Effect of in-situ tribo-oxide-layer on the non-lubricated tribological behaviours of LM27/SiC\(_p\) composites

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**ABSTRACT**

In this study, the influence of the in-situ tribo-oxide-layer on non-lubricated tribological behaviours of LM27/SiC\(_p\) composites was studied at different applied loads. For this purpose, LM27/SiC\(_p\) composites were manufactured by stir casting route with reinforcement of different amounts (3–12 wt.\%) and different sizes (fine: 1–20 µm and coarse: 106–125 µm) of SiC\(_p\). A comprehensive characterization of the friction and wear mechanisms ranging from mild to severe wear for oxidative and delamination at a contact pressure of 0.125–0.624 MPa was discussed. Results indicate that the friction and wear behaviour is strongly influenced by the morphology and the nature of the oxide scale on the wear track. In-situ formation of oxide layers on the contact region of the specimen supports the self-lubrication which supports better wear performance of LM27/SiC\(_p\) composites. However, these study portraits that composite with 12wt. % fine size SiC\(_p\) exhibits better wear performance in comparison to the other developed composites.

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1. Introduction

Aluminium matrix composites (AMCs) show great potential for producing automotive, construction, aerospace structures in rail transport and other different engineering sectors [1–4]. Generally, the calliper and rotor used in most automobile industries are typically made from ductile cast iron and grey cast iron, respectively. To reduce automobile weight and improve fuel efficiency, the auto industry has dramatically increased the demand for light vehicles in recent times [5–7]. To reduce automobile weight and improve fuel efficiency, the automobile industry has dramatically increased the demand for lighter vehicles in recent times. So, cast AMC brake rotors to fill the essential demand of automobile industries and provide 45–60% weight reduction in AMC braking system [4,8–10]. From the point of view of safety, the AMC material selected for the brake system should have stable and reliable friction and wear properties under varying conditions of load, velocity, temperature, environment and high durability. The aluminium alloy as a matrix is suitable for brake disc material selection based on these factors. Literature analysis allowed us to notice that there is a lot of research related to the review of materials for manufacturing friction and wear materials [4–10]. The method of friction and wear, including changes in the physical, chemical and mechanical properties of the composite surface, is relatively difficult [7]. At high temperatures, materials will oxidize and oxidation resistance plays an important role in the evaluation of their high-temperature properties [8]. It is known that a tribo-oxide layer typically forms on the contact surfaces, which can significantly affect the material’s tribological properties and plays an important role in severe-mild wear.
transition AMCs with reinforcements such as oxide, carbide and boron nitride [7–9]. The wear resistance of the brake lining friction materials in the automobile sector is one of the most critical parameters for better performance. It has been reported that an improvement in wear performance at high temperatures can be achieved with the development of composites with discontinuous reinforcement (e.g. Al₂O₃, B₄C, TiB₂, SiC, ZrSiO₄, etc.) [3,11].

The non-lubricating tribology of AMC is an important mechanism process-related not only to mechanical but also to other processes such as thermal and chemical interactions. In non-lubricated sliding, mechanical delamination is majorly responsible for the change in wear. However, this is not the only reason but is usually associated with various operating parameters during the wear test. The study of the topography and physically homogeneous oxide layer (glaze-layer) on the wear track surface, which significantly influence the excellent tribological behaviour of AMCs, has attracted the attention of many research scholars. Kumar et al. [12] investigated the mechanism of delamination wear based on plastic deformation of the worn surface, which is caused by subsurface crack nucleation, followed by crack propagation. Hogan and Song [13] published work on the composition and structure of the worn surface and found that the effect of wear tests manifested in the cross-section through three regions; region-I: tribo-layer, region- II: reorientation of the reinforcing particles bent towards the wearing direction and region-III is unaffected material. Panwar et al. [14] also studied dry sliding wear of hypo-eutectic Al–Si alloys incorporated with different sizes of ZrSiO₄ particles at different temperatures. They observed that the incorporation of two sizes of reinforced particles showed better wear performance than a single reinforced particle in Al–Si alloys. The wear performance of the Al–Si alloy with fine-size reinforcement was lower than that of the Al–Si alloy with coarse-size reinforcement. Wood et al. [15] investigated the evolution of the oxide layer in different stages such as the formation of a mixed transient oxide layer, preferentially formed thermodynamically stable oxide layer, formation of patches by compacting wear particles, formation of the tribo-oxide layer, cycling formation, and breakdown of tribo-oxide layer and formation of a uniform oxide layer. An increase in microhardness has been observed with the increasing size of ZrSiO₄ but the wear rate also increases with increasing the size of ZrSiO₄ particle [4]. Peterson et al. [16] and Purrzrangs et al. [17] investigated the high-temperature solid-lubrication performance of several metal oxides and found that few metal oxides exhibit low friction behaviour at elevated temperatures. Kumar et al. [12] investigated the relationship between the oxidation behaviour and tribological performance of the composites at high temperatures showing that the wear performance improved with the decreased particle size.

