Fiber Laser Welding of Al-Si-Coated Press-Hardened Steel

The mixing and formation of the Al-Si coating’s delta-ferrite phase was the key factor for premature failure of the laser-welded joints

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ABSTRACT

The effect of an Al-Si coating on the microstructure and mechanical properties of fiber laser-welded, press-hardened steel is examined in the present work. It is shown that mixing of the Al-Si coating significantly modified the fusion zone microstructure, which was found to be a combination of martensite and delta ferrite in both the welded as-received and hot-stamped conditions. In contrast, a uniform distribution of alpha ferrite and martensite was observed when the as-received sheet was welded then hot stamped. The ferrite phase was found to be rich in Al and Si, which retarded the kinetics of martensite formation, increasing the ferrite fraction after hot stamping of the welded coupon. Micro- and nano-scale hardness measurements on ferrite suggest it is a softer phase compared to martensite; therefore, inducing strain localization and premature fracture during tensile tests occurred. The strength of all-welded joints was found to be dependent on ferrite phase fractions, since this promoted premature fracture, and deteriorated joint strength and ductility.

KEYWORDS

• Fiber Laser Welds • Press-Hardened Steel • Coating Mixing • Weld Solidification • Delta Ferrite

Introduction

In recent years, automotive industries have reduced vehicle weight by replacing conventional low-strength steels with advanced high-strength steels (AHSSs), and ultrahigh-strength steels (UHSSs). The higher specific strength, thinner gauge of AHSS and UHSS sheets provide superior mechanical properties compared to the other high-strength, low-alloy steels (Ref. 1). By decreasing the thickness of the sheets, the weight of the vehicles can be reduced, improving fuel efficiency and decreasing greenhouse gas emissions to benefit the environment (Refs. 2, 3).

Among the other AHSS, press-hardened steel (PHS), such as the USIBOR® glade, offers high ultimate tensile strength (1500 MPa) in the hot-stamped condition. In the as-quenched state, PHS exhibits high strength and low formability, making it difficult to form complex parts (Ref. 1). Consequently, these steels are intended for a hot stamping process, i.e., forming at the austenite temperature and subsequently quenching to form martensite. In order to prevent surface oxidation and decarburization, PHS sheets are usually coated with various coatings, such as Al-Si, Zn, or Zn-Ni alloys (Ref. 4). Among these coating alloys, Al-Si offers better corrosion and high-temperature oxidation resistance (Ref. 4).

Automotive components commonly made with PHS are A-pillars, B-pillars, door rings, bumpers, roof rails, and tunnels (Ref. 5), all of which are usually made from laser-welded blanks (LWBs), also known as tailor-welded blanks (TWBs). Other property tailoring methods such as partial tempering, partial quenching, and post tempering of the monolithic sheets can be employed (Ref. 6). Utilization of PHS materials combined with LWB technology provides weight saving, material cost reduction, improved part performance, and simplifies the manufacturing process (Refs. 7–9). Fiber laser welding is considered to be preferred in this application compared to other forms of laser sources, considering it offers a small beam diameter, flexible beam delivery, higher precision and accuracy, less thermal distortion, and high-plug efficiency (Refs. 10, 11).

Apart from the chemical composition of steels, the presence of a coating also influences the properties of LWBs. It has been reported that the coated steels are prone to more welding defects such as weld concavity (Ref. 12) and coating mixing into the weld pool (Refs. 13, 14), compared to uncoated steels. One of the major issues associated with using coated PHS is the presence of the Al-Si coating on the surface, which affects the welding process. During laser welding, the coating is mixed into the fusion zone...
(FZ), and thereby changes the local chemistry and equilibrium phases (Ref. 15). Hence, heterogeneous phases can form and adversely affect the mechanical properties (Refs. 14, 16).

