Introduction

The use of advanced high-strength steels (AHSSs) has increasingly continued to grow in automotive applications due to their role in the manufacturing of weight-reduced vehicles with enhanced crash safety (Refs. 1, 2). The performance of the AHSSs depends on their sophisticated designed microstructures. However, the steel microstructure can be altered due to the welding thermal cycle, which can affect damage-tolerance properties of the welds. Therefore, the failure of welds during a crash is a critical issue for crash-worthiness, stiffness, and noise, vibration, and harshness (NVH) performance of the vehicle (Refs. 3–5).

Resistance spot welding (RSW), as the dominant joining process in the automotive industry, plays a critical role in vehicle manufacturing. The use of AHSS in the automotive industry comes with a challenge in their weldability. Through past research studies (Refs. 4–22), it has been identified RSW of AHSS has several challenging issues, including the following:

1) Material properties mismatch induced by complex phase transformation in the weldment. The two crucial phase transformations in AHSS welds are as follows:
   • Martensite formation in both the fusion zone (FZ) and coarse-grain heat-affected zone (HAZ) (Refs. 4–12), which can induce an adverse effect on the weld failure characteristics during some loading conditions (e.g. peel and cross-tension tests) (Refs. 4, 5). Formation of martensite in the FZ is due to the high cooling rate of the RSW process. It is reported that the cooling rate during RSW changes from roughly 3000°C/s for 2.0 mm thickness to more than 105°C/s for thicknesses less than 0.5 mm (Ref. 8). These cooling rates are sufficient for producing martensite in the FZ and the coarse-grain HAZ of the AHSSs, and even in low-carbon steel weld (Refs. 5, 8, 9).
   • Softening due to martensite tempering in the subcritical HAZ during welding of martensite-containing AHSS (e.g., dual-phase steels and martensitic [MS] steels) (Refs. 5, 13–17). This phenomenon can reduce the load-bearing capacity of spot welds compared to the strength expected from the initial base metal microstructure (Refs. 5, 14).

2) High susceptibility to interfacial failure mode. It is identified that sizing of spot welds made on AHSS based on 4t0.5 and 5t0.5 rules does not ensure the pullout failure mode in AHSS welds (Refs. 5, 8, 10, 13). Therefore, in some situations, postweld treatment (e.g., in-situ tempering) can improve the weld’s mechanical properties.

These challenges can lead to the possibility of using stainless steels as a new material for structural applications in automotive industries (Refs. 23–26). A study on the possibility of using stainless steels for automotive body structures was the subject of the next generation vehicle (NGV) program (Refs. 4–12).
23, 24). Duplex stainless steels (DSSs) exhibited great potential due to decreasing weight along with increasing safety, sustainability, and corrosion resistance of the structural automotive systems. The implementation of DSSs for the design of several critical components (e.g., A and B pillar) demonstrated the possibility of further weight reduction while ameliorating safety (Refs. 23, 24).

In automotive bodies, there are two primary crashworthiness zones with specific functions, as follows (Refs. 2, 5):

- Crumple zone. Components in this area must be capable of absorbing energy during a crash event.
- Safety cage. Components in this zone must resist deformation to prevent the vehicle structure from impacting on passengers.

The material requirements in these zones are different. The crumple zone requires materials with high energy absorption, and the safety cage needs materials with high yield strength. Therefore, a multimaterial design has been suggested for automotive bodies that composites several lightweight materials with specific properties. In a multimaterial construction, materials with proper properties are selected for the intended part’s functions (Refs. 27–29). Therefore, welding of dissimilar metal joints in the automotive bodies is inevitable.

By taking into account the current trend toward using the martensitic AHSS (as a good candidate for use in the safety cage zone) and DSS (as a good candidate for use in the crumple zone), the study of similar and dissimilar joints of these steels is a critical issue. However, there is limited research on the joining of dissimilar steels involving DSSs by RSW. In previous works, the metallurgical and mechanical behavior of similar joints involving DSSs has been studied. Arabi et al. (Ref. 30), in their study on the metallurgical response of 2304 DSS joints during RSW, found that an unbalanced microstructure in the FZ and HAZ of joints is generated due to the high cooling rate of the RSW process. Pouranvari et al. (Refs. 25, 26) found the presence of Ti in the composition of DSSs can contribute to decreasing austenite content in the FZ. Moreover, Thulin et al. (Ref. 31), Kotecki (Ref. 32), and Arabi et al. (Ref. 33) showed applying a second pulse after the first melting/solidification pulse can improve the phase balance of the FZ. However, there is no research study on the RSW of AHSS/DSS dissimilar combination. Failure behavior of dissimilar resistance spot welds can be a problematic issue due to differences in the physical, mechanical, and metallurgical properties of the weldments (Refs. 34, 35). Moreover, phase transformations in the FZ can be affected by the mixing of base metals (Refs. 36–40). This paper aims at investigating the phase transformations and mechanical behavior of similar and dissimilar joints of MS1200 martensitic AHSS and 2304 duplex stainless steel in RSW.

