Influence of Paint Baking on Microstructure and Mechanical Behavior of Resistance Spot Welds

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INTRODUCTION

As the automotive industry develops vehicles with reduced mass for enhanced fuel efficiency and improved crashworthiness for passenger safety, an advance in joining technology is necessary to take best advantage of the development of high strength steels with superior combinations of strength and ductility. Gen3 AHSSs have been developed and maintain the structural integrity of an automobile manufactured with conventional steels but at a reduced weight by utilizing down-gauged sheet1, 2. Complex microstructures that may contain ferrite, martensite, bainite, and retained austenite are created in Gen3 steels with unique processing paths and fine-tuned alloying additions, such as elevated carbon and manganese contents compared to first generation (Gen1) AHSSs3–16. While carbon facilitates the retention of stable austenite to room temperature to achieve the desired mechanical properties via the transformation-induced plasticity (TRIP) effect, reduced weld strength may also occur, as illustrated in Figure 1 (with literature data3–12, 14, 16–27). The trend of diminishing cross tension weld strength with increasing carbon content in Figure 1(a) applies to Gen1 steels, consisting of dual phase (DP) and martensitic steels, as well as Gen3 steels. The hardness of the fusion zone generated during resistance spot welding, the most common joining method used during automobile assembly28, increases with carbon content; with elevated carbon levels, fusion zone microhardness in Gen3 RSWs tends to exceed the Gen1 fusion zone microhardness in steels of comparable substrate UTS (Figure 1(b)). It is generally accepted in the literature that the martensite formed in Gen3 RSWs exhibits lower toughness than martensite in Gen1 RSWs because of the difference in carbon content6, 7, 20, 30. While carbon content does influence martensite hardness, it, along with manganese, also decreases the martensite-start temperature31, increases the fraction of austenite retained in quenched microstructures32, and increases the fraction of twinned martensite formed during quenching33. Therefore, the potential influence of Gen3 alloying additions on spot weld characteristics, and thus the difference in Gen3 spot weld performance compared to Gen1, extends beyond that of just the martensite hardness.

An end stage process in the manufacturing of automobiles is to paint the vehicle and then cure this paint in a baking oven, a step that has been linked to improvements in strength and energy absorption of Gen3 spot welds6, 15, 29, 34–40. Several investigators have reported the absence of a baking effect in DP spot welds6, 34 while others have reported conflicting results that indicate DP spot welds exhibit a minor baking effect35, 36, 41, though the effects of baking in DP spot welds are consistently less than in Gen3 steels36, 41. Tumuluru observed a 6 pct increase in tensile shear strength after baking spot welds made in a DP780 and TRIP780 steel, with the DP780 exhibiting plug failure and the TRIP780 exhibiting interfacial failure35. Eftekharimilani et al. reported a 37 and 47 pct increase in cross-tension strength of single- and double-pulse welds made in a DH1000 steel; the single-pulse welds exhibited a partial-interfacial failure before and after baking, while the double-pulse welds exhibited plug failure in both conditions37. A change in failure mode from interfacial to partial-interfacial failure was observed with an 82 pct increase in cross-tension strength of Gen3 spot welds in work done by Shamsujjoah et al.5. Similar findings were obtained by Chabok et al., with the same failure mode transition observed along with a 100 pct increase in cross-tension strength in a single-pulse weld; a double-pulse weld that did not exhibit a transition in failure mode, however, experienced a 150 pct increase in cross-tension strength29. These inconsistencies in weld failure behavior after baking complicate the identification of a baking-critical weld microstructure. To isolate one of these microstructures, mechanical characterization of simulated supercritical (1200 °C) HAZ microstructures has been performed via uniaxial tension testing of specimens prepared with a Gleeble®; baking was reported to increase the yield strength, but no change in the ultimate tensile strength or uniform elongation was reported34. Therefore, a need still exists to characterize and understand the behavior of subcritical, intercritical, and supercritical HAZ microstructures in stress states present during cross-tension testing of resistance spot welds.
The work presented herein seeks to elucidate the mechanisms responsible for the paint bake effect in spot welds of Gen3 steels and explain the absence of paint baking sensitivity in Gen1 steels. A modified lap shear geometry is used for characterization of the baking sensitivity of spot welds made in a Gen1 and a Gen3 steel. The baking sensitivity of spot welds exposed to post-weld heat treatments to simulate HAZ microstructures is then characterized with this modified lap shear geometry, enabling the subcritical, intercritical, and supercritical microstructures to be evaluated separately. Microstructural characterization coupled with Mössbauer effect spectroscopy were used to compare simulated HAZ microstructures to actual weld microstructures and to identify a mechanism. Collectively, the work tests the hypothesis that the weld HAZ, particularly the intercritical HAZ, is responsible for the differences in as-welded performance and baking sensitivity of Gen3 steels compared to Gen1 steels.