To date, limited work has been published regarding fiber laser welding of PHS. Kim et al. (Ref. 13) investigated the laser welding properties of Al-Si-coated PHS and found an intermetallic phase at fusion boundary. Other studies (Refs. 17, 18) also reported intermetallic phase formation in the FZ while welding Al-Si-coated PHS. Studies done on laser welding of Al-Si-coated PHS have claimed that a second phase formed in either the FZ or the fusion boundary was an intermetallic compound, without identifying it; however, it was shown to be related to the presence of Al content.

Recently, Kang et al. (Ref. 19) investigated the influence of beam size and hot stamping parameters (temperature and time) on the FZ microstructure formation. They found that the fraction of martensite increases in the FZ when a larger beam size, higher austenization temperature, and time are considered. Industries are now dealing with this issue by employing a dedicated welding process known as laser ablation (Refs. 17, 18), which uses a short multi-impulse laser to remove the Al-Si coating, leaving behind only an intermediate Fe₂Al₅ and FeAl₃ layer.

The formation mechanisms of producing the second phase in the FZ and their effect on mechanical properties are not well understood. In this study, the effect of the coating will be examined in terms of the formation of various phases in the FZ, and their impact on the microhardness and tensile properties. In the present study, all of the possible processing combinations of PHS joints will be considered, such as as-received welded (ARW), as-received welded then hot stamped (ARWHS), and hot stamped welded (HSW). The properties of individual phases within the FZ will be evaluated using Vickers microhardness and nanoindentation.

**Experimental Methodology**

Fiber laser welds were made on 1.00-mm-thick Al-Si-coated PHS, where the yield strength and ultimate tensile strength of the as-received sheets are about 424 and 570 MPa, respectively. The chemical composition and mechanical properties are listed in Tables 1 and 2, respectively. For hot stamping, as-received sheets were austenized in a furnace at 930°C for 5 min, and then trans-ferred (within 10 s) to flat dies, where quenching was done to ensure fully martensite microstructure (Refs. 1, 21). The cooling rate was measured with a thermocouple welded to the dies, and confirmed to be about 31°C/s, which is identical to the critical cooling rate of martensite formation of PHS as reported in the literature (Refs. 1, 22).

Welds were made using an IPG Photonics ytterbium fiber laser system (model: YLS-6000-S2); the details of the system can be found in previous studies (Ref. 23). The welds were inspected for concavity, porosity, and other defects in accordance with General Motors engineering standards (GM4485M) (Ref. 24). The minimum concavity of the joint was estimated to be within the industrial acceptable range (below 20%).

Welds were made in the bead-on-plate configuration, and the materials, hot stamping conditions, and welding parameters can be found in Table 3. The mechanically ground and polished specimens were etched with 5% Nital solution for 2 s to reveal the microstructure. Microstructures were characterized using an optical microscope and a field-emission scanning electron microscope (FE-SEM, model:...
Zeiss Leo 1550). The equilibrium binary phase diagram of the studied steel was calculated using commercial thermodynamic software ThermoCalc.

Vickers microhardness was measured under 200-g load and 15-s dwell time. In order to avoid interference from the strain fields developed by the adjacent indents, a spacing of at least three times the diagonal was maintained between indents, as recommended by ASTM Standard E384 (Ref. 26). Nanoindentation was performed on the etched specimens using 5000-N load for 20 s using a Hysitron Triboindenter TI-900 equipped with a scanning probe microscope. Transverse tensile specimens were machined as per ASTM Standard E8 (Ref. 27) with the sample dimensions shown in Refs. 28, 29. The tensile tests were carried out using a computerized tensile testing machine at a crosshead speed of 1 mm/min.

Results

Coating and Base Metal Characteristics

Figure 1 shows typical morphology of the coating in the as-received (Fig. 1A) and hot-stamped (Fig. 1B) condition. As-received material has a coating thickness of about 15–20 μm, which was transformed and diffused into the steel to a depth of about 30–40 μm during hot stamping. It was reported that the growth of the coating thickness is highly dependent on the austenitization temperature and time. Both parameters increase the coating thickness (Ref. 4).

