Influence of Al Interlayer Thickness on Laser Welding of Mg/Steel

The impact of different Al interlayer thicknesses on interfacial reactions and mechanical properties during laser welding of Mg to steel was investigated


ABSTRACT

The bonding of immiscible system Mg/steel was facilitated by adding an Al interlayer using laser welding. The influence of the added interlayer thickness on the microstructure and mechanical properties of the dissimilar joints were investigated. The Al interlayer was dissolved into the Mg fusion zone and mutual diffusion with steel, causing metallurgical bonding between Mg and steel at the interface. Fe-Al reaction layers with different thicknesses were produced along the fusion zone-steel interface. Two different reaction layers divided by the thickness of the reaction layer (2 μm) were thereafter identified by TEM analysis. The phases formed adjacent to the steel substrate were Fe(Al) solid solution and Al(Fe, Mn, Fe) phase on the Fe(Al) surface, while the Al(Fe) phase formed in the reaction layer was more than 2 μm. Three different fracture modes were distinguished from all the joints, which was closely associated with the interfacial reaction. Insufficient atomic diffusion caused interfacial failure. Fracture occurred in the fusion zone when the thickness of the Al interlayer was 0.2–0.5 mm. The maximum value of the tensile-shear fracture load reached 133 N/mm, which represented 40.3% joint efficiency with respect to the steel base metal. However, the excessive interfacial reaction resulted in cracking at the interface and a greater amount of brittle Mg-Al compounds, giving rise to decreased mechanical properties.

KEYWORDS

• Laser Welding • Dissimilar Metals • Magnesium Alloy • Microstructure • Mechanical Properties

Introduction

Vehicle design and fabrication has been increasingly innovated by automotive manufacturers and their suppliers in the effort to maximize fuel efficiency to meet the new, more stringent fuel and emission standards, while maintaining critical safety requirements (Ref. 1). To realize this goal, automakers are working to develop new lightweight materials and integrate them into vehicle designs. Magnesium alloys have been considered as promising materials for automotive applications due to their low density and high specific strength, which has drawn considerable attention in the past few years (Ref. 2). Currently, steel sheets remain the most commonly used material in the automotive industry. They are considered the most cost-effective material for vehicle applications. Therefore, the fabrication of hybrid lightweight structural components of magnesium alloys and steel, becomes a feasible method to further lower the vehicle weight. Reliable joining of Mg to steel would facilitate the increased use of magnesium alloys and expand the scope of their application in the automotive industry.

Direct joining of magnesium alloys to steel was proven to be difficult by conventional fusion welding techniques due to a huge difference in their melting points and Mg-Fe immiscibility characteristics. The melting point of steel (1450°C) is higher than the boiling point of magnesium (1091°C), which could cause catastrophic vaporization of the molten magnesium if they melt simultaneously. In addition, the maximum solid solubility of Fe in Mg is only 0.0041 at-%. Their immiscibility and low solid solubility suggests they do not react with each other or mix in the liquid state at ambient pressure. Thus, metallurgical bonding between these two materials would only be possible if another intermediate element that can interact with or possess substantial solid solubility with both of them can be added during the welding process.

The feasibility of joining Mg to steel using various processes, such as friction stir welding (FSW) (Refs. 3–8), resistance spot welding (RSW) (Refs. 9, 10), diffusion bonding (DB) (Refs. 11–13), cold metal transfer (CMT) welding (Ref. 14), hybrid laser-arc welding (Refs. 15–20), and laser welding-brazing or brazing (Refs. 21–27) have previously been investigated. The FSW process has been reported to successfully realize the bonding of Mg and steel by accelerating the diffusion of Al atoms from the Mg base metal to steel with the combined effect of external force and strong stirring. The Fe-Al phase formed, acting as the transitional compound to bond
Mg and steel. The results achieved using RSW showed low interplanar mismatching of Fe$_2$Al$_5$/Fe (0%) and Fe$_2$Al$_5$/Mg (4.8%) heterophase interfaces, indicating the Fe-Al phase could act as an interlayer between Mg and Fe.

However, without external force, utilizing various interlayers has been the main way to solve the metallurgical bonding problems of Mg/steel. The benefits of using different interlayers and elements, such as Cu, Ni, Sn, Zn, and Al-12Si, have been explored. Elthalabawy and Khan (Refs. 11, 12) tried to join Mg to steel using diffusion bonding (transient liquid bonding), in which the Cu and Ni interlayers were used. Solid-state diffusion, eutectic formation, and the formation of ternary intermetallic compounds were observed at the interface. The interfacial reaction was intense in the liquid state, inducing the excessive formation of brittle and thick layers, which was detrimental to the joint strength.

