Effect of Coating and Welding Wire Composition on AHSS GMA Welds

The roles of galvanized iron coating and filler material on weld strength were investigated for different weld configurations using the DIC technique

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ABSTRACT

Advanced high-strength steels (AHSS) such as complex-phase (CP) and high-formability (HF) steel offer weight-saving advantages for automotive applications such as chassis and frame applications. To prevent material oxidation, materials are often galvanized to protect the substrate from corrosion. However, the weldability of coated AHSS becomes challenging due to the trapping of zinc in the weld molten pool, which could lead to weld defects such as porosity and liquid metal embrittlement cracks. This work focused on the weldability of AHSS (CP800 and 980HF) using the gas metal arc welding process. The roles of both galvanized iron coating and filler material on weld strength were investigated. The welds were performed using two different filler materials: a low-strength filler (ER70S-6) and a high-strength filler (ER100S-6) material. In addition, two different joint configurations were studied: lap joints and butt joints. The results showed that the butt joint had a higher strength compared to the lap joints. Furthermore, the strength of the butt joint overmatched the base material strength in all of the tested materials (both in galvanized and uncoated). In general, lap joint strength undermatched the base material strength, which was attributed to the rotation during tensile testing that induced unaccounted bending stress on the lap joint, while using a higher-strength welding wire improved the tensile strength material in the lap joint configuration. The hardness profiles in the 980HF steel also showed a significant hardness mismatch due to the formation of a fully martensitic microstructure in the heat-affected zone, which led to suppressing the deformation across the lap joint.

KEYWORDS

- Advanced High-Strength Steel (AHSS)
- Gas Metal Arc Welding (GMAW)
- Digital Image Correlation • Tensile Strength
- Lap Shear Test • Butt Joint • Filler Material Strength

Introduction

The need to decrease the curb weight of automobiles to

reduce CO₂ emission and meet greenhouse gas emission regulations encouraged the automotive industry to use advanced high-strength steel (AHSS) for frame applications (Refs. 1, 2). This allows thinner gauges of steel to be used to manufacture automotive parts, reducing part weight and improving vehicle fuel economy while maintaining or increasing vehicle safety (Refs. 2–4). To join steels in an automotive structure, welding processes such as laser beam welding (LBW) and resistance spot welding (RSW) are the most frequently used (Ref. 5). However, due to the thickness of the sheets used for frame applications, gas metal arc welding (GMAW) is often used (Ref. 6). When AHSS is welded, the resultant weldments can exhibit heterogeneous properties where the hardness will vary greatly across the weld zone. This is due to the complex chemical composition and microstructures of the AHSS resulting from the local heating cycles associated with welding (Ref. 7). After welding, the supercritical heat-affected zone (SCHAZ) adjacent to the fusion zone (FZ) can form martensite, resulting in very high local hardness. Conversely, away from the FZ, where the peak temperature reached during welding does not exceed the Ac₁ temperature of the steel, any martensite in the base material will temper, resulting in a local hardness drop below that of the base material, known as heat-affected zone (HAZ) softening (Refs. 7, 8). Variation in local properties can control the failure location of the weld joint during tensile testing, where the exact failure location depends on the relative strength of the various areas in the weld zone (Refs. 9, 10). The galvanized coating on AHSS is important to protect the material surface from corrosion. However, the zinc coating adds complexity to the welding process. This is due to several factors, including the difficulty of achieving a stable arc, increasing spatter, and increasing porosity in the weld metal (Refs. 11–13). All of these issues relate to the low vaporization temperature of Zn (907°C), which is much lower than the melting temperature of steel (1500°C) (Ref. 14). To avoid these issues in the zinc-coated steel, several solutions have been reported in the literature, such as controlling the welding current and torch position (Ref. 12), using a flux cored welding wire specifically developed for GMAW of zinc-coated steels (Ref. 13), or leaving a small gap (0.1–0.3 mm) between the lap-joined sheet to allow the zinc fume to

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Fig. 1 — Engineering stress-strain curves for both materials: A — 980HF; B — CP800.

escape during welding (Ref. 11). With GMAW, low-strength welds are tolerant of small amounts of internal porosity (Refs. 10, 13, 15). However, in higher-strength materials, it will be more important for welds to remain pore free to maintain weld strength because the overmatching between the FZ and base material will be much less than observed in lower-strength steels (Ref. 10).

