Role of enhanced surface grain refinement and hardness improvement induced by the combined effect of friction stir processing and ultrasonic impact treatment on slurry abrasive wear performance of silicon carbide particle reinforced A356 composites

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Abstract
In this study, the slurry abrasive wear behavior of silicon carbide particle reinforced A356 composite alloy was investigated after the different surface mechanical attrition treatments. It is known that the aluminum matrix composites produced by the stir casting method have some deficiencies (e.g. unfavorable microstructure formation, particle clustering, porosity formation, etc). These kinds of drawbacks of the composites adversely affect surface mechanical properties of materials such as wear resistance. For this purpose, the surface properties of the silicon carbide reinforced A356 matrix composites fabricated through the stir casting method were improved by using friction stir processing (FSP) and ultrasonic impact treatment (UIT) in the study. The results indicated that a remarkable increase was observed in the hardness and wear resistance of the cast composite via FSP and ultrasonic impact treatment following friction stir processing (FSP + UIT). The hardness of the stir zone after FSP and FSP + UIT was determined as 82.7 ± 2 HV and 101.9 ± 3 HV0.2, respectively. The stir zone showed a similar tendency also in slurry abrasive wear resistance. FSP increased the wear resistance in the stir zone at the rate of 33.9% while it was determined as 35.5% for FSP + UIT. The microstructural modification of the cast composite that occurred after FSP was clearly demonstrated via optical microscope and scanning electron microscopy (SEM) examinations. Enhanced grain refinement after FSP + UIT was indicated especially by x-ray diffraction analysis (XRD). According to the findings, it was observed that the application of ultrasonic impact treatment following the friction stir processing can be used to obtain an enhanced microstructure and extra hardness increment in the surface of the SiC reinforced A356 alloy, thus resulting in slurry abrasive wear resistance increment.

1. Introduction

Aluminum alloys, which are frequently used in many engineering applications with their high specific strength and corrosion resistance, have a wide usage area in the aviation and automotive industry, especially to reduce fuel consumption [1]. Due to the poor wear resistance of these materials, the development of ceramic-reinforced aluminum-based composite materials has become an important issue in the literature [2–5]. In particular, many attempts have been made to improve the wear resistance of aluminum alloys with hard ceramic particles by using the stir casting method [6–9]. Unfortunately, the dendritic grain structure and coarse acicular silicon phase of these alloys manufactured via the casting method can cause unfavorable effects on important surface mechanical properties such as wear resistance [10–13]. It was suggested that the interfacial region between the
In the study, the A356 alloy was used as a matrix alloy. 2.1. Stir casting

The coarse silicon phase and the matrix initiates the formation of microcracks. It was also reported that the fine distribution of these crystals within the structure will improve the wear resistance [14]. In addition, as emphasized by different authors, ceramic reinforced metal matrix composites fabricated by the stir casting technique can have some manufacturing-based problems, such as nonuniform dispersion and agglomeration of reinforcements in the structure, insufficient wetting of reinforcement in the molten metal, matrix-reinforcement reaction probability, and also the formation of particle cluster porosity [10, 15–18]. For this reason, it is required to improve the surface mechanical properties of ceramic reinforced aluminum silicon alloys has arisen and friction stir processing has been suggested as an effective method by many researchers [19–22].

It is clearly known that an enhanced grain refinement in the microstructure can be obtained by the friction stir processing [23–30]. In this method, a non-consumable rotating tool, having a shoulder and pin, is inserted into the material surface forcibly. When adequate heat input is obtained in order to soften the material, the tool is moved throughout a specific local direction. As the tool moves, a material flow occurs from the advancing side to the retreating side via the rotation and linear movement of the tool. This material flow, under the shoulder pressure, is surrounded by the cold material and tool that was defined as the ‘extrusion chamber’ [31–33]. This intense local deformation, called ‘severe plastic deformation’, at elevated temperatures causes grain refinement in the processed zone and provides favorable new mechanical properties [34]. The application of consecutive passes via friction stir processing allows obtaining an equiaxed fine-grained microstructure and a uniformly dispersed ceramic phase over the entire surface [31, 32, 35]. Besides this, it was stated that friction stir processing enhances the plastic deformation ability of materials by modifying the grain structure. In this way, high elongation values can also be obtained for materials produced by casting [36].

