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Influence of thermomechanical processing sequence on properties of AA6082-IF steel cold roll bonded composite sheet

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Abstract

Cold roll bonded tri-layered composites consisting of AA6082 and IF steel have been produced. The goal was to find the optimal sequence of solutionizing, aging and rolling, for obtaining good bond strength and at the same time, a precipitation hardened aluminum structure. The influence of rolling temperature was also investigated. Results show that the sequence of the thermomechanical process influences the interface characteristics and the final bond strength. The different thermomechanical sequences did not have any negative effects on the aluminums aging potential.

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Keywords: Cold roll bonding; Steel-aluminum joints; Thermomechanical processing; Bond strength; Intermetallic phases

1. Introduction

Steel and aluminum are considered to be the two most important dissimilar metals to be joined into composite products [1]. In recent years several different joining techniques trying to achieve joints with high quality and strength have been studied [2]. One joining technique for producing such composites is cold roll bonding. Cold roll bonding is a solid-state joining technique, where large surface expansion due to plastic deformation is the main contributing factor for obtaining a solid joint and it is commonly assumed that the film theory explains the main bonding

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mechanism [3]. According to the film theory, the work hardened surface layer of the metals, produced by surface preparation, cracks during rolling due to the surface expansion of the metals, allowing fresh metal to extrude locally through the cracks. A bond can be formed in areas where fresh metal from each opposing metal surface meet [3, 4]. For a bond to be obtained, a minimum degree of thickness reduction is required, referred to as the threshold reduction [4]. As stated by Li et al. [5], the threshold reduction is higher when joining dissimilar metals where the differences in mechanical properties, especially the hardness, are large. Several process parameters influence the achieved bond strength of cold roll bonded composites, such as surface preparation [6-8], the degree of deformation/thickness reduction [9-12], heat treatments (pre and post) [7, 10, 12] and rolling temperature [9, 11]. Increasing bond strength with increasing degree of thickness reduction above the threshold reduction has been reported in several studies [4, 9-12]. In some studies, the aluminum has been reported to achieve a higher degree of thickness reduction in cold roll bonded composites consisting of aluminum and steel or another hard metal. This has been reported to be due to the lower yield strength found in aluminum compared to e.g. steel [13, 14]. Performing pre-heat treatments or increasing the rolling temperature result in higher bond strengths at lower thickness reductions and the threshold reduction is reported to decrease [7, 9, 11, 15]. Post-rolling annealing has also proven to increase the bond strength [10], as post-rolling heat treatment helps to decrease the composite's hardness, therefore increasing the toughness of the joint, reducing the residual stresses in the metals and attributing to atomic diffusion and short-range movement of atoms at the interface [15].

One of the main challenges when joining steel and aluminum is the formation of Fe_xAl_y intermetallic phases and the most commonly reported intermetallic phases are Fe_2Al_5 (η) and Fe_4Al_{13} (θ), often shortened to $FeAl_3$ [16-18]. However, which phases that form and the growth rate are strongly influenced by the chemical composition of the alloys, and the time and temperature reached, either during the joining process or during the post-joining heat treatment [16, 18]. In studies of joints produced with aluminum alloys containing various amounts of silicon, the addition of silicon has been reported to impede the formation of the Fe_xAl_y intermetallic phases. Fe-Al-Si phases have also been observed together with the common Fe_xAl_y phases in such joints [16, 19]. The intermetallic phases have been reported to strongly influence the bond strength of the joints, due to their hard and brittle characteristics [18, 20]. However, when the intermetallic phase layer is kept below certain thicknesses (below 5 µm or 10 µm as stated by Borrisutthekul et al. [20] and Schubert et al. [21] respectively), the layer will not have a negative influence on the bond strength of the joint.

The aluminum alloys in the AA6xxx series with silicon and magnesium as the main alloying elements are precipitation hardenable alloys, meaning that with the right heat treatment, a high strength state can be obtained, referred to as the T6-state [22]. Hence, when joining steel and AA6xxx, the thermomechanical processing steps can be decisive for the achieved bond strength, as well as the obtained strength in the aluminum sheet. In this study, the effect of the sequencing of solutionizing, aging and rolling in the thermomechanical process when producing cold roll bonded joints of AA6082 and IF steel, were investigated. The three different thermomechanical processing sequences were selected in order to determine the optimal processing route for producing joints with good bond strengths as well as a precipitation hardened structure in the AA6082. The influence of rolling temperature was also studied.