Tribological studies of AMCs with different nature, sizes and amounts of reinforcements on various Al–Si alloys are available in the literature [12–18]. However, the reported research work on the non-lubricated dry sliding wear performance of LM27 alloy with fine and coarse sized SiC_{p} reinforced is limited. Furthermore, to the best of the authors’ knowledge and available data, the effect of in-situ tribo-oxide-layer on the non-lubricated tribological behaviours of LM27/ SiC reinforced composite with fine and coarse size particles has not been reported elsewhere so far. In this research work, the emphasis is on knowing the effect of size and amount of reinforced particles on the wear mechanism, asperities behaviour and the mechanically mixed layer (MML) of the composites. The tribo-oxide layer plays an important role such as the transition from severe wear to mild wear, the resistance of a newly exposed area and the covering of the damaged surface on wear tracks, etc. We have analyzed the wear behaviour and transitions in the wear, the relationship between the tribo-oxide and the tribo-layer of the composites under different conditions.

### 2. Experimental details

#### 2.1. Materials and methods

We select a hypo-eutectic Al–Si alloy (LM27 alloy) as a metal matrix, which is mostly used in brake disc in automobiles due to good tribological performance and castability. LM27 Alloy has been purchased from Emmes Metals Private Limited, Kandivli-400067, Mumbai, INDIA. The chemical composition analysis of as-cast LM27 alloy is shown in Table 1.

Two different sizes (fine: 1–20 µm and coarse: 106–125 µm) of SiC particles (SiC_{p}) were used as reinforcement. The silicon carbide particles have been purchased from INDIAN RARE EARTHS LIMITED (A Govt. of India Undertaking) Orissa Sands Complex, Matikhalo-761045, Odisha, INDIA. The chemical analysis of SiC_{p} is shown in Table 2. The distribution of SiC_{p} within the chosen size range was analyzed with the help of CAMSIZER XT which is shown in Figure 1.

Discontinuously reinforced composites were developed by the stir casting route. All experiments on the

### Table 1. Chemical composition of LM27 alloy.

| Composition of LM27 Alloy | Si  | Fe  | Cu  | Mn  | Mg  | Zn  | Ti  | Ni  | Sn  | Al  |
|---------------------------|-----|-----|-----|-----|-----|-----|-----|-----|-----|-----|
| XRF                       | 8.0 | 0.75| 1.247| 0.598| 0.349| 1.0 | 0.16| 0.268| 0.1 | Bal.|
| Chemical Analysis         | 8.0 | 0.8 | 2.250| 0.6  | 0.35 | 1.0 | 0.2 | 0.3  | 0.1 | Bal.|
stir casting method were performed in a graphite crucible surrounded by an induction furnace. The temperature in the induction furnace was raised to 750°C to melt the LM27 alloy. A 3-blade graphite impeller was rotated for a speed of 630 RPM to create a vortex in the molten metal. Before the addition of SiCp in the molten melt, SiCp was warmed at 400°C to remove wetness and other variable substances. SiCp was added to the molten in defined proportions and properly mixed into the molten metal for uniform distribution.

In the present research work, SiCp was reinforced with fine & coarse sizes ranging from 3 to 12wt.% amount. Even after the completion of particle charging, the stirrer was continued for 5 min to confirm the uniform distribution of SiCp inside the molten melt.