Cao et al. (Ref. 14) investigated the feasibility of CMT welding of Mg to galvanized mild steel. The Zn coating was found to promote the wetting of the AZ61 filler metal on the steel surface. A sound CMT welded Mg/coated steel joint was obtained with the joint strength comparable to a Mg-Mg fusion joint.

Laser welding (brazing), as an alternative advanced welding technique, has received considerable attention due to its potential for flexible joining of dissimilar metals. Liu et al. (Refs. 15–17, 20) used hybrid laser-arc welding to bond Mg and steel with Cu, Ni, and Sn interlayers, respectively. These thin interlayers were placed in between upper Mg and lower steel sheets. Typically, laser with low power penetrated from the Mg sheet and the interlayer into the lower steel sheet. These added interlayers were heated and melted to react with Mg and steel, resulting in the formation of a transitional layer in the fusion zone and solid solution close to the steel side. The tensile-shear strength was improved because the bonding status was changed from mechanical bonding to semimetallurgical bonding. A severe evaporation of the upper Mg base metal occurred during the process, which limited the application of this welding technique.

In addition, Wahba and Katayama (Ref. 21) also reported the desirable joint could be achieved with the presence of Zn coating on the steel surface using laser conduction welding.

To minimize the vaporization of Mg sheet and improve welding flexibility, laser brazing with a preset Mg-based filler was employed. Various interlayers, including Al-12Si (Ref. 24), Ni (Ref. 26), and Sn (Ref. 27), were used. When using an Al-12Si coating, formation of ϑ-Fe(Al,Si)$_3$ was observed along the fusion zone-steel interface. In the case of the Ni interlayer, formation of Fe(Ni) solid solution on the steel surface was found to be the key for metallurgical bonding of Mg/steel immiscible systems. The joint strength was reported to be higher than that welded with Al coated steel. Afterward, they selected another viable in-
terlayer, Sn, following a review of binary and ternary phase diagrams. The results suggested that all tensile-shear specimens fractured in the steel base metal. The fracture load far exceeded that achieved with the former two kinds of interlayers. The Sn coating was reported to promote wetting between the molten filler metal and steel sheet, creating an oxide-free steel surface for metallurgical bonding. Formation of the Fe(Al) solid solution and Al₆(Mn, Fe)₃ on the Fe(Al) surface, with their low interplanar mismatch, was mainly responsible for high joint strength.

Therefore, selection of an appropriate intermediate alloying element for joining Mg to steel was vital for enhancing wetting-spreading ability and interfacial bonding between the fusion zone and steel without producing too thick brittle reaction layers. In our previous studies, the influence of different coating surfaces (Refs. 28, 29) and alloying elements from filler metal and steel (Refs. 30–32) have been investigated. The important finding was that the Al element, as a key element diffusing from the filler metal into the interface, was accelerated and induced by the chemical potential, resulting in metallurgical bonding of the Mg/steel interface. The Fe-Al phase formed at the Mg/steel interface was expected to be favorable for the mechanical properties after the aforementioned studies.

In the present work, the Al interlayer was employed as the transitional interlayer, which was expected to induce the Fe-Al reaction after melting and produce the Fe-Al phase to realize metallurgical bonding of Mg/steel. The aim of this study was to investigate the influence of different Al interlayer thicknesses on interfacial reactions and mechanical properties during laser welding of Mg to steel. The microstructure characteristics were observed and identified. The mechanical properties were evaluated. Finally, the bonding mechanism was expected to be elucidated.

### Experimental Procedures

Selected as the base metals were 1.5-mm-thick AZ31B Mg alloy (Mg-3Al-1Zn-0.2Mn, wt%) and 1-mm-thick Q235 mild steel (0.17C-0.7Mn-0.35Si-0.035S-0.035P). The sheets were cut into rectangular strips 30 mm wide and 100 mm long. The size of the Al interlayer used in the present work was 3 mm wide and 100 mm long, so the Al interlayer could totally melt and react with the Mg or steel sheets. Before welding, the surfaces of the Mg alloy and steel sheets were cleaned with abrasive paper to remove surface oxides, and then cleaned in acetone and other contaminants from the surfaces.

