



# Improved Weldability for Laser Welding Press-Hardened Steel via Thinning Coating — Part 2: Al Enrichment Inhibition Technique

**Al-Si-coated PHS increased intermetallic formation and prevented Al enrichment ensuring stable welds**

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## Abstract

Our previous work noted that thinning the coating significantly improves the weldability of Al-Si-coated press-hardened steel (PHS). However, an Al enrichment formed at the weld toe of the thin ( $\sim 16\ \mu\text{m}$ ) coating PHS during laser welding at a high welding speed of 7 m/min, originating from the coating outside the weld. In this work, a surface pre-alloying (SPA) technique was applied to the thin Al-Si coating near the weld toe before laser welding. After conducting the SPA process, the fraction of intermetallic compounds (IMCs) in the coating increased significantly, and the IMCs consisted of  $\text{Fe}_2\text{SiAl}_7$  ( $\tau_5$ ),  $\text{Fe}_2\text{Al}_5$  ( $\eta$ ), and  $\text{Fe}_3\text{Al}_2\text{Si}_3$  ( $\tau_1$ ). When the energy input to SPA exceeded 0.48 kJ/cm, the fraction of IMCs in the coating exceeded 60%. This significantly increased the coating's liquidus temperature near the weld toe due to the altered chemical composition, reducing the molten amount of the coating and preventing it from entering the weld through the weld toe. In this case, the Al enrichment at the weld toe disappeared after high-speed welding, and the tensile joints failed at the base metal. Eventually, the thin Al-Si coating PHS was stably welded at a high welding speed because of the application of SPA.

## Keywords

- Surface Pre-Alloying
- Al Enrichment
- Press-Hardened Steel
- Thin Al-Si Coating
- High-Speed Laser Welding

## Introduction

With the rapid development of automotive lightweighting, Al-Si-coated press-hardened steel (PHS) is widely used in integrated door rings (Refs. 1, 2). Laser welding is employed to join the components of integrated door rings due to the low heat input and high production efficiency (Ref. 3). However, the Al-Si coating is prone to entering the weld during welding, leading to the formation of ferrite, which deteriorates the mechanical properties of the welded joints (Refs. 4–6). To mitigate the negative effects of ferrite, measures such as removing the coating using a nanosecond-pulsed laser or filling the wire with austenitic-stabilizing elements are adopted (Refs. 7–10). However, these methods are often associated with high costs or complex procedures, and they yield low welding speeds.

Recently, we found that thinning the coating can significantly improve the weldability of Al-Si-coated PHS. In this case, the extra-thin ( $\sim 10\ \mu\text{m}$ ) coating PHS showed a wider welding speed range of 3–7 m/min, whereas the thin ( $\sim 16\ \mu\text{m}$ ) coating PHS could only achieve good tailor welding at 3–5 m/min. When the welding speed was increased to 7 m/min, an Al enrichment formed at the weld toe of thin ( $\sim 16\ \mu\text{m}$ ) coating PHS, seriously deteriorating the weld properties (Ref. 11). However, during the Al-Si coated PHS production process, the control of high precision thickness for the extra-thin ( $\sim 10\ \mu\text{m}$ ) coating was difficult, which had strict requirements for hot-dipping and air-blasting processes,

resulting in a high production cost (Ref. 12). In contrast, the  $\sim 16\text{ }\mu\text{m}$  thickness of the coating was easier to control. The production cost was relatively low, thus the PHS with a  $\sim 16\text{ }\mu\text{m}$  coating was widely applied. Additionally, high-speed laser welding increased production efficiency and reduced costs. Therefore, high-speed laser welding of the PHS with a thin coating ( $\sim 16\text{ }\mu\text{m}$ ) was significant in fabricating the door rings.

It has been demonstrated that the Al enrichment at the weld toe of thin ( $\sim 16\text{ }\mu\text{m}$ ) coating PHS originated from the coating outside the weld by selectively removing the coating (Ref. 11). In this work, a surface pre-alloying (SPA) technique

was applied to the Al-Si coating outside the weld before laser welding to modify the Al enrichment. The thin ( $\sim 16\text{ }\mu\text{m}$ ) PHS coating was laser-welded at a high welding speed of 7 m/min, and the microstructural evolution of the coatings and the metallurgical behavior of the welds under various conditions were systematically studied.

### Experimental Procedure

The experimental material utilized was PHS plates (1.2 mm) with a thin ( $\sim 16\text{ }\mu\text{m}$ ) Al-Si coating. The chemical compositions of PHS and Al-Si coating are listed in Table 1.

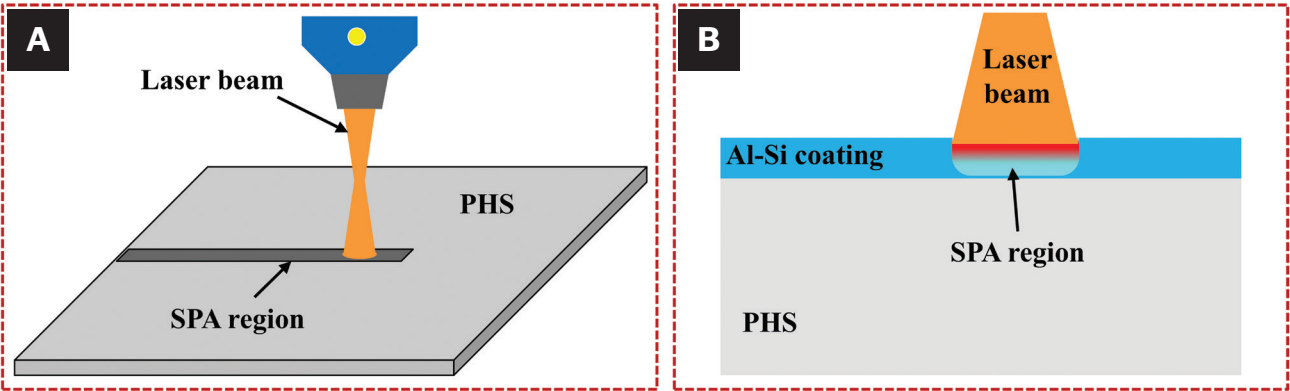


Fig. 1 – A – The schematic illustrations of the SPA process; B – cross sectional view.