Table 2 — Tensile Properties and Coating Compositions of the Studied Steels

<table>
<thead>
<tr>
<th>Conditions</th>
<th>0.2% Yield (MPa)</th>
<th>UTS (MPa)</th>
<th>Elongation (%)</th>
<th>Coating (76 g/m²)</th>
<th>Al (wt-%)</th>
<th>Si (wt-%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>As-received BM</td>
<td>420</td>
<td>570</td>
<td>15</td>
<td>90</td>
<td>10</td>
<td></td>
</tr>
<tr>
<td>As-hot stamped</td>
<td>1122</td>
<td>1601</td>
<td>4</td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

Coating of the as-received steel consists of an Al matrix with elongated pure Si particles (Fig. 1A), which formed by a eutectic reaction during the aluminizing process (Ref. 30). An intermediate layer exists between the coating, and the steel substrate has a thickness of about 4–6 μm (Ref. 31) with a composition of Fe₂SiAl₇.

In addition, a thin layer (<1 μm thickness) of FeAl₃ was also observed between the steel substrate and the Fe₂SiAl₇ layer, as shown in the inset of Fig. 1A (Ref. 30). Energy-dispersive spectroscopy (EDS) line scanning of the coating (Fig. 1C) shows a thin Si-enriched layer at the outer surface of the coating. A multilayered coating is formed during heat treatment (Refs. 4, 30–34), as shown in Fig. 1B and D. A total of five distinct microstructural regions can be found in the coating in the hot-stamped condition, with different morphologies and compositions (Refs. 30, 31, 34).

The microstructure of the as-received steel is comprised predominantly of ferrite, pearlite, and a small amount of martensite at the grain boundaries — Fig. 2A. A typical morphology of lath martensite can be found in the hot-stamped condition — Fig. 2B. Laths are parallel to each other and some of the laths contain distributed carbides, which resulted from the autotempering (Ref. 35).

Microstructure of the Welds

To investigate the influences of the Al-Si coating on the FZ microstructure, the coating was removed chemically by dipping into a solution of 10% sodium hydroxide (NaOH) in deionized water. Figure 3 shows the full weld profile and the FZ microstructure; a fully martensitic microstructure was obtained before and after hot stamping of the welded coupon.

As-Received Welded (ARW)

Figure 4 shows SEM micrographs showing microstructures at different locations of the as-received welded sample (ARW). A — Full weld profile; B — HAZ; C — intercritical HAZ; D — FZ.

As-received welds were made with the welding parameters as specified in Table 3. From Fig. 4, it is noted that the thermal cycle employed during welding has drastically changed the HAZ and FZ microstructures. Although the fiber laser welding process involves rapid heating and cooling, the FZ was not
transformed to complete martensite due to mixing of the coating into the FZ. The FZ microstructure in the coated sheet is a combination of skeletal ferrite and martensite (Fig. 4D), and the area fraction of ferrite and martensite (using ImageJ software) (Ref. 36) was estimated to be about 20 and 80%, respectively. It has been reported that the skeletal pattern ferrite is a typical structure of high-temperature delta-ferrite (Ref. 37, 38); therefore, the ferrite structure found on the ARW sample can be referred to as δ-ferrite. From the solidification pattern, it can be presumed that the δ-ferrite phase was solidified as the first solidified phase; afterward, the martensite phase appeared. The influence of the coating on the FZ microstructure modification is discussed later. The BM contains only ferrite and pearlite, which are already softer than the martensite in the FZ; therefore, no heat-affected zone (HAZ) softening was detected. However, the HAZ of the ARW sample consists of retransformed martensite and ferrite (Fig. 4C) where the peak temperature during welding was between the Ac1 and Ac3 temperatures of the steel.

