Welding Metallurgy of Duplex Stainless Steel during Resistance Spot Welding

Solidification and postsolidification solid-state phase transformation during resistance spot welding of 2304 duplex stainless steel were investigated

BY S. H. ARABI, M. POURANVAR, AND M. MOVAHEDI

ABSTRACT

This paper investigates the metallurgical and mechanical response of 2304 duplex stainless steel — as an interesting candidate for automotive body-in-white applications — to resistance spot welding. The results showed the high cooling rate associated with the resistance spot welding process suppressed the postsolidification ferrite-austenite transformation leading to improper ferrite-austenite phase balance with a reduced volume fraction of austenite and consequent precipitation of chromium-rich nitrides. The effects of welding current, as the key parameter determining the weld heat input, on the austenite volume fraction and precipitation are discussed. The minimum fusion zone hardness corresponding to the highest austenite volume fraction and minimum volume fraction of nitrides was obtained when the cooling rate was the lowest. The failure mode transition, peak load, and energy absorption of the welds were investigated. It was found that the fusion zone size is the key factor controlling peak load and energy absorption of the welds. Mechanical properties of the duplex stainless steel spot welds were compared with those of other automotive steels.

KEYWORDS

• Duplex Stainless Steel • Resistance Spot Welding • Phase Balance • Phase Transformations • Mechanical Performance

Introduction

Due to the continued rising vehicle safety and crash requirements and the need for weight reduction in the automotive industry, the use of ultrahigh-strength steels in body-in-white applications is rapidly increasing. The ferrite-martensite dual-phase steels, transformation-induced plasticity (TRIP) assisted steels, and martensitic steels are the most common advanced high-strength steels (AHSSs) that are implemented in today’s car-body design (Refs. 1–3).

The introduction of advanced high-strength steels in the automotive industry is accompanied with the challenge of their weldability. Automotive structural assemblies use groups of spot welds to transfer load through the structure during a crash. Additionally, spot welds can act as fold initiation sites to manage impact energy (Ref. 4). Vehicle crashworthiness, which is defined as the capability of a structure to provide adequate protection to its passengers against injuries in the event of a crash, largely depends on the integrity and the mechanical performance of the spot welds (Ref. 5). Spot weld failure during a crash is a critical issue for crashworthiness, stiffness and noise, vibration and harshness performance of the vehicle. Therefore, the quality, performance, and failure characteristics of resistance spot welds are important for determining the durability and safety design of the vehicles (Refs. 6–8).

Through an examination of past research (Refs. 9–20), it is recognized that resistance spot welding (RSW) of AHSSs has several challenges, including the following:

1) Complex phase transformation in the weldment: The strength/ductility properties of AHSS are governed by their sophisticated, designed microstructure. However, the stability of AHSS base metal microstructure is significantly affected by weld thermal cycle (Ref. 21). This produces significant property (strength and toughness) mismatch among the fusion zone (FZ), heat-affected zone (HAZ), and base metal, which in turn affects the load-bearing capacity and failure behavior of the AHSS spot welds. The two important phase transformations in AHSS welds are as follows:

• Martensite formation in both fusion zone and coarse-grain heat-affected zone (Ref. 9–12), which can induce an adverse effect on the weld failure characteristics during some loading conditions (e.g., peel and cross-tension tests) (Defs. 13–15).

• Tempering of the base metal martensite in the subcritical HAZ, which creates softening compared to the base metal. The HAZ softening phenomenon can take place during welding of martensite-containing AHSS (e.g., dual-phase steels and martensitic steels) (Refs. 16–18). This phenomenon can reduce the load-
bearing capacity of spot welds compared to the strength expected from the initial base metal microstructure (Refs. 19, 20).

