Original Article

A comparative study between the mechanical and microstructural properties of resistance spot welding joints among ferritic AISI 430 and austenitic AISI 304 stainless steel

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ABSTRACT

In this study, 1.5-mm-thickness ferritic AISI 430 and austenitic AISI 304 stainless steel are resistance spot-welded by combining 304/304, 304/430 and 430/430. The dissimilar 304/430 nugget displays a strong texture and consists of α-ferrite columnar grains, with some γ-austenite and α'-martensite dispersed over the grains' boundary zone. Lap-shear tests are performed for each welding combination under different parameters. The results show that the interfacial failure to pull-out failure mode transition tendency is in the following order: 304/430 > 304/304 > 430/430. This is because: (1) the nugget of 304/430 samples are harder than that of the 304/304 samples, which makes the former less likely to fracture under nugget plastic failure; (2) the 304/430 nugget samples possess higher local toughness than the 430/430 samples, a property that makes 304/430 joints resistant to fracture under nugget brittle rupture. Besides, the 304/304 joints possess high energy absorption in the lap-shear test, even for interfacial failure mode samples. This is confirmed in the digital image correlation tests which show the presence of a massive plastic deformation in the 304/304 joint fracture in interfacial failure mode.

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1. Introduction

Stainless steel materials have been widely investigated for their application in the next generation vehicle body manufacturing. This is because they possess superior corrosion resistance and energy-absorption ability compared with carbon structural steels [1-3]. For instance, austenitic stainless steel is used to design the roofs of vehicles [4]. The high nickel element content makes the traditional 300 series stainless steels expensive, which limits its application in automotive industry. One solution for this challenge is the development of the new generation of Ni-free austenitic stainless steels [5] which are often require high Mn content to create a full austenitic microstructure at room temperature. In addition, employing hybrid austenitic-ferritic stainless-steel structures presents a promising pathway that minimizes Ni-consumption [6-8]. The manufacture of this hybrid structure requires dissimilar joining process of austenitic stainless steel (ASS) with ferritic stainless steel (FSS).
The high portion of alloying element of stainless makes its metallurgical reaction more complex than its counterpart carbon structural steel, which makes the welding process more complex. Typically, the phase transformation of the stainless-steel weld can be predicted using constitution diagrams, such as Schaeffler diagram [9], WRC-1992 diagram [10], and Balmforth diagram [11]. Although significant progress has been made in the prediction of the phase composition of stainless steel welds based on constitution diagrams, this approach has some drawbacks. Typically, constitution diagrams are laid down in two-dimensional geometry for engineering convenience. The alloy components should be grouped into two equivalents, i.e., chromium equivalent (Cr<sub>eq</sub>) and nickel equivalent (Ni<sub>eq</sub>). However, numerous studies have reported that elements other than Cr and Ni may play important roles not only in ferritic-austenitic phase promoting but also in precipitation forming [12], solidification mode [13,14] and so on. In the case of dissimilar metal joining, the composition of the resultant weld is different from that of typical commercial materials, which challenges the use of constitution diagrams. Calculation of phase diagrams (CALPHAD) is a modern methodology that can predict the multi-component phase transformation based on thermodynamics [15]. Tate et al. employed CALPHAD technique to explore the solidification mode of 21Cr-6Ni-9Mn stainless steel laser beam welds (under different impurity content) [13,14]. The results showed that both the impurity content and the high cooling rate lead to the deviation of the primary solidification mode. Frank employed the CALPHAD technique to investigate the austenite reversion phenomenon in martensitic stainless steels [16]. Elsewhere, Yu et al. demonstrated the relationship between the high temperature liquid/ß-ferrite)/(γ-austenite) metallurgical reaction and the solidification crack sensitivity of austenitic stainless steel melts. These findings show the high utility of CALPHAD technique in weldability analysis [17].

