Characterization of Intermetallic Compounds in Dissimilar Material Resistance Spot Welded Joint of High Strength Steel and Aluminum Alloy

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Dissimilar materials of H220 Zn-coated high strength steel and 6008 aluminum alloy were welded by median frequency resistance spot welding. Interfacial characteristics and kinetics of growth of intermetallic compound layer at steel/aluminum interface in the welded joint were investigated. The intermetallic compound layer was mainly made up of $\eta$-Fe$_2$Al$_5$ and $\theta$-FeAl$_3$ phases, and its morphology and thickness varied with positions along the interface. The growth behavior of the intermetallic compound layer was dominated by $\eta$-Fe$_2$Al$_5$, which exhibited parabolic characteristic. The growth coefficient of $\eta$-Fe$_2$Al$_5$ could be expressed as 

$$k = k_0 \exp\left(-\frac{Q}{RT}\right)$$

with $k_0$ of 132 m$^2$/s and $Q$ of 239 kJ/mol. The kinetics of growth of the intermetallic compound layer indicated that its formation and growth were mainly driven by reactive diffusion between Fe and Al atoms, and hence the thickness and morphology of the layer were dependant on interaction time between liquid aluminum alloy and solid steel, and also interfacial temperature history during welding. The brittle intermetallic compound layer at the steel/aluminum interface was the weak zone where cracks inclined to derive and propagate during tensile shear testing. The fracture surfaces of the welded joint displayed mixed fracture morphology with both brittle and ductile features.

KEY WORDS: intermetallic compounds; resistance spot welded joint; high strength steel; aluminum alloy; growth kinetics.

1. Introduction

In order to meet anti-pollution and energy saving legislations, the automotive industry has been devoted to reducing vehicle fuel consumption and developing lighter and more energy-efficient vehicles nowadays.1–3) Aluminum alloys have been increasingly used to fabricate lightweight vehicles, due to their advantages of low density, high corrosion resistance and easy recyclability. Nevertheless, there exist some inherent problems in terms of cost and security performance for an intended all-aluminum vehicle, although it could reduce the vehicle weight by 50%.4) The introduction of aluminum alloy components to high strength steel vehicle body, i.e. multi-structure methodology, is expected to reach a compromise between cost and security performance, and hence it is recognized as an effective method to achieve automobile lightening.5) The multi-structure methodology requires the joining of high strength steel and aluminum alloy. Owing to great differences in melting point, thermal expansion coefficient and thermal conductivity between steel and aluminum alloy, it is difficult to obtain a sound fusion welded joint of the dissimilar materials as a result of the formation of defects like cavities, porosities and cracks. Moreover, the nearly zero solid solubility of iron in aluminum alloy is readily to promote the formation of intermetallic compounds.6) Several solid state welding methods have been investigated to join steel and aluminum alloy, including explosion welding,7,8) ultrasonic welding,9) magnetic pressure welding,10,11) friction welding,12–14) and friction stir welding.15,16) The solid state welding techniques are expected to restrain the growth of the intermetallic compound layer within a permissible limit, which has been identified as 10 $\mu$m thickness,17) but the adaptability of these methods is restricted in automotive industry due to equipment configuration. In the last few years, arc welding-brazing,18–21) laser reactive wetting6) and laser brazing4,5,22–25) of the dissimilar materials of steel and aluminum alloy have drawn great attentions. Lin et al.18–20) studied dissimilar metals TIG welding-brazing of aluminum alloy to stainless steel using Al–Si, Al–Cu and Al–Si–Cu filler wires. Multiple intermetallic compounds were formed at the seam interface in the joints, involving $\tau_7$-Al$_7$Fe$_2$Si, $\tau_6$-Al$_5$FeSi, $\eta$-Fe$_2$Al$_5$, $\theta$-Fe(Al,Si)$_3$ and FeSi$_2$ with Si addition, $\theta$-(Fe,Cu)Al$_3$ with Cu addition. The overall thickness of the intermetallic compound layer ranged from 3 $\mu$m to 35 $\mu$m. Laser reactive wetting could obtain the steel/aluminum alloy joint with the formation of mainly $\eta$-Fe$_2$Al$_5$ with 2–20 $\mu$m thickness at the interface.6) Besides, laser brazed joints of steel and alumi-
num alloy with Al–Si or Zn–Al filler wires also exhibited thin intermetallic compound layers, consisting of $\tau_5$-Al$_7$Fe$_2$Si, $\tau_6$-Al$_5$FeSi and $\eta$-Fe$_2$Al$_5$ with overall thickness of less than 2 $\mu$m for the former filler wire, and $\eta$-Fe$_2$Al$_5$(Zn) and $\theta$-FeAl$_3$(Zn) with overall thickness of 1–15 $\mu$m for the latter one. However, few publications concerning resistance spot welding of steel and aluminum alloy have been reported to date, and the limited studies focus on the use of transition materials such as aluminum clad steel plate to aid the resistance spot welding process. Accordingly, there is a lack of understanding on resistance spot weldability of the dissimilar materials of steel and aluminum alloy.