Despite the grain refinement and hardness increment effect of the FSP mentioned in the literature, considering the wear resistance, extra enhanced surface properties are needed for specific applications [37–39]. It was revealed that loss of material occurs due to the low hardness value of the material exposed to abrasion and plastic deformation [20]. Furthermore, it was also explained that the wear mode was converted from two-body abrasive wear to three-body abrasive wear by the worn-out particles. In the line of the mentioned evidences, ultrasonic impact treatment has been recommended by researchers in order to obtain an advanced grain size and hardness value [40–43]. The outstanding feature of this method compared to friction stir processing is that it can produce an intense plastic deformation on the near-surface region of the metal surfaces without any thermal effect to the material in the high-frequency range [44, 45].

In view of the above description; friction stir processing has been applied to eliminate the casting-related defects, distribute the particle reinforcements homogeneously in the matrix, enhance the cast microstructure, and accordingly increase the toughness and strength of the silicon carbide reinforced A356 alloy manufactured via the stir casting method. In addition to this, ultrasonic impact treatment following the friction stir processing has also been utilized to obtain a finer grain size and extra increased hardness in the near-surface region. In this way, the current study has focused on the combined surface property enhancement effects of the FSP and UIT. The surface mechanical properties and slurry abrasive wear resistance of the processed and unprocessed SiC particle reinforced A356 alloy were investigated.

2. Experimental procedure

2.1. Stir casting

In the study, the A356 alloy was used as a matrix alloy (table 1). The average size of sharp-edged silicon carbide particles selected as reinforcement is 45 μm (figure 1). The stir casting method was used to obtain A356 matrix composite plate (figure 2(a)). The matrix material was melted in a resistance heating furnace in an inert gas atmosphere (argon) by increasing the temperature to 800 °C. A K-type thermocouple was inserted in the crucible and the temperature of melt has been controlled regularly during the casting process. In order to increase the wettability of reinforcements, 2% wt. magnesium was added to molten matrix material and silicon carbide particles were kept at 550 °C for 3 h. The melt was stirred at 1000 min⁻¹ for 1 min after the addition of magnesium. Subsequently, 10% wt. silicon carbide reinforcement was carried out and the molten composite was stirred again at 1000 min⁻¹ for 3 min. The casting was carried out in a preheated steel die (550 °C) to prevent possible cracks that are triggered by rapid solidification. While casting, the temperatures of molten composite

| Si  | Mg  | Fe  | Cu  | Mn  | Zn  | Ti  | Al  |
|-----|-----|-----|-----|-----|-----|-----|-----|
| 6.5–7.5 | 0.25–0.45 | 0.2 | 0.2 | 0.1 | 0.1 | 0.1 | Balance |

Table 1. The chemical composition of A356 matrix alloy (wt %).
Figure 1. SEM image and EDS analysis of the silicon carbide particles.

Figure 2. The schematic representation of (a) Stir casting, (b) Friction stir processing and (c) Ultrasonic impact treatment.
and the steel die were 725 ± 5 °C and 500 ± 10 °C, respectively. Then, the cooling of the casting was done in the die at room temperature. The dimensions of the obtained silicon carbide particle reinforced aluminum matrix composite plates were 200 mm × 150 mm × 10 mm.

2.2. Friction stir processing
Friction stir processed sample were produced on a vertical milling machine with a conical pin of 6 mm diameter at the top and 5 mm diameter at the bottom with a flat shoulder of 20 mm in diameter was manufactured using H13 tool steel (55 HRC). The length of the pin was 3 mm. The tool tilted by 2 degrees with respect to the vertical axis of the work piece and keeps constant during the process. In order to enhance the mechanical properties of the surface, overlapped three consecutive passes were carried out by friction stir processing, (figure 2(b)). The overlap ratio was 50% and determined by [31]:

\[
\text{Overlap Ratio} = 1 - \left( \frac{l}{d_{\text{pin}}} \right)
\]

where \(l\) is the distance of the pin axes in two consecutive passes and \(d\) is the pin diameter (the top diameter was taken into consideration). By using 50% overlap ratio, partial stir zone interpenetration was achieved. A total of three successive passes were performed in the same direction by overlapping on the advancing side. Each pass was done after the previous pass cooled down and all processing was performed under ambient conditions. In order to obtain defect-free processing, the initial preheating was obtained by plunging the tool with a penetration speed of 1 m/min. In the friction stir processing, 900 min⁻¹ tool rotation and 20 mm min⁻¹ traverse speed were used as favorable process parameters. These parameters were determined to fabricate a sound composite structure with proper strength after detailed preliminary investigations. In these attempts; various tool rotation speeds ranging between 900–1200 min⁻¹ were used, while they were in the range of 80–20 mm/min for traverse speeds. The plunge depth of the shoulder was 0.7 mm.