Spot welded joints present more complex response behaviors under external load than the butt joint due to its hybrid stress state at the load bearing site [1,2,18–22]. Typically, for a resistance spot welding (RSW) structure, the metallurgical factors (strength and toughness of the fusion zone or the heat affected zone) coupled with the weld’s physical factors (nugget size or the indentation) govern the mechanical properties of the joints [1,2,23]. Thus, these physical factors render the mechanical properties of the dissimilar stainless steels RSW joints more complex to understand compared to butt joints [24,25]. Recently, it has been recognized that digital image correlation (DIC) is suitable for investigating the complex mechanical behavior of a non-homogeneous structure (specially, a welded structure) [26,27]. Kang et al. employed this technique to characterize the constitutive behavior of aluminum alloy resistance spot welds (which is difficult to extract by traditional methods) [28]. Zhang et al. designed a kind of half-cut sample to reveal the in situ stain field evolution at the cross-section of the RSW joint during the lap-shear test [29]. Thus, this method can be used to comprehensively analyze the mechanical properties of RSW joints (under dissimilar materials combination).

This work performed RSW tests among similar and dissimilar ferritic AISI 430 and austenitic AISI 304 stainless steels. The phase evolution of dissimilar 430/304 weld was analyzed by the CALPHAD technique. The evolution of strain distribution at the cross-section during the lap-shear test was measured by DIC. The mechanical and microstructural properties of RSW joints in the ferritic AISI 430 and austenitic AISI 304 stainless steel materials were compared.

2. Experimental procedure

2.1. Welding parameters

In this study, 1.5-mm-thickness wrought ferritic AISI 430 and austenitic AISI 304 stainless steels (the chemical composition is shown in Table 1) were used.

<p>| Table 1 – Chemical composition of ferritic AISI 430 and austenitic AISI 304 stainless steel (wt.%). |</p>
<table>
<thead>
<tr>
<th>Composition</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>Cr</th>
<th>S</th>
<th>P</th>
<th>Ni</th>
</tr>
</thead>
<tbody>
<tr>
<td>AISI 430</td>
<td>≤0.12</td>
<td>0.4</td>
<td>0.4</td>
<td>17.0</td>
<td>≤0.03</td>
<td>≤0.04</td>
<td>–</td>
</tr>
<tr>
<td>AISI 304</td>
<td>≤0.06</td>
<td>0.32</td>
<td>1.38</td>
<td>18.4</td>
<td>≤0.03</td>
<td>≤0.04</td>
<td>8.17</td>
</tr>
</tbody>
</table>

A 220-kW medium frequency direct current resistance spot welder was used to perform the welding tests in the similar and dissimilar ferritic AISI 430 and austenitic AISI 304 stainless steels. A pair of RWMA class II chrome C18200 spherical tip electrodes with 50 mm diameter were employed. Orthogonal trial L<sub>16</sub>(4<sup>2</sup>) for each similar and dissimilar combination was utilized to investigate the effects of the welding current I, the conduction time t and the electrodes force F on the joining quality of the workpieces. The details of these parameters are shown in Tables 2 and 3.

<p>| Table 2 – Factors and levels of the orthogonal trial. |</p>
<table>
<thead>
<tr>
<th>Factors</th>
<th>Levels</th>
</tr>
</thead>
<tbody>
<tr>
<td>Welding current I (kA)</td>
<td>3</td>
</tr>
<tr>
<td>Conduction time t (ms)</td>
<td>50</td>
</tr>
<tr>
<td>Electrode force F (kN)</td>
<td>1</td>
</tr>
</tbody>
</table>

The peak load, energy absorption and nugget size were collected as the end of the lap-shear test. The mechanical properties of stainless-steel RSW joints under each combination were compared based on the liner fitting plots of the above-mentioned data. Only the nugget size that above it makes all of the samples fractured in pull-out mode was recorded as the critical failure mode transition nugget size D<sub>c</sub>. The higher the D<sub>c</sub> value, the less tendency that the joints fractured in PO mode.

