Influencing mechanism of inherent aluminum oxide film on coach peel performance of baked Al-Steel RSW

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HIGHLIGHTS
• The fracture mode of 5754/LCS resistance spot welds changed from button to interfacial after paint baking.
• In-situ high temperature SEM was utilized to reveal the cracking process of oxide film inclusions due to baking.
• Baking induced thermal stress at the Al-oxide film and formed cracked oxide film inclusions.
• Oxide film inclusions floating on the faying interface acted as low-energy crack paths under coach-peel load after baking.

GRAPHICAL ABSTRACT

ABSTRACT
A traditional resistance spot weld process in combination with Multi-Ring Domed welding electrodes was applied to join AA5754-O to low carbon steel. The coach peel performance of welds was compared between the as-welded (unbaked) and post-paint bake condition (baked). The unbaked joints exhibited button pullout fracture mode, whereas the baked joints were severely weakened and demonstrated an interfacial fracture mode which is attributed to the baking process inducing thermal stress at the dissimilar material interface which cracked the semi-continuous oxide film inclusions at the edge of the solidified aluminum nugget to form a more continuous defect layer across the diameter of the aluminum nugget. These defects act as low energy crack paths under applied loads and reduce joint strength by promoting undesirable interfacial fracture in baked welds.

1. Introduction
The combination of steels with aluminum alloys in vehicle body structures has become attractive for mass savings in order to meet automotive greenhouse gas emission targets. Resistance spot welding (RSW) is the main joining method for automotive body assembly. However, directly joining aluminum and steel by RSW is challenging due to the different physical properties and metallurgical incompatibility of the two metals as well as the presence of oxide film on the aluminum surface. Aluminum oxide has a very high melting point, 2060 °C, compared with pure aluminum which melts at 660 °C [1]. This oxide is incorporated within the weld structure during melting and may influence the solidified joint performance [2]. A conventional approach to mitigate the negative effects of oxide films is by mechanical grinding or chemically cleaning the aluminum sheets [3]. However, these...
The application of stress to the faying interface. The application of stress after paint baking which affected joint mechanical strength by altering for AA6xxx aluminum-steel joints produced using a MRD electrode. welds has been presented in the literature. Haselhuhn, et al. [13] ob-

The effect of the paint baking procedure on dissimilar Al-steel weldability in the as-received condition of the material, however, it is necessary to characterize the post-baked weld performance as this directly affects how the welded structure will behave in service [11] [12]. The effect of the paint baking procedure on dissimilar Al-steel welds has been presented in the literature. Haselhuhn, et al. [13] observed that the hardness of the aluminum heat affected zone (HAZ) and magnitude of notch root angle at the faying interface were critical for AA6xxx aluminum-steel joints produced using a MRD electrode. They noted that the AA6xxx HAZ hardened in some Al-steel welds after paint baking which affected joint mechanical strength by altering the application of stress to the faying interface. The application of stress at the faying interface was identified as a critical parameter for dissimilar Al-steel joints as this is the location of the weak IMC layer. Shi, et al. [14] also evaluated the properties of AA6xxx-stee dissimilar RSW in the baked condition. They demonstrated that the fracture mode of baked welds in the tensile shear configuration was determined by the critical weld nugget size which was represented as a function of the shear strengths of the HAZ and IMC layer, aluminum sheet thickness, and the thickness reduction rate (indentation) of the aluminum sheet.

Fewer studies evaluating RSW of AA5xxx-steel have been presented in the literature as compared to AA6xxx-steel, despite AA5xxx aluminum alloy being widely applied in automotive body in white structures. One study evaluating AA5xxx-steel RSW discussed the effect of welding parameters on unbaked weld microstructure and mechanical properties [15]. Other work has characterized the morphology and phases of the IMC layer but the authors neglected to mention the effect of oxide film which should play an integral role in weld fracture mechanics [16]. The inherent oxide film between the AA5xxx and steel sheets was demonstrated in another study to produce expulsion and oxide film defects in the weld structure which resulted in poor weldability and low mechanical properties [17]. AA5xxx-series aluminum alloys exhibit a higher affinity for oxygen than AA6xxx alloys due to the elevated Mg content in the AA5xxx alloys [18]. Thus, there is the risk for more persistent oxide film defects in dissimilar AA5xxx-steel welds compared to AA6xxx-steel welds. Furthermore, because the yield strength of AA5754 alloys initially increases upon application of heat up to 121 °C [19], it can be more difficult to achieve the same degree of deformation in AA5xxx as with AA6xxx under the same welding conditions. Less deformation of the Al could translate to limited disruption of oxide films at the faying interface upon heating. Thus, preparation of AA5xxx-steel welds using MRD electrodes may yield high quality welds with fewer oxide defects, but no such studies have been presented in the literature. The existing studies on the mechanical performance of Al-steel RSW joints primarily focus on the fracture mechanisms of unbaked resistance spot welds or spot welds comprised of AA6xxx series aluminum with little emphasis on analogous AA5xxx series–steel welds. This study attempts to fill this gap in the literature by investigating the effect of paint bake treatment upon the mechanical performance of AA5xxx-steel RSW joints.