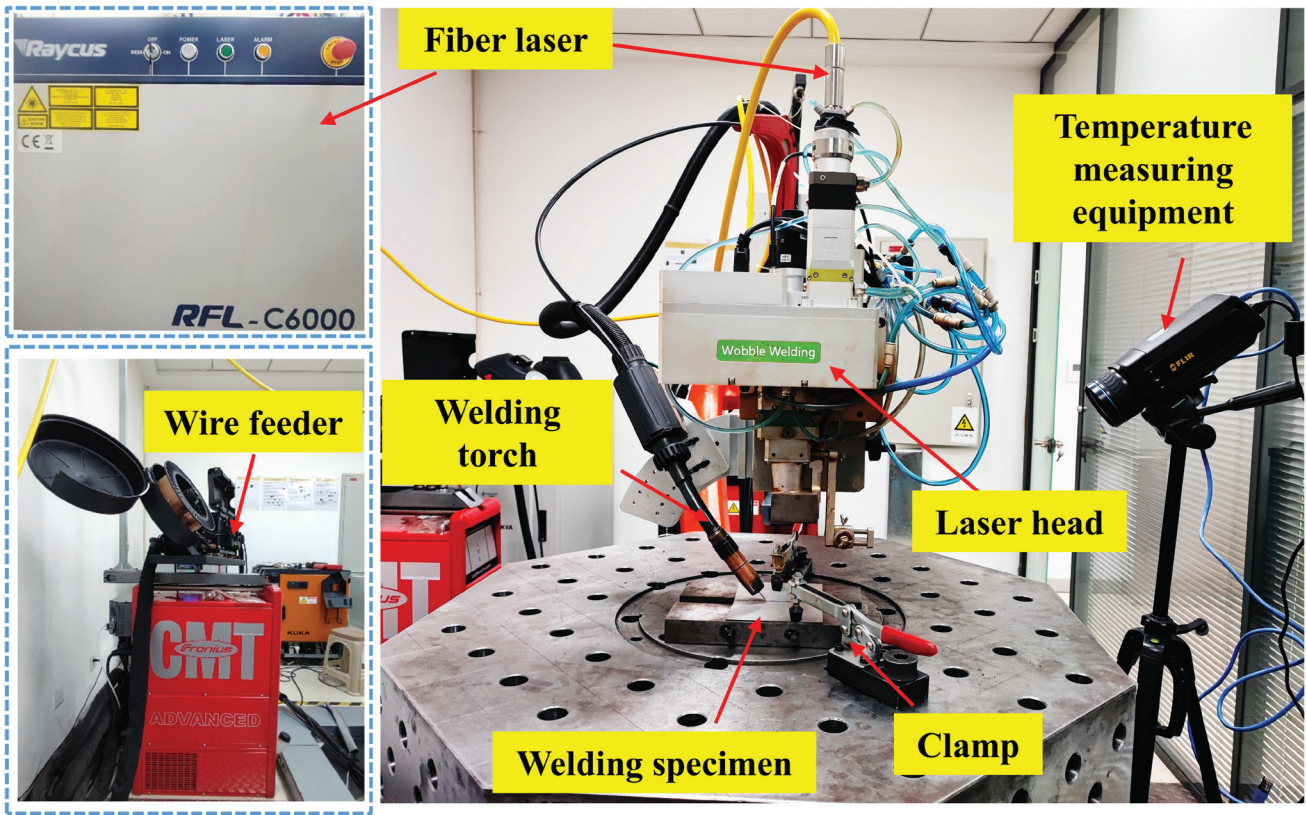


Fig. 2 – Laser welding system.

The SPA process was conducted via an RFL-C 6000 fiber laser, as shown in Fig. 1A. A defocused Gaussian laser beam was used to increase the SPA area. The scanning speed ( $v$ ) and spot radius ( $r$ ) were 5 m/min and 0.5 mm, respectively. The laser power ( $P$ ) was set to 3, 4, and 5 kW, and the corresponding energy inputs ( $Q$ ) were determined by equation 1 to be 0.36, 0.48, and 0.60 kJ/cm, respectively. Pure argon was used as the shielding gas at a flow rate of 25 L/min. During the SPA process, the surface coating of PHS was rapidly heated by a high-speed moving laser, as illustrated in Fig. 1B.

$$Q=\frac{P}{v} \tag{1}$$

Figure 2 shows the laser welding process. A butt-joint configuration was adopted and welded by an RFL-C 6000 fiber laser and a Fronius wire feeder. The fiber laser wave-

length was 1064 nm. The constant welding parameters were optimized with a laser power of 5 kW, a welding speed of 7 m/min, a wire feeding rate of 1.2 m/min, and a defocus distance of -5 mm. The surface focusing position of the focusing mirror was located 300 mm above the steel surface, and the defocus of -5 mm meant that the welding head was moved down 5 mm from the surface focusing position. The filler wire was a C-Mn wire consisting of 0.3 wt.-% C, 0.5 wt.-% Mn, and the balance was Fe. After welding, the tailor-welded blanks were heated in a 950°C muffle furnace for 5 min and then quenched to room temperature in a flat die at a cooling rate of 60°C/s. The temperatures of the SPA and welding process were recorded using a thermal imager (FLIR A400, USA) at 50 frames per second (fps), which could obtain the temperature information at any location within the field of view. The Factsage software was used to calculate the Fe-Al-Si phase diagram.

The cross sectional metallographic samples of the coatings and welds before and after press hardening were cut perpendicularly to the welding direction. The microstructure

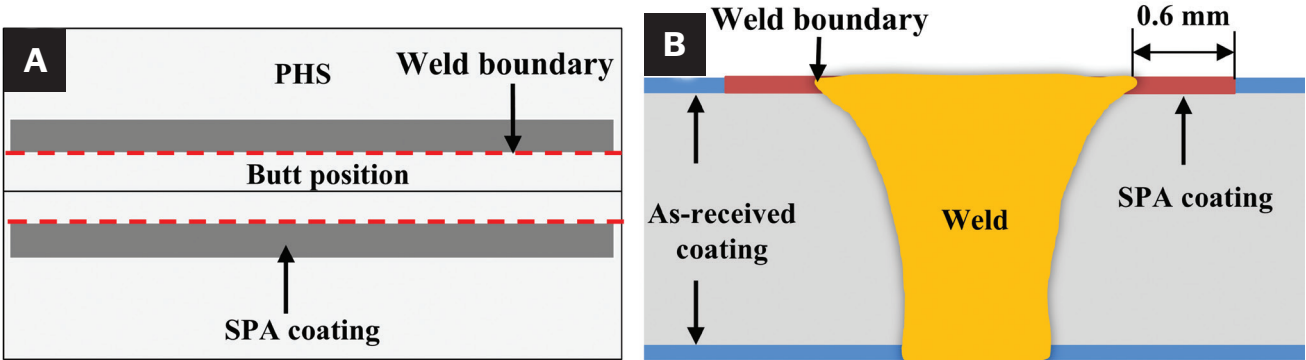


Fig. 3 – The schematic diagrams: A – SPA coating PHS before welding; B – the cross section of SPA coating PHS weld.

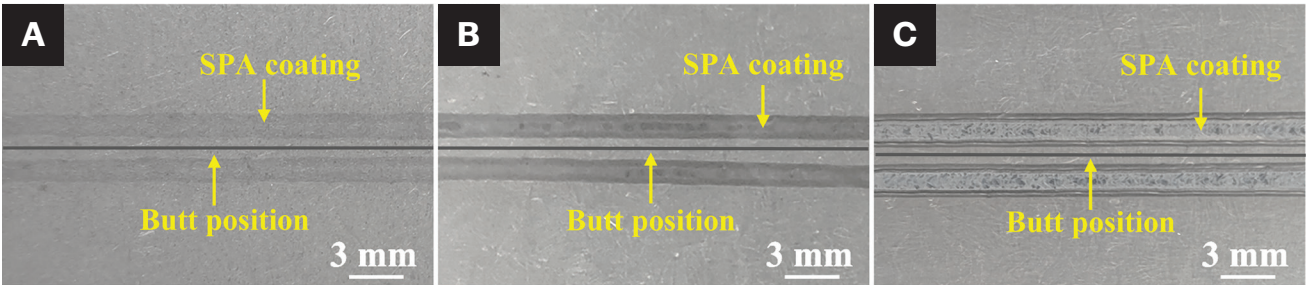


Fig. 4 – The surface images of SPA coatings at different energy inputs: A – 0.36 kJ/cm; B – 0.48 kJ/cm; C – 0.60 kJ/cm.

| Materials     | C    | Mn   | Al    | Si    | Ni | Cr   | B      | P     | S     | Fe   |
|---------------|------|------|-------|-------|----|------|--------|-------|-------|------|
| PHS           | 0.23 | 1.18 | 0.034 | 0.22  | –  | 0.16 | 0.0025 | 0.025 | 0.001 | Bal. |
| Al-Si coating | –    | –    | 88.00 | 12.00 | –  | –    | –      | –     | –     | –    |

