

Article

Microstructure, Mechanical and Thermal Properties of Al/Cu/SiC Laminated Composites, Fabricated by the ARB and CARB Processes

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Abstract: The aim of the current work is to investigate the effect of SiC particle weight percent and rolling passes on Al/Cu/SiC laminated composites, fabricated by accumulative roll-bonding (ARB) and cross-accumulative roll-bonding (CARB) processes. The optical microscopy (OM) images of composites revealed that despite the good bonding of the layers, they underwent plastic instabilities as a consequence of strain hardening of the layers. However, these instabilities occurred more in ARBed composites than in composites fabricated by the CARB process. This is because in the latter process, the composites are rolled in two directions, which leads to better strain distribution. Furthermore, with an increase in passes, SiC particles were well distributed in the matrix and interfaces. The mechanical findings showed that, by increasing passes, there was a growth in the values of strengths and elongation. This behavior is believed to be related to increased work-hardening of layers, better distribution of reinforcing particles, and an enhanced bonding of interfaces at higher rolling passes. In addition, the results of thermal conductivities showed a downward trend with an increase in passes; in fact, the increased number of Al/Cu interfaces declined the heat conduction of composites.

Keywords: laminated composite; interface; mechanical properties; thermal conductivity; SiC particles



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1. Introduction

In recent decades, there has been a rapid rise in the use of dissimilar materials due to their remarkable features, namely low weight, and high strength. Particle-reinforced composites, such as Al/SiC and Cu/SiC, are good examples, usually fabricated for specific features. The former composite has a great potential to be used in aerospace applications because of its low density, high thermal conductivity, and Young's modulus. The excellent thermal and electrical properties of the latter make this composite a good candidate for electrical packaging applications [1,2]. One of these materials is the laminated composite, in which different metals are used, and its features, thus, enable them to have applications in the field of airplanes and automobiles [3–6]. Over the last two decades, many of these composites have been fabricated, namely Ti/Al [7], Ti/Al/Steel [8], Ni/Ti [9], and Al/Cu [10]. However, in addition to metallic reinforcement, ceramic powders, including SiC [11], B₄C [12,13], and TiC, have been added to the matrix. Since these powders have a high elastic modulus and thermal conductivity, they are often used to enhance the mechanical and thermal properties of metallic composites [11]. Considering the fact that the mechanical properties of the composite are mainly controlled by the volume fraction of reinforcement, and particularly the interfacial bonding between the matrix and reinforcement, choosing the appropriate manufacturing technique is vital. Laminated

composites have been produced through several techniques, including diffusion bonding [7], magnetron sputtering [14], and severe plastic deformation (SPD) [11]. By using the SPD methods [3,5,6,15,16], the ultrafine and porosity-free microstructure can be achieved. Moreover, when powder particles are used, a better distribution across the matrix can be obtained.

Among SPD methods, accumulative roll-bonding (ARB) and cross-accumulative roll-bonding (CARB) have recently attracted many researchers [17–20]. In both methods, the thickness reduction of sandwiches in each rolling pass is kept stable while more strains are applied to each layer. However, in the CARB process, after each pass, the rolling direction of the sandwiches is rotated about 90 degrees, through which a better distribution of reinforcement can be obtained [8].

Generally, the mechanical properties of laminated composites fabricated by the mentioned processes strongly depend on the quality of the bonding between the matrix and reinforcement, distribution and instability of reinforcement [21,22]. For example, in the Al1050/Al5052/SiC composite, where SiC particles were well distributed by increasing the number of passes, its strength grew, whereas, in Cu/Al/Ni/SiC, the strength values decreased on the account of the cluster of SiC particles, which may have a negative impact [23].

Up to now, there are no scientific reports on the differences between laminated composites produced by ARB and CARB processes. Moreover, SiC particles are nowadays widely used as reinforcements in all types of composites with polymer or metal matrix. Nonetheless, there is still little research on the effect of particle size in metallic laminated composites. Hence, in this research, the effect of particle percentage on the microstructure and mechanical and thermal properties in Al/Cu/SiC composites produced by the ARB and CARB processes at different rolling passes were investigated.

2. Materials and Methods

2.1. Materials

In this research, pure sheets (99.9% purity) of aluminum (0.5 mm thickness) and copper (0.5 mm thickness) were supplied by Sahand Company (Tabriz, Iran). The same dimensions of 18 cm × 8 cm were used. SiC particles, with a size of 5 µm, were also used [23]. The mechanical properties and thermal conductivity of the used materials have been highlighted in Table 1.

Table 1. The ultimate strength and thermal conductivity of the sheets used in this work.

Material	Cu	Al	SiC
Ultimate strength (MPa)	132	98	-
Thermal conductivity (W/m·K)	380	210	490

2.2. Surface Preparation and Stacking of Sandwiches

Surface preparation consists of scratching the layers by steel brushing and then cleaning them in acetone. Sandwiches of Al/Al/Cu/Al/Al/Cu/Al/Al were stacked alternatively. Next, 2 w% and 3 w% of SiC particles were sprayed at interfaces, and it should be noted that they were sprayed at Al/Al interfaces twice as much as at Al/Cu interfaces, so that the final sandwiches of Al/Cu/2w%SiC and Al/Cu/3w%SiC were prepared.

2.3. ARB Process

In this process, by using GCr15 roll mills (50-ton load capacity, 50 cm diameter, and 10 r/min rotation speed) without any lubricant, both sandwiches of Al/Cu/2w%SiC and Al/Cu/3w%SiC were the first rolls bonded separately, to a 66% thickness reduction. After the first pass, each sandwich was cut in half and surface-prepared. Afterward, the sandwiches were rolled and bonded to a 50% thickness reduction. Following that, the

layers' surface preparation and roll bonding were repeated eight more times. Finally, eight passes of Al/Cu/2w%SiC and Al/Cu/3w%SiC composites were achieved.

2.4. CARB Process

In this process, the composites of Al/Cu/2w%SiC and Al/Cu/3w%SiC were fabricated using the conditions and rolling parameters described above; however, the only difference is that, after each pass, the sandwiches were 90° rotated. This rotation in the CARB process is always executed with the purpose of achieving a better distribution of reinforcement. Additionally, an equal number of passes were applied through the processes described above.

Both processes were carried out at room temperature, as displayed in Figure 1. However, since the composites are subject to rolling mills, the temperature will probably rise due to friction. The second, fifth and eighth pass temperatures were about 40 °C, 90 °C, and 125 °C, respectively.