2. Experimental procedure

A 3 mm thick extruded AA6082 flat profile and a 1 mm thick rolled IF steel sheet were chosen for this study. The AA6082 profile was cold rolled down from 3 to 0.9 mm thickness, before the AA6082 and IF steel sheets were cut into specimens of 30 x 120 mm. Composites were produced by following three different thermomechanical processing sequences, at two rolling temperatures (150 and 200 °C) obtained by pre-heating directly prior to rolling. The three cases are presented in detail in Fig 1. For simplicity, the cases are named according to the state of the aluminum during rolling; Case 1; AA6082-AR – as rolled, Case 2; AA6082-W – solutionized prior to rolling and Case 3; AA6082-T6 – aluminum was aged to T6-state prior to rolling. Solutionizing was conducted at 540 °C for 2 minutes and the aging was conducted at 185 °C for 90 minutes for case AA6082-W and AA6082-T6, and 120 minutes for case AA6082-AR. The selected times were based on initial testing. After each step, the specimens were water-quenched and a time gap of 30 minutes was kept between the solutionizing and aging step in all three cases. The engineering stress-strain curves in the rolling direction for the 0, 9 mm thick AA6082 in the different states (as rolled, solutionized and aged to T6) and for IF steel at room temperature and at 200 °C are presented in Fig. 2(a).

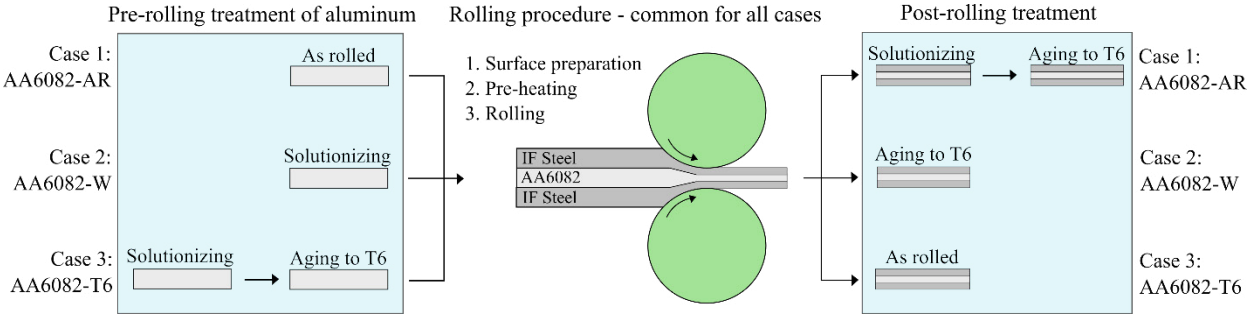

Fig. 1. Schematic illustration of three cases produced with different thermomechanical process sequences, Case 1; AA6082-AR, Case 2; AA6082-W and Case 3; AA6082-T6. Rolling was performed at two rolling temperatures, 150 °C and 200 °C after pre-heating directly prior to rolling.

Directly prior to the rolling procedure (pre-heating and rolling), the specimens were degreased with acetone and the surfaces were brushed manually with a 0.3 mm wire-steel brush. The specimens were stacked together, creating a composite with the stacking sequence; St/Al/St and fastened using aluminum rivets in front and in the back, as illustrated in Fig. 2(b), in order to prevent lateral movement between the specimens during rolling. Then, the pre-heating of the composites were conducted in an air-furnace set to 185 and 220 °C, where the composites reached the desired temperatures of 150 and 200 °C after 10 minutes. The composites were rolled directly after they were taken out of the furnace. The rolling was performed in a single pass, achieving a thickness reduction of approximately 60-65%.