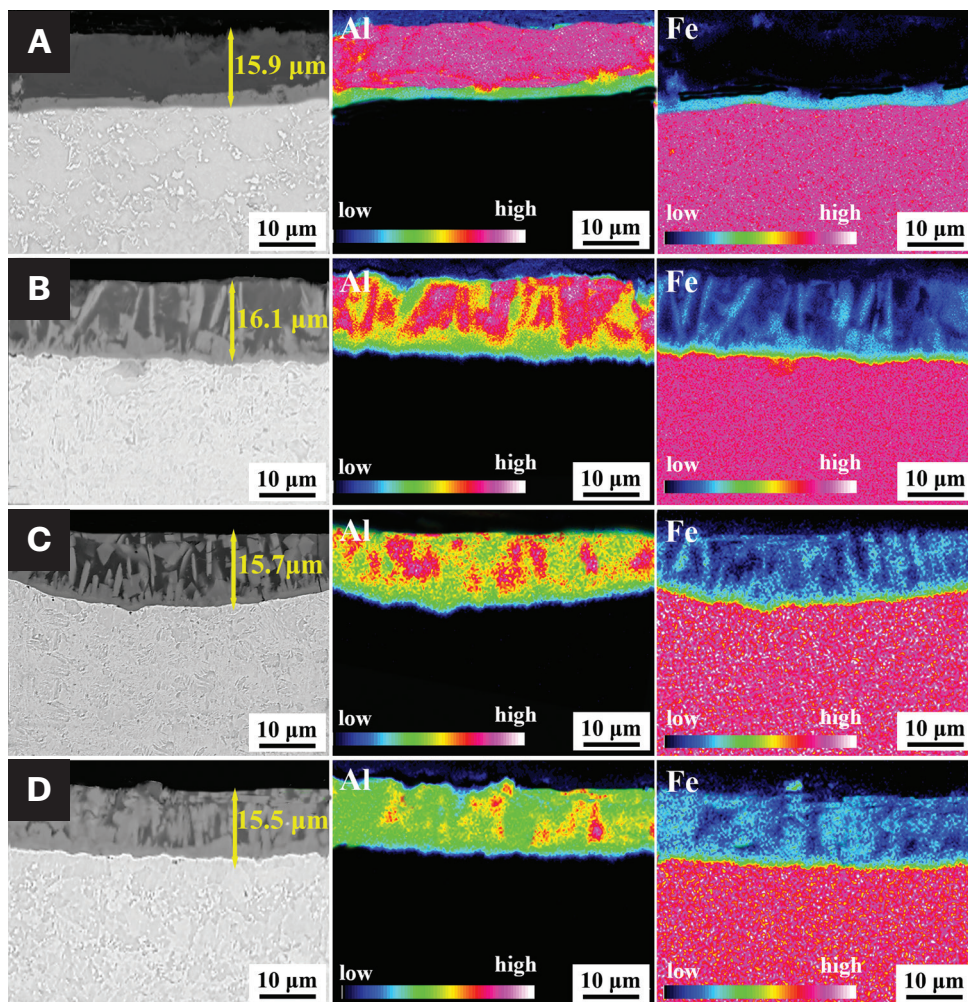


Fig. 5 — Backscattered images and EPMA element maps of the coatings: A — As-received coating; B — SPA coating at energy input of 0.36 kJ/cm; C — SPA coating at energy input of 0.48 kJ/cm; D — SPA coating at energy input of 0.60 kJ/cm.

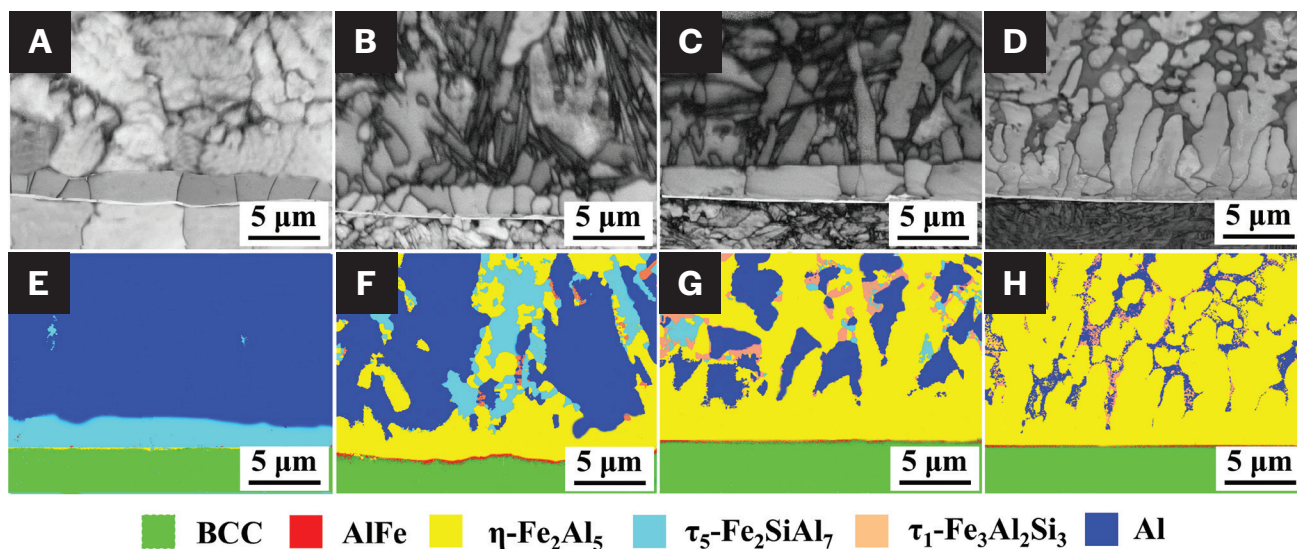


Fig. 6 — The EBSD-band contrast (BC) images, and phase maps of the coatings: A and E — as-received coating; B and F — SPA coating at energy input of 0.36 kJ/cm; C and G — SPA coating at energy input of 0.48 kJ/cm; D and H — SPA coating at energy input of 0.60 kJ/cm.