In the following study, as-welded i.e. unbaked, and post-paint baked i.e. baked, AA5xxx-steel joints made using MRD electrodes were compared in terms of coach peel strength, energy absorption, fracture mode, and crack path. The nugget size, aluminum penetration, indentation, IMC thickness, and defect distribution along the faying surface of baked and unbaked welds were measured to understand the baking induced changes on both macro and micro levels. Finally, in-situ high temperature scanning electron microscopy of the Al-steel weld faying surface was performed to reveal the structural evolution during the paint bake thermal cycle and thereby provide evidence on the root cause of performance differences between baked and unbaked welds.

### Table 1
Chemical composition of AA5754-O and HDG LCS (wt%).

| Material     | Si  | Cu  | Mg  | Fe  | Mn  | Zn  | Ti  | Al  |
|--------------|-----|-----|-----|-----|-----|-----|-----|-----|
| AA5754-O     | 0.12| 0.02| 2.70| 0.28| 0.11| 0.01| 0.01| Bal.|
| HDG LCS      | 0.003| 0.11| 0.01| 0.008| 0.005| 0.034| Bal.|

### Table 2
Mechanical properties of AA5754-O and HDG LCS at room temperature (RT) and elevated temperature.

| Material         | Elastic modulus (GPa) | Yield strength (MPa) | Ultimate tensile strength (MPa) | Elongation (%) | Coefficient of thermal expansion (×10⁻⁶ K⁻¹) |
|------------------|-----------------------|----------------------|---------------------------------|----------------|---------------------------------------------|
| AA5754-O (RT)    | 72.4                  | 92                   | 220                             | 15             | 25.7                                        |
| AA5754-O (175 °C) | 62.5 [21]             | 82                   | -                               | -              | -                                           |
| HDG LCS (RT)     | 207                   | 138                  | 275                             | 61             | 11.3                                        |

### Table 3
Multi-stage resistance spot weld schedule.

| Weld force | Pre-heat | Stage 1 (RMS) | Stage 2 (RMS) |
|------------|----------|---------------|---------------|
| 1200 lb.  | 40 ms @ 8 kA | 250 ms @ 9.3 kA | 1420 ms @ 11.3 kA |
| (5.34 kN) | 10 ms cool | 500 ms cool | 500 ms hold |
2. Experimental procedure

2.1. Materials

1.0 mm thick AA5754-O aluminum alloy and 2.0 mm thick hot dip galvanized low carbon steel (HDG LCS) sheets were joined in this study. For each material, the nominal chemical composition and mechanical properties from the supplier are provided in Tables 1 and Table 2. The yield strength of 5754 at 175 °C was measured as well using an Instron tensile machine with a crosshead rate of 2.0 mm/min. The tensile specimen was selected based on the ASTM E8 standards [20]. The thermal expansion coefficients of the Al were measured from 25 °C to 250 °C and steel were measured from 25 °C to 700 °C using a thermal expansion instrument (DIL 402 Expedis). No surface cleaning procedures were performed prior to welding and all experiments were completed with materials in the as-received condition.