**Experimental Procedure**

This paper concerns the RSW of uncoated MS1200 martensitic AHSS and 2304 DSS sheets. The thickness of both sheets was 1.5 mm. The chemical composition and mechanical properties of the base metals in this work are given in Tables 1 and 2, respectively. Resistance spot welding was performed using a programmable logic-controlled, 120-kVA, AC-pedestal-type RSW machine. Welding was conducted using a 45-deg truncated cone, RWMA Class 2 electrode with an 8-mm face diameter. Figure 1A shows the welding schedule. Squeeze, welding, and electrode holding times after current off and electrode force were kept constant at 0.6, 0.24, 0.6, and 1.5 kN, respectively.
and 0.2 s as well as 4.5 KN, respectively. The welding current was incrementally increased from 9 to 13 kA with a step size of 1 kA.

Metallurgical examinations of the joints were performed under scanning electron and optical microscopes. Marble etchant (10-g CuSO₄, 50-mL HCl, and 50-mL H₂O) was used for microstructural examination of the FZ and DSS side. Moreover, a nital etchant (100-mL C₂H₅OH, 2-mL HNO₃) for microstructural metallography of the MS1200 side was used. Vickers microhardness test was performed using an indenter load of 100 g for a period of 15 s.

The quasi-static tensile shear test samples were prepared according to the ANSI/AWS/SAE/D8.9M Standard (Ref. 41). Figure 1B shows the sample dimensions for the tensile-shear test. Mechanical tests were conducted at a crosshead of 10 mm/min with an intrusion universal testing machine. The failure mode of the spot welds was determined by observing the fractured samples.

Results and Discussion

Metallurgical/Hardness Characteristic

The metallurgical properties and hardness characteristics of the resistance spot welds play an essential role in controlling the mechanical performance of the spot welds (Refs. 4–6). In this section, metallurgical characteristics of the DSS/DSS, MS/DSS, and DSS/MS joint are studied. Figure 2A, B shows a typical macrostructure and hardness profile of the dissimilar DSS/MS RSW joint, indicating microstructural gradient across the weldment. The failure behavior of the spot welds is influenced by materials properties mismatch among the base metal, HAZ, and FZ. In the following sections, the microstructure of each zone is described and analyzed.

The Base Metals

The DS2304 duplex steel microstructure, as shown in Fig. 3A, exhibited a balanced duplex microstructure of precipitate-free ferrite and austenite. The corresponding hardness was about 225 HV — Fig. 2B. The MS1200 steel microstructure, as depicted in Fig. 3B, consisted of very fine packets of martensite lath. The corresponding hardness is 400 HV — Fig. 2B.

The Fusion Zone

Figure 4A, B shows the microstructure of the DSS/MS fusion zone. The FZ is predominantly martensite with a small amount of ferrite. Regarding the microstructure evolution of the FZ in the DSS2304/MS1200 joint, the following points should be considered:

• The chemical composition of FZ. To investigate the effect of materials mixing on composition of the FZ, the composition of the weld FZ was measured using FESEM-EDS. Figure 4C shows the Ni/Cr concentration profile across the weldment along the through-thickness direction. According to Fig. 4C, the content of nickel and chromium in the FZ was almost uniform, indicating there was a complete mixing of liquid metals, which can be attributed to the intense electromagnetic stirring force in the molten weld nugget (Ref. 42). Due to the limitation in the determination of carbon content with high accuracy using FESEM-EDS analysis, the chemical composition of the FZ was calculated using the dilution concept. According to the macrograph of the cross section of the solidified weld nugget, the melting ratio of DSS2304/MS1200 welds varied from 57:43 in welds made using the welding current of 9 kA to 50:50 in welds made using the welding current of 13 kA. Therefore, considering a melting ratio of 55:45, the chemical composition of the FZ can be determined as Fe-0.06C-12.88Cr-2.6Ni-1.59Mn-0.28Si-0.24Mo-0.08N-0.03Al.