2) High susceptibility to interfacial failure mode: Spot welds can fail in two distinct modes (Ref. 9): 1) interfacial failure (IF) mode in which the fracture propagates through the fusion zone. It is believed this failure mode has a detrimental effect on the crashworthiness of the vehicles; 2) pullout failure (PF) mode in which the failure occurs via withdrawal of the weld nugget from one sheet. Despite the fact that AWS D8.1 (Ref. 21) allows interfacial failures in AHSS, generally the PF mode exhibits the most satisfactory mechanical properties due to its higher associated plastic deformation and energy absorption. It has been shown that AHSSs exhibited a higher tendency to fail in the interfacial mode compared to the low-strength traditional steels (Refs. 21–24). It has also been shown that sizing based on $4t^{0.5}$ and $5t^{0.5}$ rules do not ensure pull-out failure mode in AHSS welds (Refs. 23–25). This can affect the energy absorption capability of AHSS welded joints. Therefore, in some situations, postweld treatment (e.g., in-situ tempering) is needed to modify the as-welded microstructure of the FZ and HAZ to improve the failure behavior of the weld (Refs. 26–28). However, this increases the time required to produce a weld with optimum mechanical properties.

These challenges can lead to the possibility of using stainless steels as a new material for structural applications in the automotive industry (Refs. 29–33). Duplex stainless steels (DSS) are excellent candidates due to a combination of high mechanical properties and corrosion resistance (Refs. 34, 35). The strain hardening and the behavior at high strain rates of duplex stainless steels (DSSs) (Refs. 36, 37) make them especially suitable in vehicles’ crumple zones for energy absorption in crashes. Moreover, the DSSs do not exhibit martensitic transformation and related problems during welding. Therefore, the use of DSSs is a good solution for the welding problems of martensite-containing AHSSs.

Implementation of new materials in automotive bodies-in-white requires thorough knowledge of their metallurgical response to the resistance spot welding thermal cycle. The austenite/ferrite phase balance and the absence of Cr$_2$N and sigma and chi ($\chi$) precipitates are two key factors controlling the mechanical behavior of DSSs (Refs. 38–42). Both factors are influenced by the high cooling rates inherent in resistance spot welding. Therefore, improving the knowledge regarding the phase transformation and failure behavior of DSSs is a priority for their successful implementation in vehicle applications.

There are limited publications on welding behavior of DSSs during resistance spot welding. Pouranvari et al. (Refs. 32, 33) investigated microstructure and mechanical properties of a high-carbon, Ti-bearing 1Cr21Ni5Ti duplex stainless steel. It was found that the presence of Ti as a strong carbide former and ferrite promoting element can remove carbon from the matrix by TiC formation, reducing the tendency for austenite formation. Thulin et al. (Ref. 43) investigated the weldability of 2205, 2304, and 2101 duplex stainless steels during resistance spot welding. They found that applying a second pulse current after the first melting/solidification pulse can

### Table 1 — Chemical Composition and Mechanical Properties of the Investigated Duplex Stainless Steel

<table>
<thead>
<tr>
<th>Chemical Composition</th>
<th>Tensile Properties</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ni</td>
<td>Cr</td>
</tr>
<tr>
<td>4.7</td>
<td>23.4</td>
</tr>
</tbody>
</table>

(a) $YS = $ Yield strength
(b) $UTS = $ Ultimate tensile strength
(c) $EI = $ Total elongation
increase the austenite content of the weld nugget. Moreover, they found that a shielding gas can be used to suppress oxidation on the top surface of the spot weld improving the aesthetic appearance. However, the effect of welding heat input on the phase transformations, failure behavior, and mechanical properties of the welds are not detailed in the previous works.

This paper details the solidification and postsolidification solid-state phase transformations during resistance spot welding of 2304 duplex stainless steel. The influence of welding heat input on austenite/ferrite phase balance and secondary phase precipitation were analyzed. Finally, the failure characteristics and mechanical properties of the 2304 DSS resistance spot welds were investigated.

**Experimental Procedure**

A 1.5-mm-thick sheet of 2304 duplex stainless steel was used as the base metal for this research. Table 1 shows the chemical composition and tensile properties of the base metal. Resistance spot welding was performed using an AC pedestal-type machine with a programmable logic controller (PLC) working at 50 Hz. Welding was conducted using a 45-deg truncated cone RWMA Class 2 electrode with an 8-mm face diameter. Squeeze time, welding time, electrode holding time after current off, and electrode pressure were kept constant at 0.8 s, 0.2 s, 0.2 s, and 4.5 kN, respectively. Welding current was incrementally increased from 6 to 18 kA with a step size of 1 kA. The welding schedule is schematically shown in Fig. 1A.