2.2. Welding process

Resistance spot welding was performed using a CENTERLINE medium frequency direct current (MFDC) welding machine integrated with a FANUC R-2000i industry robot and a 6043 WTC programmable logic controller. All welds were made using MRD electrodes [5] [6]. The electrodes were comprised of a conventional C15000 alloy with a body diameter of 19 mm and featured a 25 mm radius of curvature on the electrode cap surface. The average ring heights were measured to be 75 μm, 93 μm, 84 μm, 56 μm, and 52 μm, respectively, from inner ring to outer ring and were created using a rotary cutting blade [22]. Table 3 summarizes the weld schedule which included a preheat stage and 2 welding stages which were designed to induce multiple solidification events during welding [4,7]. All welding was performed with the aluminum sheet in the positive polarity condition. Samples were welded on a non-conductive aluminum sheet in the positive polarity condition. Samples were welded using wire electrical discharge machining (EDM). Then, the thin slice of metal was further cut along the thickness direction of the weld to obtain a 5 mm×3 mm×1 mm sample which had one boundary coincident with the weld centerline. Specimens were polished using the procedure previously described before conducting the in-situ SEM test.

Fig. 2. Illustrations of metallurgical sample: definitions of different weld dimensions and locations of IMC thickness and oxide film defects measurement regions.

2.3. Mechanical testing

Quasi-static coach peel tests were performed for the Al-steel welds in this study, refer to Fig. 1. The weld closest to the coach peel bend was the test weld, which was pulled to fracture to acquire the load-extension curves during testing. The other weld was the anchor weld, which is also sometimes referred to as the shunt weld. Quasi-static tests were conducted on a SUNS tensile machine at a crosshead rate of 5.0 mm/min. Three coach peel specimens were repeated in both the unbaked and the baked conditions.

2.4. Metallurgical examination

Metallographic samples were prepared by a standard procedure for polishing aluminum. Specimens were cold mounted in epoxy, polished to a surface finish of 0.03 μm using silica, and were etched using Keller’s solution before measuring the nugget morphology. Three welds were examined in both the unbaked and baked conditions for metallurgical analysis. Weld macro and microstructures were imaged using a LEICA DM4M digital optical microscope. For both the unbaked and the baked welds, images of the aluminum nugget at the faying interface were taken at 11 evenly spaced locations along the maximum diameter of the aluminum weld nugget. The average thickness of the IMC layer and the length of the oxide film defects were measured in each of the 11 images, refer to Fig. 2. The average IMC thickness was determined by calculating the total IMC area based on image contrast and dividing that area by the section length (corresponding to the total length of the specimen represented in the image). The percentage of oxide film defect was determined by measuring the length of the oxide film and dividing that length by the section length.

The Finer scale observations of weld defects and coach peel fracture surfaces were performed using a ZEISS scanning electron microscope (SEM) with an acceleration voltage of 15 keV. Electron Backscattered Diffraction (EBSD) of the IMC layer was performed on a TESCAN LM3 SEM using 20 keV acceleration voltage.

2.5. In-situ SEM examination sample preparation

In-situ observation of structural evolution during the weld baking process was performed using a TESCAN LYRA3 SEM with a Murano heating stage. Specimen cross-sectioning for this in-situ SEM observation is presented in Fig. 3, in which the specimen width was 1.0 mm. The 1.0 mm slice was cut from the center of the unbaked spot weld using wire electrical discharge machining (EDM). Then, the thin slice of metal was further cut along the thickness direction of the weld to obtain a 5 mm×3 mm×1 mm sample which had one boundary coincident with the weld centerline. Specimens were polished using the procedure previously described before conducting the in-situ SEM test.

Fig. 4 is a plot of the thermal histories applied to the sectioned welds for in-situ SEM analysis and the weld specimens during the experimental baking process. It should be noted that the in-situ analysis included an additional 5-min warming stage as the equipment required the specimens to be in place prior to heating. Structural evolution throughout the
heating-holding-cooling process was imaged and results corresponding to the initial phase (0 min), the end of the holding stage (40 min), and the end of the cooling stage (75 min) were acquired and correspond to the three phases marked with numbers S1-S3 in Fig. 3. “S1” represents the initial stage of the system, “S2” designates the end of the holding stage, and “S3” represents the end of the cooling stage. The weld specimens subjected to the experimental paint baking process were put in an atmospheric environment oven which was preheated to 175 °C. The specimens were heated for 35 min to simulate the paint baking process prior to natural cooling to room temperature.