• The transformation path. To determine the transformation path of the FZ, the phase diagram of the FZ was developed by ThermoCalc™ (Ref. 43). Figure 5 shows the pseudo-binary phase diagram of the FZ vs. carbon content. According to Fig. 5, the solidification and solid-state transformation path for the FZ, before the austenite decomposition, was as follows:
According to the equilibrium phase diagram, when the temperature falls below 1297˚C, ferrite is entirely transformed to austenite via $\delta + \gamma \rightarrow \gamma$. This reaction is a solid-state, diffusion-controlled reaction. Therefore, due to the rapid cooling rate of RSW, some amount of untransformed ferrite is retained in the microstructure. The subsequent transformation of austenite to martensite depends on the austenite stability, which can be determined using the martensite start temperature ($M_s$) using the following equation (Ref. 44):

$$M_s = 526 - 12.5Cr - 17.4Ni - 29.7Mn - 31.7Si - 354C - 20.8Mo - 1.34 (CrNi) + 22.41 (Cr + Mo)C$$ \hspace{1cm} (2)

where the chemical symbols indicate the weight percentage of the elements present. The $M_s$ for the FZ of DSS2304/MS1200 RSW was estimated to be 223˚C, which is well above the room temperature. This indicates that martensitic transformation occurs in the FZ. Furthermore, the martensite finish temperature, $M_f$, is approximately 100˚C below the $M_s$ value (Ref. 44). This confirms that the austenite is not stable at room temperature, and it completely transforms to martensite during cooling. Thus, the transformation path of the FZ in dissimilar DSS2304/MS1200 welds under the rapid cooling rate of RSW can be summarized as follows:

$$L \rightarrow L + \delta + \gamma \rightarrow \delta + \gamma \rightarrow \gamma$$ \hspace{1cm} (1)

$$L \rightarrow L + \delta + \gamma \rightarrow \delta + \gamma \rightarrow \gamma$$

According to the equilibrium phase diagram, when the temperature falls below 1297˚C, ferrite is entirely transformed to austenite via $\delta + \gamma \rightarrow \gamma$. This reaction is a solid-state, diffusion-controlled reaction. Therefore, due to the rapid cooling rate of RSW, some amount of untransformed ferrite is retained in the microstructure. The subsequent transformation of austenite to martensite depends on the austenite stability, which can be determined using the martensite start temperature ($M_s$) using the following equation (Ref. 44):

$$M_s = 526 - 12.5Cr - 17.4Ni - 29.7Mn - 31.7Si - 354C - 20.8Mo - 1.34 (CrNi) + 22.41 (Cr + Mo)C$$ \hspace{1cm} (2)

where the chemical symbols indicate the weight percentage of the elements present. The $M_s$ for the FZ of DSS2304/MS1200 RSW was estimated to be 223˚C, which is well above the room temperature. This indicates that martensitic transformation occurs in the FZ. Furthermore, the martensite finish temperature, $M_f$, is approximately 100˚C below the $M_s$ value (Ref. 44). This confirms that the austenite is not stable at room temperature, and it completely transforms to martensite during cooling. Thus, the transformation path of the FZ in dissimilar DSS2304/MS1200 welds under the rapid cooling rate of RSW can be summarized as follows:

$$L \rightarrow L + \delta + \gamma \rightarrow \delta + \gamma \rightarrow \gamma$$

$$L \rightarrow L + \delta + \gamma \rightarrow \delta + \gamma \rightarrow \gamma$$

where the chemical symbols indicate the weight percentage of the elements present. The $M_s$ for the FZ of DSS2304/MS1200 RSW was estimated to be 223˚C, which is well above the room temperature. This indicates that martensitic transformation occurs in the FZ. Furthermore, the martensite finish temperature, $M_f$, is approximately 100˚C below the $M_s$ value (Ref. 44). This confirms that the austenite is not stable at room temperature, and it completely transforms to martensite during cooling. Thus, the transformation path of the FZ in dissimilar DSS2304/MS1200 welds under the rapid cooling rate of RSW can be summarized as follows:

$$L \rightarrow L + \delta + \gamma \rightarrow \delta + \gamma \rightarrow \gamma$$

$$L \rightarrow L + \delta + \gamma \rightarrow \delta + \gamma \rightarrow \gamma$$

where the chemical symbols indicate the weight percentage of the elements present. The $M_s$ for the FZ of DSS2304/MS1200 RSW was estimated to be 223˚C, which is well above the room temperature. This indicates that martensitic transformation occurs in the FZ. Furthermore, the martensite finish temperature, $M_f$, is approximately 100˚C below the $M_s$ value (Ref. 44). This confirms that the austenite is not stable at room temperature, and it completely transforms to martensite during cooling. Thus, the transformation path of the FZ in dissimilar DSS2304/MS1200 welds under the rapid cooling rate of RSW can be summarized as follows:

$$L \rightarrow L + \delta + \gamma \rightarrow \delta + \gamma \rightarrow \gamma$$

$$L \rightarrow L + \delta + \gamma \rightarrow \delta + \gamma \rightarrow \gamma$$

To further understand the effect of pairing and mixing of DSS and MS steels on the weldability of the dissimilar joint, the metallurgical/hardness characteristics of the similar DSS/DSS and MS/MS joints are investigated. Figure 6 compares the hardness values of the FZ in similar and dissimilar joints. Figures 7 and 8 show the microstructure of the FZ in similar and dissimilar joints, respectively. The following points can be drawn from Figs. 6–8:

- **DSS/DSS weld.** The FZ exhibited an unbalanced microstructure with a reduced volume fraction of austenite and consequent second-phase precipitation (i.e., chromium-rich nitrides), as confirmed in previous works (Refs. 30, 46). The phase transformation path of the FZ is as follows:

$$L \rightarrow L + \delta + \gamma \rightarrow \delta + \gamma \rightarrow \gamma$$

$$L \rightarrow L + \delta + \gamma \rightarrow \delta + \gamma \rightarrow \gamma$$

where the chemical symbols indicate the weight percentage of the elements present. The $M_s$ for the FZ of DSS2304/MS1200 RSW was estimated to be 223˚C, which is well above the room temperature. This indicates that martensitic transformation occurs in the FZ. Furthermore, the martensite finish temperature, $M_f$, is approximately 100˚C below the $M_s$ value (Ref. 44). This confirms that the austenite is not stable at room temperature, and it completely transforms to martensite during cooling. Thus, the transformation path of the FZ in dissimilar DSS2304/MS1200 welds under the rapid cooling rate of RSW can be summarized as follows:

$$L \rightarrow L + \delta + \gamma \rightarrow \delta + \gamma \rightarrow \gamma$$

$$L \rightarrow L + \delta + \gamma \rightarrow \delta + \gamma \rightarrow \gamma$$

where the chemical symbols indicate the weight percentage of the elements present. The $M_s$ for the FZ of DSS2304/MS1200 RSW was estimated to be 223˚C, which is well above the room temperature. This indicates that martensitic transformation occurs in the FZ. Furthermore, the martensite finish temperature, $M_f$, is approximately 100˚C below the $M_s$ value (Ref. 44). This confirms that the austenite is not stable at room temperature, and it completely transforms to martensite during cooling. Thus, the transformation path of the FZ in dissimilar DSS2304/MS1200 welds under the rapid cooling rate of RSW can be summarized as follows:

$$L \rightarrow L + \delta + \gamma \rightarrow \delta + \gamma \rightarrow \gamma$$

$$L \rightarrow L + \delta + \gamma \rightarrow \delta + \gamma \rightarrow \gamma$$

where the chemical symbols indicate the weight percentage of the elements present. The $M_s$ for the FZ of DSS2304/MS1200 RSW was estimated to be 223˚C, which is well above the room temperature. This indicates that martensitic transformation occurs in the FZ. Furthermore, the martensite finish temperature, $M_f$, is approximately 100˚C below the $M_s$ value (Ref. 44). This confirms that the austenite is not stable at room temperature, and it completely transforms to martensite during cooling. Thus, the transformation path of the FZ in dissimilar DSS2304/MS1200 welds under the rapid cooling rate of RSW can be summarized as follows:

$$L \rightarrow L + \delta + \gamma \rightarrow \delta + \gamma \rightarrow \gamma$$

$$L \rightarrow L + \delta + \gamma \rightarrow \delta + \gamma \rightarrow \gamma$$