2.2. Microstructure analysis

The similar and dissimilar stainless-steel welded joints under No. 10 parameter combination were selected for the microstructure analysis. These samples were wire cut to expose their cross-section. After being processed by standard metallographic procedures, hydrochloric acid and iron chloride hydrochloride solution (30-ml-HCl, 50-g-FeCl<sub>2</sub> and 70-ml-H<sub>2</sub>O) were used as the etch to show the macrostructure
of the welds. In addition, sodium metabisulfite solution (10 g Na₂S₂O₅ and 100 ml H₂O) was used to reveal the ferrite and austenite within the dissimilar 430/304 nugget for high magnification microstructural observation. In this case, the ferrite appears gray whereas the austenite appears white. If present, martensite appears black [30]. The optical microstructure (OM) photos were taken by a metallographic microscope (GX51, Olympus Corp.). Vickers hardness test was carried out at the cross-section of the joints using a HV-1000A micro-hardness tester (a 200 g load for 15 s holding time).

Commercial CALPHAD software JMat-Pro was employed to perform thermodynamic calculations of the dissimilar 430/304 nugget. Firstly, phase transformation under equilibrium state was calculated by considering 50%-50% mixing of two workpiece. Next, continuous cooling transformation (CCT) diagram was calculated. A combination of the X-ray diffraction (XRD) analysis and the Bruker D8 ADVANCE diffractometer were used to qualitatively analyze the phase evolution. A μ-X360s X-ray analyzer was used to collect the large angle X-ray diffraction of the dissimilar 430/304 nugget from 360°, which shows the crystal preferential orientation of the nugget (by observing the intensity distribution of the resultant Debye ring).

### 2.3. Mechanical property test

Lap-shear test was employed to analyze the mechanical properties of each RSW joints in this study. The dimensions of the lap-shear test sample are shown in Fig. 1(a). A CSS-44100 material test system was used for these tests (under a tensile speed of 10 mm min⁻¹). For the strain measurement, the lap-shear samples were wire cut to expose the cross-section of the welds, as shown in Fig. 1(b). Specules were painted on the obverting plane. A VIC-2D digital image correlation system was used to record the in-plane strain field during the test. The tensile speed for the half-cut samples is half of that of the typical lap-shear samples. The collection frequency of the CCD camera in the DIC system was 5 frames per second. A previous work [31] showed that a biaxial stress field within the observing plane appears during the lap-shear test.

The DIC system first collected the displacement field data of the observing plane. Subsequently, the strain field evolution was calculated from these data. Von Mises equivalent strain, εₑq, was calculated based on the following form of Eq. (1).

\[ \varepsilon_{eq} = \sqrt{\frac{2}{3} \left( \varepsilon_{xx}^2 + \varepsilon_{yy}^2 + 2\varepsilon_{xy}^2 \right)} \]  

where x and y are the two axes of rectangular coordinates, εₓₓ and εᵧᵧ are the in-plane normal strain, and εₓᵧ are the in-plane shear strain.

### 3. Result and discussion

#### 3.1. Microstructure evolution

Fig. 2 shows the macrostructure of the welds under the three kinds of workpiece combination in this study. It can be seen that all the samples fully solidified in columnar dendritic mode. This caused the occurrence of a shrinkage cavity at the end of the solidification process as shown in Fig. 2(a) and (c). It has been stated [1,2] that the solidification and post-solidification transformation paths of similar 304/304 and 430/430 nugget can be represented as follows:

\[ L \rightarrow L + \delta \rightarrow L + \delta + \gamma \rightarrow \delta + \gamma \]  

\[ L \rightarrow L + \delta \rightarrow \delta \rightarrow \delta + \gamma \rightarrow \delta + \alpha' \rightarrow \alpha \rightarrow \alpha' \rightarrow C_1 \]  

where Eq. (2) refers to the phase transformation path of the 304/304 nugget. L is the liquid; δ is the high temperature ferrite; γ is the austenite. Eq. (3) refers to the phase transformation path of the 430/430 nugget. α is the ferrite at room temperature; α’ is the martensite; C₁ is the primary carbide.