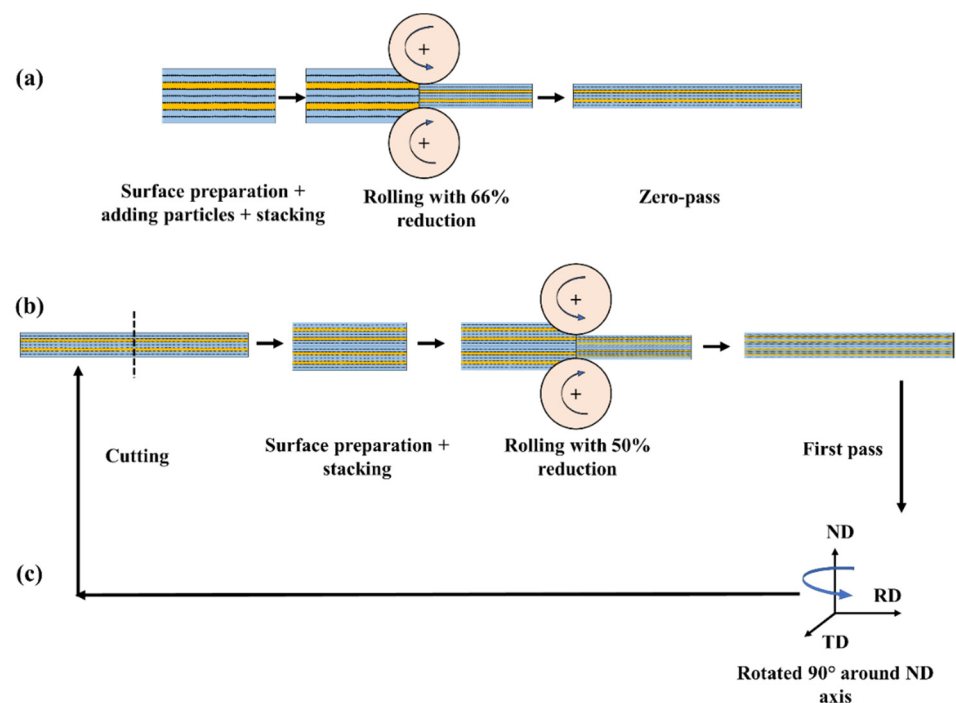


Figure 1. Schematic of (a) zero pass, (b) ARB and (c) CARB processes.

2.5. Microhardness Test

Vickers microhardness tests, with a load of 0.5 N for 10 s, were carried out at different points of the surface on the ND–RD plane of Al/Cu/2w%SiC and Al/Cu/3w%SiC samples at different passes of ARB and CARB processes. During this test, five indents were performed for each composite after each pass. Then, the average values were reported. Moreover, since the layers were completely separated and distributed. Each Vickers indent was carried out on each fragment. They were spaced about 100 µm away from each other in both X and Y directions.

2.6. Tensile Tests

To investigate the tensile behavior of Al/Cu/2w%SiC and Al/Cu/3w%SiC composites produced by ARB and CARB processes, after wire cutting the samples based on the JIS Z2201 standard, a uniaxial tensile test was performed on each sample in the direction of rolling (RD), at an $8.5 \times 10^{-4} \text{ s}^{-1}$ strain rate at ambient temperature, by using an Instron tensile test machine, equipped with an extensometer by which force displacement curves were measured.

2.7. Thermal Properties

Thermal conductivity (K) of the Al/Cu/2w%SiC and Al/Cu/3w%SiC composites was obtained by measuring the coefficient of heat diffusion. These tests were carried out on composites processed by ARB and CARB, according to ASTM E1461 by the XFA500 apparatus.

2.8. Microstructural Investigation

In order to investigate the grain sizes of initial sheets and composites, the surfaces of samples were ground on an ND–RD plane, polished and electro-chemically etched. The electrochemical etching was carried out with the Adak apparatus in Barker solution at 20 V for 120 s. In addition, the current in the first 10 s was 0.08 A and then changed to 0.02 A. However, the samples were just ground and polished to examine the layers, interfaces and distribution of particles. Finally, the microstructures of each composite were characterized using Olympus PMG3™ optical microscopy and a Cam Scan MV-2300™ scanning electron microscope (20 KV-BSE detector), to study the bonding of the layers and the distribution of SiC particles at different rolling passes of the ARB and CARB processes.

3. Results and Discussion

3.1. Microstructural Characterization

Figure 2 shows the light micrographs of the initial Al and Cu sheets. It is extensively reported that, with an increase in passes during SPD processes, the grains become elongated in the rolling direction, and at higher passes, grain refinement usually takes place [24,25].

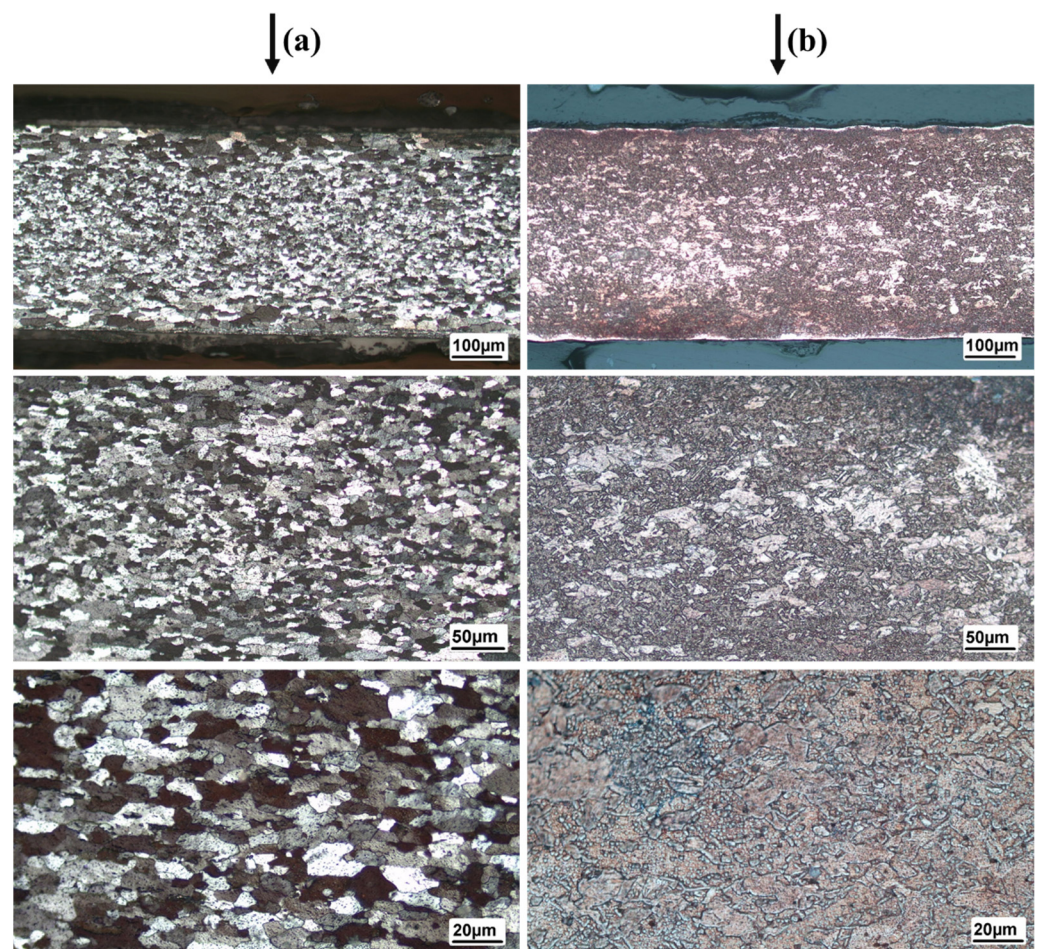


Figure 2. OM images of initial sheets of: (a) Al and (b) Cu.

Figure 3 displays the microstructures of Al/Cu/SiC composites, fabricated by ARB and CARB processes. Regarding all four composites, despite the good bonding between the layers, reinforcing layers underwent instability during plastic deformation, which appeared most after the initial pass. These instabilities consist of necking and fracturing, and are usually attributed to different plastic flow and mechanical properties of Al and Cu layers. The instabilities, thus, occurred in the harder layer. Figures 3–5 show that Cu layers, which have a higher hard-working coefficient than Al, represented necking in the second pass, but were still continuous. From the third pass onward, rupture took place in Cu layers. In other words, at a higher amount of strain, the fracture of the layers is noticeable. These findings agree with previously published papers [8–11].