To evaluate GMAW joint strength in automotive applications, tensile lap shear tests and butt joint tensile tests are usually performed to measure the strength under different loading and failure scenarios. Due to the geometry of the lap joint, there is eccentric loading during lap shear testing and the lap joint rotates, applying a bending moment, which is not accounted for during standard strength testing methods (Ref. 10). In the butt joint configuration, the loading condition is symmetric, and the joint does not rotate during testing. Ono et al. (Ref. 16) investigated the static and fatigue strength for various uncoated high-strength steel sheets. The study focused on the strength of laser beam welded lap joints, which were compared to those of spot welds in the same material. The study proposed a relation to estimate the strength and fracture position of the laser lap joint. The suggested formula considers the angle of the rotation of the lap joint, the sheet thickness, the weld width, and the hardness of the weld. However, in the study by Ono et al., the material was not coated and no filler material was used as LBW and RSW were studied.

Kapustka et al. (Ref. 17) investigated the effect of the cooling rate, material prestrain, and filler metal composition on the microstructure and mechanical properties of GMA welds in dual-phase (DP) and transformation-induced-plasticity (TRIP) steels. The welds were made in two different configurations: lap and butt joints. The study reported that filler metal strength did not affect the static and dynamic tensile properties of TRIP780 in both weld configurations, while the lap joint in DP780 had a higher strength than when welded with the high-strength welding wire. Kapustka et al. reported that increasing the cooling rate by reducing heat input (using a short-circuit transfer mode) promoted porosity and led to a significant loss in joint efficiency. Therefore, using higher heat input (low cooling rate) will help Zn bubbles escape from the molten pool before solidifi-

cation. Ahsan et al. (Ref. 1) investigated the mechanism of porosity formation, growth, and escape in GMA welds by controlling the heat input parameters. The porosity formation was explained using the viscosity of the weld pool and the vapor pressure of the zinc. However, the study results agreed with those of Kapustka et al. (Ref. 17), which showed pores unable to rise from the weld pool, leaving small pores near the weld root, although in small numbers these small pores will not affect the mechanical performance. A high-heat-input (more than 350 J/mm) Zn vapor will grow and escape the molten pool. This being said, a high heat input has side effects such as inducing high deformation due to the thermal residual stresses and may require the use of low welding speed, which will reduce productivity (Refs. 18, 19).

In this work, optimized standard industrial parameters with relatively low heat input were used to study the effect of filler metal composition and Zn coating on joint performance. A standard robotic GMAW technique was used to join two AHSS alloys: a complex phase (CP800) and high formability (980HF). The material was examined in two different conditions: bare surface and galvanized iron (GI) zinc coated. Two different joint configurations were studied: lap and butt joints. Microstructural analysis and hardness measurements were performed to understand the local variation in the mechanical properties of the failure location. Strain developed during tensile testing was mapped using digital image correlation (DIC) to evaluate the strain distribution in the local zones across the weld. The DIC results were correlated with the hardness results to understand the effect of filler metal strength and Zn coating on the joint performance. Implementing the results of this project was the first step to integrating heavy (> 2 mm) AHSS into frame applications. Increasing the use of AHSS in frame applications will allow thinner materials to be used for automotive designs, thus decreasing vehicle weight and reducing environmental impact.

Experimental Setup

In this work, two materials were investigated: HR-CP800 and 980HF. Both materials were supplied in uncoated conditions and GI coated on both sides. The coating thickness for

Fig. 2 — Tensile test coupons geometry: A — Lap joint; B — butt joint.

Fig. 3 — Optical micrographs show the base materials' microstructures.

980HF sheets was $8.5 \pm 0.7 \mu\text{m}$, while the CP800 sheet's coating thickness was $13 \pm 2.5 \mu\text{m}$. The chemical compositions, along with the carbon equivalent, of the base materials are listed in Table 1. The carbon equivalent was calculated using Yurioka's formula (Ref. 20). DIC tensile testing was performed for both materials' coated and uncoated sheets to determine the mechanical properties for the base material. The engineering stress-strain curves are shown in Fig. 1, and to compare the coated vs. uncoated properties, the mechanical properties are listed in Table 2.

Robotic GMAW was used to join the sheets. The specimen size used in this study was $300 \times 300 \text{ mm}$ ($12 \times 12 \text{ in.}$). Welding was performed parallel to the rolling direction of the base material. Welds were made in both the butt and lap joint configurations. The overlap distance used for the lap joint was 25 mm (1 in.). The butt joint was welded in the flat position, and a small gap was left between the sheets to allow the Zn fume to escape. The lap joint was welded using a 30-deg torch angle. In both weld configurations, the contact tip working distance was about 12 mm. To understand the role of FZ strength on weld mechanical properties, two different consumables were used, a low-strength welding wire (ER70S-6) and a high-strength welding wire (ER100S-6) (Ref. 21). The details for the welding parameters and consumables are listed in Table 3. It should be noted that all welds were made using standard industrial parameters that were chosen based on manufacturing robustness.