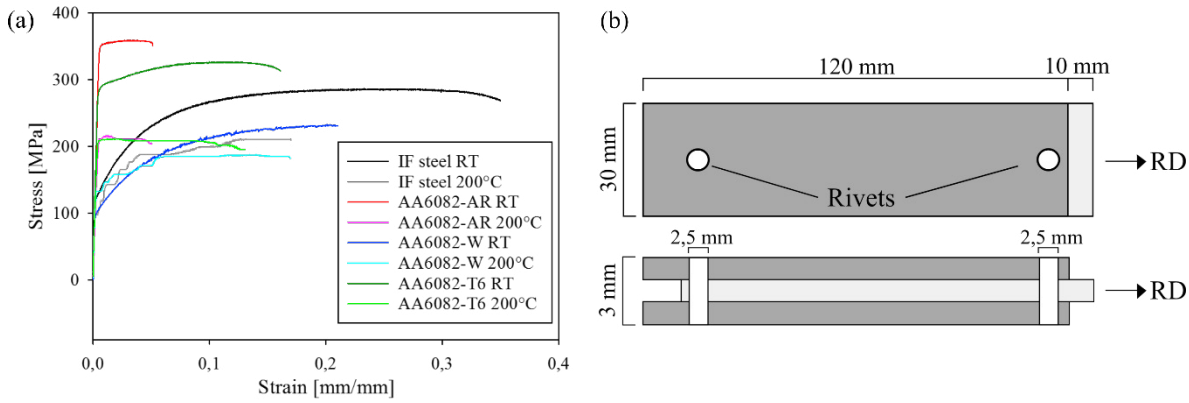

Fig. 2. (a) Engineering stress-strain curves for aluminum in three different states; as rolled AA6082-AR, solutionized AA6082-W and aged to T6 AA6082-T6, and IF steel at room temperature and 200 °C. (b) Illustration of composite with stacking sequence St/Al/St. Aluminum rivets (d: 2,5 mm) in front and back of the composite to prevent movement during rolling. RD = Rolling direction.

One part of the composite was prepared using standard metallographic preparation techniques, and interface characteristics were studied using optical microscopy, scanning electron microscopy (SEM) and transmission electron microscopy (TEM). For the TEM analysis, cross-sectional TEM lamellas were prepared with a focused ion beam of type FEI Helios G4 UX and the characterization was performed with a JEOL JEM-ARM200F operated at 200 kV. The second part of the composite was used for peel-testing in order to measure the bond strength. The peel-testing was conducted using a custome build peel-test rig, based on ASTM-D3167-10, allowing delamination with a 90-degree peeling-direction and a speed of 25 mm/min. The peel-test samples measured roughly 20 x 250 mm. After peel-testing, the fracture surfaces were studied by SEM and energy-dispersive X-ray spectroscopy (EDS). The average bond strength was determined using the following equation

$$Average\ bond\ strength\ \left[\frac{N}{mm}\right] = \frac{Average\ peel\ force\ [N]}{Sample\ width\ [mm]}\ . \tag{1}$$

3. Results and discussion

Cross-sections of the composites were studied after rolling in order to measure the achieved thickness reduction in the total composite and in the individual steel and aluminum specimens. The results from all three cases are presented in Table 1. The achieved hardness in the aluminum specimen after the final step in the thermomechanical processes are also presented in Table 1. From the hardness measurements, it can be observed that the aluminum in all cases reached high hardness values and it can be concluded that the aluminum for all cases, reached T6-state.

Table 1. Overall results after the completed thermomechanical processes for all three cases, including the achieved thickness reductions [%] and final hardness of the aluminum sheet [HV].

Case	Rolling temperature [°C]	Thickness reduction [%]			Final hardness of aluminum [HV]
		Total	Aluminum	Steel	
AA6082-AR	150	62	63	62	120
AA6082-AR	150	64	64	64	125
AA6082-AR	200	61	61	60	121
AA6082-AR	200	60	62	59	125
AA6082-AR	200	65	66	64	125
AA6082-W	150	63	64	62	128
AA6082-W	150	63	63	62	119
AA6082-W	150	62	64	61	119
AA6082-W	150	67	68	67	121
AA6082-W	150	57	57	56	124
AA6082-W	200	63	64	63	119
AA6082-W	200	66	66	65	120
AA6082-W	200	64	65	63	123
AA6082-T6	200	61	63	60	130
AA6082-T6	200	58	60	57	141
AA6082-T6	200	60	61	60	135