A fiber laser system with a maximum power of 10 kW (IPG YLR-10000) and a KUKA six-axis robot were used in this study. The laser beam had a wavelength of 1070 nm and beam parameter product of 7.2 mm mrad. It was transmitted by a 200-μm core-diameter fiber and focused by a 200-mm lens to obtain a spot size of 0.2 mm.

Figure 1 shows the schematic of laser welding. The experiments were carried out in a lap joint configuration with the Mg sheet clamped on the
steel sheet. The laser beam was irradiated on the surface of the AZ31B Mg alloy vertically. Argon shielding gas was provided at a flow rate of 20 L/min to prevent oxidation. The laser beam was defocused to irradiate a large area. The process parameters employed in the study were as follows: laser power 800 W, welding speed 0.3 m/min, and defocused distance +20 mm.

After laser welding, typical cross sections of the welded specimens were cut and mounted in epoxy resin. Standard grinding and polishing preparation procedures were then utilized. The appearances and cross sections were observed using an optical microscope (OM). The reaction layer at the interface between the fusion zone and steel were observed using scanning electron microscopy (SEM) in back-scattered electron (BSE) mode.

A transmission electron microscopy (TEM) foil of the bonded region was prepared using the focused ion beam (FIB) technique. The preparation for the FIB-TEM specimen was made using an in-situ lift out method. The TEM with a Tecnai-G2 F30 operating at a nominal voltage of 300 kV was used to characterize the microstructure in detail.

Phase identification was investigated by selected-area electron diffraction (SAED) combined with energy-dispersive spectroscopy (EDS) in scanning transmission electron microscopy (STEM) mode. A Vickers hardness measurement was performed across the fusion zone-steel interface and fusion zone adjacent to the interface, respectively. A test load of 0.1 Kgf and a dwell time of 10 s were utilized. The tensile-shear tests were performed at room temperature using an Instron 5569 at a crosshead speed of 1 mm/min.

Shims were clamped to each end of the specimens to ensure shear loads in the lap joint while minimizing bending or torque of the specimens. Joint strength was calculated via the tensile testing of at least three specimens, and the average was reported with the standard deviation provided via error bars.

Fig. 4 — Microstructure morphologies of interfacial reaction layers with different thicknesses of Al interlayers: A — 0.1 mm; B — 0.3 mm; C — 0.5 mm; and D — 0.7 mm.
Results and Discussion

Joint Appearances

Figure 2 shows the joint appearances with different Al interlayer thicknesses. No spatter or obvious defect was evidenced in the figure because of suitable heat input employed in the present work. However, a nonuniform and rough surface was first observed when the interlayer was thin, as shown in Fig. 2A and 2B. It was mainly attributed to the instability of the Mg molten pool at the action of laser energy. With the increase of Al thickness, the joint surface was improved, obtaining a smooth and uniform appearance indicated in Fig. 2C and 2D.

From the cross-sectional views shown in Fig. 3, the Al foil was not observed as a separate layer along the interface after the process, indicating it entirely melted and was dissolved into the Mg liquid adjacent to the interface. The opening was observed at the edge of the fusion zone close to the steel interface due to the bad wetting-spreading ability of Mg and steel. The opening defect was improved with the addition of Al foil. With an increasing thickness of the interlayer, the joint width became larger due to good wetting of Al and steel. It was believed that the affinity of Al and Fe was much stronger than that of Mg and Fe. Thus, the thin Al liquid adhering to the steel interface reduced the difficulty of the Mg molten pool spreading on the steel surface. Good wetting-spreading ability of Al on the steel surface led to more flowability, as demonstrated in Fig. 3E and F.

Microstructural Analysis

Figure 4 shows microstructure morphologies along the fusion zone-steel interface with different thicknesses of the Al interlayer. Formation of the interfacial reaction layer was observed from all the joints, confirming metallurgical bonding of the immiscible Mg and Fe was realized by adding the Al interlayer.

In addition, the thickness of the Fe-Al reaction layer formed at the interface was found to increase with the increasing Al interlayer. When the thickness of the Al interlayer was 0.1 mm, no obvious reaction layer was noticed, even at higher magnification as shown in Fig. 4A, suggesting most of the Al interlayer melted and was mixed with the molten Mg fusion zone immediately near the interface. The reaction layer was evidently observed when increasing the thickness of the Al interlayer to 0.3 mm. These reaction products exhibited nonuniform and faceted morphology. With a further increase of the Al interlayer to 0.7 mm, the reaction layer grew dramatically, reaching its thickness of more than 10 μm.