After that, the molten mass was finally discharged into the cast iron frame at room temperature. For similar and optimized AMCs, the amount of molten LM27 alloy, rotating time of the impeller and position of the impeller in the molten metal were kept constant.

Table 2. Chemical composition of SiC particles.

|       | Si   | SiO₂ | Fe  | C   |
|-------|------|------|-----|-----|
| SiC   | 98.5%| 0.3% | 0.5%| 0.08|
|       | 0.3% | 0.3  |     |     |

3. Material characterization

3.1. Microstructure analysis

An optical microscope (Eclipse MA-100, Nikon) was used to observe the distribution of particles in the Al–Si alloy and for the microstructure of composites. For topographic analysis of the wear track surface and debris collected after the wear test, a scanning electron microscope (Model: JEOL, JSM-6510 LV) was used at various magnifications. The chemical composition on the worn surface and debris were analyzed through the Energy Dispersive X-Ray Spectroscopy (EDS). After mechanically polishing, freshly prepared Keller’s reagent was used to obtain superior contrast in the microstructure of composite.

3.2. Microhardness test

Microhardness of the LM27/SiCp composites at different phases (matrix, particle and interface of matrix-particle) was measured by using Vickers hardness tester (Mitutoyo, Japan). The size of the notch mark that hardness values meaningfully depends on the microstructural physical topographies of the tested zone.

3.3. Wear test

Wear test of the LM27/SiCp composites occurred under dry sliding conditions at cabin temperature. Cylindrical pin-shaped specimen (30 mm × 10 mm) wear tests were conducted against EN32 steel counter disc with a hardness of 65 HRC ware. Wear monitor setup (Model TR-20 CH, Ducom, Banglore, India) was used for this purpose. Wear tests were conducted with varying applied loads from 9.8N to 49N up to a sliding distance of 3000 m with a constant sliding velocity of 1.6 m/s. Each pin sample with a flat surface and counter surface was cleaned with acetone and dried before wear testing. In addition, the contact pressure on the vehicle’s brake rotor varies in the range of 0.125–0.624 MPa. The wear test of the specimens was performed under the condition of constant

Figure 1. Graph showing particle size of silicon carbide reinforcement measured with the help of CAMSIZER (a) coarse and (b) fine size particle.
sliding speed (1.6 m/s) and contact pressure (0.125–0.624 MPa). Table 4 presents the specifications for the coefficient of friction (COF) testing with pin-on-disc specifications. The wear tests of each specimen were repeated three times with a new surface for the optimized results.

4. Results and discussion

4.1. Microstructure analysis

Microstructural topographies of the LM27/SiCp composite portrayed that SiCp is decisively managed by the homogeneous spreading of the SiCp in the Al–Si alloy matrix during solidification. Microstructural topographies of composites reinforced with 3, 6, 9 and 12 wt.% and fine and coarse size SiCp are shown in Figures 2–3, respectively. Fair uniform distribution of fine size SiCp with 3 wt.% has been shown in the LM27 matrix phase (Figure 2(a)). The modification in surface topography can be described based on SiCp being spurned by the solid–liquid interface in the course of the solidification of the melt. We found out from these fundamental results that the SiCp must occupy the space in the inter-dendritic regions.

Microscopic structural (Figure 2(a–d)) observations clearly show that eutectic Si is nucleated on SiC particles. Microstructural observations clearly show that eutectic Si is nucleated on SiC particles because it is a potential site for the nucleation of Si. The modifications in the shape and size of eutectic