**EXPERIMENTAL METHODS**

Two industrially produced steels were utilized to investigate the paint bake effect on spot welds of Gen3 steels. Both steels, a TRIP-aided bainitic ferrite steel, TBF1000, and a dual phase steel, DP1000, were available in the bare condition with a 1.0 mm sheet thickness. The chemical composition of each steel is listed in Table I. As several literature investigations have reported that DP steels exhibit no sensitivity to baking, the DP1000 was selected for comparison to the TBF1000 steel; any hypotheses regarding the mechanism(s) of the paint bake effect would also need to explain the absence of a paint bake effect in DP steels.

Resistance spot-welding was performed using a 100 kVA Taylor Winfield Type ERE-12-100 air pedestal spot welder at the Colorado School of Mines. The welding parameters used are shown in Table II. After welding, the samples to be ‘baked’ were submerged in oil at 180 °C for a duration of 20 minutes. All specimens were tested within 24 h of welding.
Mechanical characterization of spot welds was performed via modified lap shear testing on an Alliance® 20-kip screw-driven test frame at a displacement rate of 0.05 mm/s. The modified specimen geometry, a schematic of which is displayed in Figure 2, possesses a reduced coupon length and width compared to standard geometries. This coupon geometry was selected after preliminary findings revealed enhanced baking sensitivity associated with reducing the ratio of fusion zone size to coupon width from 1:10 in most standards to 1:4 in the modified geometry. Triplicate specimens for the as-welded and baked conditions in both steels were tested, and the resulting load-displacement curves were analyzed for maximum force and energy absorption obtained by measuring the area under the force-displacement curve. Representative specimens were sectioned and mounted in a resin based cold mount to avoid introduction of excessive heat that would be present if mounting specimens in Bakelite. Grinding and polishing to a final abrasive size of 1 μm was performed before etching with a 2 pct picral solution at room temperature for 5 s to reveal the weld solidification structure. Low magnification images of failed welds were acquired using an Olympus® DSX500 light-optical microscope with image stitching capability.

Figure 2. A drawing of (a) the modified tensile shear specimen geometry with (b) an illustration of the test specimen with arrows indicating the grip location and direction of force application.

To examine the baking sensitivity of the different HAZ microstructures formed in the spot welded DP1000 and TBF1000 steels, six modified lap shear specimens for each of five target temperatures were fabricated in each steel and subjected to a post-weld heat treatment, depicted in Figure 3(a)-(d). Dilatometry was performed to identify the critical transformation temperatures, Ac1 and Ac3, that would dictate the target heat treatment temperatures necessary to simulate one subcritical, three intercritical, and one supercritical HAZ microstructure. The Ac1 was measured to be 680°C in the TBF1000 and 723 °C in the DP1000 steel. The Ac3 for both steels was approximately 880 °C. The welded specimens were heated to either 650, 725, 775, 825, or 900 °C, held for 5 minutes, and quenched in water. Three “unbaked” specimens were retained in the as-quenched state, while three “baked” specimens were submerged in an oil bath at 180 °C for 20 minutes. Modified lap shear testing was then conducted according to the procedure indicated previously. Of note is the fact that simulation of HAZ microstructures using modified lap shear specimens enables testing of the HAZ microstructures with a geometry that incorporates the stress states present during actual weld testing in which baking sensitivity was previously measured. Microstructural characterization was performed on select conditions after standard metallographic preparation and etching specimens with a 2 pct nital solution for 5 s. A JEOL® 7000 field-emission scanning electron microscope was used with an accelerating voltage of 20 kV and working distance of 10 mm for characterization of the as received steels, the spot welds, and the simulated HAZ microstructures.