The crack was evidently observed at a higher magnification indicated in the inset of Fig. 4D, which deteriorated joint strength. The results were in accordance with the previous report that cracking may easily occur if the thickness of the Fe-Al intermetallic layers exceeded 10 μm, which was detrimental to mechanical properties (Ref. 33).

EDS line scanning analyses were performed to obtain concentration profiles of the main alloying elements across the interface between the Mg fusion zone and steel. Figure 5 shows the corresponding line scan results. The Mg element decreased gradually from the fusion zone to the steel side, while the Fe content was varied in an opposite way. Note that an apparent Al concentration peaked at the interface in all joints, indicating an occurrence of atomic diffusion or dissolution of the Al element, which thereafter induced interfacial reaction. Additionally, the diffusion distance became longer with the increasing Al...
thickness, which caused formation of more Fe-Al reaction layers.

Figure 6 shows the microstructure variation adjacent to the fusion zone-steel interface with the different thicknesses of the Al interlayer. The Al interlayer was heated and melted by laser irradiation during the process. Part of the molten Al interlayer was dissolved into the upper Mg molten pool close to the interface. Mutual diffusion of Mg-Al atoms then took place in a fast rate due to their liquid state. Mg-Al intermetallic compounds would thus form after solidification. Their morphology was varied with the participation of different quantities of Al atoms.

The precipitation of scattered Mg-Al intermetallic compounds (Mg₁₇Al₁₂ phase) was noticed in Fig. 6A when the thickness of the Al interlayer was 0.1 mm. These dispersed particles played an important role in the improvement of mechanical properties by dispersion strengthening (Ref. 34). Similar phenomenon and effects were reported with the formation of finer and denser Al-Cu particles when FSW Al to Cu (Ref. 35). With the increase of Al interlayer, the quantity of precipitated Mg-Al reaction products increased. When the thickness of the Al interlayer exceeded 0.5 mm, network structure formed at the fusion zone, which made the weld brittle and deteriorated the mechanical properties of the joint.

Phase Identification

Transmission electron microscopy analysis was performed to further identify the composition and structure of the reaction layers formed between the fusion zone and steel.

Figure 7 shows a typical bright field TEM image of the fusion zone-steel interface when the thickness of the reaction layer was not more than 2 μm. A TEM foil was prepared at the interface, as shown in Fig. 7A. A noncontinuous, ultra-thin reaction layer was found to exist between the Mg fusion zone and steel substrate indicated in Fig. 7B. The elemental composition at different positions indicated in Fig. 7B are listed in Table 1 based on STEM-EDS results.

Location P2 contained 13.0 at-% Mn, which was much higher than the base metals. Location P3 contained 5.5 at-% Al and 94.5 at-% Fe. As shown in Fig. 7C, a relatively long diffusion distance (a depth of about 550 nm) of Al element into the steel side was observed from the STEM-EDS line scan result. Meanwhile, the Mn element was enriched at the interface close to the Mg fusion zone side. Consequently, the Al-Mn phase and Fe-Al reaction layer was expected to form at the interface.

Figure 8 shows a TEM micrograph with SAED patterns corresponding to Fig. 7. The ultra-thin reaction products consisted of two different phases,
The phases close to the steel substrate were found to be continuous and uniform followed by noncontinuous phases toward the fusion zone. The nonuniform phases close to the fusion zone were identified as Al₈Mn₅, while some Fe element was also detected. Therefore, the Al₈(Mn,Fe)₅ intermetallic compound was exactly indexed, in which some Mn atoms were replaced by Fe atoms because the atomic radius of Mn (0.112 nm) and Fe (0.124 nm) were close (Ref. 27). The phase adjacent to the steel substrate was indexed by TEM and identified as Fe(Al) solid solution with a body-centered cubic (BCC) crystal structure, respectively. The SAED was taken along the [111] zone axis of the phase, making the atomic number of Mn dominate in the whole Fe-Al system. As a result, it was possible that the Al-rich Al-Fe intermetallic compound formed, rather than solid solution.

High magnification of the TEM image at the interface indicated the reaction layer when laser brazing Mg to Sn coated steel sheet (Ref. 27).