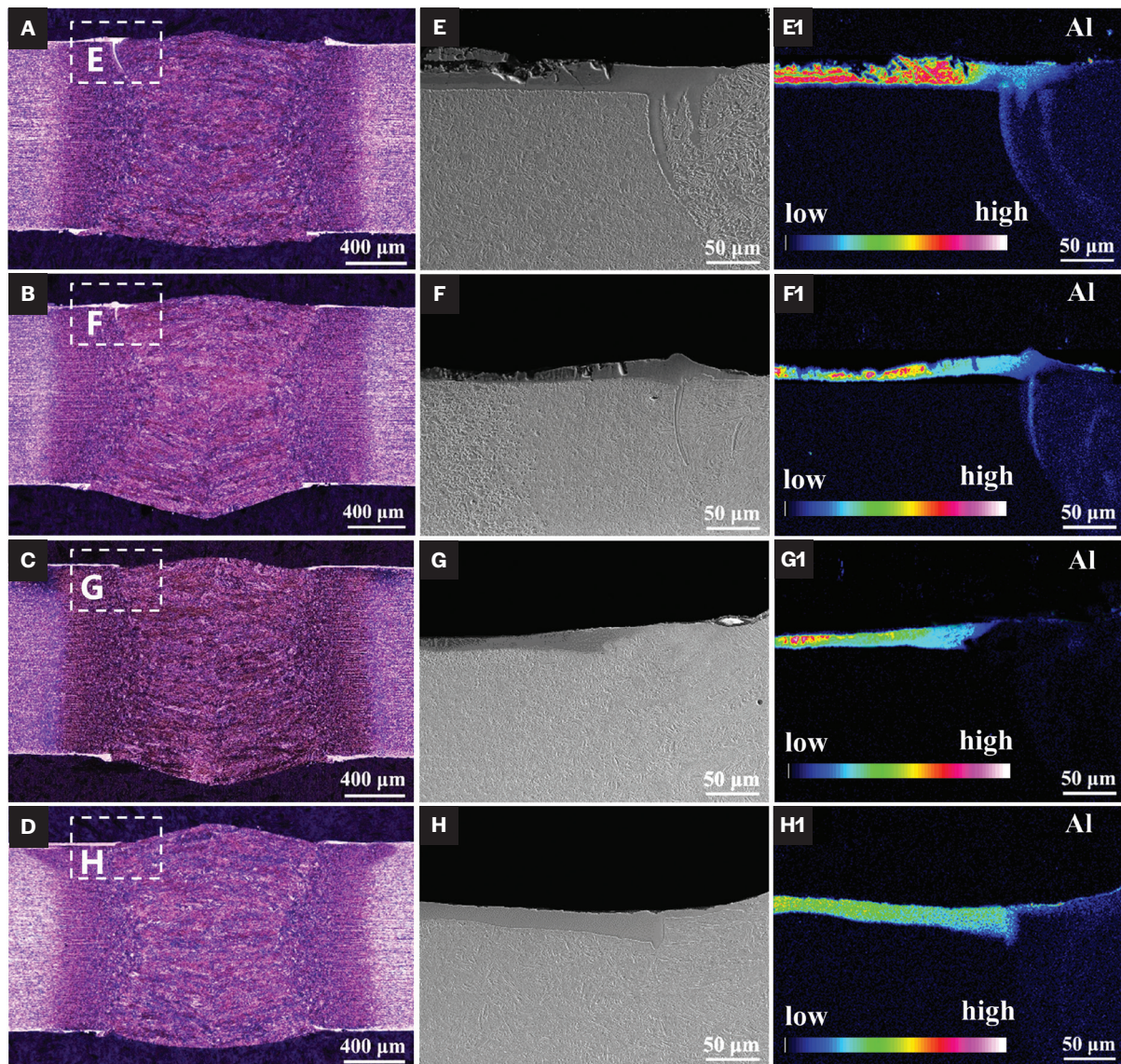


Fig. 7 — A–D — The OM images of the PHS joints; E–H — the SEM images of the weld toes; E1–H1 — the Al distribution maps of the weld toes; A, E, and E1 — as-received coating; B, F, and F1 — SPA30 coating; C, G, and G1 — SPA60 coating; D, H, and H1 — SPA90 coating.

was characterized using Olympus DSX500 optical microscopy (OM) and a Zeiss Ultra-55 scanning electron microscope (SEM) equipped with an electron backscatter diffraction (EBSD) detector. The element distribution in the coatings and welds was analyzed by JEOL-8530F electron probe micro-analyzer (EPMA). The coating heating process was recorded using laser-scanning confocal microscopy (LSCM). The OM, SEM, and EPMA specimens were mechanically polished and etched using 4% nital solution for 5 s. The LSCM samples were mechanically polished. The EBSD specimens were prepared by electro-polishing with a 90% ethanol solution and 10% perchloric acid at 25 V for 20 s. After press-hardening, the tensile specimens of the joints were machined vertically to the welding direction. Tensile tests were performed at an initial strain rate of  $1.0 \times 10^{-3} \text{ s}^{-1}$  using a GOTECH AI-7000 tensile machine at room temperature. The displacement of the gauge section was measured using a TRViewX video

extensometer. Three samples were tested for each condition, and their average value was applied.

## Results

### The Microstructure of the SPA Coatings

According to part 1 of the study (Ref. 11), the Al enrichment at the weld toe obtained at a welding speed of 7 m/min was from the coating outside the weld. To modify the Al enrichment, the coating outside the weld was subjected to a SPA treatment before welding, as shown in Fig. 3. Figure 4 shows surface images of the SPA coatings at different energy inputs. It was clear that the SPA coatings darkened, and the coating color variation became more pronounced as the SPA energy input increased.

**Table 2 — The Phase Fraction in Al-Si Coatings Under Various Conditions**

| Coatings                    | Al-Si (%) | $\tau_5$ -Fe <sub>2</sub> SiAl <sub>7</sub> (%) | $\eta$ -Fe <sub>2</sub> Al <sub>5</sub> (%) | $\tau_1$ -Fe <sub>3</sub> Al <sub>2</sub> Si <sub>3</sub> (%) | AlFe (%) |
|-----------------------------|-----------|-------------------------------------------------|---------------------------------------------|---------------------------------------------------------------|----------|
| As-Received Coating         | 85.4      | 13.4                                            | 1.2                                         | —                                                             | —        |
| SPA Coating<br>(0.36 kJ/cm) | 69.4      | 12.9                                            | 15.4                                        | 0.6                                                           | 1.7      |
| SPA Coating<br>(0.48 kJ/cm) | 41.5      | 2.4                                             | 52.5                                        | 2.1                                                           | 1.5      |
| SPA Coating<br>(0.60 kJ/cm) | 9.7       | —                                               | 87.4                                        | 1.5                                                           | 1.4      |

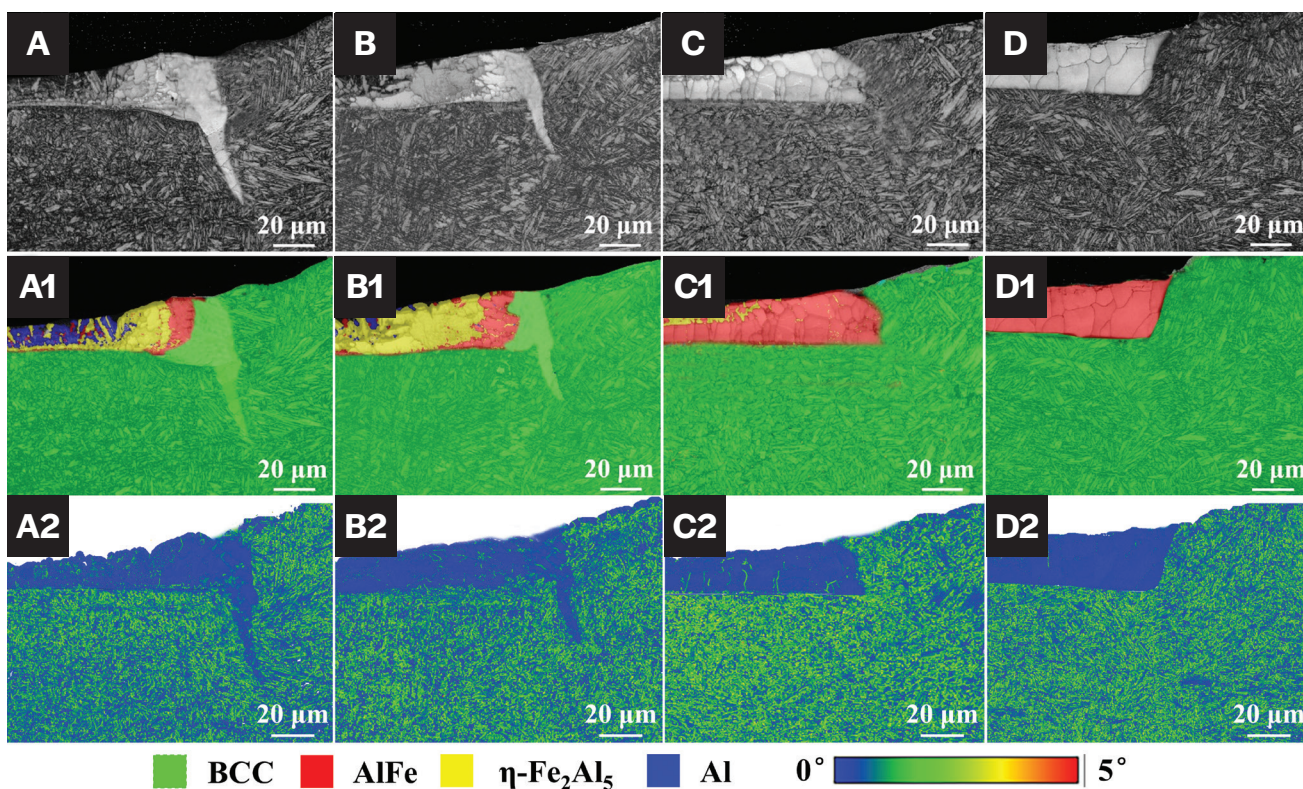


*Fig. 8 — The EBSD-BC, BC + phase, and KAM images of the PHS weld toes: A-A2 — As-received coating; B-B2 — SPA30 coating; C-C2 — SPA60 coating; D-D2 — SPA90 coating.*

To investigate the effect of the SPA treatment on the microstructure of the coating, EPMA backscattered electron imaging and elemental mapping were performed on the coatings under various conditions, as shown in Fig. 5. For the as-received coating, a continuous gray layer (2~3 μm) appeared between the Al-Si alloy and steel substrate, which was enriched in Fe element and the Al content in the region was low (Fig. 5A). After the SPA process, the area of Fe-enriched regions within the coating increased, which continued to rise as the SPA energy input increased (Figs. 5B–D). Additionally, the morphologies of the Fe-enriched regions in SPA coatings were irregular, unlike the

multilayered structure observed in coatings obtained via conventional heat treatment.