Sample Preparation

Samples were cut from the welded sheets (using a waterjet) to perform the mechanical testing (lap shear and tensile test) as well as to characterize the microstructure and perform the hardness test in different zones through the weld joint. Standard uniaxial tensile and lap shear tests were performed at room temperature using an initial displacement rate of 1 mm/min. Testing was carried out on an MTS frame with a maximum capacity of 100 kN. Tensile samples were prepared to the dimensions shown in Fig. 2 to match the testing suggested by the Auto/Steel Partnership (Ref. 22) and the work done by Min Kuo (Ref. 23) to be comparable to previous studies conducted on these materials. During testing, sample elongation was measured using a noncontact DIC measurement technique. This enabled the measurement of elongation across the gauge length. Furthermore, the DIC technique was also able to provide full-field strain measurement and mapping strain distribution in two- or three-dimensions. To use the DIC technique, samples were prepared by applying a random speckle pattern to their surface by painting the sample with white paint followed by speckling them with black spray paint. To reduce load eccentricity during lap shear tensile testing, the test coupons were shimmed.

To analyze the weld microstructure and hardness, cross sections of the welds were cut and hot mounted in conductive bakelite. The samples were prepared for metallographic examination according to ASTM E3-11, *Standard Guide for*

Table 1 — Chemical Composition and Carbon Equivalent for Base Materials

Material	CP800-UC	CP800-GI	980HF-UC	980HF-GI
C	0.09	0.08	0.20	0.23
Mn	1.19	1.19	2.14	1.97
Si	0.16	0.17	1.40	1.02
Other Elements	Cr, Ti, Ni, B, Al	Cr, Ti, Ni, B, Al	Al, Cr	Al, Cr, V, B
C _{eq}	0.345	0.337	0.622	0.644

Table 2 — The Mechanical Properties for the Base Materials

Material	CP800-UC	CP800-GI	980HF-UC	980HF-GI
Thickness-mm	2.4	2.5	1.2	1.2
YS-MPa	722	735	615	660
UTS-MPa	795	802	987	998
Elongation-%	14.0	15.0	20.6	22.0

Table 3 — Welding Parameters for Butt and Lap Joints

Butt Joint Welding Parameters							
Material	ID	Wire Type	Travel Speed (in./min)	Wire Feed Speed (in./min)	Trim	Voltage Volt	Welding Mode
CP800-UC (2.4 mm)	G1	70S	27	275	0.90	—	Pulse
	G2	100S	34	270	—	22	Precision pulse
CP800-GI (2.5 mm)	H1	70S	27	260	—	24	Precision pulse
	H2	100S	29.5	260	—	22	Precision pulse
980HF-UC (1.2 mm)	K1	70S	36	145	—	19	Precision pulse
	K2	100S	34.5	160	—	20	Precision pulse
980HF-GI (1.2 mm)	L1	70S	30	140	—	20	Precision pulse
	L2	100S	22	90	0.86	—	Power mode
Lap Joint Welding Parameters							
Material	ID	Wire Type	Travel Speed (in./min)	Wire Feed Speed (in./min)	Trim	Voltage Volt	Welding Mode
HR-CP800-UC (2.4 mm)	B1	70S	35	210	1.0	—	Pulse
	B2	100S	27	270	0.95	—	Pulse
HR-CP800-GI (2.5 mm)	C1	70S	20	240	1.0	—	Pulse
	C2	100S	30	280	1.0	—	Rapid-X
980HF-UC (1.2 mm)	D1	70S	35	150	1.0	—	Pulse
	D2	100S	30	200	1.0	—	Pulse
980HF-GI (1.2 mm)	E1	70S	28	190	—	19.5	Precision pulse
	E2	100S	25	220	—	17.0	Precision pulse

the Preparation of Metallographic Specimens (Ref. 24). The mounted samples were ground using silica grit paper starting with 120, 240, 320, 400, 600, and 1200 sizes. Automated diamond polishing was used to ensure a high-quality surface finish with different particle sizes starting at 9, 3, and then 1 μ m. After polishing, the samples were etched using a 5% Nital solution for 5 s and rinsed with water followed by alcohol. The microhardness was measured using an automated Vickers microhardness tester (Clemex Hardness Tester). The Vickers hardness tests were performed according to ASTM E384, *Standard Test Method for Microindentation Hardness of Materials* (Ref. 25). The hardness indentation load was 0.5 kgf, and the spacing between the indents was at least three times the indent diagonal to avoid interac-

tion of adjacent plastic deformation fields. The dwell time was set to be 10 s.