The static tensile-shear test samples were prepared according to the ANSI/AWS/SAE/D8.9-97 standard (Ref. 44). Figure 1B shows the dimensions of the test specimen. The tensile-shear tests were performed at a cross-head speed of 10 mm/min. Peak load (measured as the peak point in the load-displacement curve) and failure energy (measured as the area under the load-displacement curve up to the peak load) were extracted from the load displacement curve. Failure modes of spot welds were determined by observing the weld fracture surfaces.

Metallographic specimens were prepared using standard metallography procedures. Optical microscopy was used to examine macro/microstructure of the weldments. Marble etchant (50 mL H₂O, 50 mL HCl, and 10 g CuSO₄) was used for macrostructural observations. Braha’s etchant (85 mL H₂O, 15 mL HCl, 1 g K₂S₂O₅) was used for microstructural observation. Volume fraction of ferrite and austenite were determined using a Clemex image analyzer. Kalling’s no. 1 (33 mL H₂O, 1.5 g CuCl₂, 33 mL HCl, and 33 mL C₂H₅OH) was used to reveal very fine precipitates within the ferrite grains. Vikers microhardness test was used to assess the hardness values of the weldment, using a Bohler microhardness tester. An applied load of 100 g and a time of 10 s were used.

**Results and Discussion**

Mechanical performance and failure mode of resistance spot welds are affected by weld physical macrostructural attributes (i.e., FZ size and electrode indentation depth) and weld metallurgical attributes (i.e., microstructure/hardness of the FZ and HAZ). Figure 2A shows a typical macrostructure of DSS spot welds indicating a heterogeneous metallurgical structure. According to Fig. 2B, DSS base metal exhibited a balanced duplex microstructure of ferrite and austenite. The ferrite grains are precipitate-free. It was necessary to investigate whether resistance spot welding might lead to precipitation of deleterious phases. The 2304 resistance spot
welds exhibited a very narrow HAZ (100–200 μm depending on the welding current) due to the high power density of RSW (Ref. 45) as well as low thermal conductivity of the base metal. The next section emphasizes the phase transformation in the fusion zone during RSW.

**Phase Transformations in the Fusion Zone**

The fusion zone, which experienced melting and resolidification during the weld thermal cycle, showed a columnar grain cast structure zone with an equiaxed grain structure zone in a narrow region of the weld metal adjacent to the fusion boundary — Fig. 2C. To explain the presence of the equiaxed grains at the fusion boundary (Fig. 2D), two possibilities need to be considered. The first is heterogeneous nucleation of the originally present intermetallic phase in the base metal, as has been reported in the case of welding of Al-Cu-Li alloys (Ref. 46). Because the 2304 base metal is precipitate-free, this mechanism can be dismissed. The second mechanism is based on the formation of a chilled zone at the fusion boundary (Refs. 47, 48), where the cooling rate is high-
est. Consequently, many small grains were nucleated and grew in random directions at the nugget edge to form an equiaxed grains region, similar to the chilled zone that is formed on the surface of the castings and ingots, e.g., in the vicinity of the metallic molds.

Competitive growth between equiaxed grains causes those grains with their easy growth direction antiparallel to the heat flow direction to grow toward the center of the nugget, leading to the formation of the columnar grains in this region.

Figure 2E shows the microstructure of the FZ center. The FZ consisted of an unbalanced microstructure of ferrite and austenite with extensive precipitates within the ferrite grains. Regarding the solidification and postsolidification phase transformation in the FZ of the DSS resistance spot welds, the following points should be considered:

1) **Solidification mode.** The solidification mode strongly depends on the chemical composition of the FZ. It is shown that welds with a Cr\textsubscript{eq}/Ni\textsubscript{eq} ratio (expressed as in the WRC-1992 diagram) higher than 1.85 will solidify in primary ferrite mode (Ref. 49). According to the chemical composition of the investigated DSS, the Cr\textsubscript{eq}/Ni\textsubscript{eq} ratio for the FZ is 2.69. Therefore, the FZ will solidify as 100% ferrite.