**As-Received Welded then Hot Stamped (ARWHS)**

The as-received material was welded using similar welding parameters as used for ARW sample (Table 3). The welded coupons were then austenized and quenched between dies, which resulted in the microstructures shown in Fig. 5. It can be noted that after hot stamping, the HAZ microstructure is completely transformed to martensite similar to the hot-stamped BM; therefore, identical microhardness can be expected, since the microhardness is dependent on martensite fraction. It is suggested that the hot stamp-

### Table 3 — Materials, Hot Stamping Conditions, and Welding Parameters Used in This Study

<table>
<thead>
<tr>
<th>Conditions</th>
<th>Power (kW)</th>
<th>Focal Length (mm)</th>
<th>Spot Size (mm)</th>
<th>Speed (m/min)</th>
<th>Stamping Conditions</th>
</tr>
</thead>
<tbody>
<tr>
<td>ARW</td>
<td>4</td>
<td>200</td>
<td>0.6</td>
<td>12</td>
<td>As-received welded</td>
</tr>
<tr>
<td>ARWHS</td>
<td>4</td>
<td>200</td>
<td>0.6</td>
<td>12</td>
<td>As-received welded then hot stamped</td>
</tr>
<tr>
<td>HSW</td>
<td>4</td>
<td>200</td>
<td>0.6</td>
<td>12</td>
<td>Hot-stamped welded</td>
</tr>
</tbody>
</table>
ing process could not transform the steel into a fully martensite structure due to the mixing of the coating alloy into the FZ. The FZ microstructure was found to be a combination of ferrite and martensite. The peak temperature (930°C) in the hot-stamping process is identical to the intercritical HAZ temperature (between Ac1 and Ac3); therefore, during hot stamping, the \(\delta\)-ferrite and martensite structure can transform to ferrite (\(\alpha\)-ferrite) and austenite (\(\gamma\)). This would explain the ferrite phase observed in the FZ of the ARWHS sample, which can be referred to as \(\alpha\)-ferrite. This will be discussed in a later section. The area fraction of the \(\delta\)-ferrite and martensite were estimated to be about 40 and 60%, respectively, compared to 20% \(\delta\)-ferrite, and 80% martensite in the ARW condition — Fig. 4D.

**Hot-Stamped Welded (HSW)**

Unless coating mixing occurs at the fusion boundary, the FZ is also composed of 100% martensite, which has a similar BM lath microstructure — Fig. 6D. However, the martensite laths are clearly thinner and finer in the FZ, which suggests the cooling rate was much faster during welding. The HAZ microstructure can be divided into three distinct zones termed as subcritical HAZ, intercritical HAZ, and supercritical HAZ, depending on the temperature ranges. Among these zones, the subcritical HAZ (Fig. 6C) experienced temperatures between room temperature and the Ac1 temperature, such that the BM martensite structure underwent tempering. During tempering, the martensite structure is decomposed to ferrite and cementite \((\text{Fe}_3\text{C})\), which has a lower strength compared to a fully martensite structure (Refs. 39–42), which degrades the tensile, fatigue, and formability behavior of the welded joints (Refs. 23, 28, 29).

The next zone is the intercritical HAZ, where partial austenization occurred between the temperatures of Ac1 and Ac3 line. The microstructure was found to be a combination of ferrite and retransformed martensite — Fig. 6B. As the temperature increased in the HAZ, the fraction of martensite increased as well, thereby improving the mechanical properties in contrast to the subcritical HAZ. When the local peak temperature during welding exceeds the Ac3 temperature, the microstructure becomes fully martensitic (with the highest micro-hardness) and is therefore referred to as the supercritical HAZ. The FZ structure was found to be fully martensitic, whereas the fusion boundary microstructure is composed of martensite and a network of \(\delta\)-ferrite phase.

During laser beam welding, a higher vapor pressure pushed away the molten metal from the keyhole center (Ref. 43) and the weld metal mixed with the Al-Si coating generated a vortex flow at the upper sides of the molten pool. Therefore, due to limited solubility of the hot-stamped coating to the weld pool, the coating was not mixed with the molten metal homogeneously. Consequently, the local Al concentration was higher at the fusion boundary, which stabilizes the high-temperature ferrite phase (\(\delta\)-ferrite) during rapid solidification.