2.3. Ultrasonic impact treatment
Ultrasonic impact treatment was performed on the surface of friction stir processed specimens with a head that has one pin (figure 2(c)). The diameter of the pin was 5 mm. The used frequency and amplitude of the pin during treatment were 24.8 kHz and 17.5 μm, respectively. Ultrasonic impact treatment has been carried out under steady treatment conditions for 15 min in order to prevent obtaining a nonhomogeneous treated surface. The mass of the ultrasonic impact treatment tool was used as a counterforce of formed pin impact forces. After the treatment, the surface of the specimen had a bright view and uniformly dispersed indentations on it. For these attempts; various tool rotation speeds ranging between 900–1200 min⁻¹ were used, while they were in the range of 80–20 mm/min for traverse speeds. The plunge depth of the shoulder was 0.7 mm.

2.4. Characterization
In order to carry out the microstructural and macrostructural investigations, metallographic test samples were extracted from the friction stir processed region with respect to the normal of consecutive pass directions that allowed examining the pass cross-sections. 80, 120, 400, 800, 1200, 2000 and 2500 grit silicon carbide papers were used to grind the cross-section surfaces of the samples. Polishing was performed by using Colloidal Silica to obtain mirror finish. Samples were etched with 6 ml hydrofluoric acid, 3 ml nitric acid, and 150 ml distilled water mixture and examined with an optical microscope and scanning electron microscope to investigate the variation in size and shape of the grains and distribution of silicon carbide particles on the surface. For macrographic examination, Tucker’s reagent was used.

In order to reveal the formed dislocation density in samples x-ray diffraction analysis (XRD) were done by using Cu Kα radiation. Cross-section hardness profiles of the samples were carried out along a line at 0.5 mm depth from the top surface by using a Micro-Vickers hardness test machine. The hardness measurements were obtained at a load of 200 g and a dwell time of 15 s. The distance between indentations were 0.5 mm. The indentation distance for the hardness profile throughout the depth from the surface in the FSPed + UITed sample was 1 mm.
2.5. Wet (slurry) abrasive wear test

Wet (Slurry) abrasive wear tests were carried out according to ASTM G105-20 standard. In order to obtain the slurry, sand and deionized water were mixed in the ratio of 1.5:1. In slurry abrasive wear tests, a 222 N force has maintained between the wear sample positioned in the sample holder and the rubber wheel. Three different rubber wheels that have different hardness values (50, 60, 70 Shore A) were used. The wear tests of each sample were carried out by using the rubber wheels in the order of the increase in their hardness. Each rubber wheel was rotated at 245 min$^{-1}$, and every 1000 revolutions the rubber wheels were changed, and wear loss of the sample depending on the used wheel was noted. All wear tests were done at room temperature.

3. Results and discussions

The cross-section macrostructure of the friction stir processed zone consisted of three distinct zones as base material (BM), thermo-mechanically affected zone (TMAZ) and stir zone (SZ). Both surface morphology and cross-section of the friction stir processed zone showed optimum and favorable results (figures 3(a)–(b)). The process parameters used in friction stir processing provided obtaining adequate heat input in the metal matrix composite plate (figure 3(c)). Thus, the deformation ability of the composite plate increased and as a result, a defect-free (such as wormhole, porosity, silicon carbide clusters, etc) and sound processed zone were obtained (figure 3(b)). It was observed that the width of the stir zone throughout the depth reduced. Due to the geometry of the used tool, different flow characteristics occur at the shoulder and pin zone of the tool [10]. The combined
effect of pin-driven extrusion and shoulder-driven stirring in stir zone attracted the attention of several researchers and has been clarified in many studies [31, 46–49].