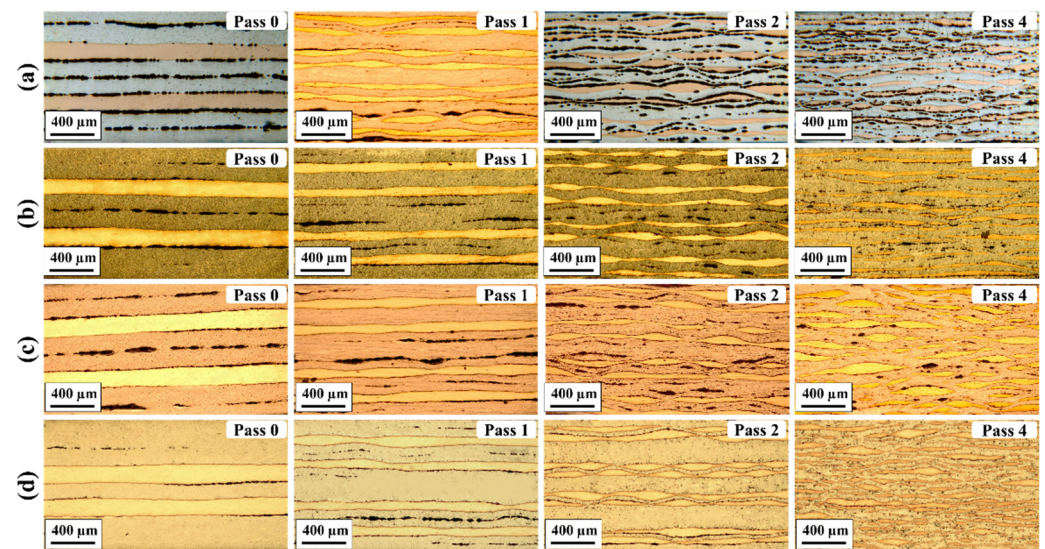


Figure 3. OM images of laminated composites: (a) ARBed-Al/Cu/3wt%SiC, (b) CARBed-Al/Cu/3wt%SiC, (c) ARBed-Al/Cu/2wt%SiC, and (d) CARBed-Al/Cu/2wt%SiC.

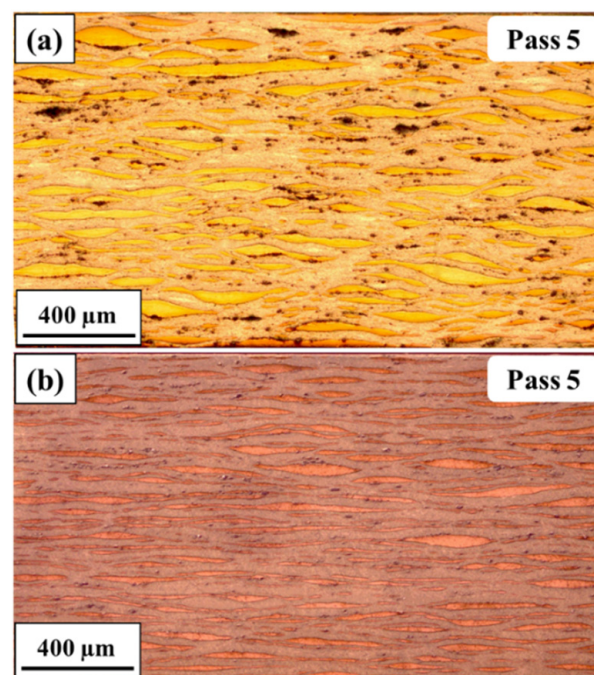


Figure 4. OM images of laminated composites at the fifth pass: (a) ARBed-Al/Cu/3wt%SiC, (b) CARBed-Al/Cu/3wt%SiC.

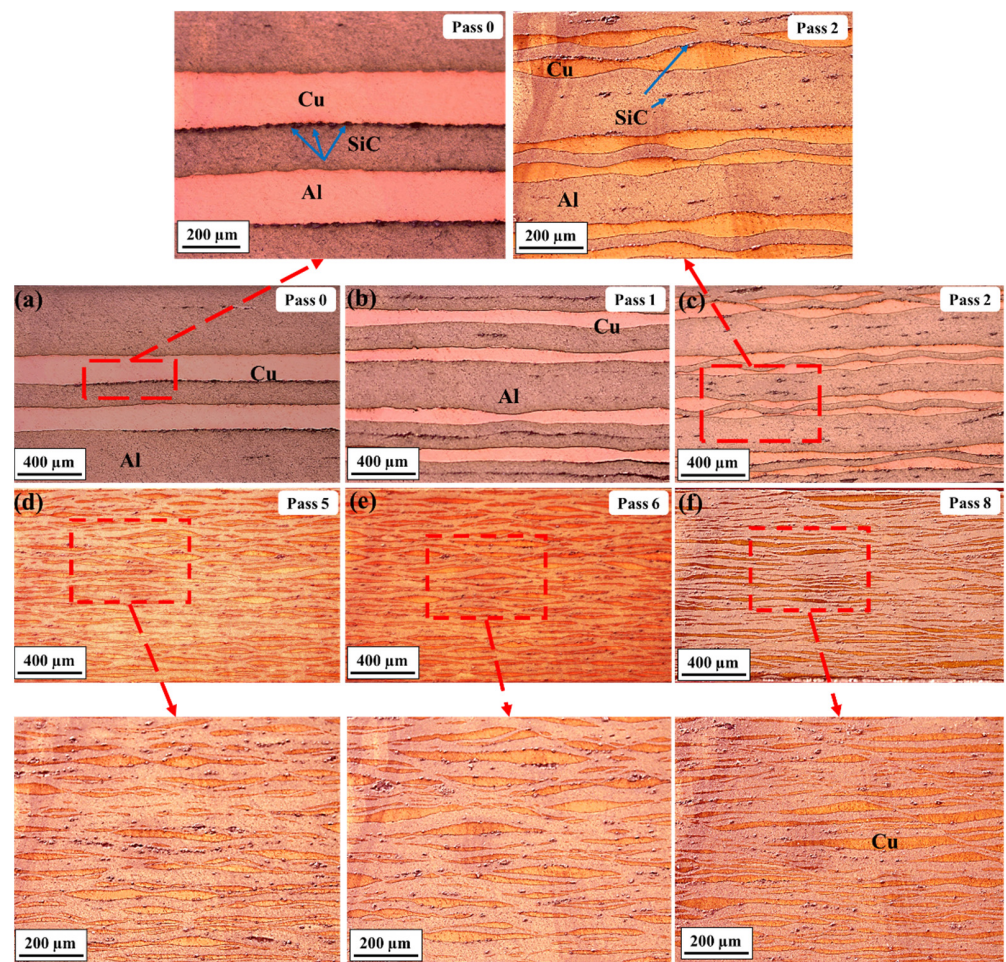


Figure 5. SiC particle distribution at different passes of the CARBed Al/Cu/2wt%SiC composite.

Comparing the microstructure of an ARBed composite with that of a CARBed composite indicates that more instabilities may exist in the former on the account of the non-uniform distribution of strain during this process, while in the CARB process, as was mentioned latter. The rolling direction changes after each pass, which may lead to a better strain distribution [8,26]. This, in turn, brings about a better distribution of reinforcing layers within the matrix, as shown in Figures 3–5.

By increasing the number of passes, the number of layers within the composites increased, enabling the better distribution of SiC particles. As a result, as can be seen in Figures 3–5, the SiC powders were distributed in two different regions: (a) in the Al matrix and (b) at the Al/Cu interface.