where the chemical symbols indicate the weight percentage of the elements present. The $M_s$ for the FZ of DSS2304/MS1200 RSW was estimated to be 223˚C, which is well above the room temperature. This indicates that martensitic transformation occurs in the FZ. Furthermore, the martensite finish temperature, $M_f$, is approximately 100˚C below the $M_s$ value (Ref. 44). This confirms that the austenite is not stable at room temperature, and it completely transforms to martensite during cooling. Thus, the transformation path of the FZ in dissimilar DSS2304/MS1200 welds under the rapid cooling rate of RSW can be summarized as follows:

$$L \rightarrow L + \delta + \gamma \rightarrow \delta + \gamma \rightarrow \gamma$$

$$L \rightarrow L + \delta + \gamma \rightarrow \delta + \gamma \rightarrow \gamma$$

where the chemical symbols indicate the weight percentage of the elements present. The $M_s$ for the FZ of DSS2304/MS1200 RSW was estimated to be 223˚C, which is well above the room temperature. This indicates that martensitic transformation occurs in the FZ. Furthermore, the martensite finish temperature, $M_f$, is approximately 100˚C below the $M_s$ value (Ref. 44). This confirms that the austenite is not stable at room temperature, and it completely transforms to martensite during cooling. Thus, the transformation path of the FZ in dissimilar DSS2304/MS1200 welds under the rapid cooling rate of RSW can be summarized as follows:
The high cooling rate associated with RSW suppressed the postsolidification ferrite-austenite transformation (stage III) leading to ferritization of the FZ. Therefore, the higher hardness of the FZ compared to the base metal can be attributed mainly to the formation of second phase precipitations.

- **MS/MS welds.** The FZ exhibited almost martensitic microstructure. This is due to the high cooling rate as well as high hardenability of the MS1200 steel (Refs. 5, 14). The average hardness of the FZ is slightly higher than the base metal, which can be related to the light tempering of the steel during the manufacturing process.

- **DSS/MS welds.** The high hardness of the DSS/MS FZ is attributed to the presence of a high volume fraction of martensite in the weld nugget. Despite the presence of alloyed martensite in the DSS/MS FZ, compared to the unalloyed martensite in MS/MS FZ, the presence of ferrite in the former makes it slightly less hard than the latter.

As the FZ size is the key in determining the failure mode and mechanical properties of the welds (Refs. 5, 47), the effect of welding current, the main factor affecting the welding heat input, on the FZ size for the similar and dissimilar combination was examined — Fig. 9. The plots follow general trends. The differences in the FZ size are attributed to the difference in physical properties (electrical resistivity and thermal conductivity) of the steels (Ref. 5). The drop in FZ size in high welding currents is due to expulsion.

The Heat-Affected Zones

The hardness variations in the HAZ prove that the metallurgical responses of dissimilar DSS2304/MS1200 joint to weld thermal cycle are analogous with those of similar joints. The phase transformations in the HAZ of the MS1200 and 2304 DSS are well described elsewhere (Refs. 14, 30).

<table>
<thead>
<tr>
<th>Zone</th>
<th>Microstructure</th>
<th>FZ Transformation Path</th>
</tr>
</thead>
<tbody>
<tr>
<td>DSS</td>
<td>δf + γ + Cr,N</td>
<td>(I) L → L + δ + γ → δ + γ + precipitates</td>
</tr>
<tr>
<td>UCHAZ</td>
<td>M</td>
<td>(II) δ + γ → δ + γ + precipitates</td>
</tr>
<tr>
<td>SCHAZ</td>
<td>TM</td>
<td>(III) δ + γ + precipitates</td>
</tr>
<tr>
<td>MS</td>
<td>M + αf</td>
<td>(IV) δ + γ + precipitates</td>
</tr>
<tr>
<td>ICHAZ</td>
<td>M</td>
<td>(V) δ + γ + precipitates</td>
</tr>
</tbody>
</table>

formation path in the FZ and HAZ of similar and dissimilar joints. The softening observed in the HAZ of the MS side is due to ferrite formation in the intercritical HAZ and martensite tempering in the subcritical heat-affected zone (SCHAZ). The minimum hardness of the HAZ in the MS side was a function of the welding current. Increasing welding current from 8 to 13 kA decreases the SCHAZ hardness from 266 to 205 HV due to more severe tempering of martensite at higher welding heat input.