3. Results and discussion

3.1. Weld coach peel performance

Unbaked spot welds exhibited an average peak load of 441 N and an average total energy absorption of 7.2 J as defined by the area under the load-extension curve, refer to Fig. 5 and Fig. 6. However, the baked specimens demonstrated a 33% lower average peak load of 294 N and a 56% reduction of total energy absorption to 3.1 J compared to those of the unbaked welds. The baked welds also exhibited reduced stiffness compared to the unbaked welds as evidenced by the lower initial slope of the load-deflection curve. The greater energy absorption of the unbaked spot welds is a combined result of greater peak loads where the onset of cracking occurred (on average at 12 mm) followed by significant extension until complete separation of the weld with an average extension of 21 mm. In contrast, the peak load of the baked specimens and complete separation of the weld occurred coincidentally at significantly lower loads and at an average extension of 14 mm indicating that the baked welds exhibited a less ductile behavior and are less desirable.

Fractography was undertaken to identify any differences in fracture surfaces between unbaked and baked coach peel samples, refer to Fig. 7 and Fig. 8, respectively. The unbaked weld exhibited button pullout fracture, so defined because of the remaining portion of the aluminum weld nugget attached to the steel in the center of the spot weld, refer to Fig. 7 (a) and (b). The backscattered electron image presented in Fig. 7 (a) exhibited dark grey and lighter grey regions on the fracture surface. The dark grey regions were aluminum and oxide rich whereas the lighter grey surfaces were iron and IMC according to the EDS spot scan results provided in Table 4. It could be deduced from observation of Fig. 7 (c)–(f) and corresponding EDS results in Table 4, that the crack initiated outside of the aluminum nugget region at the interface between the aluminum and steel sheets before entering the aluminum nugget.
refer to Fig. 7 (c). Then cracks propagated through the aluminum nugget along aluminum with oxide-rich films and the IMC layers parallel to the faying interface, leaving fractured aluminum blended with IMC fragments, refer to Fig. 7 (d) and (e) and Table 4. After the crack propagated for a certain distance, it deflected upward into the aluminum weld rather than continuing to move parallel to the faying interface. This action initiated the aluminum button formation. Following this, the crack propagated around the aluminum nugget following a path along the faying interface that was observed to be rich in iron and iron-aluminum intermetallics, until complete fracture of the weld occurred, refer to Fig. 7(f).

The baked welds exhibited interfacial fracture mode in which the weld structure fractured along the faying interface, refer to Fig. 8(a). The backscattered electron images presented in Fig. 8 exhibit the large dark grey fracture surface rich with residual oxide film and aluminum near the faying surface, refer to Fig. 8(d) and (f). There were scattered areas of lighter grey which indicated the presence of higher atomic number elements, refer to Fig. 8(c) and (e), such as IMC phase according to the EDS spots scanning results provided in Table 5. During coach peel testing, the crack propagated through the aluminum nugget along aluminum with rich oxide and the IMC layers parallel to the faying interface like the initial path as in the unbaked welds. However, the crack path of the baked weld did not deflect upward through the aluminum like the crack in the unbaked weld. Rather, the crack in the baked weld continued to propagate along the aluminum with rich oxide or IMC across the entire nugget diameter and resulted in weak, brittle interfacial fracture of the coach peel specimens, refer to Fig. 8(b).
Table 4
Chemical composition of various locations in Fig. 6(c) to (f).

| Figure Position | Element (at.%) | Anticipated phases |
|-----------------|----------------|--------------------|
|                  | Al  | Fe  | O    | Mg  | Si  | Zn  |
| 6(c) 1           | 76.11 | 16.01 | 5.69 | 1.76 | 0.30 | 0.13 | Aluminum |
| 6(d) 3           | 82.62 | 3.11  | 6.68 | 6.38 | 2.12 | 0.20 | Aluminum |
| 6(e) 5           | 86.16 | 1.30  | 9.61 | 2.60 | 0.23 | 0.00 | Aluminum(oxide-rich) |
| 6(f) 7           | 67.87 | 27.28 | 3.69 | 0.88 | 0.00 | 0.28 | IMC |

Table 5
Chemical composition of various locations in Fig. 7(c) to (f).