According to the optical micrographs, the base material has a coarse-grained and relatively heterogeneous cast microstructure (figure 4(a)). The cast microstructure comprises of high-volume percentage of \(\alpha\)-aluminum dendrite, intergranular aluminum-silicon eutectic, and second-phase silicon carbide particles. In friction stir processing, by its very nature, friction raises the temperature, and tool rotation-traverse speed couple cause severe plastic deformation. As a result, high temperature and severe plastic deformation result in dynamic recrystallization in the stir zone [10, 31, 50–54]. Accordingly, the coarse-grained base material structure was turned into a fine equiaxed grain structure in the stir zone by friction stir processing. The reason is that the grain growth-inhibiting of continuous stirring action of the tool during dynamic recrystallization, and the grain disintegrating effect of the second-phase silicon carbide particles by causing nonuniform local deformations around themselves [10, 55]. It is evident that the silicon carbide clusters in the base material were annihilated and these hard particles were uniformly distributed in the stir zone via friction stir processing. Simultaneously, the porosity formation caused by particle clusters during solidification was also removed during the method (figure 4(d)). Furthermore, no swirl patterns (in other words onion rings) that are commonly observed in friction stir welded/processed aluminum alloys were observed in stir zone. This situation is good for obtaining a fully homogenized microstructure in the whole stir zone. Throughout the transition from the composite base material to the stir zone (figure 4(b)), it was observed that there was no obvious distinction between microstructures of the heat-affected zone (HAZ) and base material. A parallel result related to HAZ-BM microstructure similarity was also observed in a study carried out to fabricate ceramic particle reinforced A356
In the thermo-mechanically affected zone, elongated grains were formed, and the structure of the eutectic phase was distorted by having a flow pattern between BM-SZ transition (figure 4(c)).

It can be observed from energy dispersive x-ray spectroscopy maps of the FSPed sample and base material that the intergranular aluminum-silicon eutectic phase in the base material was completely removed through the friction stir processing (figures 5(a)–(b)). Although the silicon carbide clusters in the base material endeavored to prevent the growth of the eutectic phase via particle pushing phenomena, the coarse eutectic phase has still continued to form between the aluminum dendrites (figures 5(d)–(e)) [11, 56, 57]. This phenomenon clarifies the obstruction effect of silicon carbide growth during solidification. According to this, the second-phase silicon carbide particles is rejected by the solidification front and thus particles pushed away by dendrites until they trapped in intergranular regions in the structure of aluminum matrix composite. Consequently, hard second-phase silicon carbide particles prevent or retards the growth of aluminum-silicon eutectic in intergranular regions (figures 5(d)–(e)). Friction stir processing changed the size and existence form of this coarse eutectic phase. Friction stir processing fragmented and distributed the eutectic phase uniformly to the whole stir zone by causing severe plastic deformation in the structure (figure 5(b)). Severe plastic deformation showed also the same effect on silicon carbide particles. It can be clearly observed from SEM micrographs that, silicon carbide particles were disintegrated (figure 1, figure 5(b)) and uniformly distributed (figure 5(b)) in the stir zone while the morphologies have been the same as in the received
condition (sharp-edged). The fracture of aluminum–silicon eutectic and hard carbide particles by severe plastic deformation depends on their brittle natures and thereby the properties of not able to deform plastically under force \[10, 58, 59\]. In higher magnification of stir zone, the bonding between aluminum matrix and silicon carbide particle can be clearly seen, (figure 5(c)). There was no microporosity formation observed around the particle. Matrix material wrapped the particle completely and thus obtained composite via friction stir processing showed preferable mechanical properties in comparison with base material.

It can be seen from the micrographs that the FSPed + UITed sample has a relatively slight undulating surface in a microscale (figures 4(g)–5(f)). Different layers occur through the depth with enhanced properties in the ultrasonic impact treated surface (near-surface region) similar to the methods such as air blast shot peening and ultrasonic shot peening [60, 61]. These are nanocrystallization layer, work hardened (plastic deformation) layer, and residual stress layer, respectively [62]. In the nanocrystallization layer, grain boundaries cannot be clearly seen in the optical microscope due to the nano-scale-sized grains. In the intermediate layer, known as the work hardened layer, the grain orientation forms along the normal of the plastic deformation direction. As a result of plastic deformation, the grains are crystallographically reoriented in this layer by slip and twinning micromechanisms [63]. The undermost layer is the residual stress layer which is induced by the work hardening. The existence of these layers can be easily understood from the high hardness values compared to the base material or FSP’ed sample [64]. The ultrasonic impacts cannot cause a significant deformation in this layer, and thereby apparent microstructural changes. But ultrasonic impact treatment induces to form considerable compressive residual stresses in this layer [65]. These formed high-stress values cause a hardness increase in this layer compared to the base material. It was also observed from the optical and SEM micrographs that, in the near-surface region, some extra fractures occurred in the hard second-phase carbides and the aluminum–silicon eutectic in the stir zone due to the UIT. The brittle natures that prevent undergoing deformation of the silicon carbide and eutectic particles caused the ultrasonic impact treatment broke them into small pieces (figures 4(e)–(g)–5(f)). This situation can be stated as an additional fragmentation of carbide and eutectic particles that was carried out by ultrasonic impact treatment.