2) **Transformation path.** The transformation path can be explained with the help of Fig. 3A (Ref. 50). As can be seen, at the end of solidification, the DSS is fully ferritic. When the temperature falls below the ferrite solvus temperature, the solid-state transformation of ferrite to austenite starts. The subsequent transformation of austenite to martensite during cooling to room temperature depends on the austenite stability. The austenite stability can be assessed based on the martensite start temperature (M\textsubscript{s}).

The following equation can be used to estimate M\textsubscript{s} (in °C) in stainless steels (Ref. 49):

\[
M_s = 526 - 12.5Cr - 17.4Ni - 29.7Mn - 31.7Si - 31.7Mo - 354C - 20.8(Mo)C + 22.41(Cr+Mo)C
\]

where the chemical symbols indicate the weight percentage of the elements present. For the investigated DSS, the M\textsubscript{s} was calculated as -45°C, which is well below the room temperature which confirms that the austenite is stable. After continuous cooling below approximately 1000°C (Ref. 49), the precipitation reactions took place over a temperature range. Therefore, the transformation path of the FZ in 2304 WELDING RESEARCH
DSS resistance spot welds can be summarized in four stages, as follows:

1. **Stage I**: The formation of the weld nugget,

2. **Stage II**: The solid-state transformation of ferrite to austenite,

3. **Stage III**: The solid-state transformation of austenite to ferrite and austenite,

4. **Stage IV**: The precipitation of phases and the formation of the HAZ.

The phase balance in the FZ can be influenced by the welding heat input, which affects the cooling rate. The cooling rate can be estimated using the model of Gould (Ref. 10):

\[
\frac{\partial T}{\partial t} = \frac{\alpha}{4t_s^2} \left( T - T_p \right)
\]

where, \( \alpha \) is the thermal diffusivity of the steel sheets, \( T_p \) is the maximum temperature experienced in the FZ during the welding process, \( t_s \) is the sheet thickness, and \( t_e \) is the electrode face thicknesses (i.e., the distance of the water-cooled hole to the electrode surface). \( k_s \) and \( k_e \) are the thermal conductivities of the steel and the electrode, respectively, and \( x \) is the position through the spot weld in the sheet thickness direction.

Figure 3B shows the calculated cooling rates for RSW of 2304 DSS with different thicknesses for the temperature range of 400–1400°C. The cooling rates in the range of 800–1200°C are critical for postsolidification solid-state transformation. The cooling rates in this critical temperature range for 1.5-mm-thick sheet are 3000–3700°C/s. Therefore, the rapid cooling rate of the RSW process hinders the nucleation and growth process of the austenite formation resulting in improper phase balance (i.e., low austenite content).

It should be noted that the austenite volume fraction at the weld nugget edge was lower than that in the weld nugget center. This can be attributed to the higher cooling rate at the weld nugget edge due to its closer distance to the water-cooled Cu-based electrodes as well as more effective heat dissipation via heat sink into the base metal sheets. It should be added that the phase balance was also destroyed in the HAZ (see Fig. 2D).