Failure Mode of DSS/MS Dissimilar Welds

During the quasi-static tensile-shear test, dissimilar DSS/MS RSW joints failed in two distinct modes, described as follows:

- Interfacial failure (IF) mode. Figure 10A shows a typical fracture surface of DSS2304/MS1200 spot weld, which failed in the IF mode. In this mode, the fracture propagates through the weld nugget centerline, and the sample undergoes little plastic deformation during the failure process.

- Partial thickness-partial pullout (PT-PP) failure mode. Figure 10B exhibits a typical PT-PP failure mode in which slant crack propagates into the FZ, and part of the mating sheet thickness is removed during separation.

The failure mode transition from the IF to PT-PP for both DSS/DSS and DSS/MS combinations occurred at a welding current of 10 kA. However, IF-to-PF transition for the MS/MS combination occurred at a welding current of 13 kA. Note that the corresponding critical fusion zone sizes leading to the non-IF mode in these three combinations are different.

Mechanical Behavior of DSS/MS Dissimilar Spot Welds

To investigate the mechanical properties of DSS/MS dissimilar spot welds, peak load and energy absorption ob-
tained in the tensile-shear test were measured at various welding currents. According to Fig. 12A, B, increasing the welding current from 8 to 10 kA significantly improved the mechanical properties of the joint in terms of peak load and energy absorption of the welds. This improvement is a function of enlarging the weld nugget size and the dependency of peak load on the square of the FZ size in the interfacial mode. However, when the welding current increased beyond the 10 kA, no significant improvement was observed in the mechanical properties of the DSS/MS welds. This is a function of their failure mode and the failure location hardness. The DSS/MS welds made using welding current higher than 9 kA failed in the PF mode. Therefore, their strength is a function of weld nugget size and PF location hardness (i.e., SCHAZ of the MS side). Thus, despite the slight increase in the FZ size by increasing the welding current from 10 to 13 kA, the peak load of the welds remained essentially constant due to increasing the HAZ softening in the MS side.

Comparison of Mechanical Performance of Similar and Dissimilar Welds

Failure Mode Transition

The failure mode of the similar and dissimilar combinations was determined after the tensile-shear test. Figure 9 shows the effect of welding current on the FZ size and the failure mode of the similar and dissimilar welds. Maximum weld size leading to the IF mode and minimum weld size leading to the non-IF mode for each combination can be determined using Fig. 9. The critical FZ size (Dc) is located between these two values. The susceptibility of the joints to interfacial failure decreases in the order of MS/MS, DSS/DSS, and DSS/MS combinations. It is well known that hardness ratio of the FZ to the minimum hardness of the weld circumference (base metal or HAZ) is one of the key controlling factors in determining the susceptibility of the spot welds to fail in the IF mode (Refs. 5, 6, 10). Increasing the hardness ratio reduces the tendency to fail in the IF mode. The hardness ratio for DSS/DSS welds is the ratio of FZ hardness to base metal hardness. The hardness ratio for MS/MS welds is the ratio of FZ hardness to SCHAZ hardness. The average hardness ratio in the MS/MS and DSS/DSS joints are 1.9 and 1.4, respectively. However, Dc for the DSS/DSS joint is smaller than that of the MS/MS joint. This can be explained by considering the high tendency of DSS to rotate during tensile-shear loading (see Fig. 11B), which reduces the effective shear stress acting on the sheet/sheet interface. The relationship between the degree of rotation and the failure mode of the spot welds during the tensile-shear loading is previously reported (Refs. 35, 48). The high strength of the martensitic steels impedes the easy rotation of the welded joint during the tensile-shear test and increases its susceptibility to fail in interfacial failure. The DSS/MS joint exhibited the lowest Dc. Although the hardness ratio of the DSS/MS joint is similar to the MS/MS joint, its higher rotation during testing prompted the non-IF failure mode.