Mössbauer effect spectroscopy was implemented to quantify fractions and carbon contents of carbide and retained austenite phases pre- and post-baking in simulated HAZ microstructures. Specimen preparation followed the procedure detailed by
Euser et al., in which specimens were ground to a thickness of 80 µm and subsequently thinned in a solution of 10 parts deionized water, 10 parts hydrogen peroxide, and 1 part hydrofluoric acid until a thickness of 20–30 µm was achieved. Mössbauer spectra were acquired at room temperature with a 57Co–Rh source using a spectrometer operating in the triangular constant acceleration mode and a Wissel data acquisition module (CMCA-500 USB). Samples were run until high precision counting statistics were obtained to allow detection of small quantities of austenite and carbide phases. Velocity calibration and the isomer shift zero value were established with a pure bcc-Fe foil. Subspectra were fitted using Lorentzian line shapes with WinNormos V3.0 coupled with IGOR Pro V6.3 software. Thickness corrections and recoilless fraction corrections were part of the quantitative analysis. Further details are provided elsewhere.

RESULTS

Shown in Figure 4(a) and (b) are the force-displacement results obtained after modified lap shear testing RSW specimens in the as-welded (dashed lines) and baked (solid lines) conditions for the DP1000 and TBF1000 steel, respectively. The as-welded energy absorption and peak load of the DP1000 RSWs were 23.0 J and 13.4 kN, with no change observed after baking. An increase in energy absorption from 15.6 to 34.2 J was accompanied by an increase in peak load from 9.8 to 12.6 kN following baking in the TBF1000 RSWs. Comparing the energy absorption and strength of the as-welded and baked conditions in both steels, it appears that the TBF1000 RSWs exhibit reduced strength after welding that is mitigated by the baking cycle, as the baked strength is comparable to the as-welded strength of the DP1000 RSWs and the energy absorption is greater than DP1000 after baking.

Figure 4. Force-displacement curves obtained during modified tensile shear testing of the as-welded (dashed lines) and baked (solid lines) welds made in the (a) DP1000 and (b) TBF1000 steels.

Figure 5 presents transverse views at the mid-width position of failed welds after modified tensile shear testing; the opposite side of the welds (not shown) experienced some deformation without fracture, or a fracture similar to that displayed. All images are similarly orientated relative to the applied load during testing, with the ligaments to which load was applied indicated by the arrows. The sectioned views of the as-welded and baked DP1000 specimens after modified tensile shear testing are displayed in Figure 5(a) and (b), respectively. No difference in failure mode is observed for DP1000, which is to be expected given the similarity in the force-displacement curves shown in Figure 4. The as-welded and baked TBF1000 specimens are displayed in Figure 5(c) and (d), respectively, and reveal a significant difference in failure behavior induced by the baking process for this steel. In the as-welded condition, fracture initiates at the sheet interface along the fusion boundary. The crack propagates a short distance along the fusion boundary and then deviates from the fusion boundary; at point α in Figure 5(c), the fusion boundary curves to the right, but fracture does not follow the fusion boundary. Rather, the flat fracture continues along its original trajectory, through the supercritical HAZ. In the baked specimen, fracture appears to have initiated at the sheet interface in the supercritical HAZ and progressed through the intercritical HAZ and into the base metal, in a manner similar to that observed for the as-welded and baked DP1000 welds.

The results from the modified lap shear testing of the simulated HAZ microstructures are summarized in Figure 6 for each of the five post-weld heat treatment temperatures. Each data point represents the average peak force measured during testing or the area under the force-displacement curve (energy absorption) measured for three specimens, with error bars representing the standard deviation of that average. Dashed and solid lines represent specimens tested in unbaked and baked conditions, respectively. Figure 6(a) and (b) present the summarized energy absorption data for the DP1000 and TBF1000 specimens, respectively, and reveal a minimum in the energy absorption that occurs within the intercritical temperature range for the unbaked condition of both steels. In the DP1000 steel, no significant change in energy absorption occurs due to baking at any of the post-weld heat treatment temperatures, while the TBF1000 specimens heat-treated at 725, 775, 825, and 900 °C exhibit
increases in energy absorption of 15, 475, 335, and 41 pct, respectively. Similar trends are observed in the summarized peak load measurements for the DP1000 and TBF1000 in Figure 6(c) and (d), respectively. Again, baking is shown to have no influence on the peak load measured in any of the simulated HAZ microstructures formed in the DP1000 steel, while peak load increases of 85 and 93 pct are measured for the 775 and 825 °C conditions, respectively, of the TBF1000 steel.