As far as the growth mechanism of Fe–Al intermetallic compounds is concerned, continuous efforts have been made to describe the kinetics of growth of the compounds during hot dip aluminizing, immersion and liquid metal corrosion process. The kinetics of growth of the Fe–Al intermetallic compounds is dependant on nucleation conditions, chemical reactions, and diffusion coefficient. Denner et al. detected the formation of $\eta$-Fe$_2$Al$_5$ and $\theta$-FeAl$_3$ compounds in the case of the interaction between liquid aluminum and solid iron during hot dip aluminizing. The parabolic kinetics of growth of the two intermetallic compounds was proposed in references, whereas negative deviations from the parabolic relationship was observed after long reaction times between the liquid aluminum saturated with Fe and solid iron. The contradiction was resolved by Dybkov who confirmed the presence of a linear growth stage after the prior parabolic growth, i.e. paralinear kinetics of growth during the interaction process between liquid aluminum and 18Cr–10Ni stainless steel. However, the data associated with the kinetics of growth of the Fe–Al intermetallic compounds is still incomplete and contradictory to a certain extent. Furthermore, few studies about the growth mechanisms of intermetallic compounds formed at steel/aluminum interface during resistance spot welding have been reported till now.

The aim of the work is to understand microstructure characteristics and the kinetics of growth of the interfacial intermetallic compounds in the resistance spot welded joint of the dissimilar materials of high strength steel and aluminum alloy, and investigate the effect of the intermetallic compounds on the mechanical behavior of the welded joint.

2. Experimental Procedure

H220 high strength steel sheets with thickness of 1.0 mm and 6008-T66 aluminum alloy sheets with thickness of 1.5 mm were used in the work. Their chemical compositions in mass percentage are Fe–0.007C–0.09Si–0.51Mn–0.007S–0.05P–0.01Ti–0.02Nb for the steel and Al–0.45Mg–0.56Si–0.15Cu–0.19Fe–0.07Mn–0.007Zn–0.02Ti–0.08V for the aluminum alloy. Both sides of the steel sheet were coated with zinc layer in 7 $\mu$m thickness by hot dip galvanizing. The dimensions of sheet specimens for the steel and aluminum alloy are 100 mm x 25 mm. Lap-welded joints were employed by assembling the sheet specimens. Before welding, the aluminum alloy specimens were polished by abrasive paper first and then cleaned in acetone, whereas the steel specimens were only degreased in acetone. The dissimilar material resistance spot welding was carried out using a median frequency direct current (DC) resistance spot welding machine equipped with a welding gun. The electrical schematic diagram of the median frequency DC welding system is shown in Fig. 1. The median frequency resistance spot welding machine used in the study firstly supplies 1 000 HZ frequency single-phase alternative current (AC) to the secondary circuit by an inverter, then outputs DC as welding current by means of a single-phase full-wave rectifier. An electrode cap (DIN 5821-F16×20) with 16 mm tip sphere diameter that made by CuCrZr alloy was used. The schematic diagram of the experimental set-up is shown in Fig. 2. The reason for the use of median frequency DC resistance spot welding lied in that it could supply more stable welding current compared with industrial frequency alternative current resistance spot welding and favored dissimilar material welding. Figure 3 illustrates real-time welding parameter curve obtained in the study. As can be seen, a stable welding current was provided by the median frequency resistance spot welding. The electrode force and keeping time were fixed as 2 kN and 200 ms, respectively. The weld-
ing current varied from 4 kA to 11 kA and the welding time was changed from 50 ms to 300 ms. After welding, the welded joints were sectioned through the nugget center normal to the specimen plane, and then mounted and polished in sequence, followed by etching with Keller’s reagent to attack microstructure of the aluminum alloy and 4% HNO₃ in ethanol solution to reveal microstructure of the steel in the welded joints for metallographic examination.