Peak Load and Energy Absorption

To provide a better basis for comparison between mechanical properties of similar and dissimilar joints, peak load and failure energy of spot welds were compared at two different FZ sizes, 5t0.5 (i.e., 6.1 mm). Welding currents used to produce spot welds with the target weld size of 5t0.5 and resultant weld attributes are given in Table 4. To account for the small differences in FZ size, the values of peak load and failure energy were normalized by dividing to the FZ size (D). Figure 13 shows the normalized peak load and energy absorption of three combinations. The following points can be drawn from Fig. 13:

- MS/MS joint exhibited the highest peak load. Despite its interfacial failure mode, the high peak load of the MS/MS joint can be attributed to the high hardness of the failure location (i.e., FZ). The peak load of the DSS/DSS joint, which failed in the PT-PP mode, was lower than that of the MS/MS joint due to its lower failure location hardness (i.e., base metal). The
The pairing of MS and DSS resulted in welds with comparable peak load to the DSS/DSS joint. The slightly lower peak load of DSS/MS welds compared to the DSS/DSS can be attributed to the HAZ softening phenomenon in the MS side.

- The similar DSS/DSS joint exhibited the highest energy absorption. This is because of its PF mode as well as the higher energy absorption capability of the DSS base metal. The lower energy absorption of the MS/MS joint can be related to its IF mode and the low ductility of the failure path. Pairing of MS and DSS steels resulted in a joint with a lower energy absorption than the DSS/DSS joint.

Conclusions

In this paper, metallurgical and mechanical properties of dissimilar resistance spot welds between 2304 duplex stainless steel and MS1200 martensitic advanced high-strength steel are studied and compared to those of similar joints of the mentioned steels. The following conclusions can be drawn from this research:

1) Investigation of equilibrium phase transformation path of the FZ predicts a full martensitic nugget for the dissimilar weld. However, a dual-phase microstructure of martensite and ferrite was observed in the FZ. The formation of the ferrite phase in the FZ can be attributed to the rapid cooling rate of RSW, which suppresses the solid-state transformation. The FZ hardness values of the similar and dissimilar welds increased in the order of DSS/DSS, MS/MS, and DSS/MS.

2) The critical FZ size for avoidance of the interfacial failure increased in the order of DSS/MS, DSS/DSS, and MS/MS. The lower susceptibility of the dissimilar joint compared to the similar welds is due to a combination of three factors, including high hardness of the FZ, presence of the heat-affected softening in the MS side, and higher tendency of the joint to rotate during the tensile-shear loading compared to the MS/MS joint.

3) The failure location of dissimilar welds in PT-PP failure was a function of minimum hardness value in the hardness profile, ultimate tensile strength, and differences in work hardening behavior of base metals. It was found the failure location of DSS/MS is controlled by the work hardening behavior of the base metal, and despite its higher ultimate tensile strength, the failure is initiated from the MS side.

4) The FZ size at sheet/sheet interface is the key macrostructural feature controlling the load-bearing capacity and energy-absorption capability of the DSS/MS dissimilar weld as well as the fraction of the pullout failure during PT-PP mode.

5) The pairing of MS and DSS resulted in welds with comparable peak load to DSS/DSS joint and lower peak load compared to MS/MS joint. The slightly lower peak load of DSS/MS welds compared to the DSS/DSS can be attributed to the HAZ softening phenomenon in the MS side. The dissimilar joint exhibited increased the energy absorption capability compared to similar MS/MS welds.

<table>
<thead>
<tr>
<th>Joint Type</th>
<th>FZ Size</th>
<th>Welding Current</th>
<th>Failure Mode</th>
<th>Failure Location</th>
<th>$H_{FZ}$ (HV)</th>
<th>$H_{BM}$ (HV)</th>
</tr>
</thead>
<tbody>
<tr>
<td>DSS/DSS</td>
<td>6.1</td>
<td>10</td>
<td>PF</td>
<td>BM</td>
<td>300</td>
<td>220</td>
</tr>
<tr>
<td>MS/MS</td>
<td>6</td>
<td>11</td>
<td>IF</td>
<td>FZ</td>
<td>410</td>
<td>410</td>
</tr>
<tr>
<td>DSS/MS</td>
<td>6.2</td>
<td>11</td>
<td>PP-PT</td>
<td>SC HAZ of MS1200</td>
<td>398</td>
<td>215</td>
</tr>
</tbody>
</table>

(a) Fusion zone hardness
(b) Failure location hardness
Fig. 13 — Comparison of mechanical properties of similar and dissimilar combinations at a weld nugget size of 5t0.5: A — Peak load; B — failure energy.

References


SAEID SOBHANI and MAJID POURANVARI (pouranvari@sharif.edu) are with the Department of Materials Science and Engineering, Sharif University of Technology, Tehran, Iran.