In some parts of the composites, where there are agglomerated particles, the distance between them represents porosities that cause weak contact at the interfaces. The size of porosities varies depending on the size of clusters and the distance between the particles in them. However, during severe plastic deformation, the motion of the particles can break the oxide on the aluminum layer, which has a negative impact on aluminum/aluminum and aluminum/copper bonding, thus enhancing the quality of bonding at higher passes. It is believed that the breakage of clusters, and also particles, are mainly caused by triaxial stresses during shear deformation [27–29]. Nonetheless, another reason that has been reported to contribute to this breakage is the residual plastic strain which arises from the different coefficients of the thermal expansion, pertaining to the composite constituents [27–29]. As mentioned earlier, the measured temperatures showed that, during severe plastic deformation, the friction between the surface of the samples and rolling mills

increased the temperature. This increased temperature is likely to cause residual plastic strain, which in turn results in cluster breakage.

As mentioned earlier, since the powders were sprayed at Al/Al and Al/Cu interfaces, the layers of powders exist. These layers can be seen in the first three passes of ARBed composites and the first two passes of CARBed composites (Figure 3). This difference indicates that a more uniformly applied strain can lead to a sooner occurrence of powder fragments. On the other hand, several mechanisms are responsible for the layer-to-fragment changes. The shear flow during the rolling process at each pass is likely to cause the flow of the matrix to pass the fragment. Moreover, powders in fragments at Al/Al and Al/Cu interfaces are divided into two types: powders close to their interfaces with the matrix, and powders in the middle of the fragments. The former are susceptible to moving on account of the shearing effect of the matrix at the fragment/matrix interfaces. Therefore, with an increase in passes, the size and shape of the accumulated powders may change. Further, as more matrix flows around the fragments of powders at higher passes, a smaller size of cluster, and thus more uniform distribution, can be achieved. From Figure 4, it can be seen that SiC powders and reinforcing layers have been well distributed across the microstructure, even though better distribution without fewer clusters were achieved in a composite processed by the CARB process (Figure 4b).

Figure 5 displays the distribution of these particles in different passes, but Figure 6 shows a closer examination of them in both regions. The clusters of particles in aluminum and at the aluminum/copper interface are evident. Although these clusters have different sizes, they are larger than 2 μm . They also have porosities which in turn can be stress concentration sites [30]. However, at higher passes, these clusters are broken into smaller ones when the virgin metal matrix between the particles is extruded as a result of plastic strain. Comparing Figure 5b,c reveals that continuous layers of particles (the second pass) were separated into clusters (the third pass) in the Al matrix. The formation of the cluster is usually due to having a high surface-to-volume ratio, as well as high surface energy. The separation and distribution of particles trapped in the clusters depend on applying strain. In addition, from Figure 6a, it can be seen that the size of the clusters on both sides of the copper is different even though, as mentioned earlier, the same content of particles was sprayed at Al/Cu interfaces. This phenomenon is probably attributed to the distribution of strain from one surface to the other one; in fact, when no lubricant is used for roll-bonding, the surfaces might experience severe strain. The amount of strain in each pass declines from one surface to the middle of the sample and, again, increases to the other surface. Hence, for this reason, the distribution of the particles around the copper layers is not even, as shown in Figure 6. The particles close to the surfaces of the samples are subjected to more strain, which causes more movement of the particles, thus reducing the size of clusters. It means that applying rolling passes can lead to the breakage of the clusters and a better distribution of the particles. Despite this phenomenon, the cluster of particles is still present at higher passes. Accordingly, the number of imperfections decreased [8].

As reported by Naseri et al. [28], the dispersed particles may lead to an accelerated grain refinement at higher passes, which, based on the Hall–Petch relationship, increases the strength of the composites. Ceramic particles are generally believed to form clusters because of having Van der Waals interactions. Nevertheless, the distance between the particles increases as more strain is employed during the ARB and CARB processes.

As reported by many researchers [9,25], the formation of shear bands is typical of severely plastic-deformed composites. These bands, with angles of around 35–45° with respect to rolling direction (RD), are evident in Figure 7. It is also believed that these bands take place as a result of non-uniform strain during SPD processes. In fact, because of the difference in the work-hardening behavior of Al and Cu, which causes localized shear banding, the rupture of the layers often occurs. It is believed that this is a response to the flow instability of the composite constituents [25].

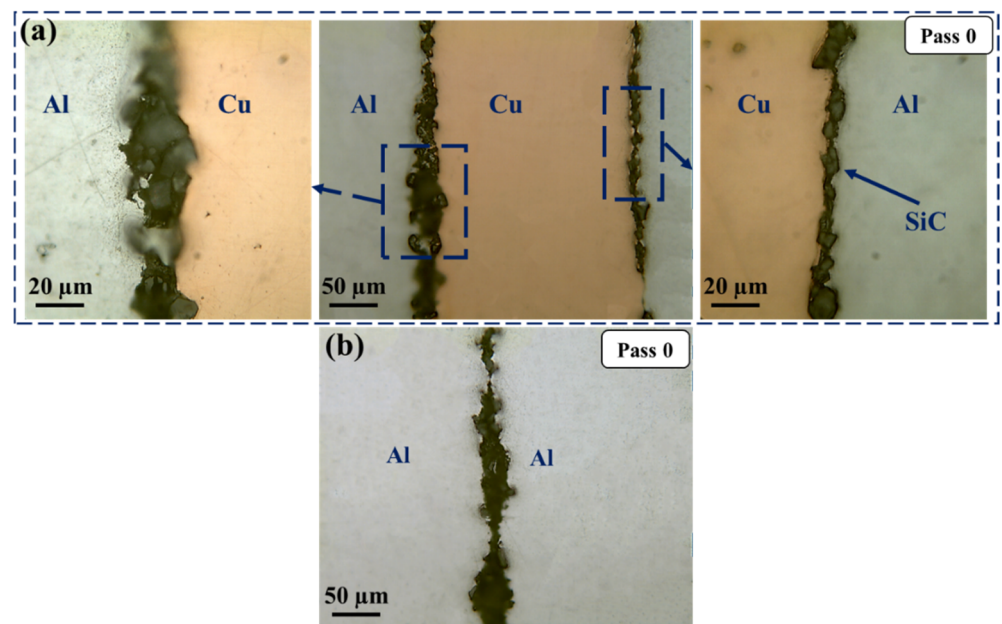


Figure 6. The cluster of SiC particles: (a) in Al/Cu and (b) at Al/Al interfaces.