To confirm the phase constituents and distributions of the coatings under various conditions, the EBSD investigations were carried out, and the results are shown in Fig. 6. For the as-received coating, the continuous Fe-enriched layer between the Al-Si alloy layer and steel substrate was composed of an ultra-thin Fe<sub>2</sub>Al<sub>5</sub> (η) layer and a Fe<sub>2</sub>SiAl<sub>7</sub> (τ<sub>5</sub>) layer with a thickness of 2~3 μm (Figs. 6A and E), indicating that these were intermetallic compound (IMC) layers. When the energy inputs of the SPA process were 0.36 and 0.48 kJ/cm, the continuous IMC layers at the bottom of the coatings transformed into an ultra-thin AlFe layer and an η layer. Furthermore, the

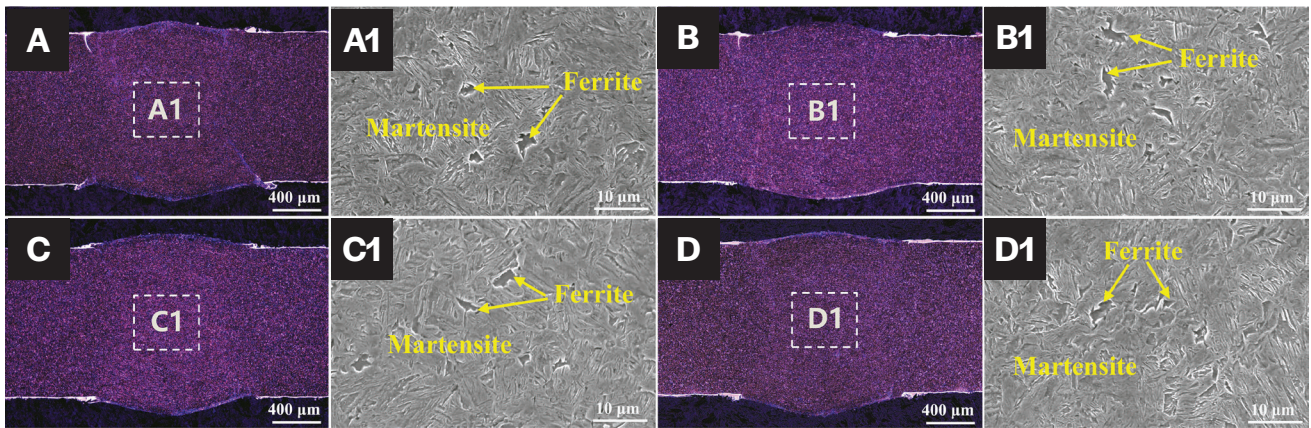


Fig. 9 – The OM macrographs and the SEM images of the PHS joints after press-hardening: A and A1 – As-received coating; B and B1 – SPA30 coating; C and C1 – SPA60 coating; D and D1 – SPA90 coating.

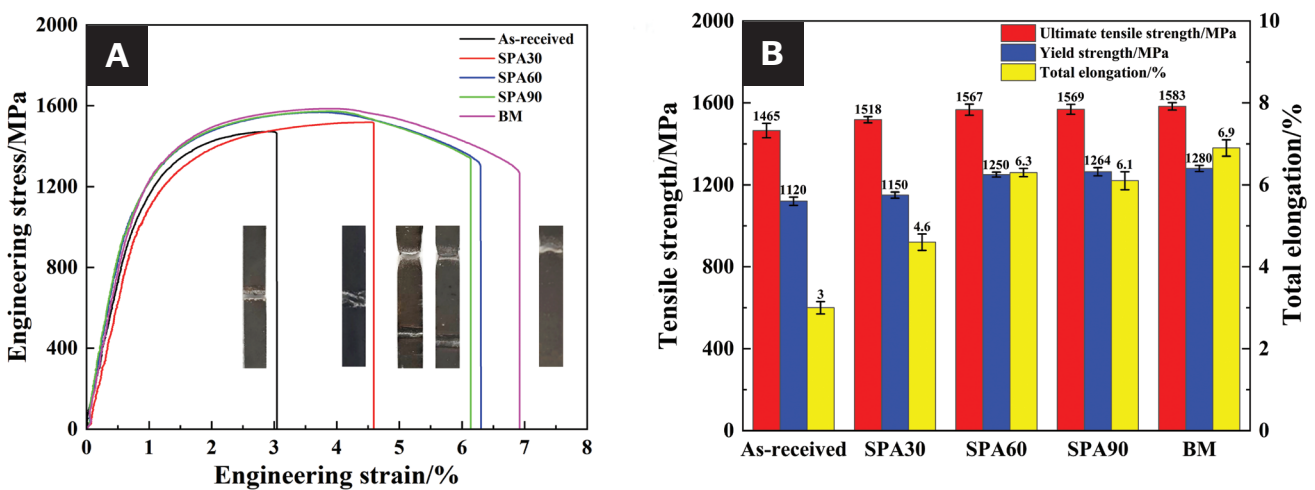


Fig. 10 – A – Engineering stress-strain curves of the BM and joints under various conditions; B – tensile properties of the BM and joints under various conditions.

$\eta$  layer of the SPA coating at energy input of 0.48 kJ/cm was thicker than that of the SPA coating at energy input of 0.36 kJ/cm. In addition to the continuous IMC layer, the Fe-enriched regions within the Al-Si alloy were also determined to be IMCs. According to the EBSD results, the IMCs consisted of  $\eta$ ,  $\tau_5$ , and  $\text{Fe}_3\text{Al}_2\text{Si}_3$  ( $\tau_1$ ), which were unordered and showed different morphologies (Figs. 6F and G). However, as the energy input of the SPA process increased to 0.60 kJ/cm, the continuous IMC layer at the bottom of the coating disappeared, and IMCs in the coating contained  $\eta$  phase with columnar or block structures, as well as  $\tau_1$  phase (Figs. 6D and H). Moreover, the fractions of different IMCs within the SPA coatings could be obtained by EBSD detection, and the statistical results are shown in Table 2. When the SPA energy inputs were 0.36, 0.48, and 0.60 kJ/cm, the total fractions of the IMCs within the coatings were 30.6%, 58.5%, and 90.3%, respectively. According to the IMCs fractions, the SPA degrees were defined as 30%, 60%, and 90%, i.e., SPA30, SPA60, and SPA90. As the SPA degree increased from 30% to 90%, the fraction of  $\tau_5$  reduced, and the fraction of  $\eta$  increased. Meanwhile, the  $\tau_1$  phase was observed at the grain boundaries of the  $\eta$  phase.