Results and Discussion

Base Material Microstructure and Hardness

The microstructure of the base metal (BM) is shown in Fig. 3. The complex-phase steel CP800 BM exhibited a very fine microstructure consisting of ferrite, bainite, and martensite. The hardness of the base material was 255 and 248 HV_{0.5} for the uncoated and GI-coated CP800 sheets, respectively. Figure 3 also shows the 980HF microstructure, which contains fer-

Fig. 4 — The cross section shows the macrostructure for butt joints.

rite, bainite, and some martensite. It should be noted that this structure is heavily banded due to the rolling process that is usually involved to produce AHSS sheets. The BM hardness was 305 and 310 HV_{0.5} for the uncoated and GI-coated 980HF sheets, respectively.

Macro- and Microstructure for Butt Joints

The macrographs in Fig. 4 show the cross section of the butt joints for both materials in the galvanized and uncoated conditions. It can be seen that some of the GI-coated welds exhibited porosity (the most severe pore has been shown here), although it should be noted that this was not universal. Cluster porosity has been reported in the literature. Cluster porosity in amounts up to 10% had no significant effect on joint strength, and that is discussed in the Mechanical Testing section (Ref. 26). The porosity did not follow trends with welding wire strength or material. Misalignment could also be observed in all 980HF welded samples. Finally, incomplete fusion was also observed in the CP800 GI-coated sheets.

Weld Metal/HAZ Microstructure

The weld metal microstructure is shown in Fig. 5 for all welding conditions. The use of the high-strength welding wire

(ER100S-6) refined the weld structure and increased the volume fraction of acicular ferrite. However, it should be noted that these changes may not be due to a change in filler material alone because different welding parameters were used for welds made using the different welding wires. Therefore, it is unclear which of these two factors had a stronger influence on the change in structure. The impact of this on properties is discussed in more detail in the next section. It was observed that the FZ microstructure of the uncoated materials differed from that of the GI-coated materials. For example, the 980HF uncoated samples had a fine microstructure, composed of a mixture of acicular ferrite, martensite, and bainite when welded with both welding wires. Conversely, the FZ microstructures of the GI-coated 980HF sheets welded with both welding wires were fully martensitic. It is believed that this was due to the differences in the base material compositions of each material (see Table 1). The composition and carbon equivalent will strongly influence the FZ and HAZ microstructures.

The microstructure of the coarse-grained HAZ (CGHAZ) was very heavily influenced by the base material microstructure and chemical composition of the base material and not affected by welding wire type (Fig. 6). The CGHAZ of the CP800 was a mixture of upper bainite and some Widmanstätten ferrite. The volume fraction and the morphology of the upper bainite and Widmanstätten ferrite varied based on the chemi-

Fig. 5 — Optical micrographs show the weld metal for each condition.

Fig. 6 — Optical micrographs show the CGHAZ microstructure for each condition.

cal composition of the base material (GI coated vs. uncoated). Due to the high carbon equivalent of the 980HF base material, the microstructure of the CGHAZ was fully martensitic, which led to an increase in the hardness and strength mismatch and controlled the failure location.

Hardness Maps across the Butt Joint

Hardness was measured for all welds using a hardness mapping system. As one would expect, the FZ of welds made using ER100S-6 were harder than those made using the lower-strength ER70S-6 welding wire (Fig. 7). Welding wire and base material composition had a large influence on FZ hardness. For example, the average hardness of the FZ of the UC-CP800 welded with ER100S-6 was 248 HV_{0.5}, whereas the hardness of the GI-coated CP800 welded with the same welding wire was 275 HV_{0.5}. It is also interesting to note that the HAZ hardness of all CP800 welds softened below that of the base material. For example, the minimum hardness in the HAZ of the UC-CP800 welded with the ER70S-6 welding wire was 230 HV_{0.5}, whereas it had a BM hardness of 255 HV_{0.5}. HAZ softening was not seen in all welds in the 980HF. The hardness maps from all the welds may be seen in Fig. 7, and the data are summarized in Table 4.