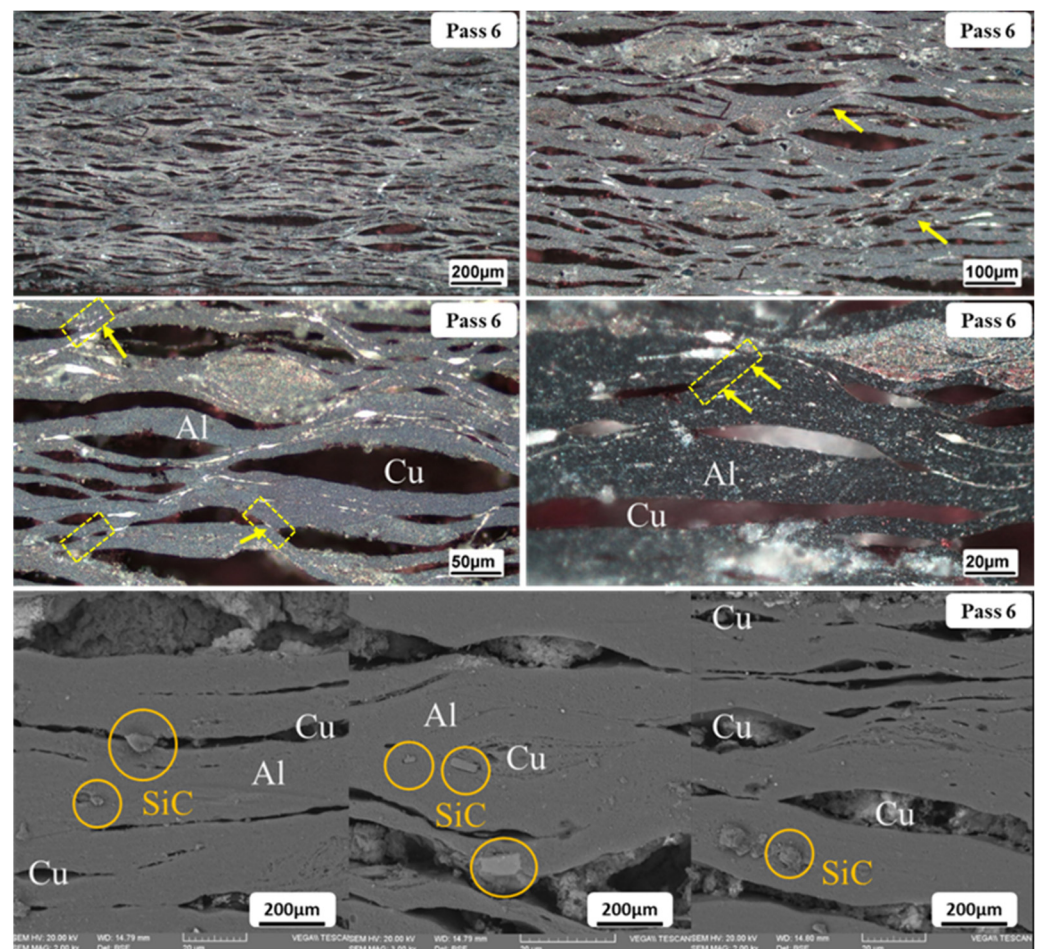


Figure 7. SEM images taken from different regions of the sixth pass of ARBed Al/Cu/3wt%SiC composite.

3.2. Mechanical Properties

3.2.1. Microhardness Test

Figure 8 shows how the hardness value of aluminum and copper layers vary from pass 0 to pass 8. Evidently, by increasing the passes, the hardness of the layers grew. Each layer was work-hardened due to severe plastic deformation. There was a sharp growth from pass zero to the first pass, but it increased modestly from the third pass. This is due to the fact that, in ARB and CARB processes, the rate of work-hardening of the first two passes is higher than the other ones. In fact, in the first passes of ARB and CARB, a high density of dislocations leads to severe work-hardening. Additionally, in higher passes, dynamic recovery is believed to be an essential factor in controlling the gradual increase [11]. During each rolling pass, the temperature increases. Therefore, it is expected that final passes experience a higher temperature value. The amount of generated heat in the next passes depends on the previous passes' cold work. The used energy is mainly dissipated through heat formation and increased temperature. Therefore, the low-temperature dynamic recovery is highly likely to happen during SPD processes, especially at higher passes. Further, this phenomenon, depending on the melting points of the constituents, usually occurs in the aluminum matrix. In fact, with an increase in passes, the density of dislocations rises, while the low-temperature dynamic recovery takes place. Due to this reason, there is not a sharp rise in the hardness values at higher passes. The noticeable increase occurred in the first two passes, which is attributed to strain hardening, whereas grain refinement at large grains does not increase the hardness values much [24]. However, the hardness values of each composite, produced by the ARB and CARB processes, are similar on the account of subjecting to the same amount of strain. Nevertheless, the hardness values of aluminum and copper layers were higher in Al/Cu/2wtSiC than Al/Cu/3wtSiC, indicating a better strain distribution.

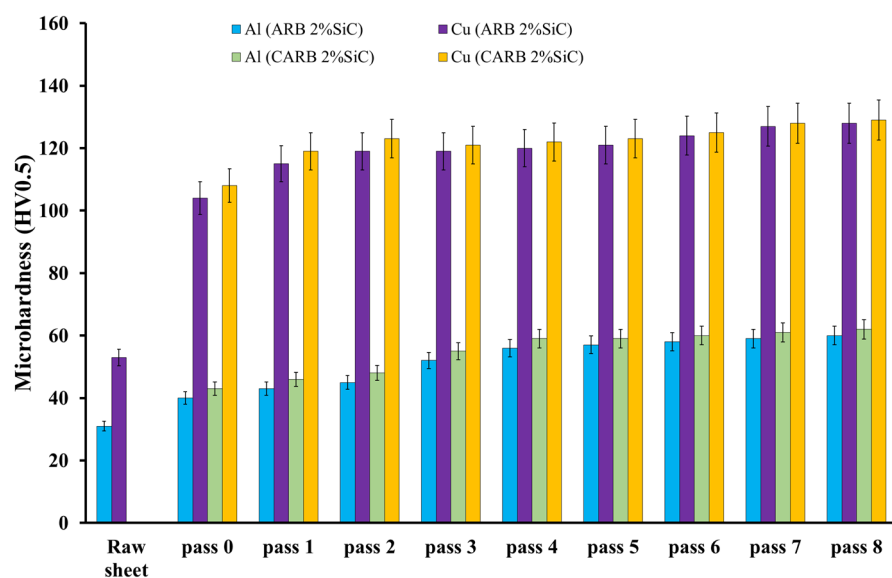


Figure 8. The variation in the hardness value of aluminum and copper layers at different passes of ARB and CARB processes.

3.2.2. Tensile Properties

Figure 9 shows the engineering stress–strain curves of Al/Cu/2wt%SiC and Al/Cu/3wt%SiC composites, fabricated by the ARB and CARB processes. Different trends are evident in these curves, in that the values of strengths and elongation, concerning four types of composites, first declined and increased gradually. As was mentioned earlier, the strength and elongation values dropped in the second pass, which might be due to the weak bonding of interfaces of particle/layers and

layer/layer. Moreover, the existence of a cluster with many porosities may weaken the strength of the composite. It is well known that the main reasons for the upward behavior of strength values are related to the work-hardening of the layers, reinforcing particles, interfaces, and thermal expansion of layers [11]. Work-hardening of the layers during severe plastic deformation is the dominant factor in increasing the strength of composites.