For the 304/304 nugget, the liquid pool solidified in the ferritic-to-austenitic mode. Under equilibrium state, all the ferrite at high temperature is transformed to austenite. Considering that the cooling rate of RSW is relatively high (7800–8000 K s⁻¹) [1,2], the segregated elements cannot migrate completely, and thus some ferrite remains at the grains’ boundary of austenite when the nugget is cooled to room temperature. For the 430/430 nugget, the liquid pool solidified in the fully ferritic mode. The subsequent solid-state phase transformation depicted by the equilibrium phase diagram shows a portion of high temperature δ-ferrite may be transformed to γ-austenite. The primary carbide C₁ is precipitated from this amount of γ-austenite during the subsequent cooling process due to the decrease in carbon solubility. Similarly, due to the high cooling rate of RSW, this γ-austenite undergoes transformation into α’-martensite. Typically, this α’-martensite is found at the grains’ boundary zone of α-ferrite (transformed directly from the high temperature δ-ferrite) [1,2].

The macrostructure of the dissimilar 304/430 welds (Fig. 2(b)) reveals that there is no obvious macro-segregation in the nugget. This is due to the strong electromagnetic stirring of the liquid pool during RSW [32]. Fig. 3 shows the phase transformation path of the dissimilar 304/430 nugget. Under equilibrium state, the calculation result shows that the nugget

| Table 3 – Parameters combination of the orthogonal trial (L=64²). |
|------------------|------------------|------------------|
| Exp. | Welding current I (kA) | Conduction time t (ms) | Electrode force P (kN) |
| 1 | I | I | I |
| 2 | I | II | II |
| 3 | I | III | III |
| 4 | I | IV | IV |
| 5 | II | I | II |
| 6 | II | II | I |
| 7 | II | III | IV |
| 8 | II | IV | III |
| 9 | III | I | III |
| 10 | III | II | IV |
| 11 | III | III | I |
| 12 | III | IV | II |
| 13 | IV | I | IV |
| 14 | IV | II | III |
| 15 | IV | III | II |
| 16 | IV | IV | I |
Fig. 1 – (a) Dimensions of the typical lap-shear test sample; (b) Dimensions of the DIC test sample.

Fig. 2 – Macrostructure of three kinds of welds: (a) 304/304; (b) 304/430; (c) 430/430.

may be a kind of $\alpha$-ferrite/$\gamma$-austenite duplex microstructure. Based on the CCT diagram, the primary carbide Cr$_23$C$_6$ cannot be precipitated from the $\gamma$-austenite under the cooling rate of RSW (Note that the cooling curves faster than 10 °C/min do not across the carbide precipitating curves). It can be inferred that large-scale martensitic phase transformation will occur in the dissimilar 304/430 nugget. The 360° large angle X-ray diffraction of the dissimilar 304/430 nugget is shown in Fig. 3(b) and (c). These results confirm that the fully columnar dendritic solidification mode forms a stronger texture at this site. The OM figure under high magnification (Fig. 3(d)) exhibits that the dissimilar 304/430 nugget mainly consists of $\alpha$-ferrite columnar grains (the gray phase), with some $\gamma$-austenite (the white phase) and $\alpha'$-martensite dispersed over the grains’ boundary zone.

Further details on the phase composition are shown in Fig. 4. The peaks of austenite (PDF#44-1293), ferrite (PDF#33-0397) and martensite (PDF#03-0411) are shown.

Fig. 5 shows the hardness distribution of the three kinds of joints in this study. The data shows that the hardness of the nugget of 304/430 and the 430/430 samples is higher than their base metals. This is due to the generation of $\alpha'$-martensite at the nugget. The FZ’s hardness is lower than that of the base metal in the 304/304 samples. Moreover, a distinct soft zone can be seen in the heat affected zone (HAZ) of 430/430 sample. This phenomenon was reported by Alizadeh et al. [1,2,33], which is
Fig. 3 – Microstructural characterization of the dissimilar 304/430 nugget: (a) phase evolution under equilibrium state; (b) the Debye ring; (c) profiles of the intensity distribution curves; (d) OM microstructure; (e) CCT curves of the nugget.

Fig. 4 – XRD pattern of the dissimilar 304/430 welds.