Table 4 — Average Hardness Values Measures in Zones across Welds for Butt Joints

Material	CP800				980HF			
	Uncoated		GI Coated		Uncoated		GI Coated	
Filler Metal	70S	100S	70S	100S	70S	100S	70S	100S
Base Metal	255		248		305		310	
SCHAZ	230	245	235	235	315	330	340	360
CGHAZ	260	270	250	245	460	500	500	480
Fusion Zone	225	248	230	275	310	350	360	430

Tensile Testing for Butt Joint Configuration

Tensile testing was performed for all butt joints. The results along with the DIC strain maps are shown in Fig. 8. The tensile strength for all joints met or exceeded the base material strength. The ultimate tensile strength (UTS) of the welds in uncoated CP800 was 900 MPa regardless of the welding wire used (see Fig. 8). This is not surprising as both welds had similar HAZ microstructures (Fig. 6, G1 and G2) and both failed in the HAZ. The UTS of the coated CP800 was influenced by filler material where the welds made with the ER70S-6 had a strength of 850 MPa, and the welds made

Fig. 7 — Hardness maps for the butt joints.

with the ER100S-6 welding wire had a strength of 970 MPa. The influence of the welding wire, in this case, may be explained by examining both the strain maps and the microstructures of the welds. When the GI-coated material welded with the low-strength welding wire was pulled, high strain occurred adjacent to the FZ, therefore the mechanical properties of the FZ influenced thinning (see Fig. 8). Also, some pores were observed on the fracture surface of the weld made using the low-strength welding wire, which explains the drop in the strength of the weld. However, when the high-strength welding wire was used, thinning in the area adjacent to the FZ was restrained, shifting thinning to the harder area further away from the FZ (Fig. 7), increasing the joint strength. The effect of the coating on the tensile strength can be seen in Fig. 8. The coated material exhibited a higher tensile strength compared to the uncoated material when the same filler material was used. It should be noted that there was no improvement in the tensile strength when the high-strength welding wire was used.

All 980HF butt joints exceeded the minimum tensile strength of the base material seen in Table 2. From the DIC strain maps and strain distribution, the maximum local strain was measured in the HAZ on both sides with a maximum value of 10% for coated and uncoated welds (Fig. 9),

although it is interesting to note that in all cases the hardness of this zone was higher than that of the FZ (Fig. 7). The higher strain in the harder HAZ may be due to the larger material thickness (due to reinforcement) of the FZ. Interestingly, the coated material displayed slightly higher strength values; this may be due to the higher carbon equivalent of this material (0.644), as shown in Table 1, which would have contributed to the higher hardness of this zone (see Fig. 7 and Table 4). It should be noted that in the case of the 980HF welding wire, strength did not significantly increase the strength of the joint. Unlike the case of the CP800, increasing FZ hardness did not shift failure in the butt joint away from lower-strength areas of the weld. Also, internal porosity was observed in the 980HF coated welds when low-strength welding wire was used in the weld. The porosity reduced the peak strength and total strain by 2% (see Fig. 9).

The summary of the measured butt joint strength is presented in the bar charts shown in Fig. 10. Using high-strength welding wire did not increase the global strength of the joint, as strain concentration in the softened HAZ determined the global stress-strain behavior. However, the porosity and lack of fusion in some of the coated sheet welds led to a decrease in the butt joint strength and ductility when

Fig. 8 — Stress-strain curves along with DIC strain maps for CP800 butt joint welds.

Fig. 9 — Stress-strain curves along with DIC strain maps for 980HF butt joint welds.

Fig. 10 — Bar charts compare the peak tensile strength values. A — All butt joints produced using different filler metals; B — the effect of porosity on joint performance.

Table 5 — The Average Hardness Data Measured for Each Material across the Welds for Lap Joints

Material	CP800				980HF			
	Uncoated		GI Coated		Uncoated		GI Coated	
Filler Metal	70S	100S	70S	100S	70S	100S	70S	100S
Base Metal	255		248		305		315	
SCHAZ	270	265	245	248	275	277	300	275
CGHAZ	305	296	260	270	470	485	550	485
Fusion Zone	255	258	225	237	285	295	295	300

low-strength welding wire was used. For example, the strength decreased by 8% for CP800-GI-70 and 2% for 980HF-GI-70. Note that the average strength reported in Fig. 10A included the samples that showed welding defects.

Macrostructure for Lap Joint Welds

All the lap joint welds were free of porosity, regardless of whether the base material was GI coated or uncoated (Fig. 11). It should be noted that some of the joints presented an inconsistent gap between the sheets, such as in samples C2, D2, and E2, which may affect the joint performance. However,

comparing the strength values to the recommended values in the standard procedure showed these values are acceptable. It was seen that although some samples had large gaps, the measured joint strength and joint efficiency were high. Therefore, although it is acknowledged that a gap will increase stress during testing, it seems that the presence of a gap will not necessarily degrade joint strength. In addition, high heat was used to weld the GI-coated materials to allow zinc vapor bubbles to escape and prevent porosity, as suggested in previous literature (Refs. 15, 17). This had the effect of increasing joint penetration, as shown in Fig. 11, which was more visible in 980HF-GI-coated welds.