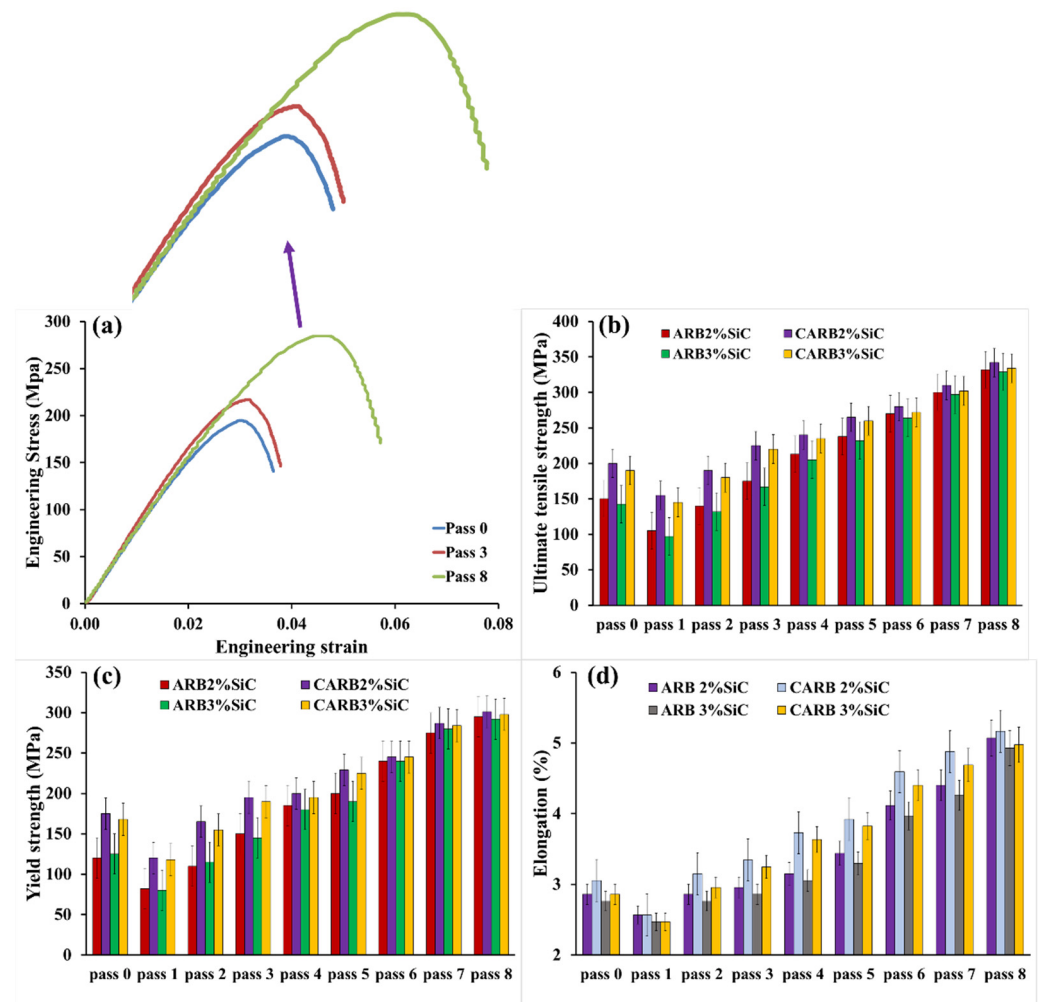


Figure 9. Mechanical properties of composites at different passes: (a) stress–strain curves of CARBed Al/Cu/2wt%SiC, (b) ultimate tensile strength and (c) yield strength, (d) elongation.

Furthermore, depending on Cu and particle distribution, the values of elongation (elongation after fracture and total elongation at fracture) and strength vary. Interestingly, in both ARB and CARB processes, by increasing the rolling passes, a better distribution of the particles can be obtained, which subsequently leads to enhanced strength. However, since higher volume fractions of particles tend to form clusters with more porosities, Al/Cu/3%SiC composites are likely to have a lower strength and elongation than Al/Cu/2%SiC composites. Regarding composites with reinforcing particles when produced by ARB and CARB, by increasing rolling passes, a better distribution of SiC particles and Cu layers led to increased values of strength and elongation [11,31,32]; however, the difference between values in composites with different weight percent of particles may be attributed to the increased number of interfaces at higher passes, in which the mentioned cluster may have a detrimental effect on interfaces. In fact, nucleated cracks during the tensile test can move easier through the cluster of particles and then lead to a faster failure of the composite, which causes a declined strength and elongation.

As can be seen in Figure 9a, with an increase in passes, the stress–strain curves show more discontinuities, such as load drops and jerkiness, during the tensile test. The serrated flow of stress has recently attracted much attention, because it affects the mechanical behavior of a material. It is believed that experimental conditions of mechanical tests, such as temperature, strain, and strain rate, probably have an impact on the serrated flow. It has been reported that serrations decreased with an increase in the test temperature and increased with an increase in the strain rate. However, in the current work, all these conditions were constant during tensile tests. A couple of mechanisms are attributed to serrations in ductile materials, namely, phase transformations induced by stress and strain, order–disorder phase transformations, and particularly the dislocation pinning by solute atoms. When the motion of interstitial and substitutional solutes is decelerated near the dislocation cores, the dislocations are pinned by strain fields. After that, by increasing the stress, the dislocation migration occurs until the next pinning. Therefore, serrations (or the serrated flow) can occur during a material's dynamic strain aging (DSA) while undergoing mechanical testing. As shown in Figure 7, shear bands occurred in the sample microstructure at higher passes. This phenomenon indicates a relatively non-uniform deformation during ARB and CARB, in which the temperature of composites may increase on the account of frictional stress between the rolling mill and deforming material. It usually appears at higher rolling passes, where the composites undergo much accumulative deformation. Moreover, the temperature rise in each pass impacts the hot-workability of the composites. Therefore, the dynamic strain aging (DSA) process is expected to occur for these reasons, resulting in the Portevin-le Chatelier effect, appearing as a serrated flow [33,34].

As mentioned earlier, during the CARB and ARB processes, the repeatedly applied plastic deformation results in a better distribution of particles by eliminating the layers of particles and reducing the size of the clusters. Furthermore, with an increase in passes, these changes improve the metal/metal bonding and subsequently increase the strength and elongation of composites. This improvement was more noticeable in CARBed composites than in ARBed composites, due to having more uniformly applied strain.

In addition to the effect of shear deformation on the work-hardening of the layers, it is also believed that the interaction between the particles can lead to the more significant work-hardening of composites [29]. However, particles positively impact the yield strength by hindering the movement of dislocations and making Orowan loops. Indeed, dislocation can pass particles and leave these loops behind during further deformation, thus enhancing the yield strength [27,29]. At higher passes, when the distance between the particles becomes shorter, as a result of the Orowan mechanism, more stress is required for the further movement of the dislocations through the particles. Nevertheless, the comparison of fabricated composites with a different content of SiC particles reveals that more significant clusters are likely to exist in Al/Cu/3wt% SiC than in Al/Cu/2wt% SiC. Thus, more cracks can be detected in these imperfections, and mechanical properties are expected to decline.

Figures 10 and 11 illustrate the fracture surfaces of Al/Cu/2wt%SiC and Al/Cu/3wt%SiC composites. Due to the ductile nature of aluminum and copper layers, many voids can be seen on all four types of composites. Nonetheless, the river-like pattern on some regions of the copper layers represents a shear mode of fracture, which is often caused by its low ductility. Another feature of these surfaces is delamination, which is usually observed in laminated composites because of the different mechanical characteristics of its constituents. Delamination often occurs during the uniaxial tensile test, when the necking of the layers takes place [8,23,26]. However, in these composites where SiC particles are present at the interfaces, during tension, cracks can easily nucleate and propagate through them [35,36]. As was mentioned earlier, even at higher rolling passes, where a better bonding of interfaces and distribution of ceramic particles can be achieved, the distance between layers because of these particles is susceptible to cracks, which in turn decreases the elongation of composites. More importantly, the distance between the layers, where a cluster of particles exists, increases the probability of cracking [11,23].