As observed from Table 1, the steel and aluminum specimens achieved an equal degree of thickness reduction in all the produced composites, independently of achieved thickness reduction, rolling temperature and the state of the aluminum during rolling. These observations are opposed to what has previously been reported in the literature [13, 14], where aluminum, has been reported to achieve the highest degree of thickness reduction. The engineering stress-strain curves after pre-heating the metals to 200 °C are presented in Fig 2(a). However, for the available tensile testing machine with heating chamber, these measurements were obtained in the lower range of the load cell's resolution, resulting in large steps on the curves with an accuracy of about ±10 MPa. The curves for AA6082-AR and AA6082-W at 150 °C were only slightly higher compared to 200 °C, and for the IF-steel the curves at 150 and 200 °C were quite similar. These curves are, hence, not included. During the pre-heating to the desired rolling temperatures, the metals become softer, achieving more similar flow-characteristics. This explains the equal plastic deformation behaviour, which leads to the even degree of thickness reduction observed in this study. The specimens were stacked together using aluminum rivets in both the front and rear end of the samples as shown in Fig. 2(b). The strength of these rivets are limited, however, they did not break during the rolling process and may thereby have contributed to the achieved similar thickness reductions between the two metals. As steel has been reported to be the layer achieving the lowest degree of thickness reduction [13, 14], one can assume that the steel is the limiting factor for the achieved bond strength when an uneven degree of thickness reduction is achieved, assuming the film theory is the main bonding mechanism. In the study by Buchner et al. [7], a higher bond strength was obtained after only pre-heating the steel prior to rolling. The individual thickness reductions of the metals are not reported in the study, but it is stated, that the pre-heating of the steel leads to a more similar flow characteristics between the steel and aluminum, resulting in a higher bond strength. This indicates that a more even degree of thickness reduction in the dissimilar metals, as achieved in this study, might be beneficial for the joint strength.

In all three cases, composites were attempted produced at both rolling temperatures (150 °C and 200 °C). However, in case AA6082-T6, it was not possible to produce sound joints with 150 °C as rolling temperature. Therefore, only samples produced with a rolling temperature of 200 °C could be studied. Due to the initial mechanical properties of the aluminum in these samples, sound joints could not be produced in composites with lower thickness reductions than 60%, which indicates that for case AA6082-T6, the threshold reduction is quite high, although it is decreasing with the rolling temperature, consistent with the literature [9, 11, 15]. Buchner et al. [7], concluded that due to the high strength of the aluminum when in T6-state, the formability decreases significantly and they suggested that in a production process, the composite should be formed prior to aging. The observations from this study support this conclusion, as the composites in case AA6082-W could be produced at a lower rolling temperature, had a lower threshold reduction and the aluminum was successfully aged to T6-state after rolling.

3.1. Interface characteristics

Comparisons are only made between composites from the three cases produced with a rolling temperature of 200 °C in this paper, as no clear differences in the interface characteristics could be observed at the two rolling temperatures for case AA6082-AR and AA6082-W. The composites in case AA6082-AR were solutionized and aged post-rolling, which resulted in an intermetallic phase layer forming along the steel-aluminum interface. A SEM-image of the interface is presented in Fig 3(a). The intermetallic phase layer is observed to be discontinuous with an uneven thickness, varying from 0-3 µm. The discontinuous growth behaviour is in line with the film theory, where bonds are formed in areas where fresh metal from each opposing metal surfaces meet [3, 4]. During the post-joining heat treatment, the intermetallic phases will first start to form in these bonded areas due to the direct contact between iron and aluminum. As the solutionizing was carried out for only a short time period, it was not sufficient for the intermetallic phase layer to become continuous. Similar observations were made in the study by Xu et al. [17] and Jindal & Srivastava [23], where discontinuous islands of intermetallic phases were initially formed, which merged into a continuous irregularly thick layer with time.