Hardness Profiles for Welds Made in the Lap Joint Configuration

Hardness profiles were measured across each lap joint. These were constructed by averaging the results of two hardness profiles and are reported in Fig. 12. The CP800 weld showed slight variation in the hardness profiles, and the FZ hardness was comparable or slightly softer than the base material hardness (Fig. 12). The base material hardness for uncoated CP800 sheets was 255 HV_{0.5}, while the GI-coated sheets were slightly softer (248 HV_{0.5}) due to the slight

Fig. 11 — The cross section shows the macrostructure for lap joints.

variation between the carbon equivalent of the two materials (C_{eq} is 0.345 for the CP800 uncoated material vs. 0.337 for the CP800 GI-coated material). When high-strength welding wire was used (with its associated change in welding parameters), the hardness of the FZ was similar to that of the base material. It should be noted that the HAZ near the BM SCHAZ showed hardening behavior. This is due to the transformation of pre-existing carbides found within the bainite structure and usually occurs at temperatures around 500 °C (Ref. 27). It should also be noted that softening was not seen in the HAZ of the CP800 lap joint, unlike the observations from the butt joints. The 980HF lap joint hardness profiles are shown in Fig. 12; the BM hardness for 980HF is 305 and 315 HV_{0.5} for uncoated and GI-coated sheets, respectively. The FZ hardness of the 980HF joint welded using both consumables (low-strength and high-strength welding wires) was slightly softer than that of the 980HF base material hardness (see Table 5 for more details). The hardness values of 285 to 300 HV_{0.5} in the FZ were due to the change in the welding parameters (lower travel speed and higher welding wire feed speed — see Table 3). Using a high-strength welding wire slightly improved the weld metal hardness for both coated and uncoated welds, but the hardness values undermatched the BM hardness. The hardness of the CGHAZ reached up to 550 HV_{0.5} for the 980HF-GI-coated plate welded using the low-strength welding wire. This was due to the formation of a fully martensitic microstructure, as shown in Fig. 6.

The hardness mismatch between the weld metal/HAZ in

Fig. 12 — Hardness profiles for the lap joint configurations.

the 980 HF welds was very high (300 HV_{0.5}/500 HV_{0.5}), which promoted failure at the FZ/HAZ interface. Softening was also seen in the subcritical HAZ near the transition to the BM due to the martensite from the BM tempering. However, due to the hardness mismatch and a high-stress concentration that occurred near the weld interface, failure did not occur in the softened HAZ. Similar behavior was reported in previous literature (Ref. 10). The summary of the hardness results for each zone/material is shown in Table 5.

Tensile Testing for Lap Joint Configuration

The lap joints were prepared and tested using a tensile frame with a DIC system to map the strain for both materials, similar to the butt joint configuration. To observe the sample rotation and strain distribution on the side view using the DIC system, the tensile frame grips were rotated 90 deg. The stress-extension curves for the CP800 lap shear

Table 6 — The Average Peak Load and Strength with Standard Deviation for Both Weld Configurations Measured by Tensile Test

Material Configuration	CP800-UC		CP800-GI		980HF-UC		980HF-GI	
	ER70	ER100	ER70	ER100	ER70	ER100	ER70	ER100
Butt Joint Peak Load-kN	82 ± 0.4	83 ± 0.22	78 ± 2.9	82 ± 0.04	49 ± 0.35	49 ± 0.23	50.2 ± 0.05	51 ± 0.16
Butt Joint Strength-MPa	878 ± 5	884 ± 4	802 ± 24	952 ± 10	1059 ± 6	1041 ± 5	1067 ± 11	1087 ± 3
Lap Joint Peak Load-kN	85 ± 2.7	95.5 ± 0.2	81.2 ± 0.8	94.3 ± 1.1	51.9 ± 1.4	57.9 ± 1.9	56.9 ± 0.2	60.2 ± 0.6
Lap Joint Strength-MPa	707 ± 41	792 ± 0.5	666 ± 5	765 ± 9	885 ± 18	972 ± 24	965 ± 6.5	1033 ± 12

Fig. 13 — Stress-extension curve for CP800 lap shear tests along with the last DIC frame of strain distribution.