**4. Effect of welding heat input on the phase balance.** According to the preceding discussion, the cooling rate can affect the phase balance in the FZ. Because the cooling rate is influenced by the heat input, the effect of welding current, as the key parameter affecting the amount of heat generation during RSW on the phase balance was studied. Figure 4 shows the effect of welding current on the FZ microstructure. Figure 5A gives the effect of welding current on the average volume fraction of austenite in the weld nugget center, measured using image analyzer software. Increasing the welding current from 6.5 to 16 kA increases the austenite volume fraction from 4 to 18%. This can be attributed to the decreased cooling rates at higher heat input (i.e., higher welding current). The higher heat input increases the available time for the \( \delta \rightarrow \gamma \) solid-state phase transformation to occur. The improvement of the phase balance at higher heat input has an upper limit when intensive expulsion occurs. In-
tensive expulsion leaves a deep indentation causing a significant reduction in sheet thickness. The reduction of sheet thickness increases the cooling rate due to more effective heat dissipation via water-cooled Cu-based electrodes. As can be seen in Fig. 3B, decreasing sheet thickness from 1.5 (nominal sheet thickness) to 0.9 mm (reduced sheet thickness due to indentation) increases the average cooling rate in 800°-1200˚C temperature range from 3400° to 8400˚Cs⁻¹. The unbalanced microstructure of the weld nugget calls for designing a proper postweld heat treatment to increase the amount of austenite as indicated by Kotecki (Ref. 51) and Thulin et al. (Ref. 43).

**5. Effect of welding current on the austenite morphology.** Two different austenite morphologies were dominant in the FZ including grain boundary austenite (GBA) and intragranular austenite (IGA). According to Fig. 4, it can be seen that the formation of austenite at δ-δ grain boundaries was dominant in all welds made using various welding currents. In welds made at a higher welding current, the formation of IGA is highlighted.

The transformation of δ to γ begins with the formation of a continuous thin layer of allotriomorphic austenite decorating the boundaries of the columnar δ-grains. It is believed this type of austenite forms by reconstructive transformation. The GBs of δ grains, as a high-energy site, are the preferred sites for formation of austenite. It is believed that the formation of IGA is due to partitioning of solutes within the δ-grains during solidification and postsolidification. The δ-ferrite grain interior is the least preferred site for formation of austenite. However, as it is observed in the microstructure of weldments, the IGA had the highest contribution to the austenite volume fraction of the FZ. This can be related to the large initial δ-ferrite grains. It is reported that a larger δ-ferrite grain size promoted formation of IGA since more space is then available for intragranular nucleation and growth reactions, which are otherwise stifled by events originating at the δ-ferrite grain boundaries (Refs. 52, 53).

**6. Prediction of ferrite content of the FZ using constitution diagrams.** The ferrite content of the fusion zone of the DSS can be predicted by the WRC-1992 diagram developed by Kotecki and Siewert (Ref. 54), the function fit model proposed by Babu et al. (Ref. 55) (accessible online at Ref. 56), and the Oak Ridge Ferrite Number (ORFN) model proposed by Vitek et al. (Refs. 57, 58) (accessible online at Ref. 59). Figure 6 shows the predicted ferrite number (FN) values using the above-mentioned models. Because the small scale of the weld nugget induced some difficulty in measuring FN using a Magne-Gage or Feritscope, the volume fraction of ferrite was obtained using the metallographic technique. Thereafter, the methodology developed by Kotecki in 1997 was used to convert FN to volume fraction of ferrite (Ref. 60). WRC-1992 predicts a FN of 90, corresponding to an austenite volume fraction of 0.39. In function fit mode, a FN of 83
(corresponding to austenite volume fraction of 0.43) was predicted. Both model predictions are well above the measured austenite volume fraction.

The WRC-1992 diagram and function fit model were developed based on arc welding data. The underestimation of FN using these models can be attributed to the high cooling rate of RSW compared with that of arc welding processes, which suppresses solid-state ferrite to austenite transformation resulting in a higher nonequilibrium residual ferrite content in the FZ. The ORFN model was able to predict a FN in stainless steels welds as a function of cooling rate and composition. According to Fig. 3B, the average cooling rate of DSS welds in the range of 800°-1200°C, where ferrite-to-austenite transformation occurs, is approximately 3400°C/s. According to the ORFN model, considering the cooling rate effect, the predicted FN is 110 (corresponding to austenite volume fraction of 25.4), which is much closer to the measured FN.