**Coating Mixing into the Fusion Zone**

In the current studies, it was found that the presence of an Al-Si coating influences the formation of \(\delta\)-ferrite phase in the FZ (Figs. 4–6), since no evidence of \(\delta\)-ferrite was observed in the welded sample where this coating was
chemically removed — Fig. 3. It was also observed that the presence of the coating on either surface may be mixed into the FZ; however, a coating on the top surface of the sheet has the most prominent effect on the δ-ferrite formation. Coating mixing was most noticeable along the fusion boundary, where δ-ferrite was formed as a continuous structure due to the effect of Marangoni convection flow (Refs. 44, 45).

Figures 7 and 8 distinguish the coating mixing and resultant microstructure among the ARW and ARWHS conditions. Although both of the FZ consist of a ferrite (δ-ferrite or α-ferrite) and martensite structure, the martensite fraction is clearly higher in the ARW condition (about 80%). It is worth mentioning here that the higher fraction of martensite is associated with higher heating parameters, i.e., temperature and time, during hot stamping, as reported by Kang et al. (Ref. 19). Their calculation shows that when the Al and Si content increases, the ferrite to austenite phase transformation temperature also increases.

Martensite formed in the FZ of the ARW sample resulted from the higher austenization temperature and rapid cooling (~10,000°C/s (Refs. 23, 46)) during welding, whereas the ARWHS specimen showed a decreased fraction of the martensite (about 40%) because of the lower austenization temperature (930°C) and slower cooling rate (~31°C/s) in the hot-stamping process. Another distinguishable feature is that samples in the ARW condition contain a larger fraction of δ-ferrite along the fusion boundary (Fig. 7D); conversely, a uniform distribution of the δ-ferrite and martensite can be identified throughout the FZ of the ARWHS sample — Fig. 8D.

Unlike the above two cases, the HSW condition exhibits coating mixing only along the fusion boundary — Fig. 9. There is no trace of δ-ferrite at the center of the FZ, where the microstructure was fully martensite with a high microhardness (505 ± 23 HV). A higher fraction of localized δ-ferrite at the fusion boundary indicates that there was limited mixing of the coating in the weld pool. As a result, the local concentration of Al at the fusion boundary was higher compared to the center area of the FZ. The δ-ferrite phase at fusion boundary was found to have a skeletal-like continuous structure (Fig. 9C and D) rather than a grain-like structure — Fig. 8D.

The compositional analysis of the martensite and ferrite phase was performed using the EDS line scanning technique as presented in Fig. 10. In addition, the chemical composition of each phase was measured (Table 4); the result shows rich Al content (1.70 to 2.64 wt-%) in the ferrite phase. The influence of Al content on phase transformation and δ-ferrite formation is predicted using a ThermoCalc Fe-Al binary phase diagram (Fig. 11), and is discussed later. However, Si content was found to be at the same level in all three different sample conditions.

**Mechanical Properties**

**Hardness**

Nanohardness of the individual...
phases of martensite, and ferrite (α-ferrite and γ-ferrite) were further estimated by using instrumented nanindentation tests. The indent impressions from individual grains on each phase in three different samples are presented in Fig. 12. A total of 12 to 16 indents were made on each phase, i.e., on martensite and ferrite in the FZ (ARW and ARWHS sample) and the fusion boundary (HSW sample). The average nanohardness of the martensite and ferrite phase ranges from 7.1 to 7.9 GPa and 4.5 to 4.9 GPa, respectively, for all three sample conditions.

Hardness impression on the martensite islands of the ARWHS sample exhibit the lowest nanohardness (7.1 ± 0.6 GPa) among the martensite phases on the two other samples, which is due to the lower cooling rate in the hot-stamping process and smaller stress-free martensite grain. Conversely, nanohardness measured on the ferrite phase is fairly in the same range, which indicates that the strength of the ferrite phase was not influenced by the hot-stamping process. It should be concluded that the second phase found in the FZ of all material conditions definitely was not consistent with the hardness of an intermetallic compound as reported in earlier literature (Refs. 13, 17, 18). The nanohardness of the intermetallic compound phase (Fe(Al, Si), Fig. 1B) is much higher (10.1 ± 0.5 GPa) compared to the hardness of ferrite (4.1–4.9 GPa) phase.

