tend to propagate along the boundaries of ferrite and martensite, which is the result of interfacial bond damage caused by strain incompatibility between ferrite and martensite [23]. The fracture surface is irregularly serrated, as shown in figure 13(a). Tempered martensite and ferrite are observed in the fracture microstructure of KT-HAZ and KT-FZ specimens, as shown in figures 13(b), (c). Fewer microcracks were found at the interface between ferrite and martensite (compared with the base metal). This is related to the weak deformation resistance of tempered martensite. During the tear process, the grains were elongated along the fracture surface, causing the micropores to be filled by the deformed structure. In addition, the crack path of KT-HAZ and KT-FZ is straighter than that of KT-BM. The profile of the fracture surface is basically consistent with the fracture morphology.

4. Conclusion

(1) The welding of DP980 steel was successfully realized by fiber laser welding. The FZ consisted of fully lath martensite with a higher microhardness than the BM. The tempering of martensite resulted in a softened zone in the subcritical HAZ.

(2) The welded DP980 steel exhibited a continuous yield behavior. The joint showed a superior tensile strength with a joint efficiency of 98%. The strain hardening rate of the joint in the second stage is lower than that of DP980-BM.

(3) The tear toughness of the laser-welded specimen is clearly decreased for DP980 steel. As a result of the asymmetrical plastic zone at the crack tip, the crack initiated from the FZ gradually deviated from its original growth direction after a stage of steady growth toward the HAZ.
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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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