7. Precipitation reaction in the FZ. Extensive precipitation was evident within the ferrite phase of the FZ — Fig. 7A. While accurate analysis of the precipitates requires transmission electron microscopy (TEM) study, considering the low carbon content (0.03 wt-%) and comparatively high nitrogen content (0.15 wt-%) and high chromium content (23.4 wt-%) of the base metal, it can be concluded that the observed precipitates are Cr-rich nitrides. It has been reported that high nitrogen content tends to shift carbide precipitation to longer times (Ref. 42). Moreover, considering the rapid cooling rate of the RSW, the formation of intermetallic compounds (e.g., sigma phase, etc.) is hindered. The precipitation of Cr,N was linked to the unbalanced microstructure of the FZ. The solubility of nitrogen in ferrite was much less than that of the austenite. Therefore, considering the high volume fraction of ferrite in the FZ, provides high content of supersaturated nitrogen in the ferrite phase. This fact coupled with the high affinity of chromium for nitrogen leads to precipitation of Cr-rich nitrides.

8. Precipitation-free zone. As can be seen in Fig. 7, there is a precipitate-free zone in the ferrite adjacent to the austenite in the FZ. Due to the high solubility of austenite for nitrogen, it can act as a sink for nitrogen at elevated temperature. Short-range diffusion of nitrogen from ferrite into the austenite reduces the local concentration in the ferrite, and upon cooling through the precipitation range, there is little or no driving force for precipitation.

9. Effect of welding heat input on the precipitation in the FZ. According to the preceding discussion, the extent of the Cr,N precipitation in the FZ was a function of ferrite-austenite balance. Therefore, the lower austenite proportion, the higher volume fraction of Cr-rich nitride precipitates. Because the volume fraction of austenite is controlled by the cooling rate, the precipitation of Cr-rich nitride was also affected by the cooling rate. As can be seen in Fig. 7, increasing welding current decreased the extent of the Cr,N precipitation due to increasing the volume fraction of austenite, the sink for nitrogen, reducing the available nitrogen for precipitation as Cr,N.

10. Hardness evolution of the FZ. Figure 5B shows the effect of the welding current on the FZ hardness. The FZ hardness depends on 1) the proportion of the ferrite-austenite and their individual hardnesses (the ferrite is slightly harder than the austenite phase in duplex stainless steels (Ref. 61)), 2) volume fraction of precipitates, and 3) strengthening due to grain and phase boundaries. The minimum FZ hardness roughly corresponds to the welds with maximum volume fraction of austenite and hence minimum volume fraction of Cr,N precipitates within the ferrite.

Failure Mode

Generally, failure of resistance spot
welds can be categorized into four distinct modes, namely (Ref. 9)
a) Interfacial failure (IF) mode in which fracture propagates through the fusion zone (FZ).
b) Pullout failure (PF) mode in which failure occurs via withdrawal of the weld nugget from one sheet. In this mode, fracture may initiate in the base metal (BM), heat-affected zone (HAZ), or HAZ/FZ depending on the metallurgical and geometrical characteristics of the weld zone and the loading conditions.
c) Partial interfacial failure (PIF) mode in which the fracture first propagates in the fusion zone (FZ) and is then redirected toward the thickness direction.
d) Partial thickness-partial pullout (PT-PP) mode in which fracture initiates in a manner similar to PF mode. However, some part of mating sheet thickness is removed by a slant crack through the FZ during final crack propagation around the circumference of the weld nugget.

Failure mode of spot welds depends on several factors including FZ size, relative material properties of FZ, HAZ and BM, and loading condition as well as sample size (Refs. 9, 11, 22–24). Examination of the fracture surface of the failed samples showed the DSS spot welds failed in two modes during the tensile-shear test: interfacial mode and partial thickness-partial pullout mode.

1) **Interfacial failure mode.** Figure 8A shows a typical fracture surface of a spot weld that failed in the interfacial failure mode. In this mode, crack propagates through the weld nugget centerline and separates the weld nugget into two sections. As is observable in Fig. 8A, little plastic deformation was accompanied by this failure mode, which implies that a low energy absorption capability is allocated to this failure mode. The interfacial failure mode was observed at low welding current — Fig. 8B. Figure 8C shows the macrograph of the weld cross section loaded just after peak load. The relative displacement of upper and lower sheets indicates the failure is controlled by shear stress. The flow direction of the grains in the FZ also implies that the IF failure was driven by the shear plastic deformation localization of the FZ.