| Figure Position | Element (at.%) | Anticipated phases |
|-----------------|----------------|--------------------|
|                  | Al  | Fe  | O    | Mg  | Si  | Zn  |
| 7(c) 1           | 9.14 | 84.65 | 5.75 | 0.05 | 0.00 | 0.41 | Iron |
| 7(d) 3           | 84.31 | 4.06  | 8.08 | 3.41 | 0.14 | 0.00 | Aluminum(oxide-rich) |
| 7(e) 5           | 79.00 | 1.84  | 12.45 | 6.55 | 0.16 | 0.00 | Aluminum(oxide-rich) |
| 7(f) 7           | 67.31 | 26.56 | 5.34 | 0.32 | 0.39 | 0.08 | IMC |
| 8               | 87.02 | 3.90  | 5.98 | 2.67 | 0.00 | 0.43 | Aluminum |
3.2. Weld defect distribution along the faying surface

There were no discerning differences in the unbaked and baked weld macrostructures. Welds from both conditions exhibited an aluminum nugget approximately the shape of a half-ellipse with indentation primarily in the Al sheet and an intact faying interface between the aluminum and steel sheets, refer to Fig. 9 and Fig. 10. No significant differences were observed between the unbaked and baked aluminum nugget diameter, nugget penetration, or the amount of aluminum indentation, indicating that the baking temperature was insufficient to change the weld macrostructure, as expected since the bake temperature was significantly lower than the aluminum melting temperature.

Fig. 9(a)–(f) and Fig. 10(a)–(f) present higher magnification images of selected locations at the faying surface of the unbaked and baked welds, respectively. In both the unbaked and baked structures, IMC layers were present at the faying interface. The IMC layer was comprised of two phases: the thinner phase adjacent to the aluminum was FeAl₃ which had a smaller grain size and the thicker phase with a large grain size adjacent to the steel was Fe₂Al₅, refer to the Fig. 11, which is consistent with the IMC bi-layer structure identified by Hu, et al. [17] and Shi, et al. [23] in analogous AA6xxx Al-steel welds. The IMC structures were thinnest at the weld nugget periphery and increased in thickness at the center of the weld nugget with growth of the IMC layer occurring into the aluminum weld nugget. The greater thickness could be explained by the higher temperature in the nugget center compared to the weld periphery which produced a greater driving force for diffusion-based IMC growth in this region [24]. At the periphery of the unbaked and baked aluminum weld nuggets, defects appeared as semi-continuous or continuous black lines above the IMC layer in the aluminum, refer to Fig. 9(a), (f), Fig. 10(a), (e) and (f). Structures of this type have been identified as being rich in oxygen and magnesium and were remnants of a nascent oxide film layer that was originally present on the aluminum sheets surface as described by Sigler, et al [4] Due to the repetitive heating and cooling nature of the weld schedule [7], the uneven oxide films at the faying interface are typically broken into smaller particles and absorbed into the aluminum nugget or may remain at the faying interface resulting in oxide film inclusion and micro cracks where they could act as a low-energy crack path. These kind of structures concluded oxide film inclusion and micro cracks were designated as oxide film defects in the rest of the paper.

In the baked AA5xxx-steel welds these oxide film defects were more continuous across the faying interface of the weld and extended further toward the middle of the nugget from the nugget edge than seen previously for AA6xxx-steel welds. Gas porosity and shrinkage porosity were also observed in the center of the aluminum nugget near the faying interfaces for both the unbaked and the baked welds, refer to Fig. 9(c), (d) and Fig. 10(c), (d). These defects do not influence the initial crack propagation in the welds under externally applied loads and would not influence weld behavior in a button pullout fracture regime, but could affect cracking behavior at later stages of crack propagation during interfacial fracture as the crack approached these defects. Neither the size nor morphology of these gas or shrinkage pores changed as a result of the baking process.

To quantitatively characterize the distributions of IMC thickness and oxide film defects along the faying interface of the weld structure, an IMC thickness distribution chart was developed, refer to Fig. 12. The oxide film defect percentage in each section is shown in the chart in different colors. Green represents no defects, yellow corresponds to approximate 50% of the imaged section exhibiting oxide film defects, the red represents that oxide defects covered the entire section length.
and black is reserved for the two nugget edges where it could be difficult to discern the structures.