Fig. 5 – Vickers hardness distribution.
attributed to the excessive coarsening of grains in the presence of free martensite at this local site. Note that the hardnes is not significantly different between the HAZ and BM of 304. This is ascribed to the annealed state of the sheet used in this study.

3.2. Mechanical properties

Fig. 6 shows the peak load and energy absorption of three kinds of samples subjected to the lap-shear test. Under equivalent nugget size, the peak load is comparable between 304/304 and 304/430 samples both of which are higher than that of 430/430 to some extent. The energy absorption was significantly different among the three kinds of joints, decreasing in the following order: 304/304 > 304/430 > 430/430. As the nugget size increases, the RSW joint's fracturing mode transforms from interfacial failure (IF) mode to the nugget pull-out failure (PO) mode. Compared with the 304/304 joints (Dc \( \approx 7 \text{ mm} \)), the dissimilar 304/430 joints (Dc \( \approx 4 \text{ mm} \)) are more prone to fracture in the PO mode as shown in the test. Based on the parameters selected in this study, the critical diameter for the 430/430 joints could not be identified (note that some joints fractured in IF mode possessed larger nugget size than those of the PO failure mode). Thus, the IF-to-PO transition tendency follows the order: 304/430 > 304/304 > 430/430 in this study. Currently, the analytical models used to determine the failure mode transition in the lap-shear test are derived from Eqs. (4)–(6) [34–36].

\[ P_{IF} = 1.44 \cdot d \cdot K_{eq} \cdot \sqrt{t} \]  
\[ P_{IF} = \frac{\pi}{4} \cdot d^2 \cdot \sigma_{IF} \]  
\[ P_{PO} = \pi \cdot d \cdot t \cdot \sigma_{PO} \]  

where \( P_{IF} \) and \( P_{PO} \) are the peak load in IF and PO mode, respectively; \( d \) is the nugget diameter; \( t \) is the thickness of workpiece; \( K_{eq} \) is the equivalent stress intensity factor; \( \sigma_{IF} \) is the shear strength of the fusion zone; \( \sigma_{PO} \) is the ultimate tensile strength of the local site where initial cracking in PO mode occurs. When the value of \( P_{IF} \) is above that of \( P_{PO} \), the fractured mode transition occurs. Note that Eqs. (4) and (5) are cited from different literatures that depict the same peak load of IF mode but in different forms. The former model (fracture-toughness-based, Eq. (4)) shows that a sharp notch exists near the nugget at the workpiece/workpiece interface. When the equivalent stress intensity factor at this notch exceed a critical value, the crack propagates along the faying surface of the RSW joint (which means IF failure happens) [36]. The later model (strength-based, Eq. (5)) reveals that the failure mode of the RSW joint depends on the local load-bearing ability competition between the fusion zone and its peripheral materials [34,35].

According to the strength-based model, the higher hardening ratio between the fusion zone and base metal, the higher the probability that the joints failure in PO mode occurs. This is based on the assumption that the local strength is proportional to its local hardness. Pouranvari et al. [34,35] supported this concept by summarizing the results of lap-shear tests of similar DQSK, DP600, DP780, DP980 and AISI 304 RSW joints. They concluded that AISI 304 joints are prone to fracture in the IF mode due to its soft fusion zone. However, in this study, the 304/304 joints present lower tendency of IF failure mode compared with the 430/430 joints although the later have a higher hardening ratio of the fusion zone. This phenomenon was also observed in martensitic stainless steel RSW joints, which is due to the low fracture toughness of its fusion zone [36]. There are two kinds of IF failure mechanisms for RSW joints: (1) The strain concentration at fusion zone, which triggers the IF failure when this strain reaches the plastic limit of the local site. This mechanism is suitable for the workpiece that possesses low hardening tendency such as mild steel, interstitial free steel or austenitic stainless steel. Numerous studies have characterized the signs of grains distortion at the cracking path under this circumstance [23,35]. (2) The notch at the nugget edge can be considered as a crack tip for RSW joints. For some high hardening tendency materials, when the equivalent stress field intensity factor exceeds a critical value, the IF failure may occur. Under this condition, the nugget is very brittle, which prevents plastic deformation of the grains [36].