Figure 5. Cross-sectioned welds after modified tensile shear testing in (a) the as-welded and (b) baked DP1000 welds and the (c) as-welded and (d) baked TBF1000 welds (light optical micrograph, 2 pct picral etch).

Figure 6. Summarized results from modified lap shear testing of simulated HAZ microstructures in post-weld heat treated spot welds tested in the unbaked (dashed lines) and baked (solid lines) conditions; energy absorption data are summarized for

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the (a) DP1000 and (b) TBF1000 welds while peak load measurements are given for (c) the DP1000 and (d) the TBF1000. Data are plotted against the post-weld heat treat temperature that indicates the peak temperature at which specimens were held in a furnace for 5 minutes prior to water quenching.

Because of the significant increase in strength and energy absorption after baking in the specimens heat treated at 775 and 900 °C for both steels, the microstructures in these specimens were examined to characterize microstructural evolution and to compare the simulated microstructures with those observed in actual spot welds. First, the spot weld microstructures along with the base metal for each steel are displayed in Figure 7. The base metal in the DP steel, Figure 7(a), exhibits a relatively coarser microstructure consisting of intercritical ferrite (F) and martensite (M). The base metal in the TBF steel, Figure 7(b), contains intercritical ferrite (F), carbide-free bainite with retained austenite (CFB+RA), and martensite (M). The intercritical microstructures are displayed for both steels in the unbaked and baked conditions in Figure 7(c) and (d) in the DP steel and Figure 7(e) and (f) in the TBF steel, respectively. Each of the intercritical microstructures contains two primary constituents: ferrite (F) and martensite or martensite austenite (MA) constituent. Of note is that some of the MA constituent in the unbaked TBF weld exhibits a smooth appearance while the MA constituent in the baked TBF weld along with both the unbaked and baked DP welds exhibit a textured appearance. The difference in etched appearance suggests a difference in carbon distribution between the former and three latter conditions. The coarse interlath features in the TBF weld microstructures are likely MA rather than austenite films expected in the CFB regions of the base metal. The supercritical microstructures are provided in Figure 7(g) and (h) for the unbaked and baked DP steel and Figure 7(i) and (j) for the unbaked and baked TBF steel, respectively. The supercritical microstructures appear to be fully martensitic. The martensite in the unbaked DP and TBF welds appears to exhibit a greater fraction of smooth regions compared to the baked conditions in each steel, suggesting some microstructural change (i.e., tempering reactions) may also occur in these microstructures during baking.
Figure 7. The as-received (a) DP and (b) TBF microstructures for comparison to (c) the unbaked and (d) baked DP intercritical microstructures, the (e) unbaked and (f) baked TBF intercritical microstructures, the (g) unbaked and (h) baked DP supercritical microstructures, and the (i) unbaked and (j) baked TBF supercritical microstructures. Phases are identified as ferrite (F), martensite (M), martensite-austenite constituent (MA) and carbide-free bainite with retained austenite (CFB+RA).

Conditions are labelled as as-welded (DP-AW) and baked (DP-B) in the DP steel and as-welded (TBF-AW) and baked (TBF-B) in the TBF steel. (secondary electron micrograph, 2 pct nital etch).

The simulated HAZ microstructures produced at 775 and 900 °C for modified lap shear testing were compared with actual spot weld microstructures to identify any microstructural evolution that occurred during baking. The microstructures produced after heating to 775 °C are shown in the unbaked and baked conditions for the DP steel in Figure 8(a) and (b) as well as the TBF steel in Figure 8(c) and (d), respectively. In the DP specimens, a reduced volume fraction of MA is observed in comparison to the TBF specimens, indicating a difference in HAZ microstructure evolution in the intercritical region for this steel. Further, possible presence of carbides (small spherical particles) is observed in the simulated intercritical specimens of the DP steel while such particles were not observed in the weld microstructures. It is possible that the use of a furnace to simulate weld microstructures contributed to the difference in microstructural evolution observed here, i.e., more tempering (carbide formation) occurred during heating prior to intercritical temperatures being reached. In the TBF specimens, the same change in MA constituent etching response after baking is observed for the simulated microstructures as was observed in the actual weld microstructures in Figure 7(e) and (f). The supercritical microstructures are presented in Figure 8(e) and (f) for the unbaked and baked DP steel and in Figure 8(g) and (h) for the unbaked and baked TBF steel; each of the microstructures appears fully martensitic, and no visible changes after baking are evident.