At the edges of the aluminum nugget, both unbaked and baked welds exhibited a thin IMC layer approximately 3.0 μm thick. The IMC thickness increased gradually from the nugget edge to the center of the nugget and reached a maximum value of approximately 8 μm in both welds. The IMC thickness distribution did not change significantly as a result of baking, refer to Fig. 12 (a) and (b). However, the distribution of oxide film defects near the faying interface changed significantly after baking. In the unbaked weld, only four positions exhibited oxide film defects with the length ratio of less than 50% and these defects were primarily located near nugget edges. In the baked weld, the oxide film defects were present across nearly the entire faying interface with 6 positions exhibiting 100% defect and 2 exhibiting 50% defect.

While the baking process did not cause changes in the weld macrostructure or in the IMC thickness, it significantly increased the amount of observable oxide film defects adjacent to the faying interface and made the distribution of oxide film defects more continuous along the faying interface.

The crack paths in unbaked and baked 5754-HDG LCS welds are summarized in Fig. 13. In the unbaked welds, cracks initiated at the
nugget edge in the oxide film defects regions and then propagated along oxide film defects and IMC layers in the aluminum nugget parallel to the faying interface, refer to Fig. 13(b). Since the oxide film defects were discontinuous, it was difficult for cracks to propagate fully into the middle of the nugget where less oxide film was present, under an upward peel force. Therefore, the crack propagated upward through the aluminum at sites of gas or shrinkage porosity or where aluminum sheet was thinned deeply by MRD indentation, forming a button fracture mode, refer to Fig. 13(c). For the baked welds, more continuous oxide film defects in the aluminum parallel to the faying surface provided a low-energy path for cracks to preferentially propagate, refer to Fig. 13(d) and (e). The crack propagated through the aluminum nugget parallel to the faying interface until the entire aluminum nugget was torn away, resulting in interfacial fracture, refer to Fig. 13(f).

3.3. Oxide film defect evolution during paint baking

In-situ high temperature SEM was conducted to understand how the thermal cycles during the paint bake process influenced the oxide film defects since these defects produced a profound effect upon mechanical performance, refer to Fig. 14 which contains still images edited from a video capturing the entire in-situ bake cycle. Three locations adjacent to the faying interface, B, C, and D in Fig. 14, exhibited discontinuous oxide film inclusions and micro cracks prior to baking, and were the chosen locations for observation. Images (b1)–(b3), (c1)–(c3), and (d1)–(d3) of Fig. 14 present the microstructural evolution of the locations B, C, and D during the heating-holding-cooling thermal cycle described in Fig. 4. At location B, refer to Fig. 14(b1), oxide film inclusions can be observed above the IMC layer before baking. After baking for 35 min, the microstructure in location B did not show any visible changes, refer to Fig. 14(b2). However, after cooling to room temperature, the region of the weld with oxide film inclusions shifted perpendicular to the interface indicating that cracking had occurred, refer to the white arrow in Fig. 14(b3), in which the lighter region was more obvious. Oxide film exposure and crack formation led to charging of these higher energy boundaries by the equipment’s electron beam, forming the lighter regions in the image. The oxide film inclusions in location C behaved similarly as the oxide inclusion in location B as the oxide film exhibited little change during the heating and holding stages but cracked after the weld cooled to room temperature, refer to Fig. 14(c1)–(c3). While oxide-associated micro cracks were also observed in the weld structure, refer to Fig. 14(d1)–(d3), these micro cracks did not exhibit any changes during the heating and cooling thermal process. These cracks acted as prefabricated gaps between portions of the weld material which can release part of the thermally-induced stress and reduce stress concentration during the baking process. Therefore, micro cracks did not cause baking degeneration.

To understand this behavior, the thermal stresses at the aluminum metal-oxide interface were calculated assuming the influence from the steel sheet was negligible as the oxide film inclusions were entirely surrounded by aluminum, refer to microstructures in Figs. 9 and 10. Aluminum oxide (Al₂O₃) naturally forms on the aluminum alloy surface in atmospheric conditions. The primary alloying element of the 5754 aluminum welded in this study is magnesium; thus, MgO also forms on the aluminum sheet surface. The thermal expansion coefficients of Al₂O₃ and MgO are provided in Table 6. Cracking of the oxide film inclusions during the cooling stage can be explained by thermal stresses.
induced by differences in the thermal expansion of the aluminum and oxide films at the faying interface, refer to Tables 2 and 6. The thermal expansion coefficient of aluminum is nearly 4 and 2 times larger than that of the Al₂O₃ and MgO, respectively. During the baking process, the heating induced thermal stress (σ) in the aluminum adjacent to the Al-oxide film interface can be calculated assuming aluminum was in an elastic phase using relative thermal stresses