The microhardness distributions across the three different welded samples are presented in Fig. 13. The as-received and hot-stamped BM hardness values are about 244 ± 4, and 575 ± 8 HV, respectively. Hardness profiles obtained from the ARW and HSW conditions exhibited the maximum hardness in the supercritical HAZ, which was due to the solid-state transformation, rapid cooling, and a larger fraction of the martensite phase compared to other locations. Although there are significant differences in hardness among the as-received and hot-stamped steels, the average FZ hardness of the ARW and HSW samples were similar, about 516 ± 25 and 505 ± 23 HV, respectively.

A larger deviation of the hardness in the FZ was believed to be the effect of the coating mixing in the fusion boundary. For instance, the circular marked point on the hardness profile of the HSW sample (Fig. 13B) indicates the fusion boundary, where there is a higher fraction of network-like ferrite — Fig. 9C. The average micro- and nanohardness of the fusion boundary and ferrite were measured to be about 404 ± 63 HV and 4.5 ± 0.1 GPa, respectively. Unlike the HAZ of the ARW condition, the HAZ of the HSW sample undergoes tempering because of the martensite microstructure present in its BM. It has been reported that tempering of martensite has reduced the strength of the joint in all types of welding processes (Refs. 23, 28, 29).

In this study, it was found that the HAZ hardness was reduced by about 35% (from 575 ± 8 HV (BM) to 374 ± 25 HV (HAZ)) compared to the hot-stamped BM hardness. Similar softening was reported by Jia et al. (Ref. 47) in hot-stamped welded 22MnB5 steel. In addition, they also found that with increasing welding speed, the HAZ and FZ hardness increases due to lower heat input. While comparing hardness distribution of the ARWHS sample with other conditions, there is no softening or variation of the hardness between BM and HAZ. Conversely, a sharp drop in hardness was measured in the FZ (Fig. 13A); the BM hardness was 575 ± 8 HV, which decreased to 394 ± 24 HV, representing about a 32% reduction of hardness, in contrast to the BM and HAZ. The degree of softening in the FZ of the ARWHS sample was the same as that observed due to the HAZ softening in the HSW sample — Fig. 13B.

**Tensile Properties and Failure Locations**

Typical engineering stress vs. strain curves of the BMs and three different welded samples are presented in Fig. 14. The corresponding mechanical properties and the failure macrographs of the joints can be found in Table 5 and Fig. 15, respectively.

The hot-stamped BM (HSBM) shows the highest yield strength (1122 MPa) and ultimate tensile strength (1601 MPa), which can be expected from the
fully martensitic microstructure (Fig. 2B). While comparing the as-received BM (ARBM) with its welded condition (ARW), there is a negligible difference in the yield strength and the ultimate tensile strength. However, the as-welded sample shows a slight increase in the ultimate tensile strength due to the formation of martensite in the FZ (Fig. 4D), which exhibited higher microhardness — Fig. 13. It can be expected that a higher fraction of martensite in the FZ will reduce the joint elongation as half of the joints were yielded; therefore, necking will likely occur in either one half of the joint, and as a result, the joint elongation is reduced.

In both of the cases, the welded joints failed at the BM (Fig. 15A), which suggests the welding condition does not influence the mechanical properties of the joints. When the samples were hot-stamped after welding (ARWHS samples), the ductility of the joint was reduced dramatically, which is due to the fact that the ferrite and pearlite in the BM (Fig. 2A) were transformed completely to martensite; the FZ microstructure remained a mixture of ferrite and martensite — Fig. 5D.