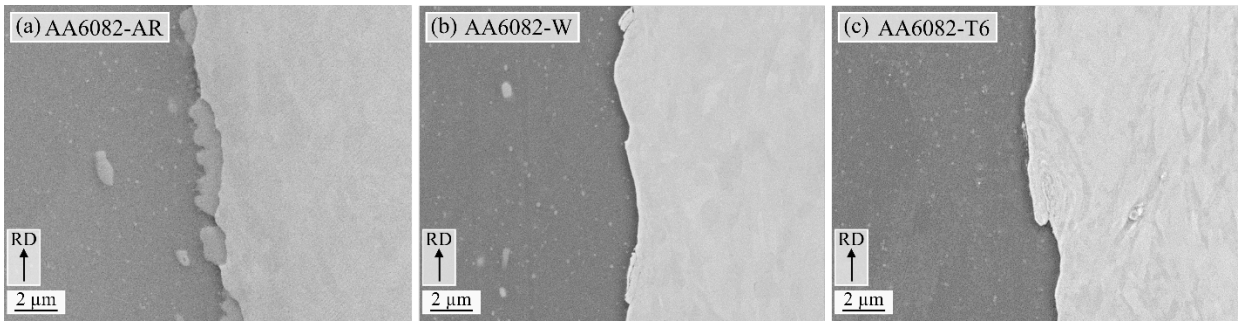

Fig. 3. SEM images of interface regions for composites joined at rolling temperature 200 °C. The aluminum is to the left, and steel to the right in the images. (a) shows the discontinuous intermetallic phase layer observed in the AA6082-AR composites. (b) and (c) shows the interface region in the AA6082-W and AA6082-T6 composites respectively, where steel fragments in various sizes and cracks in the steel matrix can be observed.

TEM analysis of the interface of the composite from case AA6082-AR was performed to characterize the observed intermetallic phases. A bright field TEM image of the interface region is presented in Fig 4(a). The intermetallic phase region consists of three distinct phases, and to determine these, selected area electron diffraction was used and a diffraction pattern from one grain in each layer are shown in Fig. 4(b)-(d). It was concluded that the intermetallic phase layer consists of Fe_2Al_5, Fe_4Al_{13} ($FeAl_3$), and cubic α-FeAlSi [24]. The phase with highest Fe-content is located closest to the IF-steel, consistent with the observation reported in the literature [16-18]. The same phase sequence has been observed in joints of steel and Si-containing aluminum [19, 25]. The interface characteristics of the composites produced in case AA6082-W and AA6082-T6 are very similar, as seen in Fig. 3(b) and (c). In both cases, steel fragments and cracks in the steel base metal can be observed along the interface and no intermetallic phases were observed in the SEM for both rolling temperatures.

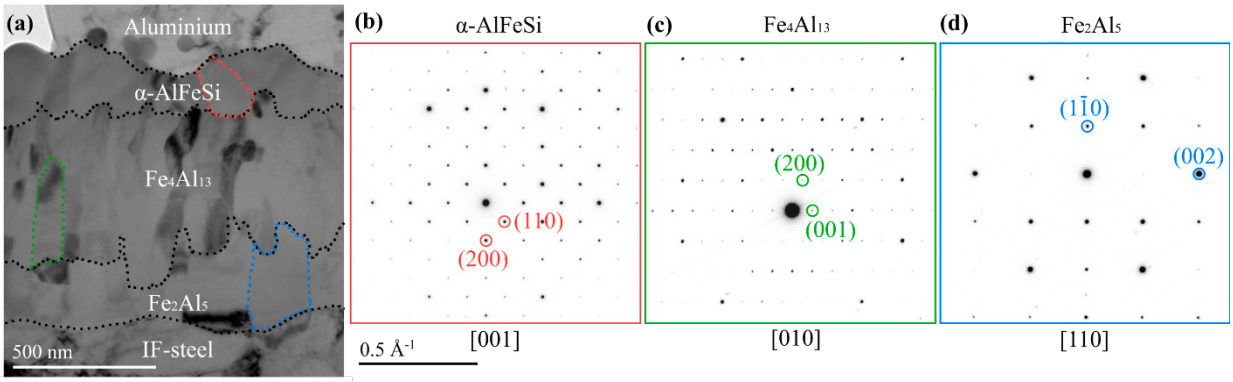

Fig. 4. (a) Bright field TEM image of interface region of the case AA6082-AR composite, where three distinct intermetallic phases are observed. The borders between the phases are marked with black dashed lines. The coloured dashed lines outline the grains from which the selected area electron diffraction patterns of the intermetallic phases (b) cubic α-AlFeSi, (c) Fe_2Al_5 and (d) Fe_4Al_{13}, were taken. The scale bar shown under (b) is valid for all the patterns.