\[ \sigma = \varepsilon E_{Al} = \Delta \alpha \Delta T E_{Al} \]

where \( \varepsilon \) is the thermal strain, \( \Delta \alpha \) is the difference in thermal expansion coefficients between aluminum and oxide film layers, \( \Delta T \) is the temperature difference between room temperature (25 °C) and the baking temperature (175 °C) and \( E_{Al} \) is the elastic modulus of aluminum at 175 °C, which was calculated using JMatPro [21]. Using this approach, the thermal stresses in the aluminum were calculated to be 176 MPa adjacent to the Al-Al₂O₃ interface and 121 MPa adjacent to the Al-MgO interface. The thermal stresses adjacent to the Al-oxide film interfaces exceeded the yield stress of the AA5754-O sheet at 175 °C, which was 82 MPa, indicating that sufficient energy was present for plastic deformation to occur. In addition, the relative thermal stresses tore the oxide film which has very weak ultimate strength (14.9 MPa) at 100 °C [25]. Thus, during the heating stage the thermal expansion of aluminum was constrained by the oxide film which produced plastic strain in the aluminum and the cracks started to form within the oxide film inclusions where stress was concentrated. During the cooling stage, the aluminum experienced plastic deformation as it attempted to contract to its original shape. The aluminum could not perfectly close to the original shape since the thermal stress exceeded the aluminum's yield stress and the oxide film inclusions were broken. This resulted in visible cracking at the Al-oxide film interfaces which produced more continuous defects cracks across the faying surface.

In actual weld specimens, thermal expansion and contraction occur in three-dimensions which complicates the stress state. The addition of these added constraints may worsen the bake-induced cracking of oxide films in real welds compared to the weld sections observed in the in-situ SEM tests. Although successful in AA6xxx-steel welds [4],

### Table 6

| Material   | Thermal Expansion Coefficient (x10⁻⁶ K⁻¹) |
|------------|------------------------------------------|
| Al₂O₃      | 6.9 [26]                                 |
| MgO        | 12.8 [27]                                |

Fig. 14. Microstructural changes during the heating-holding-cooling thermal cycle observed by high temperature in-situ SEM. (b1)–(b3) location B, (c1)–(c3) location C, (d1)–(d3) location D.
the combined use of the MRD electrodes and multiple solidification weld schedules was insufficient to break apart the robust oxide films in AA5xxx-steel welds which particularly degraded weld behavior after baking. Further research is required to mitigate the negative effect of oxide film defects in AA5xxx-steel welds through welding process improvement.

4. Conclusions

This body of work investigated the combined influence of aluminum oxide films and baking thermal cycles on the coach peel performance of AA5754 to low carbon steel resistance spot welds. The coach peel performance, weld morphology, IMC characteristics, and microstructural defect distribution of the unbaked and baked welds were systematically compared. In-situ high temperature SEM analysis was performed to observe the evolution of defects during the baking process. The conclusions drawn from this study are as follows:

- The Al-steel spot weld subjected to the post weld paint baking process exhibited a 33.3% degradation in coach peel strength and 56.1% lower energy absorption compared to the unbaked welds. In addition, the baked welds exhibited lower stiffness than the unbaked welds.
- The fracture mode changed from button pullout mode in the unbaked welds, in which cracks propagated through the aluminum sheet, to interfacial mode in the baked welds, in which the cracks propagated along the oxide film defects and IMC layers adjacent to the weld paying interface.
- The baking process was not observed to influence weld macrostructural features or the weld IMC thickness. However, the semi-continuous oxide film defects concentrated at the nugget edge in the unbaked welds were observed to be more continuous and were present across a greater length of the faying interface in the baked welds.
- The oxide film incursions cracked during cooling from the baking temperature due to thermal stresses between the aluminum and oxide film induced by the vastly different thermal expansion behavior of the two substances. The damaged oxide film structure presented low-energy crack paths during mechanical testing which resulted in weaker coach peel strengths for the baked welds compared to the unbaked welds.

Declaration of Competing Interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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