Microstructure of steel/aluminum interface in the welded joints was investigated by means of optical microscopy, scanning electron microscopy (SEM) and transmission electron microscopy (TEM). Chemical compositions across the steel/aluminum interface were examined using energy dispersive spectroscopy (EDS). The phase compositions of the fracture surfaces of the welded joints were detected using X-ray diffraction (X-RD). Besides, material testing system was used to examine tensile shear load of the welded joints at a crosshead speed of 0.5 mm·min⁻¹. The tensile shear load was determined by the average value of over three measurements per condition.

3. Results and Discussion

3.1. Characteristics of Interfacial Intermetallic Compounds

Figure 4 shows the cross-section of the resistance spot welded joint of H220 Zn-coated high strength steel and 6008-T66 aluminum alloy, which was obtained at welding current of 9 kA and welding time of 250 ms. An apparent interface between the high strength steel and the aluminum alloy was observed. During welding, the aluminum alloy in the welded joint was molten, i.e. aluminum nugget, whereas the high strength steel retained solid. The welded joint was achieved by means of wetting and spreading of liquid aluminum alloy on solid steel surface. Therefore, it can be regarded as a brazed joint. Figure 5 shows the SEM images acquired in the steel/aluminum interface regions. It indicated that an intermetallic compound layer was formed at the steel/aluminum interface regions. It indicated that an intermetallic compound layer was formed at the steel/aluminum interface, and that the thickness and morphology of the intermetallic compound layer presented great variations at different regions along the interface. As is illustrated, an intermetallic compound layer with thickness of about 5.6 µm was formed at the interface in the weld center region (Fig. 5(a)). From Figs. 5(a)–5(d), a decreased tendency for the thickness of the intermetallic compound layer along the interface was observed with approaching towards the weld periphery, and a fairly thin layer with thickness of 0.4 µm was produced in the weld periphery (Fig. 5(d)). The intermetallic compound layer also exhibited different morphologies as the locations varied along the interface. The intermetallic compound layer displayed a dual-layer characteristic in the thickness direction in the weld center, which was named as layer I beside the high strength steel and layer II beside the aluminum alloy nugget (Fig. 5(a)). The layer I, being of 3.6 µm thickness, showed a lath-like morphology with a planar shaped interface between the layer I and steel, while layer II with thickness of 2 µm turned out to be a coarse needle-like (or serrated-like) morphology, its needles (or serrations) orienting into aluminum nugget. However, evolutions for microstructure of the intermetallic compound layer from lath-like to tongue-like morphology adjacent to
high strength steel side and from coarse needle-like (or serrated-like) to fine needle-like morphology adjacent to aluminum alloy nugget side were observed when the locations varied from weld center to weld periphery (Figs. 5(a)–5(d)). Table 1 shows EDS compositions of the intermetallic compound layer at the steel/aluminum interface with analysis positions marked as A1 and A2 in Fig. 5(a), B1 and B2 in Fig. 5(b) and D1 in Fig. 5(d). It indicated that the lath-like and tongue-like phases had the similar chemical compositions and the needle-like (or serrated-like) phases contained more Al element. According to the Fe–Al binary phase diagram, the former two phases might be $\eta$-Fe$_2$Al$_5$ and the latter one might be $\theta$-FeAl$_3$. In order to certify the detailed phase compositions of the intermetallic compound layer, the interfacial regions were examined by TEM. Figure 6 shows the TEM bright field image and the selected area electron diffraction patterns of the interface (region C in Fig. 4). Dual-layer structure composed of tongue-like and serrated-like phases was observed (Fig. 6(a)), which was similar to the result obtained by SEM (Fig. 5(c)). From the electron diffraction patterns shown in Figs. 6(b) and 6(c), it could be concluded that the tongue-like phase was $\eta$-Fe$_2$Al$_5$ and the serrated-like phase was $\theta$-FeAl$_3$. Nevertheless, only $\theta$-FeAl$_3$ was confirmed in the interface of the weld periphery (region D in Fig. 4).