Figure 8. Microstructures obtained in post-weld heat treated specimens heated to temperatures of 775 °C in the (a) unbaked and (b) baked DP steel and the (c) unbaked and (d) baked TBF steel; the microstructures of specimens heat treated to 900 °C are displayed for the (e) unbaked and (f) baked DP steel as well as the (g) unbaked and (h) baked TBF steel. Phases are identified as ferrite (F), martensite (M), and martensite-austenite constituent (MA). Conditions are labelled as as-welded (DP-AW) and baked (DP-B) in the DP steel and as-welded (TBF-AW) and baked (TBF-B) in the TBF steel. (secondary electron micrograph, 2 pct nital etch).

Mössbauer effect spectroscopy was performed on the specimens heat treated at 775 and 900 °C in the unbaked and baked conditions of the TBF steel to help quantify microstructural changes associated with the improvements in strength and toughness as well as the apparent tempering reactions observed after baking in these conditions. Figure 9 presents the results obtained from Mössbauer effect spectroscopy and provides the fraction of the iron atoms that are in austenite as well as the carbon content of that austenite in the unbaked and baked conditions for each temperature. The data indicate that baking increased the carbon content of the retained austenite in both the 775 and 900 °C conditions. In the 775 °C condition the fraction of austenite was measured to be $2.7 \pm 0.1$ and $2.8 \pm 0.1$ at pct Fe in the unbaked and baked conditions, respectively; after baking, the carbon content of the austenite increased from $1.5 \pm 0.7$ to $2.2 \pm 0.5$ at pct carbon. In the 900 °C conditions, the fraction of austenite was significantly lower at $0.8 \pm 0.1$ at pct Fe in both conditions; the austenite in this condition exhibited an increase in carbon content from $2.2 \pm 1.8$ to $6.3 \pm 2.4$ at pct carbon. Though not shown, the results obtained for transition
Carbides and cementite fractions were at or below the detection limit for this technique in all conditions except the baked 900 °C specimen, in which 0.5 ± 0.3 at pct Fe was measured as \( \eta \), indicating a slight amount of precipitation during baking in this condition. Another notable result is the measurement of total carbon accounted for, as carbon in solution in the martensite or trapped at defects is not detectable by Mössbauer spectroscopy, and so this measurement provides a qualitative comparison of the martensite carbon supersaturation. In the unbaked 775 °C specimen only 18 ± 13 pct of the carbon was accounted for via the transition carbide or austenite, and this increased to 48 ± 27 pct carbon in the baked condition. In the 900 °C specimens, an increase from 41 ± 27 pct to 61 ± 27 pct was measured after baking. These data suggest that the unbaked 775 °C condition may contain the greatest amount of carbon in solution in the martensite and would thus exhibit the greatest driving force for tempering reactions during the baking cycle.

Figure 9. Mössbauer effect spectroscopy results quantifying the fraction of austenite and the carbon content of austenite present in HAZ microstructures simulated at (a) 775 °C and (b) 900 °C for unbaked and baked TBF specimens.

**DISCUSSION**

Following a baking treatment, the energy absorption and strength of TBF spot welds increased and the failure location changed. In contrast, DP spot welds exhibited no change in strength, energy absorption, or failure mode in response to baking, though each of these characteristics of the DP welds exceeded that of the TBF welds in the unbaked condition. Simulated HAZ microstructures in the TBF steel exhibited strength and toughness troughs within the intercritical range that were mitigated following the baking treatment, whereas the DP steel exhibited a less prominent toughness trough that was less affected by baking. Microstructural characterization revealed an apparent tempering of MA constituent in the intercritical HAZ of simulated TBF weld microstructures and actual TBF spot weld microstructures, while MA constituent in the corresponding microstructures for the DP steel appeared tempered prior to baking. Tempering of MA constituent during baking has been previously associated with improved properties of spot welds made in a TRIP1180 steel46 as well as a QP1000 steel41; these previous findings in addition to the results of the present work support the hypothesis that differences in the intercritical microstructure may be responsible for the baking effect observed in the TBF steel and the absence of the baking effect observed in the DP steel.