3.2. Bond strength and fracture surface characteristics

The achieved bond strengths are presented in Fig. 5, and show that the sequence of the thermomechanical process influences the final bond strengths. Overall, a slight increase in bond strength with increasing thickness reduction can be observed, consistent with the literature [4, 9-12]. However, no clear increase in bond strength with increasing rolling temperature can be observed. For case AA6082-W, a large spread in the obtained bond strengths can be observed and the average highest bond strength is found in case AA6082-T6. Overall, the bond strengths obtained in this study are low compared to other studies on cold roll bonded steel-aluminum [6, 8, 9, 12]. The surface preparation is very important to obtain a good bond strength and in this study, the surface preparation prior to rolling was performed manually with a steel-wire brush. As observed from the interface images in Fig. 3(b) and (c), steel particles of various sizes can be observed along the interfaces. During peel-testing, these fragments break off from the steel matrix and are seen embedded in the aluminum in the aluminum fracture surface. This can be observed in the EDS-mapped images of the aluminum fracture surfaces, for case AA6082-W and AA6082-T6 in Fig. 6(e) and (d), respectively. Similar observations have been made where steel wire brushes were used [6, 8]. Wang et al. [6] concluded that the steel fragments along the interface were the reason for the low bond strength due to the large difference in mechanical properties between the steel matrix and the fragments.

A clear difference in the characteristics of the residual aluminum on the steel fracture surfaces from the three cases can be seen in Fig. 6(a)-(c). These differences are in line with the obtained bond strengths, as well as the interface characteristics shown in Fig. 3(a)-(c). For case AA6082-AR, the intermetallic phases grew discontinuously along the interface. This is reflected by the appearance of the steel fracture surface, Fig. 6(a), where the residual aluminum covers large areas. However, there are areas in between where no residual aluminum is observed, indicating no intermetallic phases formation that area. In the study by Buchner et al. [7], on cold roll bonded IF steel and 6xxx aluminium, a thin intermetallic layer had no influence on bond strength. In this study, the intermetallic phase layer is below what has been reported as the critical thickness (5-10 µm [20, 21]) and should therefore not influence the bond strength. Fig. 5, however, shows that the composites from case AA6082-AR have the lowest bond strength that is similar for both rolling temperatures, indicating that the intermetallic phase layer has influenced the bond strength.

In the study by Akramifard et al. [12], the intermetallic layer is only present on the steel fracture surface after debonding, indicating that the crack has occurred along the interface between aluminum and the intermetallic phase layer. However, in this study, large areas with residual iron can be observed on the aluminum fracture surface in Fig 6(d). Since significant amounts of residual aluminum and iron are detected on both fracture surfaces, it indicates that the fracture has occurred inside the intermetallic phase layer, likely between two of the different phases present in the layer. On the steel fracture surface for case AA6082-W and AA6082-T6, the residual aluminum has a band-like shape, perpendicular to the rolling direction and a dimple-like structure, indicating that a ductile fracture has occurred in the aluminum base metal during peel-testing. These band-like structures perpendicular to the rolling direction are

consistent with the plane strain deformation mode during rolling, where the surface expansion only occurs in the rolling direction [13], only allowing cracks to develop perpendicular to the rolling direction. Similar fracture surface characteristics have been observed in other studies [8, 15].

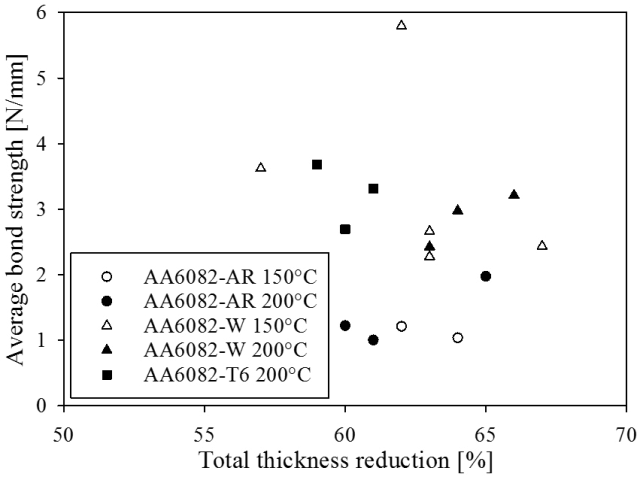

Fig. 5. Average bond strength [N/mm] as a function of the total reduction in thickness [%] for the three different cases; AA6082-AR, AA6082-W and AA6082-T6 for rolling temperatures 150 °C and 200 °C.