Fig. 7 shows the typical load-displacement curves of the samples in this study. Note that the scale of the displacement is different for each type of joint. It can be seen that the total displacement at which the 304/304 sample reaches its initial fracturing is markedly long, which involves a large plastic deformation. The degree of the plastic deformation is lower in the 304/430 sample and 430/430 sample. This is in accordance with the energy absorption data shown in Fig. 6.

3.3. DIC analysis of the joints

Due to the unbalanced external force moment, mixed normal/shear strain appears at the cross-section of RSW joints during the lap-shear tests [31]. Von Mises equivalent strain \( \varepsilon_{eq} \) was calculated from the normal and shear strain data in this study, based on Eq. (1).

Fig. 8 shows the strain field evolution of the 304/304 sample during the lap-shear test. The analysis demonstrates that a strain concentrate zone appears at the nugget center. This zone spread to the whole nugget area as the external load increases. The massive plastic deformation involved increases the energy absorption capacity of the 304/304 sample, although the samples failure is in the IF mode. The DIC data support the idea that the fracturing mechanism of 304/304 RSW joint in IF mode involves the strain concentrated in the soft nugget area that exceeds the plastic value.

Fig. 9 shows the strain field evolution of the 304/430 sample during the lap-shear test. It can be observed that no strain occurred at the nugget area. This is because the martensite formed under the high cooling rate of RSW increases the local strength of dissimilar 304/430 nugget. When the external load is above a critical value, necking occurs at the sheet metal (AISI 430 side) at the nugget periphery. This triggers the joint fracture in PO mode.

Fig. 10 shows the strain field evolution of the 430/430 sample during the lap-shear test. Only a very limited degree of strain could be observed at the nugget area, implying that this local site possesses poor ductility. This strain evolution feature
supports the observation that the IF fracturing mode of the 430/430 joint is derived from the low toughness of the nugget, which means Eq. (4) is more suitable to represent the IF failure of this kind of joint than Eq. (5).

Fig. 11 shows the final fracturing morphology of the corresponding samples from Fig. 8 to Fig. 10. The data sets indicate that the 304/304 and 430/430 sample is fractured in IF mode whereas the 304/430 sample is fractured in PO mode. Although the 304/304 sample fractured in IF mode, the presence of a massive plastic deformation (Fig. 8), increases the energy absorption ability. For the 430/430 sample, the nugget is very brittle which is splintered to cause the IF failure of the joint. The dissimilar 304/430 welds are tougher than the 430/430 welds, making the joint be fractured in PO mode.

4. Conclusion

This study compares the mechanical properties of the 304/304, 304/430 and 430/430 resistance spot welding joints. The key points are summarized as follows:

(1) Due to its columnar dendritic solidification mode, the nugget of the dissimilar 304/430 nugget has a strong texture consisting of α-ferrite columnar grains, with some γ-austenite and some α'-martensite dispersed over the grains’ boundary zone. This mixed-phase nugget displays a similar Vickers hardness with the 430/430 nugget but possess higher fracture toughness than its counterpart.
Unlike the differences in the peak loads, the energy absorption of the three kinds of joints varies significantly. The 304/304 joints have high energy absorption although fractured in IF mode. This phenomenon is associated with the multiple fracturing mechanisms of the IF failure mode.
(3) The digital image correlation tests of the three kinds of samples show that: (i) the IF failure of the 304/304 samples occurs when the strain concentrated in the nugget area exceeds its local plastic limit. This involves a massive plastic deformation which in turn increases energy absorption leading to IF failure. (ii) The IF failure of the 430/430 samples occurs when the equivalent stress field intensity factor at the notch between two workpieces exceeds a critical value. (iii) The nugget of 304/430 samples is harder than that of the 304/304 samples and has higher local toughness than the 430/430 samples. This makes the dissimilar joints less prone to fracture via the two IF failure mechanisms, which increases the tendency to be fractured in PO mode.

Conflicts of interest

The authors declare no conflicts of interest.

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