The mechanical characterization of simulated HAZ microstructures in Figure 6 revealed a trough in strength and toughness for simulated intercritical HAZ microstructures in both steels, though only the TBF microstructures exhibited any significant improvement after baking. While little attention has been given to the intercritical microstructures formed in spot welded sheet steel, these microstructures have been extensively researched for pipeline applications with particular emphasis on the detrimental impact of high-carbon MA constituent on toughness. Investigations of welds in thicker plate steels have revealed a toughness trough in the intercritical HAZ47, 48, and Davis and King reported that the greatest loss in toughness was associated with a connected network of MA constituent. The authors hypothesized that the buildup of residual stress from the martensitic transformation, along with the difference in hardness between ferrite and MA constituent, led to the loss in toughness48. In the present work, a connected network of MA constituent was observed in the intercritical HAZ of the DP and TBF steels, though the characteristics of the MA constituent differed significantly between the two steels.

The differences in MA constituent appearance in Figure 7 and Figure 8 suggest a difference in local chemical composition, i.e., the distribution of carbon within the MA constituent, in the unbaked DP and TBF microstructures. It is presumed that the elevated manganese (and perhaps carbon) levels in the TBF steel decrease the martensite-start (\( M_s \)) temperature relative to the
DP steel. Martensite in the DP steel would have transformed at a higher temperature, promoting autopartitioning or autotempering during cooling that would explain the textured appearance of the MA constituent in the unbaked DP weld (Figure 7(c)). The lower silicon concentration in the DP steel also promotes autotempering/autopartitioning in this steel. In the TBF steel, the martensite formed at lower temperatures may be more supersaturated in carbon after cooling of the weld, as evidenced by the smooth appearance. An increase in carbon supersaturation of the martensite in the TBF welds would be associated with higher hardness (greater difference in hardness between ferrite and MA) which has been previously linked to low toughness\(^{48, 49}\). The paint baking mechanism relieves carbon supersaturation in the martensite, reducing the hardness difference between ferrite and MA constituent and also reducing residual stresses in the microstructure, prompting an increase in toughness.

The results obtained via Mössbauer characterization of the simulated TBF microstructures indicate that carbon partitioning to austenite occurs during baking of simulated intercritical and supercritical microstructures. A greater fraction of austenite was measured in the intercritical specimens, though greater carbon enrichment after baking was measured in the supercritical specimens. The carbon enrichment of austenite likely occurs in parallel with a reduction of carbon supersaturation in martensite, both which might influence spot weld mechanical properties. In the simulated HAZ microstructures, the TBF specimens exhibited lower toughness in the unbaked condition but higher toughness in the baked condition compared to the DP specimens. Carbon partitioning to the austenite during baking may have stabilized austenite and facilitated an extended TRIP effect, leading to an increase in toughness. Such an observation was made by Hou et al., in which the rate of austenite decomposition during tensile deformation was reduced following a baking cycle, leading to an increase in strength and toughness\(^{50}\). Similar improvements in mechanical performance after baking have been attributed to carbon partitioning to austenite as quantified by \textit{in situ} synchrotron high-energy X-ray diffraction by Li et al.\(^{51}\) as well as 3D atom probe tomography by Lu et al.\(^{52}\). In the previously mentioned studies, baking sensitivity was associated only with steels possessing elevated manganese contents\(^{50, 52}\), but the mechanism involving carbon enrichment of austenite during baking might reasonably be hypothesized to contribute to the toughness improvements in the TBF steel examined here.

**CONCLUSIONS**

A comparison of spot weld mechanical property evolution in response to a simulated paint baking treatment was conducted using a modified lap shear geometry. The results reinforced prior findings that DP spot welds exhibit no sensitivity to baking, while Gen3 TBF spot welds exhibit an increase in weld strength and energy absorption in addition to an improvement in failure mode after baking. To understand this behavior, simulated HAZ microstructures were tested \textit{via} the modified lap shear geometry; this testing identified a trough in toughness and peak load for specimens heat treated at intercritical temperatures. After baking, the trough was mitigated for the TBF specimens, but no change was measured for the DP specimens. Microstructural characterization of actual weld microstructures as well as simulated HAZ microstructures revealed that the degree of tempering of the MA constituent in the intercritical microstructure is consistent with improvements in mechanical performance. Mössbauer effect spectroscopy revealed carbon partitioning to the austenite occurring during baking of both intercritical and supercritical microstructures; this mechanism could potentially also contribute to energy absorption.

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