2) **Partial pullout-partial thickness.** Figure 9A shows the fracture surface of the welds made at a welding current of 11 kA failed in PP-PT. To determine the failure phenomena during PP-PT failure, the tensile-shear test was interrupted immediately after peak load. The macrograph of the weld loaded until this stage (Fig. 10A) indicates the necking and subsequent
cracking at nugget circumference at the base metal. It was found that after initial failure initiation, the crack redirected through the FZ as a slant crack, causing some part of the mating sheet thickness to be removed upon final separation — Fig. 10B.

According to Fig. 9A, the fracture surface has two distinct zones: the nugget fracture zone and pulled-out zone. Figure 9A–C shows the effect of welding current on the fracture surface of welds that failed at the PP-PT mode. Figure 10B–D shows the effect of the welding current on the weld macrograph after final fracture. By examining the fracture surfaces, the fraction of the pulled-out zone is calculated as the ratio of $A_{np}$ (the area of the weld that was pulled out during the tensile-shear test) to $A_t$ (the total area of the weld fracture surface). Figure 11 shows the variation of fraction of the pulled-out zone as a function of the welding current. As can be seen, increasing welding current from 11 to 18 kA increases the PF fraction of the failure from 21 to 73%. This is the result of increased weld nugget size at higher welding currents, which impedes the propagation of cracks into the FZ.

Considering the higher energy absorption associated with PF mode, the IF to PF failure mode transition is an important issue. According to Fig. 8C, D, and Fig. 10A, the failure mode of the welds can be considered as the competition between shear plastic deformation of the FZ (IF mode) and necking of the base metal (PT-PP mode). For the given material properties at the FZ and BM, the fusion zone size determines the failure mode. Enlarging the FZ above a critical value, the failure location changes from the FZ into the BM due to lowering the shear stress acting on the sheet-sheet interface and increasing the tensile stress acting on the BM in the through-thickness direction. According to Fig. 8B, increasing welding current beyond 11 kA, which corresponds to a weld nugget size larger than 6 mm, changes the failure mode from IF to PF. While sizing based on the $4t^{0.5}$ recommendation (Ref. 62) is not sufficient to obtain PF mode, the $5t^{0.5}$ recommendation (Ref. 62) can produce welds with PF mode.

**Mechanical Behavior**

Figure 12 shows the welding current has a profound effect on the load-displacement characteristics of the welds. The peak load and energy absorption of the welded samples were extracted from the load-displacement curves. The peak load represents the load-bearing capacity of the joint. The failure energy is a measure of the joint energy absorption capability with higher value that demonstrates an increase in weld performance reliability against impact loads (e.g., during severe accident). It has been shown that there is a direct relationship between the failure energy in the static tensile-shear test and the impact tensile-shear test (Ref. 62). Figure 13A shows the effect of the welding current on the peak load and energy absorption. While the peak load improved monotonically by increasing welding current, there is an optimum welding current to achieve best energy absorption. Figure 13B shows the effect of FZ size on the peak load of the welds. As can be seen, the peak load of the welds is more sensitive to FZ size in IF mode than in PF mode. This is due to the fact the bonding area of the welds that failed in IF mode is proportional to the square of the FZ size, while that in PF mode is linearly proportional to the FZ size (Ref. 23). The reduction of failure energy at high welding current (when heavy expulsion occurred) can be related to the large electrode indentation that lowers the strain energy required to failure initiation.

It is of note that the unbalanced microstructure of the FZ is not a critical factor for the mechanical behavior of the spot welds. In welds that failed in IF mode (made using welding current lower than 11 kA), the strength was controlled by the FZ size (Fig. 13B) and the FZ hardness. In this welding current range, the FZ hardness did not change significantly (ca. 10%). Moreover, the maximum hardness in the FZ was 350 HV, which does not produce brittleness in the FZ. In welds that failed in PF mode (made using welding current higher than 10 kA), the strength was controlled by the FZ size (Fig. 13B) and the strength of the pullout failure location. Because the failure initiation site of the DSS
welds in PF mode was BM, the δ-ferrite volume fraction of the FZ did not affect the peak load of the welds.