The x-ray diffraction patterns and full width of half maximum (FWHM) values of the samples illustrates surface dislocation density of the samples (figure 6(a)). As can be clearly seen, slight peak shifting to right and broadening in (111) peak occurred after the friction stir processing. When compared to the FSPed sample, peak shifting and broadening took place more prominently for FSPed + UITed sample (figure 6(a)). Similar to the literature, increasing micro strain and enhanced grain refinement resulting from the high frequency impacts produces a compression of bonds between atoms and decrease the lattice spacing [66]. This situation is observed with the shifting and broadening in the patterns. As a result of the high density of dislocation arrangements
induced by ultrasonic impact treatment, full width of half maximum value raised up to 0.22399 after the treatment, while it increased from 0.20128 to 0.20451 after the friction stir processing (figure 6(b)).

The microhardness profiles of the cross-sections of the FSPed and FSPed+UITed samples showed similar profiles (figure 7(a)). It has been observed that both friction stir processing and ultrasonic impact treatment following friction stir processing remarkably increased the hardness of the processed zones in comparison with base material which has an average hardness of 55.4 ± 1.8 HV 0.2. The overlapped stir zone showed fluctuations in hardness distribution and the average hardness value in stir zone was measured as 82.7 ± 2.0 HV 0.2. The hardness fluctuations across the stir zone may be attributed to the formed complex material flow around the processing tool [10, 67] and inhomogeneous local deformations produced by silicon carbide particles during material flow.

Due to the difference in material flow, the formed deformation and increased temperature are not identical exactly in the whole stir zone. Thus, this may result in minor variations in hardness along with the effect of silicon carbide particles existence in the matrix. On the other hand, the hardness increase in stir zone can be explained via the uniform distribution of hard second-phase carbide particles in the matrix material and grain refinement of cast aluminum microstructure [68, 69]. According to the Hall-Petch equation, grain refinement has a significant effect on the increase of hardness in stir zone [31]. Furthermore, in overlapped stir zone, it is thought that the modification and homogeneous distribution of the silicon particles existing within the eutectic phase of the cast microstructure has also a role in the increment of hardness [10, 69]. Thermo-mechanically affected zone has also a notable hardness value in comparison with the base material. The hardness obtained in stir zone has a downward tendency throughout the thermo-mechanically affected zone in both advancing and retreating sides.

Ultrasonic impact treatment following friction stir processing has led to a significant hardness increase in the near-surface region of the FSPed+UITed sample. This situation can be attributed to the materialized work hardening and the grain refinement as a result of high volume deformation [60, 61, 64]. Ultrasonic impact treatment enhanced the hardness in stir zone and base material regions at the rate of 23.2% and 33.8%, respectively. The average hardness of stir zone in FSPed+UITed sample was measured as 101.9 ± 3.0 HV 0.2.

The change in hardness profiles from the surface throughout the thickness in the FSPed+UITed sample showed the effect of ultrasonic impact treatment on surface mechanical properties (figure 7(b)). It was observed that the hardness increase in the FSPed+UITed sample gradually decreased from the surface inwards. It was also determined that the surface hardness of the FSPed+UITed sample which was exceeding the 1.5 mm depth is higher than the FSPed sample. After this depth, both the FSPed+UITed and FSPed samples showed the same hardness characteristics. According to the hardness profile of the FSPed+UITed sample through the depth, it can be stated that the total average thickness of the plastic deformation and residual stress layer obtained by ultrasonic impact treatment is approximately 1.5 mm in the FSPed+UITed sample (figure 7(b)).

It has been already known that hard second-phase ceramics (silicon carbide, aluminum oxide, titanium carbide, boron carbide, etc) improves the wear resistance of aluminum alloys if a favorable bonding at the ceramic-matrix interface is obtained [70–73]. Therefore, wear behaviors of three different samples that were all reinforced with silicon carbide particles investigated in the study. According to the weight losses accepted as wear resistance, the wear rate was drastically increased in all samples by the hardness of the rubber wheels increased (figure 8). In most of the metal matrix composites, material loss significantly takes place in the softer
matrix material due to the load-bearing capacity of hard carbides [74, 75]. Besides this, poor bonding between the matrix and hard second-phase carbides also causes an increase in wear rate by breaking off the reinforcement from the worn surface and starting a severe three-body abrasion on the surface.