Tensile test results of the ARWHS condition exhibited the lowest elongation (0.77%) and lowest yield strength (1050 ± 60 MPa) and ultimate tensile strength (1131 ± 73 MPa) compared to other hot-stamped conditions; a 30% reduction of joint efficiency was estimated in this case (Table 5). Tensile fracture occurred consistently between the fusion boundary and FZ — Fig. 15C. Conversely, the HSW condition shows higher tensile properties (joint efficiency 82%) compared to the other welded conditions, which is due to the formation of a larger fraction of martensite in the FZ. All of the welded hot-stamped steel reached a maximum of 0.85% elongation. As explained in the earlier section, in this case, coating mixing occurs only at fusion boundary, which acted as a site for strain localization, where the softer ferrite phase promoted fracture to occur in this region and reduced ductility. This is clearly noted by the negligible necking observed in Fig. 15C and D. Although the subcritical HAZ has the lowest microhardness value (374 ± 25 HV) (Fig. 13), the soft ferrite has the lowest nanohardness (4.5 to 4.9 GPa) compared to the martensite (Fig. 12) and tempered martensite phase (Ref. 48).

**Discussion**

**Coating Mixing and Weld Solidification**

Al-Si coating mixing was detected for all three different weld conditions as confirmed by the composition analysis (Fig. 10 and Table 4). The results exhibited a rich AI percentage at the ferrite phase compared to the martensite phase. However, the Si concentration did not vary significantly among the two phases, with all three samples indicating similar uniform Si concentration profiles. A higher Al concentration is observed in the ARW (Fig. 10A and B) and HSW samples (Fig. 10E and F); conversely, the Al distribution did not fluctuate much in the ARWHS sample — Fig. 10C and D.

While comparing the FZ microstructures of ARW and ARWHS samples, it can be noted that martensite grains are larger in the ARW sample (Fig. 4D), which is also due to the effect of heating parameters and cooling rate. The morphology of the martensite structure is also different. In the ARW sample, the nucleation and growth of the martensite was faster, increasing the grain size; an elongated slender martensite was observed after hot stamping — Fig. 5D.

**Table 5 — Mechanical Properties of the Joints of Three Different Conditions**

<table>
<thead>
<tr>
<th>Conditions</th>
<th>Yield Strength (MPa)</th>
<th>Ultimate Tensile Strength (MPa)</th>
<th>Elongation (%)</th>
<th>Joint Efficiency (%)</th>
<th>Fracture Locations</th>
</tr>
</thead>
<tbody>
<tr>
<td>ARW</td>
<td>412 ± 10</td>
<td>579 ± 16</td>
<td>15.6</td>
<td>101</td>
<td>BM</td>
</tr>
<tr>
<td>ARWHS</td>
<td>1050 ± 60</td>
<td>1131 ± 73</td>
<td>0.77</td>
<td>71</td>
<td>FB-FZ</td>
</tr>
<tr>
<td>HSW</td>
<td>1140 ± 8</td>
<td>1307 ± 15</td>
<td>0.85</td>
<td>82</td>
<td>FB</td>
</tr>
</tbody>
</table>

BM: base metal, FB: fusion boundary, and FZ: fusion zone

content in the ferrite decreases due to diffusion of Al to the bulk material. In the current investigation, a similar phenomenon was observed. Al content was found to decrease from 2.52 to 1.70 wt-% after hot stamping of the ARW sample (Table 4). Since Al is a strong ferrite stabilizer, an increase in Al can shrink the single-phase austenite region and promote the formation of a two-phase region (Ref. 49). According to the calculated phase diagram presented in Fig. 11, during cooling, a ferrite phase should be formed (L → + L) first from the liquid phase (Thermo-Calc Scheil solidification calculation also confirmed that the ferrite would be the first phase); afterward, the liquid (L) phase is completely transformed into austenite through a peritectic reaction (L + L → + ).

The compositional analysis (Fig. 10 and Table 4) shows a higher percentage (1.7 to 2.52%) of Al on the ferrite phase. It was reported that an Al content of more than 1.2 wt-% is sufficient to prevent the austenitic transformation (Ref. 50). The Fe-Al binary phase diagram (Fig. 11) indicates if the Al content exceeds 0.78%, it prevents single austenite phase formation, and if higher than 1.05 wt-%, then it stabilizes only ferrite phase. Therefore, all three different samples show a combination of martensite and ferrite structure in either the FZ or the fusion boundary.