## Microstructure and Properties of PHS Joints

The as-received and SPA coatings PHSs were laser welded at a high welding speed of 7 m/min. To study the influence of the SPA process on the weld microstructure, OM imaging, SEM imaging, and elemental mapping were conducted on the welds under various conditions, as depicted in Fig. 7. Notably, all welds were well-formed without any defects (Figs. 7A–D). However, a strip-like phase was observed at the weld toes of both the as-received and SPA30 coating PHSs (Figs. 7A and B). Moreover, the strip-like phase at the weld toe of the SPA30 coating PHS was notably smaller than that of the as-received coating PHS (Figs. 7E and F). Based on the Al distribution maps in Figs. 7E1 and F1, there was Al enrichment in the strip-like phase. Surprisingly, there was no strip-like phase with a high Al content at the weld toes of the SPA60 and SPA90 coating PHSs. The Al enrichment phenomenon at the weld toe gradually weakened as the SPA degree increased, and it disappeared when the SPA degree exceeded 60%.

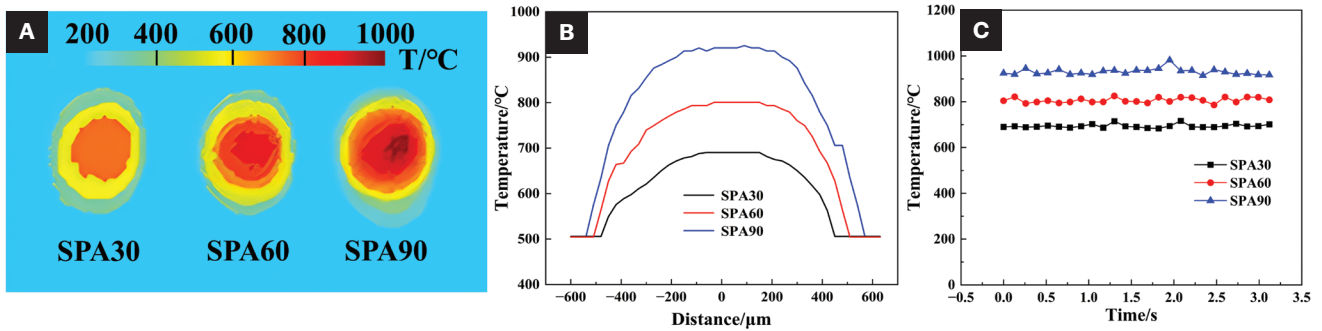


Fig. 11 – A – Temperature map distribution; B – temperature distribution in the width direction; C – peak temperature of the SPA process under various conditions.

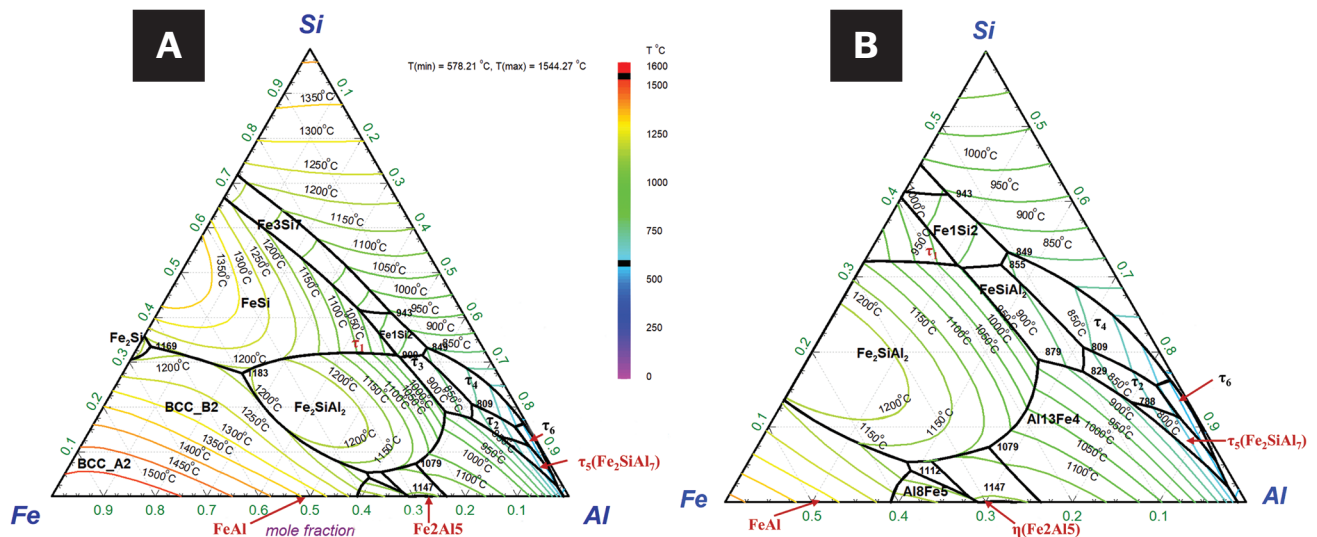


Fig. 12 – A – Fe-Al-Si ternary phase diagram; B – enlarged view of Al corner.

Figure 8 shows EBSD-BC, BC + phase, and kernel-average misorientation (KAM) images of the weld toe under different conditions. The BC images showed various contrasts across different phases. Moreover, the KAM value was used to analyze the dislocation density, and a low KAM value indicated a low dislocation density (Ref. 13). As can be seen, a large-sized ferrite with a low KAM value appeared at the as-received coating PHS weld toe and extended into the weld along the fusion boundary (Figs. 8A–A2). Moreover, as the distance from the weld toe increased, the coating phases outside the weld were AlFe,  $\eta$ , and Al-Si alloy, respectively. When the coating was SPA30, the ferrite grain size at the weld toe was reduced (Figs. 8B–B2). Furthermore, the fractions of  $\eta$  and AlFe in the coating outside the weld increased, and the fraction of Al-Si alloy decreased. Interestingly, the ferrite at the weld toes was inhibited when the coatings were SPA60 and SPA90 (Figs. 8C–D2). Additionally, the coating outside the weld of the SPA60 coating PHS contained AlFe and a small amount of  $\eta$ , and the coating outside the weld of the SPA90 coating PHS was fully transformed to AlFe. Nevertheless, the  $\tau_1$  phase was not observed in the coatings outside the welds under various conditions.

After the press-hardening process, the OM macrographs and SEM images of the PHS joints under various conditions

are shown in Fig. 9. It was clear that the ferrite remained at the weld toes of both the as-received and SPA30 coating PHSs, indicating that the press-hardening process could not eliminate the ferrite. Additionally, the microstructure in the center of the welds of the as-received and SPA30~90 coatings PHS was similar, which consisted of martensite and small amounts of ferrite (< 5%) (Figs. 9A1–D1).

Figure 10 shows the engineering stress-strain curves and the tensile properties of the base metal (BM) and the PHS joints. The ultimate tensile strength (UTS), yield strength (YS), and total elongation of the BM were  $1583 \pm 13$  MPa,  $1280 \pm 15$  MPa, and  $6.9 \pm 0.5\%$ , respectively. For the as-received coating PHS, the UTS, YS, and total elongation of the joint decreased to  $1465 \pm 20$  MPa,  $1120 \pm 20$  MPa, and  $3.0 \pm 0.3\%$ , respectively, and the fracture occurred at the fusion boundary. When the coating was SPA30, the tensile specimen also failed at the fusion boundary, although its UTS, YS, and total elongation were improved to  $1518 \pm 10$  MPa,  $1150 \pm 17$  MPa, and  $4.6 \pm 0.5\%$ , respectively. When the SPA degree exceeded 60%, excellent tensile properties were obtained, with the fracture occurring in the BM. The UTS and YS of the PHS joints for the SPA60 and SPA90 coatings were comparable to those of the BM, while their total elongations were slightly lower than those of the BM. Based on