It was observed that FSPed and FSPed+UITed samples showed higher wear resistances against to slurry abrasive wear conditions in comparison with the base material. In the first phase of wear tests done with 50 durometer wheel hardness, the least material loss (87.6 mg) was determined in FSPed+UITed sample due to the high surface hardness of the sample obtained by ultrasonic impact treatment. However, the increase in the severity of wear conditions by changing rubber wheel gave rise to wear away the hard surface layer of FSPed + UITed sample. Thereafter, this sample acted as FSPed sample and similar wear rates have been determined in these samples. The wear resistances of the FSPed and FSPed+UITed samples remarkable higher than base material. The underlying reasons are that the homogeneous distribution of silicon carbide particles in the dense stir zone of the composite plate, good bonding at the silicon carbide particle-matrix interface, and hardness increase provided by hard reinforcements and dynamic recrystallization induced grain refinement. It was determined that the majority of material loss in base material did not come into existence only in the matrix material (figure 9). Silicon carbide particle clusters and the porosity that existed in the base material caused an excessive material loss and increased the wear rate. In the zones of existing silicon carbide clusters, both brittle structured cast matrix material and carbide particles have broken off from the surface together and large craters have occurred in the surface (figure 9). Consequently, excessive material loss due to existing porosity in the aluminum matrix composite plate and heterogeneous clustering of silicon carbide particles have become dominant and provoked wear rate.

By its very nature, friction stir processing eliminates various discontinuities such as porosity, unfavorable microstructure, inclusions, etc that are arising from manufacturing methods [31, 76, 77]. By this means, large craters or porosity formations were not determined in the processed surfaces of FSPed and FSPed+UITed samples with dense and high-strength properties (figure 9). Thereby, considering the slurry abrasive wear test conditions and physical properties of the abrasives, it was determined that the wear mechanism consisted of three-body rolling and grooving. The worn-out surfaces of all samples showed similar morphologies except the groove widths and depths. As the worn surfaces were examined that the multiple indentations occurred.

Figure 8. Weight loss of the samples versus 50, 60, and 70 durometer wheel hardness in slurry abrasive.
Additionally, in the base material, debonded silicon carbide assisted scratching and cracking triggered brittle chip formation was also observed to become predominant and participated in the wear mechanism during the test.

4. Conclusions

In the study, stir casted 10% silicon carbide particle reinforced A356 composite was successfully fabricated and then friction stir processed in order to eliminate manufacturing-based defects/discontinuities and drawbacks, such as unfavorable microstructure and carbide particle clustering. Furthermore, ultrasonic impact treatment was also applied on the friction stir processed surface to obtain finer grains and elevated hardness in the surface. Thus, the slurry abrasive wear behaviors of the surfaces of FSPed and FSPed+UITed samples with their sufficient properties as having dense structure, adequate bonding at the silicon carbide particle-matrix interface, and high hardness were compared to cast base material. The following conclusions are given:

1. Friction stir processing properly refined the coarse-grained nonhomogeneous cast microstructure of base material and increased the hardness in stir zone.

2. Friction stir processing caused the fragmentation of the silicon carbide particles and aluminum-silicon eutectic phase and distributed them in whole SZ homogeneously.

3. Ultrasonic impact treatment increased dislocation density in the surface and provided extra hardness increment in stir zone by 23.2%.

Figure 9. Worn surfaces of the samples (yellow arrows depict the sliding direction).
4. Sufficient bonding between silicon carbide particles and aluminum matrix was observed in FSPed and FSPed-UITed samples. On the contrary, relatively poor bonding between reinforcement and matrix led to the break off particles heavily from the surface of the base material, thereby increasing the wear rate during wear tests.

5. Both FSPed and FSPed+UITed samples showed favorable slurry abrasive wear resistance at the rate of 33.9% and 35.5% respectively in comparison with the base material. Casting-based discontinuities (silicon carbide clusters and porosities) caused a drastic material loss in the base material.

Data availability statement

All data that support the findings of this study are included within the article (and any supplementary files).

Conflict of interest

The authors declare no financial or commercial conflict of interest.

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