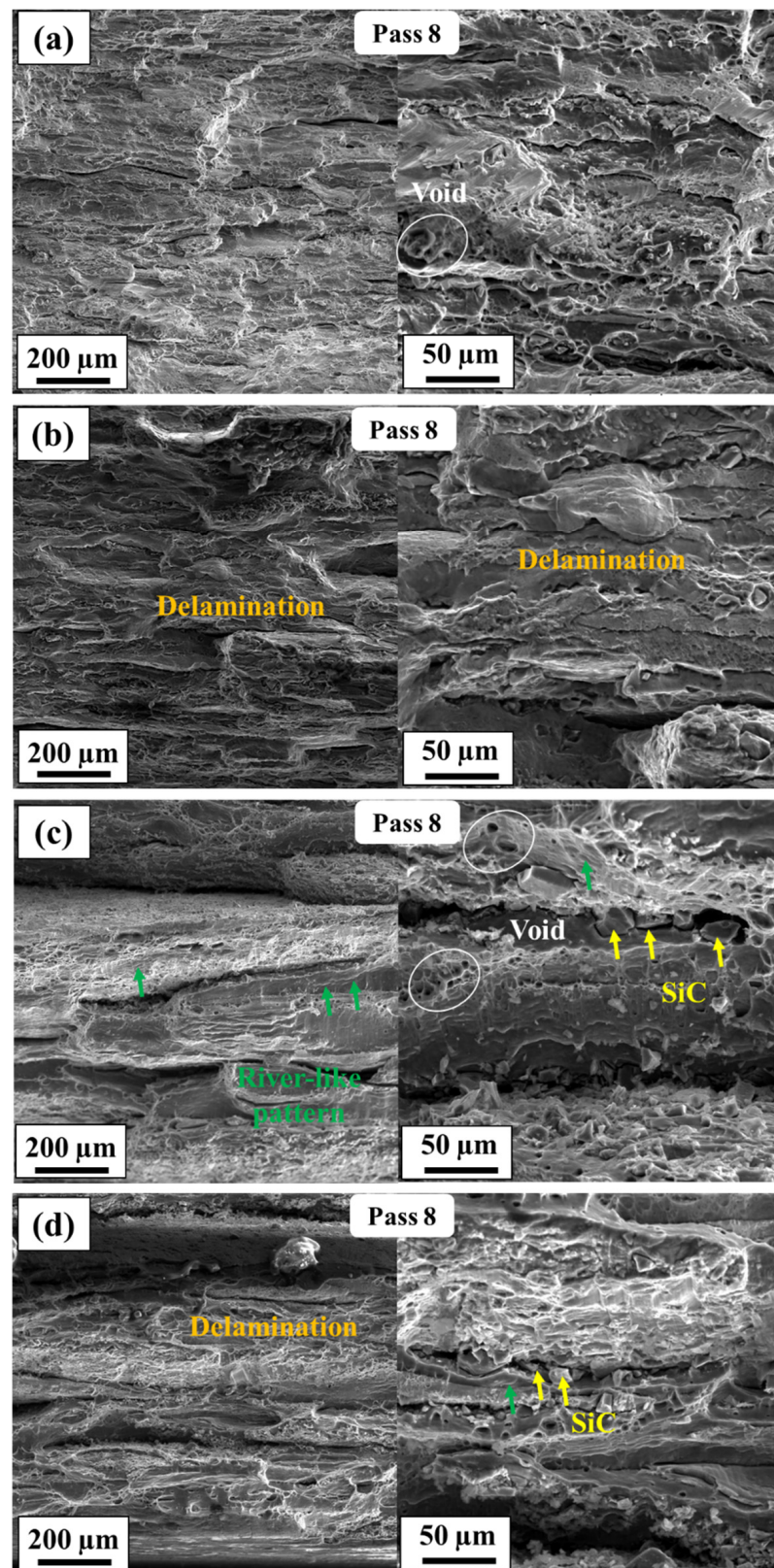


Figure 10. Fracture surfaces of composites fabricated by: (a) ARBed Al/Cu/2wt%SiC, (b) CARBed Al/Cu/2wt%SiC, (c) ARBed Al/Cu/3wt%SiC and (d) CARBed Al/Cu/3wt%SiC.

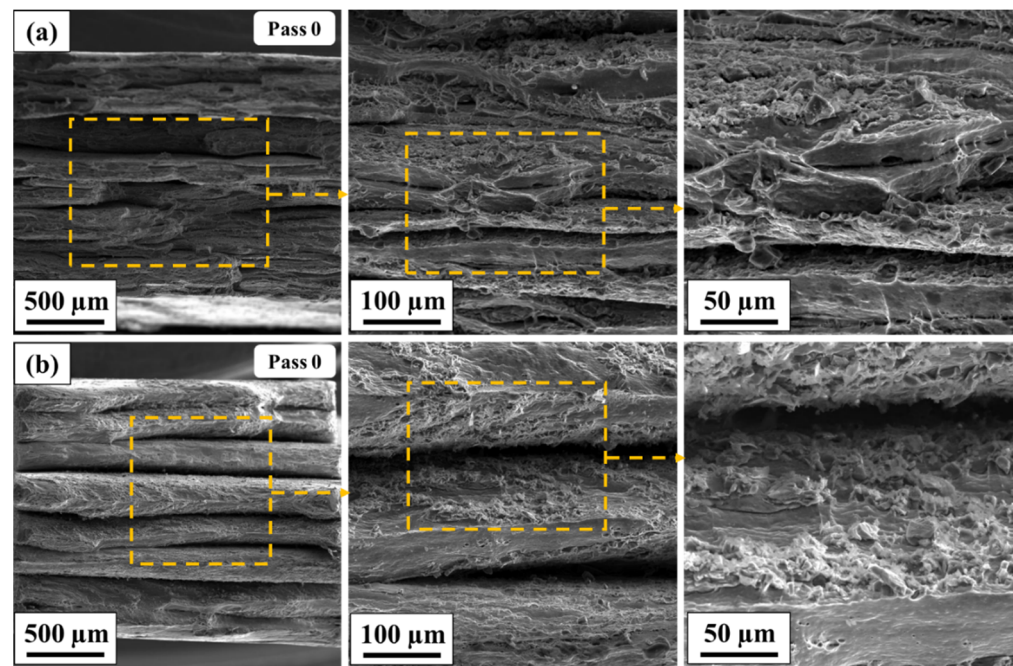


Figure 11. Fracture surfaces of Al/Cu/3wt%SiC composites fabricated by: (a) ARB and (b) CARB process.

Moreover, from Figure 10, particles exist near delamination and particularly dimples, indicating the void formation near SiC/Al and SiC/Cu interfaces. The type, size, and volume fraction of the particles may influence the fracture properties of composites. Moreover, failure of the particle-reinforced composites can be caused by the decohesion of metal/particle interfaces [27,29,37]. Figures 10 and 11 show that this failure mechanism is more evident in Al/Cu/3wt%SiC composites than in Al/Cu/2wt%SiC composites, due to the probability of cracking of bigger agglomerated particles.

Comparing the fracture surfaces in Figures 10 and 11 indicates that dimples on the surfaces of composites produced by the CARB process are symmetric, because the 90° rotation around the normal axis results in a more uniform distribution of strain. In contrast, the dimples on composites produced by the ARB process are asymmetric [23,26].

3.3. Thermal Analysis

Figure 12 displays the variation of thermal conductivity of Al/Cu/2wt%SiC and Al/Cu/3wt%SiC laminated composites, fabricated by the ARB and CARB processes at different passes. Since the thermal conductivity of the copper layer (375 W/m·K) is higher than that of aluminum (227 W/m·K) and SiC powders (155 W/m·K), the measured values are probably related to this layer.