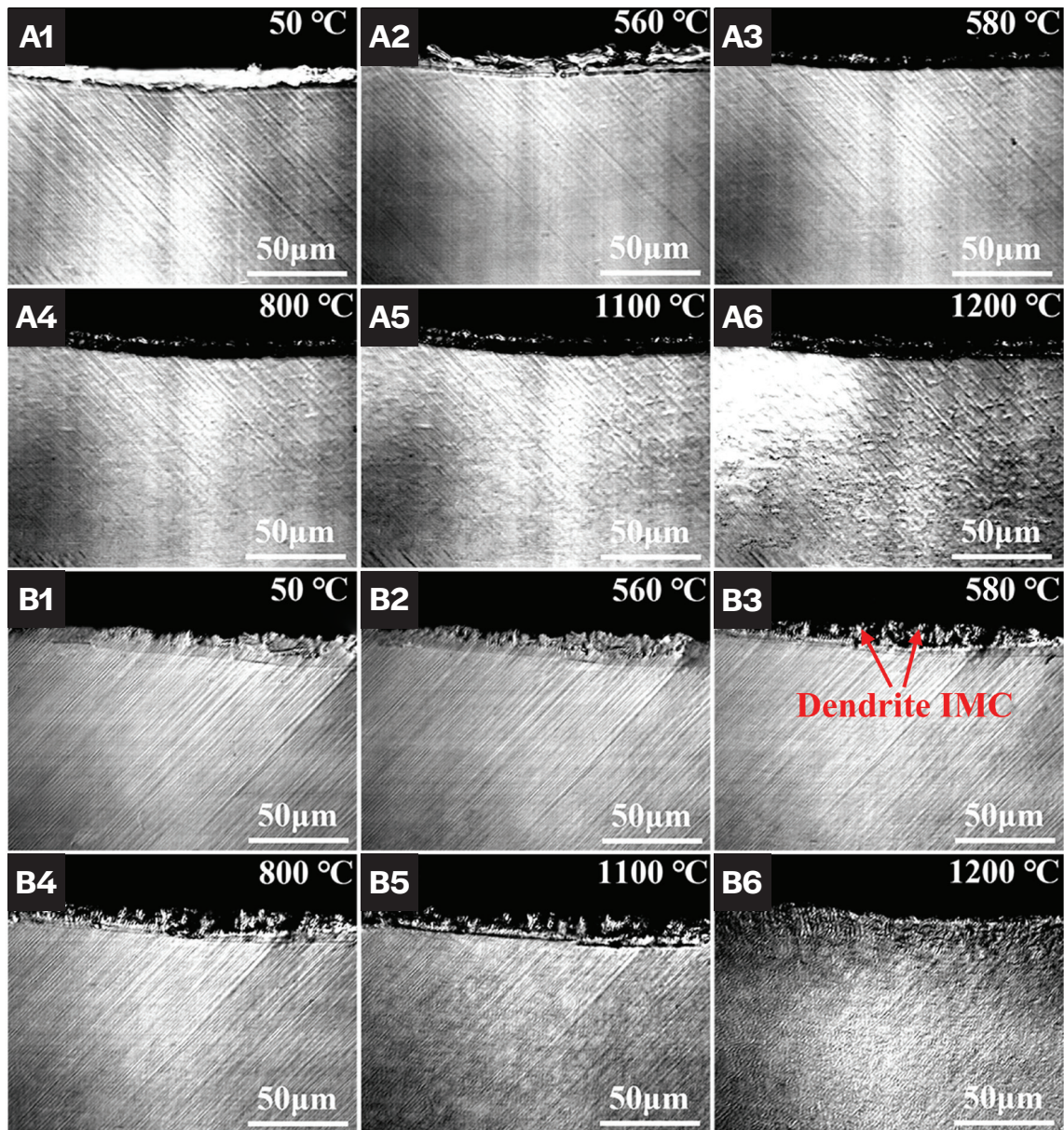


Fig. 13 – The images of the coating heating processes: A1–A6 – As-received coating; B1–B6 – SPA60 coating.

the above experimental results, the mechanical properties of the thin Al-Si coating PHS joints were improved when the SPA technique was applied to the coating outside the weld.

## Discussion

### Microstructure Evolution of the SPA Coatings

As shown in Fig. 5, large amounts of IMCs are generated in the coating via SPA treatment, which depends on the element transfer. Figure 11 shows the temperature measurements in the SPA region under various conditions. It could be seen that the peak temperatures of SPA30, SPA60, and SPA90 coatings were ~680, ~800, and ~930 °C, respectively. In this

case, the Fe element diffused from the steel substrate into the coating during the thermal cycle, leading to the formation of IMCs. Moreover, as the energy input increased, the temperature in the SPA region rose, and the diffusion of Fe into the coating increased. Thus, the fraction of IMCs in the coating gradually increased (Fig. 5), and the coating's color variation became more pronounced (Fig. 4).

In Fig. 6, a continuous IMC layer was present at the bottom of the SPA30 and SPA60 coatings, whereas it disappeared in the SPA90 coating (Figs. 6B–D). The as-received coating contained an Al-Si alloy layer and a continuous  $\tau_5$  layer, as shown in Figs. 6A and E. Previous studies have demonstrated that the liquidus temperatures of the Al-Si alloy and  $\tau_5$  phase were 577 and 830 °C, respectively (Refs. 14, 15). During the SPA30 and SPA60 processes, the peak temperatures were ~680 and ~800 °C, respectively, which were all higher than

the  $T_L$  of the Al-Si alloy but lower than that of the  $\tau_5$  phase, resulting in only the Al-Si alloy melting and the  $\tau_5$  layer being retained. By contrast, the peak temperature of the SPA90 process was  $\sim 930^\circ\text{C}$ , which was higher than the liquidus temperatures of the  $\tau_5$  layer and Al-Si alloy. In this case, both the Al-Si alloy and  $\tau_5$  layer melted, causing the continuous IMC layer to disappear.

It was observed from Figs. 6F–G that the morphologies of the IMCs in SPA coatings were irregularly shaped, and the  $\eta$  and  $\tau_5$  were intertwined. Due to the melting of the Al-Si alloy during the SPA process, element transfer in the liquid coating involved both diffusion and convection. However, the heating and cooling rates in the SPA process were fast, and the residence time at high temperatures was short, leading to insufficient diffusion and convection of the elements. Therefore, the elemental distribution in the SPA coatings was highly non-uniform, leading to the irregularly shaped IMCs. Jenner et al. also demonstrated that the Al-Fe phase composition in the coatings was complex and the interfaces between IMCs were very irregular when the heat treatment time was short (150 s) (Ref. 16).

When the SPA degree was low, the amount of Fe diffusion was small. In this case, the  $\tau_5$  phase grew toward the inside of the Al-Si alloy, and the original  $\tau_5$  layer transformed to the  $\eta$  layer (Fig. 6F). As the SPA degree increased, the Fe element further diffused, the fraction of the  $\tau_5$  phase decreased, and that of the  $\eta$  phase increased. Moreover, the  $\tau_1$  phase was formed in the SPA coatings. According to the works of Yun et. al and Takata et. al, the  $\eta$  phase had a low solubility ( $< 5\%$ ) of Si, so the  $\tau_1$  phase was precipitated along the grain boundaries of the  $\eta$  phase as the Si element was rejected from the  $\eta$  phase (Refs. 17, 18).

## Correlation Between Microstructure and Properties of Welds and SPA Degree

For the as-received coating PHS, after laser welding, there was a large ferrite at the weld toe, corresponding to the location of Al enrichment (Figs. 7E1 and 8A1). Al was a ferrite-formation element, which decreased the stabilization of austenite. The high Al content led to the transformation of austenite to ferrite at higher temperatures, promoting the formation of ferrite (Refs. 19, 20). Compared with the as-received coating PHS, the Al enrichment in the SPA30 coating PHS weld toe was reduced, but it still promoted ferrite formation. When the SPA degree exceeded 60%, the ferrite induced by Al enrichment disappeared at the weld toe. The results above indicated that the microstructure at the weld toe was significantly affected by the SPA treatment. Conversely, the microstructure in the center of the welds under various conditions was similar (Fig. 9A1–D1) because the SPA treatment was applied to the coating outside the weld and only modified the metallurgical behavior of the weld toe, thereby not affecting the weld center.