### 3.2. Kinetics of Growth of Intermetallic Compound Layer

During resistance spot welding of H220 Zn-coated high strength steel and 6008-T66 aluminum alloy, the temperature at steel/aluminum interface was higher than melting point of the aluminum alloy and lower than melting point of the steel, and hence the welding process involved wetting and spreading of liquid aluminum alloy on solid steel at the interface. The interaction between liquid aluminum alloy and solid steel induced the formation of Fe–Al intermetallic compound layer.$^{24,31,33}$ Figure 7 shows distributions of Fe and Al across the interface. It indicated that the growth of the intermetallic compounds was controlled by the reactive diffusion between the liquid aluminum alloy and the solid steel. According to some thermodynamic data,$^{29,40}$ the standard Gibbs free energy of formation of 1 mole FeAl$_{2.5}$ and 1 mole FeAl$_3$ could be expressed as follows respectively:

$$\Delta G^0(FeAl_{2.5}) = -126985.5 + 49.26T \quad \cdots \cdots (1)$$

$$\Delta G^0(FeAl_3) = -142770.5 + 50.58T \quad \cdots \cdots (2)$$

where $T$ was absolute temperature. Hence, the formation of FeAl$_3$ was superior to that of Fe$_2$Al$_5$ (i.e. FeAl$_{2.5}$) as a result of a lower standard Gibbs free energy of formation for FeAl$_3$. However, the thermodynamic calculation with Eqs. (1) and (2) associated with formation of the compounds was done in equilibrium state, and it could not determine the growth speed and even whether the compounds would be formed in non-equilibrium state, such as in the resistance spot welding process. Therefore, it became indispensable to study kinetics of growth of the intermetallic compounds. As suggested by the previous literatures,$^{31,35,36}$ in immersion process, FeAl$_3$ was formed first by interfacial reaction between the liquid aluminum alloy and the solid steel in a quasi-linear kinetics mode of growth, and then Fe$_2$Al$_5$ was formed by Al atoms diffusing through FeAl$_3$ phase and reacting with steel substrate in a parabolic kinetics mode of growth. We assumed that the transient behavior of kinetics of growth in resistance spot welding was identical to that of

<table>
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<td>B2</td>
<td>59.57</td>
<td>75.35</td>
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<td>D1</td>
<td>57.21</td>
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the early stage in the immersion process to some extent, although the welding process was more complex due to the temperature and interaction time varying with positions along the steel/aluminum interface. The kinetics of growth for Fe$_2$Al$_5$ and FeAl$_3$ could be expressed using the following equations:

$$X_{\text{Fe}_2\text{Al}_5} = \sqrt{2kt} \quad \cdots \quad (3)$$

$$X_{\text{FeAl}_3} = k't + A \quad \cdots \quad (4)$$

$$k = k_0 \exp \left( -\frac{Q}{RT} \right) \quad \cdots \quad (5)$$

where $X_{\text{Fe}_2\text{Al}_5}$ and $X_{\text{FeAl}_3}$ were thickness of Fe$_2$Al$_5$ and FeAl$_3$ layers, respectively; $t$ was interaction time, $k$ and $k'$ was growth coefficient of Fe$_2$Al$_5$ and FeAl$_3$, respectively, $A$ was temperature-dependent constant which could not be certified by theoretical method and should only be estimated by experimental measurements at an equilibrium state,$^{36}$ $k_0$ was pre-exponential factor, $Q$ was the activation energy, $R$ was gas constant and $T$ was absolute temperature. The parameters would be clarified in the following discussions.

Hence, the total thickness of Fe$_2$Al$_5$ and FeAl$_3$ layers increased with increasing interficial temperature and interaction time, which has been confirmed by the experimental results to be stated as follows. Figures 8 and 9 show thickness distribution of the intermetallic compound layer in the welded joint and interficial microstructure in weld center at different welding currents and welding times, respectively. The overall thickness of intermetallic compound layer reached 2.5 μm at the weld center at welding current of 7 kA and welding time of 250 ms (Fig. 8(a)). The value decreased gradually with approaching towards the weld periphery. The similar variation tendency was obtained at welding current of 9 kA and welding time of 100 ms (Fig. 8(b)), and welding current of 9 kA and welding time of 250 ms (Fig. 8(c)). The intermetallic compound layer was thicker at the weld center due to its higher interficial temperature and longer reactive diffusion time. According to the studies by Bouché et al.,$^{31}$ the coefficient of growth kinetics of η-Fe$_2$Al$_5$ (e.g. 2.30·10$^{-4}$ m·s$^{-2}$, 800°C) was always larger than that of θ-FeAl$_3$ (e.g. 0.12·10$^{-4}$ m·s$^{-2}$, 800°C), thereby the growth rate was slower for θ-FeAl$_3$ than that of η-Fe$_2$Al$_5$. Hence, the thickness of θ-FeAl$_3$ was thinner than that of η-Fe$_2$Al$_5$ in the study, and the growth behavior of the intermetallic compound layer was predominantly regulated by η-Fe$_2$Al$_5$. On the other hand, the thickness of intermetallic compound layer in weld center varied with the thickness of weld.

Fig. 8. Intermetallic compound layer thickness distribution at different welding currents and welding times: (a) 7 kA, 250 ms, (b) 9 kA, 100 ms, (c) 9 kA, 250 ms.

Fig. 9. Microstructure of intermetallic compound layer of weld central region at different welding currents and welding times: (a) 5 kA, 250 ms, (b) 7 kA, 250 ms, (c) 9 kA, 250 ms, (d) 9 kA, 100 ms, (e) 9 kA, 200 ms, (f) 9 kA, 300 ms.

parameters. The thickness of intermetallic compound layer was only 1.5 μm at 5 kA and 250 ms. A rapid increased tendency for the thickness of intermetallic compound layer was observed with increasing welding current (Figs. 9(b) and 9(c)), the maximum thickness of about 5.6 μm being obtained at 9 kA and 250 ms. As far as welding time effect was concerned, a sharp increased trend for the thickness of intermetallic compound layer in weld center was also acquired with the increase in welding time at welding current of 9 kA. An intermetallic compound layer with thickness of 2.9 μm was formed at 9 kA and 100 ms. The thickness of intermetallic compound layer increased to 13 μm at 9 kA and 300 ms. Generally, as welding current and welding time increased, interficial temperature and reactive diffusion time between liquid aluminum alloy and solid steel increased, which resulted in the formation of thicker intermetallic compound layer. In order to certify the kinetics of growth clearly, numerical simulation by using ANSYS software was adopted to study the temperature-time history which the intermetallic compound layer experienced during welding. Figure 10 shows the temperature distribu-
In the welded joint at the welding current of 9 kA and welding time of 300 ms. As can be seen, a decreased tendency was obtained for the interfacial temperature from weld center to periphery, and the maximum temperature of about 1120°C was achieved. Figure 11 shows the thermal cycle of the weld center (T(t)) during welding process at different welding parameters. It exhibited that the temperature experienced a rapid increase at the early stage with current feeding, and then it declined gradually after the power was turned off. According to Eqs. (3) and (5), the following formula could be obtained:

\[ dX_{Fe_2Al_5} = 2k_0 \exp \left( -\frac{Q}{RT} \right) dt \] ........................ (6)

in which \( dX_{Fe_2Al_5} \) was the differential of \( X_{Fe_2Al_5} \), and \( dt \) was the differential of \( t \). Furthermore, the melting point of 6008-T66 aluminum alloy being 610°C was recognized as the starting point for reactive diffusion, the sustainable time upon which was integration interval. Through integrating at the different welding time stated above, the parameters could be calculated to be about 132 m²/s for \( k_0 \) and 239 kJ/mol for \( Q \). It should be noted that no precise description concerning the relation between growth coefficient of \( \theta \)-FeAl₃ and temperature has been reported till now, and that the data could hardly be obtained during resistance spot welding in which temperature and iteration time varied rapidly.

In addition, the morphology of the interface between intermetallic compound layer and steel exhibited inconsistencies at different positions in the weld, i.e. planar morphology in the weld center and tongue-like morphology in the weld periphery. This tongue-like interface was obtained due to favorable possibilities for Al atoms diffusing through structural vacancies in the c-axis direction of the formed \( \eta \)-Fe₂Al₅ orthorhombic structure. \(^{31,35}\) A mathematical model was employed to evaluate the diffusion coefficient (D) of aluminum atoms through the formed \( \eta \)-Fe₂Al₅, and it was expressed as follows:

\[ D = D_0 \exp \left( -\frac{Q_0}{RT} \right) \] ........................ (7)

in which \( D_0 \) is the pre-exponential factor, \( Q_0 \) is the activation enthalpy. According to the discussion about diffusion numerical analysis during the interaction between liquid aluminum alloy and steel in literature, \(^{35}\) \( D_0 \) and \( Q_0 \) could be clarified as about 2.55×10⁻⁴ m²/s and 259 kJ/mol by means of the least squares method, respectively. With increasing interfacial temperature and reactive diffusion time, the diffusion coefficient increased and more Al atoms diffused through \( \eta \)-Fe₂Al₅ and reacted with steel substrate, facilitating the transitions from tongue-like to planar morphology for \( \eta \)-Fe₂Al₅/steel interface. Moreover, the evolution of morphology of \( \theta \)-FeAl₃ from fine needle-like to coarse needle-like (or serrated-like) was also due to the higher temperature and longer reactive diffusion time. In the weld periphery, as the interfacial temperature was too low to provide sufficient energy to form \( \eta \)-Fe₂Al₅, only \( \theta \)-FeAl₃ was identified at the interface.

Above all, the growth behavior displayed especial characteristics due to the great discrepancy for the growth condition between resistance spot welding and other thermal process with constant temperature. The intermetallic compound layer exhibited unequal thickness along the steel/aluminum interface, with the maximum thickness being obtained at weld center, which was due to its higher interfacial temperature and longer reactive diffusion time. A rapid increased tendency for the thickness of intermetallic compound layer was observed with increasing welding current, i.e. from 1.5 μm at 5 kA and 250 ms to 5.6 μm at 9 kA and 250 ms. Meanwhile, the thickness of intermetallic compound layer also displayed a sharp increase from 2.9 μm at 9 kA and 100 ms to 13 μm at 9 kA and 300 ms. Furthermore, the growth coefficient of \( \eta \)-Fe₂Al₅ could be expressed as \( k = k_0 \exp \left( -\frac{Q}{RT} \right) \) with \( k_0 \) of 132 m²/s and \( Q \) of 239 kJ/mol in the experimental condition of this study.