By employing the equations [38] below, the thermal conductivity of these composites was calculated.

$$\rho_c = \rho_m \times V_m + \rho_r \times V_r \quad (1)$$

$$C_c = \frac{C_m \times V_m \times \rho_m + C_r \times V_r \times \rho_r}{\rho_c} \quad (2)$$

$$k = \alpha \times \rho_c \times C_c \quad (3)$$

where ρ_c , V , and C are the density, volume fraction, and the specific heat, respectively, and the indexes of m and r refer to the matrix and reinforcement. Additionally, α and k are the components' thermal diffusivity and thermal conductivity, respectively. After calculating the values of theoretical density (ρ_c) and specific heat of components (C_c) using Equations (1) and (2), the third equation obtained the thermal conductivity of each laminated composite, and the values are presented in Figure 12.

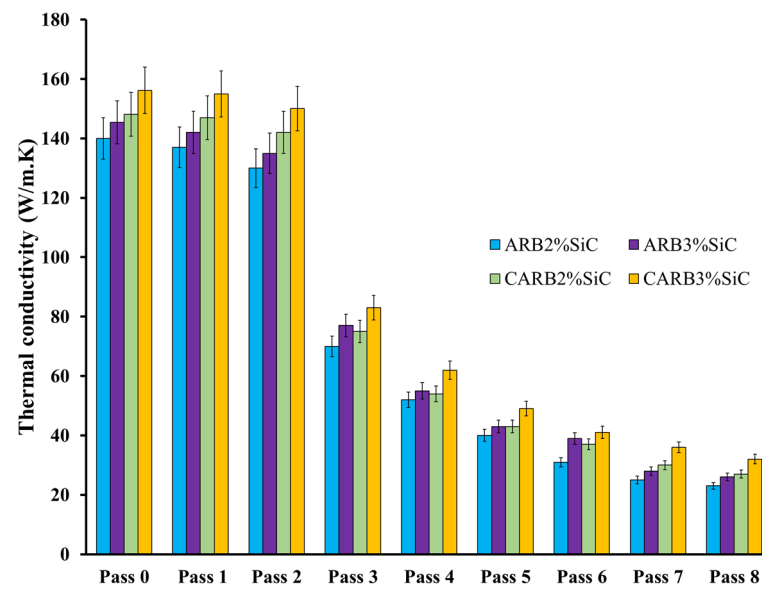


Figure 12. Thermal conductivity of composites at different passes.

However, all four types of composites undergo a decline in their values, due to the applied severe plastic deformation, which causes instabilities in the copper layers. In other words, with an increase in the number of passes, reinforcing layers may gradually undergo necking and rupture. Then, the increased number of interfaces declines the heat flux through the composites. However, these composites can be compared to the manufacturing process and the SiC powders. Regarding the former, as previously mentioned, it is well known that when the CARB process produces composites, the layers experience less plastic instability as a result of rolling the sandwiches in RD and TD directions, leading to a better strain distribution in both directions. Therefore, in all rolling passes, heat flux in composites produced by CARB is highly likely to be higher than the one made by ARB. This difference can be observed in Figure 12. Alongside that, concerning the weight percent of SiC powders, Al/Cu/2%wt SiC composites show lower thermal conductivity than Al/Cu/3%wt SiC composites, which is attributed to more interfaces between SiC powders and the other constituents, including aluminum and copper. In fact, the motion of electrons controls the heat transfer through the composite, and when the number of interfaces increases, heat flux might be restricted [39]. Furthermore, the porosities between particles and layers have a detrimental effect on heat conduction, mainly when a cluster of particles exists within the matrix and at the aluminum/copper interfaces. Nonetheless, at higher passes, they are better distributed, while a higher number of interfaces may reduce the heat flux.

It is generally known that both electrical and thermal conductivity depends on the motion of the thermal energy carriers, such as electrons and phonons. Phonon scattering usually occurs on dislocations, grain boundaries, and imperfections, such as impurities. Phonon scattering on dislocations can be a long-range or short-range interaction. In the former, it occurs through an elastic strain field of dislocation lines, whereas in the latter, it occurs on the core of the dislocation lines. In both cases, the thermal conductivity is inversely proportional to the dislocation density. This means thermal conductivity declines with an increase in passes. It is believed that during severe plastic deformation, the thermal reduction is caused by the scattering of phonons, due to deformation mechanisms. In the current research, thermal conductivity declined due to grain refinement and the dislocation high density and grain boundary density. With an increase in the number of rolling passes, the deformation mechanisms transform from dislocation-based deformation to grain-boundary-based deformation. At higher passes of the ARB and CARB processes, grain refinement takes a place that is composed of a high-volume fraction of grain boundaries.

Moreover, thermal conductivity is strongly dependent on microstructural defects. Since cold ARB and CARB processes were performed in this work, SiC particles did not chemically bond to Al and Cu. Therefore, weak interfaces with a lot of porosities near particles can be another primary source of phonon scattering [40–43]. As shown in Figures 3 and 5, the increase in passes led to a significantly increased number of interfaces, such as Al/Cu, SiC/Al, and SiC/Cu. When chemical reactions do not occur at interfaces, phonon scattering mainly takes place in these regions because of the differences in the physical properties of the constituents. Interfaces are, thus, considered to be thermal resistant. Furthermore, as reported, at higher passes of the ARB and CARB processes, a shear banding type of deformation often occurs, which causes a discontinuity in the microstructure and the subsequent reduction in thermal conductivity. This feature can be seen in Figure 7. Thus, all the mentioned reasons are responsible for the decline in thermal conductivities. These factors also account for the difference between the experimental values and those obtained from predicting models.

4. Conclusions

In this paper, laminated composites of Al/Cu/2wt%SiC and Al/Cu/3wt%SiC were fabricated by the ARB and CARB processes separately, at different passes. Their microstructures, microhardness, tensile behavior, and thermal conductivity were investigated. The findings are summarized as follows:

1. According to OM images, composites had good bonding. With an increase in passes, the copper layers underwent instabilities, and their fragments were distributed in the matrix. However, CARBed composites represented less instability on the account of the relatively even distribution of strain on the layers, which results from rolling in two directions. Regarding the SiC particles, they first appeared in continuous layers in the first passes, while with an increase in passes, they created big clusters at the Al/Cu interfaces and within the matrix. However, at higher passes, they were quite well distributed in the composites.
2. By increasing the number of passes, the hardness of aluminum and copper layers increased gradually, while this increase between the zero and first passes was noticeable because of a higher rate of work-hardening in the first passes. The layers in the composites with the higher weight percent of ceramic particles represented lower hardness, because they are highly likely to make clusters within composites which impede the even strain distribution on the layers.
3. The values of yield, and ultimate strength, experienced an upward trend, while the elongation of composites showed an initial decrease, followed by an increase. This behavior in the first and second passes is attributed to the possible weak bonding of the interfaces of SiC particle/Al, SiC particle/Cu, and Al/Cu. However, the increased values of strength are caused by the work-hardening of the layers, a better distribution of reinforcing particles, and an enhanced bonding of interfaces.
4. Based on the images of the fracture surfaces, both ductile and cleavage modes of fracture were detected. In addition to aluminum, copper layers also represent many voids, although river-like patterns could be seen on the copper layers at higher passes, due to having low ductility. Moreover, delamination was noticeable due to the different mechanical behavior of aluminum and copper layers during tension. It was also caused by the ceramic particles at Al/Cu interfaces.
5. This study has gone some way towards enhancing our understanding of the differences pertaining to the ARB and CARB processes. Based on the investigated microstructures, mechanical and thermal properties, applying the CARB technique can result in favorable composites. In addition, depending on the need of the industry, a specific number of passes should be applied to optimize the properties.

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