To prove the degree of solid solution Al in Fe, the interatomic misfit and interplanar mismatch between two phases can be calculated by

$$\delta = \frac{|a - a_0|}{a_0} \times 100\%$$

where \( |a - a_0| \) is the difference between interatomic or interplanar spacings of the two phases and \( a_0 \) is the interatomic or interplanar spacing of the steel substrate. The \( a_0 \) was 0.287 nm. The misfit was 4.2%; it reduced to 0.5% with the distance from the steel interface of 2000 nm after calculation. The results corresponded to the long diffusion of Al into the steel substrate obtained in Fig. 7C.

Figure 9 shows TEM images taken at the fusion zone-steel interface when the thickness of reaction products was more than 2 μm. The reaction layer was clearly evidenced at the interface. The relatively thick Al interlayer melted, providing more Al atoms at the interface, making the atomic number of Al dominate in the whole Fe-Al system. As a result, it was possible that the Al-rich Al-Fe intermetallic compound formed, rather than solid solution.

High magnification of the TEM image at the interface indicated the reaction layer was distinctly different from the fusion zone/steel substrate. The hardness of the joint produced with an Al thickness of 0.2 mm was measured, and the result is shown in Fig. 11A. The average hardness of the fusion zone and steel substrate was approximately 100 and 209 HV, respectively. Note that an abrupt change in the hardness value at the interface was noticed, which was higher than the fusion zone and neighboring steel substrate, reaching the maximum value of 587 HV. The distinct rise in the hardness at the interface was closely associated with the presence of interfacial reaction products corresponding to the observation in Fig. 4.

Figure 11B presents the hardness distribution profile at the fusion zone near the interface. The hardness of the AZ31 Mg alloy base metal was about 60 HV when without Al foil. The thicker the added Al interlayer, the higher the hardness close to the fusion zone. It was evident that the hardness increased toward the interface due to the precipitation of more Mg-Al compounds.

Figure 12 shows the tensile-shear fracture load of the laser welded magnesium-steel joints and the thickness of the reaction layer with the variation of thicknesses of the Al interlayer and the fracture load with a function of thickness of reaction layer. The joint strength was given here as fracture load because it had both tensile and shear stresses during tests. Three different fracture modes were distinguished in the present work. The fracture behavior depended upon the interfacial bonding force and brittleness of the fusion zone. The fracture load of the joint produced with an Al thickness of no more than 0.1 mm was quite low, resulting in interfacial failure. The interfacial bonding was weak due to insufficient atomic diffusion or interfacial reaction.

Fracture occurred at the faying surface of the Mg and steel sheets. With increasing thickness of the Al interlayer (0.2–0.5 mm), the thick-
ness of the reaction layer formed at the interface was 2.3–5.7 μm. The fracture load was enhanced to maximum value of 133 N/mm, which represented 40.3% joint efficiency with respect to the mild steel base metal, as shown in Fig. 11B.

Joint efficiency was calculated by comparing the maximum load of the joint when fractured in lap shear tensile testing with the maximum load-carrying capacity of the weaker base sheet section. The calculated joint efficiency was reported in a previous study (Ref. 30). In this case, the fracture location was at the Mg fusion zone, indicating the Mg/steel interface was not the weak point.

Compared to high interfacial bonding, the brittle precipitated Mg-Al intermetallic compounds made the crack tend to initiate and propagate at the fusion zone. The fracture load decreased sharply with further increasing the thickness of the Al interlayer. The crack formed in between the interface of the Fe-Al reaction layer and steel substrate, which became the weakest part of the joint despite the presence of brittle Mg-Al compounds in the fusion zone close to the interface.

Figure 13 presents the fracture surface morphologies of joints welded at different thicknesses of the Al interlayer. For the joint with interfacial failure mode, some residual Mg fusion zone was attached to the steel substrate, as shown in Fig. 13A. Higher magnification indicated the fracture surface was characterized by the tearing ridge — Fig. 13B.

When the joint fractured at the Mg fusion zone, the fracture surface exhibited dimples together with some cleavage-like flat facets demonstrated in Fig. 13D, which was similar to the observation of Mg-Mg fusion welding (Ref. 36). The flat surface suggested brittleness of the Mg-Al intermetallic compounds in the fusion zone. It was believed that the more brittle the microstructure, the flatter the fracture surface. With the increasing thickness of the reaction layer, the fracture was found to occur along the interface between the reaction layer and steel substrate due to the presence of cracks formed in between them after the process.

Smooth fracture surface without adequate plastic deformation was observed on the steel side, as shown in Fig. 13E. Cracking was noticed at the higher magnification arrowed in the inset of Fig. 13E. Remnants of the fusion zone adhering to the fracture surface shown in Fig. 13F were found to be more brittle compared to the result in Fig. 13D.