### 3.3. Mechanical Behavior of Intermetallic Compound Layer

The maximum tensile shear load of the welded joints reached 3.3 kN at welding current of 9 kA and welding time of 250 ms. The fracture mainly occurred at the steel/aluminum interface during tensile shear testing, which meant that the interface was the weak link of the welded joint. It was found that \( \eta \)-Fe₂Al₅ in the steel/aluminum interface tended to act as original sites for cracking, as shown in Fig. 12. This is because \( \eta \)-Fe₂Al₅ is harder and more brittle com-
pared with $\theta$-FeAl$_3$, steel substrate and aluminum nugget. Figure 13 shows the fracture surface morphology of the welded joint obtained at welding current of 9 kA and welding time of 250 ms. Three different regions, indicated as I, II and III, were observed in the fracture surface on the steel side (Fig. 13(a)). As can be seen, region I exhibited a mixed fracture features with dimples and cleavage plane, region II presented brittle fracture features and region III displayed both brittle and ductile fracture features with some secondary cracks. Three equivalent regions of the fracture surface on the aluminum side exhibited the similar features to that of the fracture surface on the steel side, as shown in Fig. 13(b). The reason for the mixed fracture feature lied in that the aluminum alloy bulging into the steel substrate at the weld center favored to lessen stress concentration, and meanwhile the intermetallic compound layer was thin at the weld periphery, thereby its detrimental effect to the mechanical resistance being restrained. Figure 14 shows the X-RD profile of the fracture surface on the steel side. As can be seen, both $\eta$-Fe$_2$Al$_5$ and $\alpha$-Al were detected on the fracture
surface. Therefore, the resultant fracture cracks propagated through intermetallic compound layer and aluminum alloy fusion zone near the interface.

Based on above results, the brittle intermetallic compound layer was the weak zone where cracks were apt to derive and propagate. Hence, further efforts should be made to prevent formation and growth of the intermetallic compounds and improve the mechanical properties of the welded joint of high strength steel and aluminum alloy.

4. Conclusions

(1) The intermetallic compound layer at steel/aluminum interface in resistance spot welded joint of H220 Zn-coated high strength steel and 6008 aluminum alloy was mainly composed of \( \eta\)-Fe2Al5 and \( \theta\)-FeAl3 phases, and the morphology and thickness of the intermetallic compound layer varied with the locations along the interface in the weld.

(2) The formation and growth of the intermetallic compounds (\( \eta\)-Fe2Al5 and \( \theta\)-FeAl3) were controlled by reactive diffusion between solid steel and liquid aluminum alloy during resistance spot welding. The intermetallic compound layer exhibited unequal thickness along the steel/aluminum interface, with the maximum thickness being obtained at weld center, attributing to its higher interfacial temperature and longer reactive diffusion time. A rapid increased tendency for the thickness of intermetallic compound layer was observed with increasing welding current, i.e. from 1.5 \( \mu \)m at 5 kA and 250 ms to 5.6 \( \mu \)m at 9 kA and 250 ms. Meanwhile, the thickness of intermetallic compound layer also displayed a sharp increase from 2.9 \( \mu \)m at 9 kA and 100 ms to 13 \( \mu \)m at 9 kA and 300 ms. The growth coefficient of \( \eta\)-Fe2Al5 could be expressed as \( k = k_0 \exp \left( \frac{-Q}{RT} \right) \) with \( k_0 \) of 132 m²/s and \( Q \) of 239 kJ/mol. The diffusion coefficient of aluminum atoms through the formed \( \eta\)-Fe2Al5 could be expressed as \( D = D_0 \exp \left( \frac{-Q}{RT} \right) \) with \( D_0 \) of 2.55×10⁻¹⁴ m²/s and \( Q_0 \) of 259 kJ/mol.

(3) The welded joint exhibited interfacial failure mode during tensile shear testing, and the brittle intermetallic compound layer at the steel/aluminum interface was the weak zone where cracks were apt to derive and propagate. The fracture surface of the welded joint displayed both brittle and ductile feature.

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