Fig. 14 — Micrographs show the fracture location for the CP800 weld in the HAZ.

tests are shown in Fig. 13. It should be noted that using a high-strength welding wire improved the peak strength and ductility of CP800 slightly. In addition, the rotation in the CP800 was higher (20 to 22 deg), which imposed higher bending stress, compared to the 980HF lap joint rotation (10 to 15 deg).

All samples showed strain localization in the HAZ. The strain maps for the uncoated CP800 welds showed that the weld failure occurred when local strain in the HAZ reached about 10% engineering strain; however, failure occurred near the weld interface in the FZ/HAZ boundary for the CP800 welds, as shown in Fig. 14. This is in agreement with previous work from Biro et al. (Ref. 10), who showed that when the BM had moderate to high hardness and the weld had low hardness, the joint failed in the FZ/HAZ boundary.

The stress-extension curves and DIC strain maps for 980HF welds are shown in Fig. 15. In this weld, strain localized in the HAZ, and the maximum local strain was only 5%. In all of the 980HF lap joints, the fracture occurred in the FZ near

the weld interface (see Fig. 16) even though the FZ did not contain any pores or other defects. This behavior may be explained by the hardness profiles, which showed that the FZ had a lower hardness of about 300 HV_{0.5} compared to the surrounding HAZ (490 HV_{0.5}). This sharp transition in the hardness profile due to the heterogeneity of the microstructure leads to deformation restriction in the FZ and rotation near the FZ/HAZ boundary, so the failure occurs in the weak area: the FZ. The strain map did not show any effect of coating on the strain distribution. In terms of the coating effect on the strength of the joint, no clear effect was observed on weld strength. While using a high-strength welding wire did not improve the joint strength, however, the ductility of the joint improved when a high-strength welding wire was used. Finally, the low angle of rotation (10 to 15 deg) for the 980HF weld samples could be attributed to the high hardness of the HAZ, which restricts the deformation in the FZ.

When the strength of all of the tested joints were compared to each other, it was clear that the butt joints exhibit-

Fig. 15 — Stress-extension curve for 980HF lap shear tests along with the last DIC frame of strain distribution.

Fig. 16 — Micrographs show the failure location for the 980HF weld in the weld zone.

ed a higher strength than the lap joints, as shown in Fig. 17. In general, using a high-strength consumable (ER-100S-6) slightly improved the joint strength in the butt joint configuration while it significantly improved the joint strength in the lap joint configuration,

Figure 18 shows the failure location for each configuration. It was also noted that the failure of the butt joint in the CP800 welds exhibited the traditional failure behavior through the BM where the failure was oriented 45 deg to the tensile force. Failure occurred on the maximum shear plane, while the 980HF failure followed a straight path through the HAZ-BM boundary. The lap joint failure of the CP800 welds for all samples was through the HAZ, while the 980HF lap

joint welds failed in the boundary between the HAZ-FZ due to the excessive strength mismatch between these two zones, as shown in the hardness profiles in Fig. 12.

Table 6 summarizes the peak load in kN for both configurations when tested in two different configurations. To calculate the stresses, the peak load value was divided by the cross-section area. It should be noted that at least three mechanical tests were conducted for each condition. The load needed to break the 980HF samples was lower due to the smaller thickness of the sheets compared to the CP800 samples.

To understand the effect of coating and welding wire strength, the weld strength was processed to visualize how each factor affects lap joint mechanical performance. In the

Fig. 17 — Bar chart compares the peak tensile strength values for the tested materials in both butt and lap joint configurations using different filler materials.

case of the CP800 steel, the joint strength was higher for the uncoated materials regardless of the welding wire used — Fig. 19. Conversely, in the case of the 980HF, the strength was higher for the GI-coated steel regardless of the welding wire strength. It is thought this behavior may be explained by the welding parameters used to weld the CP800, where higher heat input was used to weld the GI-coated steel. This would have resulted in lower cooling rates and a lower hardness value (which may be seen in the hardness profiles in Fig. 12). Also, the hardness mismatch between the weld/HAZ was higher for the 980HF welds, which would promote failure in the FZ and lead to a drop in the peak strength compared to the base material.

Welds made in the butt and lap joint configurations had different characteristic strengths. This is not surprising because the welds in the butt joint configuration are only subjected to tensile stresses, and the welds in the lap joint configuration are subjected to both tensile and bending stresses due to the rotation associated with the asymmetrical loading (Ref. 16). Figure 20 shows the effect of the welding wire strength on the lap joint strength. When lap joint strength results are compared to the BM strength, the tensile strength of the lap joints always undermatches the base material strength except in the 980HF-GI-condition weld using a high-strength welding wire (ER100S). In general, using a high-strength welding wire improves the lap joint strength by approximately 10% in all cases.