**Bonding Mechanism**

Based on the analyses above, the bonding mechanism of Mg to steel...
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with the addition of the Al interlayer was elucidated.

Figure 14 shows the schematic diagram of the joining mechanism. First, melting of the Mg sheet and Al interlayer occurred simultaneously when suffering the laser beam irradiation, as shown in Fig. 14B. The liquid Mg atoms and Al atoms dissolving into each other. At the same time, some Al atoms close to the liquid/solid interface diffused into the steel surface. Because of the laser irradiation, the atoms in the steel adjacent to the interface were also activated at a high temperature and slightly dissolved into the liquid. The diffusion distance was relatively short due to fast thermal cycle during laser welding. As a result, different percents of the main alloying elements Mg, Al, and steel were mixed from fusion zone to the liquid/solid interface. The different precipitation modes were distinguished with the variation of the thickness of the Al interlayer.

Upon cooling, as the temperature decreased to about 1000°C, the solid solubility of Al in Fe at the solid/liquid interface was saturated inducing crystallization of Fe(Al) when the Al interlayer was very thin, as indicated in Fig. 14C. A thin layer of Fe(Al) formed first from the liquid by solid-state diffusion of Al atoms in the liquid filler metal into the steel. Upon further cooling, some Mn atoms diffusing from the steel substrate were then bonded with Al atoms, giving rise to the Al$_2$Mn$_x$ phase under the temperature of 680°C (Ref. 37).

The phase nucleated and grew on the first precipitated Fe(Al) surface, as shown in Fig. 14D. Part of the Mn atoms were replaced by the Fe atoms due to their similar atomic radius and metallurgical characteristics. When the temperature decreased to 650°C or below, α-Mg first precipitated from the remaining liquid.

Finally, a eutectic reaction occurred in the liquid, producing a eutectic structure ($α$-Mg + Mg$_{0.5}$Al$_{0.5}$). In the case of the thick Al interlayer, more Al atoms segregated at the front of the liquid/solid interface. Therefore, different Fe-Al phases formed depending on the ratio of Al and Fe atoms. The content of Al and Fe reached the stoichiometric composition of Al$_2$Fe, which was precipitated when the temperature decreased to 820°C. Some of the Fe atoms were replaced by a small amount of Mn atoms diffusing from the steel. After that, the eutectic structure was produced in the molten pool, which was much denser than that formed with the thin Al interlayer.

**Conclusion**

1) With the addition of the Al interlayer, successful joining of Mg to steel was realized by laser welding. A visually acceptable and uniform joint was achieved with the assistance of a good affinity of Al and steel. The best joint quality was obtained using the following process parameters: 800-W laser power, 0.3-m/min welding speed, +20-mm defocused distance, and 0.3-mm-thickness of Al interlayer.

2) Al interlayer was dissolved into the Mg fusion zone and mutual diffusion with steel, causing metallurgical bonding of Mg and steel at the interface, and more precipitation of the Mg-Al compounds. A reaction layer formed along the interface of the fusion zone-steel joint. The thickness increased with an increasing thickness of the Al interlayer. Cracking was observed between the reaction products and steel substrate when the thickness was larger than 10 μm. Precipitation of the Mg-Al intermetallic compounds was noticed at the fusion zone close to the interface. The quantity of the compounds increased with the increase of the Al interlayer. The excessive precipitation caused the brittleness of the fusion zone and decreased the fracture load.

3) Two different reaction layers were identified by TEM analysis. When the thickness of the reaction layer was less than 2 μm, the phases close to the steel substrate were the Fe(Al) solid solution and Al$_{(3,5)}$(Mn, Fe)$_{0.5}$ phase on the Fe(Al) surface, while the Al$_2$Fe phase formed when the reaction layer was more than 2 μm.

4) Three different fracture modes were distinguished from all the joints produced with different thicknesses of the Al interlayers, which was closely associated with the interfacial reaction. Insufficient atomic diffusion caused interfacial failure with the fracture at the faying surface of the Mg and steel sheet. The suitable interfacial reaction oc-
curred when the thickness of the Al interlayer was 0.2–0.5 mm, resulting in fusion zone fracture. The maximum value of tensile-shear fracture load reached 133 N/mm, which represented 40.3% joint efficiency with respect to the steel base metal. However, the excessive interfacial reaction resulted in the cracking at the interface and greater amount of brittle Mg-Al compounds, giving rise to decreased mechanical properties.

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