While comparing the FZ microstructures of ARW and ARWHS samples, it can be noted that martensite grains are larger in the ARW sample (Fig. 4D), which is also due to the effect of heating parameters and cooling rate. The morphology of the martensite structure is also different. In the ARW sample, the nucleation and growth of the martensite was faster, increasing the grain size; an elongated slender martensite was observed after hot stamping — Fig. 5D.

The ferrite phase in the ARW sample shows an irregular shape, which seems to be highly stressed by the surrounding martensite. It can be noted that while a martensite plate forms and grows, it introduces a large compressive stress on the surrounding ferrite phase (Ref. 35). When the highly stressed ferrite phase was austenitized at 930°C for 5 min, the compressive stress was released and an unstressed ferrite phase was formed via recrystallization.
**Structure-Property Correlation**

This section will deal with the hotspot stamped version of the steels, i.e., AR-WHS and HSW specimens. In both cases, the minimum hardness was estimated to be at the FZ and subcritical HAZ, respectively. It was also observed that hardness degraded in a similar amount (ARWHS: 32% and HSW: 35%) in contrast to the BM — Fig. 13. In addition, hardness also dropped at the fusion boundary due to coating mixing and formation of ferrite, which accounts for a 30% reduction of the hardness compared to the BM. In addition to concavity, which acts as a physical crack initiation zone, the large variation of hardness between the HAZ and FZ will effectively provide a microstructural notch during tensile tests. It was interesting to observe that failure did not occur at the subcritical HAZ where the highest hardness drop (35%) was measured, but rather the origin of failure occurred at the fusion boundary. This is a result of the continuous network structure of the fusion boundary ferrite, which introduces a larger metallurgical notch and promotes failure.

Therefore, the HSW sample experiences failure along the fusion boundary and leaves the FZ completely intact. As in the HSW sample, the Al-Si coating mixing predominantly occurred on the top surface and extended to half of the sheet thickness; therefore, it can be suggested that half of the weld thickness was fractured along the soft ferrite as suggested by the fracture path (point 1, Fig. 15D) and the other half failed by tearing of the martensite phase (point 14, Fig. 15D) and the other half failed by tearing of the martensite phase (point 1, Fig. 15D). Similarly, the fracture and martensite phase dual structure shows the lowest microhardness (about 394 ± 24 HV) at the FZ.

**Conclusions**

The microstructure and the mechanical properties of fiber laser-welded PHS were characterized considering the effect of Al-Si coating mixing in the weld pool. The coating mixing role on weld solidification was examined and the major findings of this study can be summarized as follows:

1. The FZ microstructure was modified due to the effect of coating mixing. Significant coating mixing occurred in the FZ of the ARW sample, and promoted ferrite formation. Therefore, the FZ microstructure was found to be a combination of ferrite and martensite. Conversely, the HSW sample indicates a continuous network-like ferrite at the fusion boundary, and a fully martensitic microstructure in the center of the FZ.

2. A higher percentage of Al prompted ferrite formation. In addition, ferrite formation was also influenced by the hot stamping process. Lower cooling rate increases the ferrite formation; therefore, the fraction of martensite was found to be decreased after hot stamping of the ARW sample (i.e., ARWHS sample). In this case, the FZ microstructure is a combination of ferrite and martensite.

3. Vickers microhardness and nanoindentation on the fusion boundary and ferrite revealed significantly lower hardness values, 404 ± 63 HV and 4.5 to 4.9 GPa, respectively, as compared to other regions of the weldment. Likewise, the ferrite and martensite dual phase structure shows the lowest microhardness (about 394 ± 24 HV) at the FZ.

4. The strength of the fiber laser-welded PHS is dependent on the fraction of ferrite phase formation, the microhardness decreased by about 32% as a result of coating mixing and leading to failure across the FZ in the ARWHS sample, whereas the HSW specimen failed at the fusion boundary. Therefore, an insignificant effect of martensite tempering on the joint performance was observed.

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