Surface coating and filler type also did not seem to affect the difference between the strength of the two joint types. Therefore, it is believed that this difference was due to geometric attributes alone and further work should be done to determine how changing tensile geometry affects joint strength, as these values will be used to model the part strength. The DIC strain maps for the lap joint showed the free end rotation very clearly with the plastic strain value reaching up to 1%, which could be used in future studies to find the unaccounted bending stress associated with the lap shear sample rotation. Using the lower strengths of lap shear joints could result in overly conservative results. Figure 21 compares the lap and butt joint peak strength with

Fig. 18 — Failure location for each configuration.

Fig. 19 — Zinc coating effect on the lap joint strength.

the base material strength. The butt joint overmatched the base material in all cases for both materials while the lap joint undermatched the base material except in the 980HF-GI condition. In addition, the butt joint strength exceeded the lap joint strength by 20% for welds in CP800-UC and 15% for welds in 980HF-UC when welded using a low-strength welding wire — Fig. 21A. When welded using a high-strength welding wire, the strength difference between the butt and lap joint was 12% for the CP800-UC and 8.5% for 980HF-UC — Fig. 21C. The GI-coated-material welds showed a similar trend when welded with a low-strength welding wire. The difference between the butt and lap joint was 22% for the CP800 GI and 10% for the 980HF-GI (Fig. 21B). When a higher-strength welding wire was used, the difference between butt and lap joint strength was 21% for the CP800 GI while it reduced to 5.5% for the 980HF-GI, as shown in Fig. 21D.

Conclusion

This work focused on the weldability of AHSS (CP800 and 980HF) using the GMAW process. The materials investigated in this project were in both uncoated and GI-coated conditions. The welds were performed using two different filler materials: low-strength filler (ER70S-6) and high-strength filler (ER100S-6). The following conclusions can be drawn from this study:

Fig. 20 — Comparison between lap joint strength and base material strength when different filler materials are used in: A — Uncoated materials; B — galvanized materials.

Fig. 21 — Comparison between butt and lap joint strength and base material strength: A — Lower-strength welding wire used without coating; B — lower-strength welding wire used with coating; C — high-strength welding wire used without coating; D — high-strength welding wire used with coating.

1) **Material and filler metal effect.** The weld strength increases as the material strength increases in both butt and lap joint configurations. However, the increase in the lap joint when a high-strength welding wire is used is higher than the increase in the butt joint strength.

The sharp hardness transition between the CGHAZ and the FZ in the 980HF welds led to a failure occurring in the FZ. More work is needed to optimize the welding parameters for high-strength material.

2) **Zinc coating effect.** No clear effect of coating was observed on weld strength. Weld strength is more influenced by metal composition. For example, the 980HF coated material showed better performance in terms of joint strength compared to the uncoated material.

The porosity present in the FZ of some butt welds in the GI-coated material welded with a low-strength welding wire

led to the decrease in average strength. The CP800 welds showed an 8% (70 MPa) reduction in strength compared to the uncoated materials. A 2% (25 MPa) drop was seen in the average strength of the 980HF butt weld samples due to porosity.

No porosity was observed in the lap joint welds. More work is needed to understand how porosity affects joint strength.

3) **Joint configuration effect.** Joints welded in a butt joint configuration were stronger than those welded in a lap joint configuration and generally overmatched the BM strength.

- Butt joint strength exceeded lap joint strength by 20% for welds in CP800-UC and 15% for welds in 980HF-UC when welded using a low-strength welding wire.

- When welded using a high-strength welding wire, the

strength difference between the lap and butt joint was 12% for the CP800-UC and 8.5% for 980HF-UC.

- When a low-strength welding wire was used to join the GI-coated steel, the butt joint strength exceeded the lap joint strength by 22% for CP800 and 10% for 980HF.
- Increasing the welding wire strength led to an improvement in the lap joint strength for the 980HF-GI material. The difference between the butt and lap joint decreased from 10% to 5%.
- It is believed that there is an interaction between welding wire strength and rotation, where increased FZ strength prevents rotation during shear testing, thereby reducing the bending stress and increasing the overall joint strength.
- Due to the prevalence of gaps and their known influence on weld quality, a sensitivity analysis should be done using these AHSS.

4) **Sample rotation.** The authors believe the lower strength of the lap joint is due to the unaccounted bending stress because of the sample rotation during the tensile test. Future work will focus on developing a new test methodology to prevent sample rotation during the lap shear test.

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