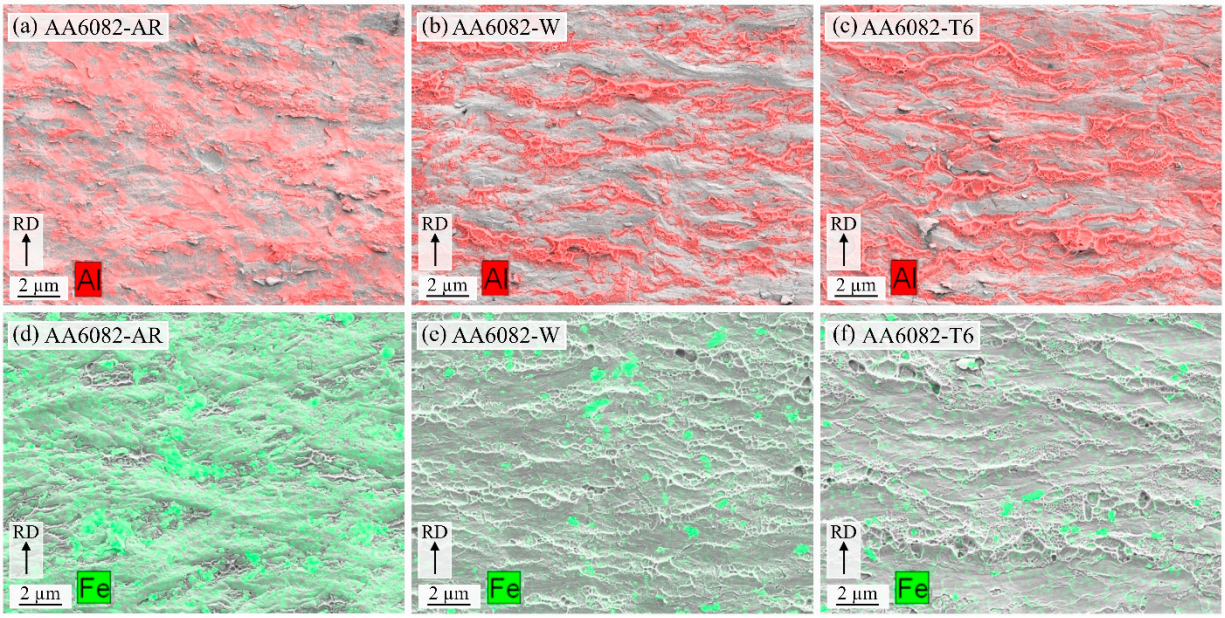

Fig. 6. Fracture surfaces after peel-testing of the three cases AA6082-AR, AA6082-W, AA6082-T6 with rolling temperature 200 °C. (a)-(c): SEM images with corresponding EDS color maps overlaid on the steel fracture surface. Red areas indicate residual Al. (d)-(f): SEM images with corresponding EDS color maps overlaid of the aluminum fracture surface. Green areas indicate residual Fe.

4. Conclusions

1. Equal degrees of thickness reductions were achieved in the steel and aluminium layers in the composites, independent on rolling temperature and the state of aluminium during rolling.
2. The sequence of the thermomechanical processing was found to influence the overall bond strength. The average highest bond strength was obtained in case AA6082-T6. However, large variation was observed in the obtained

bond strength between the composites, especially in case AA6082-W.

3. The results suggest that case AA6082-W is the most optimal process route, as composites can be produced at lower rolling temperatures and have a lower threshold reduction compared to case AA6082-T6.
4. A discontinuous and uneven intermetallic phase layer (0-3 µm) was formed in the composites produced in case AA6082-AR. The intermetallic layer consisted of Fe_2Al_5, Fe_4Al_{13} ($FeAl_3$) and α-AlFeSi, and is assumed to be the main reason for the low bond strength in these composites. The fracture has most likely occurred in the intermetallic phase layer, as equal amounts of residual Fe and Al were found on the opposing fracture surfaces.

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