To provide a basis for comparison, mechanical strength of the 2304 DSS spot welds are compared with those of some other automotive steel — Fig. 14. The peak loads of various automotive AHSSs including dual phase steel (DP600, DP780, and DP980) (Ref. 5) and martensitic steels (Ref. 63) are compared with mechanical properties of 2304 DSS welds. All data shown in Fig. 14 are for welds with FZ sizes larger than 5t0.5 made on 1.5-mm-thick sheets. Peak load data were normalized according to the following formula:

\[
\text{Normalized Peak Load} = \frac{P_{\text{max}}}{D \sigma_{\text{UTS}}}.
\]

where \(P_{\text{max}}\) is the peak load of the spot welds, \(D\) is the weld nugget size, \(t\) is the sheet thickness, and \(\sigma_{\text{UTS}}\) is the ultimate tensile strength of the base metal. According to Fig. 14A, DSS spot welds exhibited the highest normalized peak load. Figure 14B shows the effect of the base metal tensile strength on the normalized peak load of the spot welds. As can be seen, increasing the martensite content in the base metal decreases the normalized peak load of AHSSs due to HAZ softening caused by martensite tempering. However, the welding thermal cycle of RSW does not induce softening or significant hardening in the HAZ. Therefore, the joint efficiency of DSS spot welds is higher than that of high volume fraction martensite-containing AHSS.

Conclusions

The understanding of the physical/mechanical welding metallurgy of duplex stainless steels is crucial for application of these steels in the automotive industry. This paper investigated the effect of the resistance spot welding thermal cycle on the metallurgical phase transformations in the weldment and the mechanical behavior of 2304 duplex stainless steel. The following conclusions can be drawn from this work:

1) Unlike advanced high-strength steels (e.g., dual-phase, TRIP, and martensitic steels), duplex stainless steels experience neither significant overmatching in the fusion zone nor softening in the heat-affected zone. However, the well-balanced ferrite-austenite microstructure of the base metal was destroyed in the fusion and heat-affected zones, which can affect the corrosion behavior of the joint.

2) An unbalanced microstructure with high ferrite proportion was developed in the fusion of 2304 duplex stainless steel resistance spot welds.
This phenomenon was a function of the inherently rapid cooling rate of the resistance spot welding process that suppresses the postsolidification transformation of ferrite to austenite. As a consequence of the unbalanced FZ, extensive precipitation of chromium-rich nitrides was observed within ferrite grains.

3) The volume fraction of austenite (fv) in the fusion zone was varied from 4 to 18 vol.-%, depending on the welding current. The maximum fv in the FZ corresponded to a critical welding current, when the decrease in the cooling rate attributed to increased Joule heating at higher welding current was less than the increase in cooling rate owing to the decrease in sheet thickness associated with excessive expulsion.

4) The minimum FZ hardness associated with maximum austenite volume fraction and minimum volume fraction of nitrides was obtained at a critical welding current, which corresponded to the lowest cooling rate.

5) The failure of 2304 resistance spot welds was a competition between shear plastic deformation of the fusion zone (i.e., interfacial failure mode) and the through-thickness necking of the base metal (i.e., pullout failure mode) with the latter preferred at higher welding current.

6) The FZ size was the key factor controlling peak load of the 2304 DSS spot welds. It was found that there is an optimum welding current to obtain the best energy absorption. Increasing welding current beyond the critical value not only does not improve mechanical properties, it reduces the austenite volume fraction of the FZ, which can degrade the corrosion resistance of the DSS weld. It is of note that the δ-ferrite volume fraction of the FZ does not play important role in mechanical performance of the DSS welds. Due to absence of HAZ softening in DSS welds compared to the high volume fraction martensite containing AHSS welds, the joint efficiency of the former is higher than that of the latter.

References