Based on the results of our previous research (Ref. 11), the Al enrichment at the weld toe was caused by the coating outside the weld entering the weld. The coating's movement indicated that it melted during welding. It was well known that the melting of a metal is related to its liquidus temperature ( $T_L$ ). In this work, the SPA technique altered

the coating's microstructure outside the weld, leading to a change in its  $T_L$ . Furthermore, the SPA coating was a multiphase structure rather than a single phase, and the liquidus temperatures of the phases differed. Therefore, the temperature at which the SPA coating was completely transformed into a liquid phase ( $T_{LC}$ ) depended on the Al-Si alloy and IMCs. According to the experimental results in Fig. 6, the IMCs in the coating contained  $\tau_1$ ,  $\tau_5$ ,  $\eta$ , and AlFe. However, the fractions of  $\tau_1$  and AlFe in the coatings under various conditions were very low, so their effects on the TLC of the coatings were negligible. Therefore, the coating's TLC was mainly determined by the Al-Si alloy, the  $\tau_5$  phase, and the  $\eta$  phase. A previous work systematically calculated the thermodynamic parameters of the Fe-Al-Si ternary IMCs (Ref. 15). Based on the information in Ref. 15, the Fe-Al-Si ternary phase diagram was calculated using the Factsage software, as shown in Fig. 12. As can be seen, the  $T_L$  of  $\tau_5$  and  $\eta$  were  $\sim 830$  and  $\sim 1150^\circ\text{C}$ , respectively, which were consistent with the previous reports (Refs. 14, 15, and 21), and were higher than that of the Al-Si alloy ( $577^\circ\text{C}$ ). In this case, the SPA technique increased the coating's  $T_{LC}$ . Moreover, as the SPA degree increased, the coating phase transformed from Al-Si alloy to  $\tau_5$  and then to  $\eta$  (Fig. 6), which indicated that the TLC of the coating increased with the SPA degree.

Figure 13 recorded the melting behaviors of the as-received coating and SPA60 coating during the heating process. As can be seen from Figs. 13A1–A6, at a temperature of  $560\sim 580^\circ\text{C}$ , the morphology of the as-received coating changed, and its color darkened, indicating that the coating had melted. It was consistent with the  $T_L$  of Al-Si alloy ( $577^\circ\text{C}$ ). In contrast, for the SPA60 coating, it partially melted at  $560\sim 580^\circ\text{C}$ , and the retained dendrite structure comprised IMCs. With the temperature increasing to  $1100\sim 1200^\circ\text{C}$ , the dendrite structure melted. Obviously, the  $T_{LC}$  of the SPA60 coating increased, consistent with the analysis of the phase diagram above. The increase in  $T_{LC}$  of the coating reduced the molten amount of the coating outside the weld. Furthermore, the retained dendrite IMCs with a high  $T_L$  hindered the flow of molten coating and prevented it from entering the weld through the weld toe. Consequently, the Al enrichment at the weld toe was inhibited.

When the SPA degree was low (30%), the fraction of IMCs was small, and the IMCs contained more  $\tau_5$  phase with a lower  $T_L$  ( $830^\circ\text{C}$ ), resulting in the coating's  $T_{LC}$  remaining low. In this case, the molten amount of the coating adjacent to the weld toe was still relatively large, which entered the weld, forming an Al enrichment at the weld toe. As the SPA degree reached 60% or higher, the fraction of the  $\eta$  phase with high  $T_L$  significantly increased, reducing the molten amount of the coating adjacent to the weld toe and eliminating Al enrichment at the weld toe.

In addition to the ferrite at the weld toe, the SPA technique also altered the phase constituents of the coatings near the weld toe (Fig. 8). During laser welding, the elements in the coating outside the weld reacted with each other due to the thermal cycle, causing the coating to undergo phase transformations. For the as-received coating PHS, as the distance from the weld toe increased, the phases of the coating outside the weld were AlFe,  $\eta$ , and an Al-Si alloy, respectively (Fig. 8A1). Clearly, the Fe content in these

phases decreased gradually, attributed to higher temperature and a longer high-temperature residence time in the region near the weld toe, which accelerated Fe diffusion. Moreover, as the SPA degree increased from 30 to 90%, the phase of the coating outside the weld first changed to the  $\eta$  phase and then gradually to the AlFe phase (Figs. 8B1–D1). This phenomenon was attributed to a higher fraction of Fe-rich IMCs formed under a high SPA degree. These IMCs elevated the initial Fe content of the coating before welding, thereby resulting in a higher Fe content after experiencing the same weld heat effect. Additionally, the  $\tau_1$  phase was not observed in the coatings near the weld toe under various conditions. The maximum Si content of the AlFe phase was high (18%), and the AlFe phase became more stable than the  $\tau_1$  phase; thereby, the  $\tau_1$  phase transferred to the AlFe phase (Ref. 22).

The mechanical properties of the PHS joint were closely related to its microstructure. The results in Fig. 9 indicated that the ferrite did not disappear after the press-hardening process. Although the elements diffused to some extent during the press-hardening process, the Al content at the weld toe remained high, resulting in ferrite remaining there. Similar results were observed in previous studies (Refs. 23, 24). The ferrite was soft and had a low dislocation density, making it prone to cracking during tensile testing (Refs. 25, 26), resulting in fracture at the fusion boundary. For the SPA30 coating PHS, the ferrite size at the weld toe decreased. In this case, the tensile properties of the joints improved, while cracking remained at the fusion boundary. When the coating outside the weld was SPA60 and SPA90, the ferrite at the weld toe was eliminated. Therefore, the tensile properties of the PHS joints were significantly enhanced, and the fracture was transferred from the fusion boundary to the BM.

## Conclusion

In this study, a SPA technique was applied to the coating outside the weld, and the thin Al-Si coating PHS was laser-welded at a high welding speed of 7 m/min. The microstructure evolution of the SPA coatings and the metallurgical behaviors of the welds under various conditions were systematically studied. The following conclusions were drawn:

The Fe element of the steel substrate diffused into the Al-Si coating during the SPA process, which accelerated the formation of IMCs. With increasing SPA energy input, the fraction of IMCs in the coating increased from 30 to 90%. The IMCs consisted of  $\tau_2$ ,  $\eta$ , and  $\tau_1$ . As the SPA degree increased, the  $\eta$  phase in the coating grew rapidly, accompanied by a reduction in  $\tau_2$  and  $\tau_1$  formed at the grain boundaries of the  $\eta$  phase.

After the PHSs under various conditions were welded at a high welding speed of 7 m/min, a large-sized ferrite caused by Al enrichment appeared at the as-received coating PHS weld toe, deteriorating the tensile properties of the weld, and the tensile fracture occurred at the fusion boundary. Even though the size of the ferrite at the SPA30 coating PHS weld toe was reduced, it still resulted in the joints failing at the fusion boundary. In contrast, the ferrite at the SPA60 and SPA90 coating weld toes was significantly inhibited. In this case, excellent tensile properties were obtained. The UTS were  $1567 \pm 12$  and  $1569 \pm 15$  MPa, the YS were  $1250 \pm 10$

and  $1264 \pm 15$  MPa, and the total elongations were 6.3 and 6.1%, respectively.

As the fraction of IMCs in the coatings increased from 30 to 90%, the TLC of the SPA coating increased, so the molten amount of the coating outside the weld reduced. Moreover, the IMCs with high  $T_L$  hindered the flow of molten coating and prevented it from entering the weld through the weld toe. The Al enrichment region remaining at the weld toe was significantly inhibited.

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