![Figure 2. The optical micrograph of composites with difference amounts of fine size SiCp (a) LM27/SiC3F, (b) LM27/SiC6F, (c) LM27/SiC9F and (d) LM27/SiC12F.](image-url)
Si in the presence of SiC particles can be attributed to the heterogeneous nucleation of Si on SiC. The aciform eutectic topography of Si is notified in the nearby region of the SiC$_p$. This can be described on the basis that reinforced particles which offer preferential sites for the heterogeneous nucleation of eutectic silicon [19]. Several researchers [19–22] have reported on the heterogeneous nucleation of eutectic Si on SiC$_p$. The lower thermal conductivity and heat diffusivity of SiC$_p$ compared to liquid aluminium leads to retention of heat in the SiC$_p$. As a result, their temperature is higher than that of the surrounding liquid, delaying its solidification. Nucleation of the α-Al phase will start in the liquid at a distance away from the SiC$_p$ where the temperature is lower. The growth of the α-Al nuclei leads to the enrichment of Si in the areas where SiC is located. As Si can nucleate at a higher temperature compared to aluminium, SiC$_p$ can act as heterogeneous nucleation sites for Si. When the growing Al dendrites push the SiC$_p$ in the eutectic liquid, the relative amount of SiC$_p$ there is higher causing a severe refinement of the eutectic Si. In the cases where the Al–Si eutectic nucleation frequency has decreased because of the addition of solute Mg with a sufficiently high concentration to form pre-eutectic intermetallic, the final eutectic silicon morphology has changed to aciform eutectic and fibrous silicon morphology which is shown in Figure 2(a–d). An increased amount of SiC$_p$ in the LM27 alloy deferred the cooling rate in the locality of the particle which accelerates the nucleation of α-Al away from the particles. These modified α-Al dendrites lead to the pile-up of Si with the transformation of non-acicular to aciform in the remaining melt zone around the SiC$_p$. The well-modified, unmodified, or partially modified eutectic Si regions are always observed close to the α-Al dendrites. This phenomenon is a consequence of the distribution of Si concentration within the α-Al dendrites. The irregularity in modification may be caused by the inhomogeneous transfer of thermal energy when the temperature gradient is high. On account of this, atom clusters that are not broken into smaller nuclei will eventually result in the formation of non-acicular Si particles upon solidification [13,19–21]. The acicular, larger-sized Si particles were observed at the edges of the α-Al dendrites. Tolui and Hellawell [21] and Hogan and Song [13] have reported that the Si inter-particle spacing decreases with an increase in solidification rate and vice versa. The eutectic Si particles are completely transformed from long, non-acicular plates to well-modified acicular and fibrous particles. The very fine particle size results

![Figure 3](image-url)

**Figure 3.** The optical micrograph of composites with difference amounts of coarse size SiC$_p$ (a) LM27/SiC$_{3C}$, (b) LM27/SiC$_{6C}$, (c) LM27/SiC$_{9C}$ and (d) LM27/SiC$_{12C}$. 
in a significant increase in the Si particle density. In this regard, the casting technique is the factor by which the form and size of Si particles can be controlled. The form of non-acicular to acicular and fibrous Si morphology also affects the mechanical properties of a particular ductile material.

This improvement of Si morphology improves wear resistance and provides better mechanical properties [12]. However, comparing Figure 2(a–d) with Figure 3(a–d), it can be understood that this modification in Si morphology (aciform) increases with decreasing cooling rate which is directly related to the thermal gradient at the interface of particle and matrix [12]. Optical micrographs of LM27/SiC composites containing various amounts ranging from 3wt.% to 12wt.% particles are shown in Figure 3(a–d). Figure 3(a) shows that the LM27/SiC3C is more evenly dispersed in the matrix than the fine-size particles because the coarse size particles experience a higher shear force during stirring. The uniform dispersion nature of 12wt.% SiCp in the matrix is worse than that of 3wt.% SiCp. Figure 3(b,c) shows the effect of SiCp on the size and aspect ratio of eutectic Si, respectively. From Figure 3(d), it is clear that the addition of 12 wt.% SiCp leads to the refinement of eutectic Si and this effect is more pronounced at higher cooling rates. In addition to the refining of Si, SiC particles also affect the morphology of eutectic Si and the aspect ratio of eutectic Si decreases with increasing volume fraction of SiCp dispersion in the composite. The continuous interface between the reinforced and matrix indicates strong wettability (Figure 3(d)). The sleek interface offers superior mechanical and tribological properties as a transfer of applied load comes about via the interface [14]. Sharma et al. [23] in their research work reported a similar modification in the refinement of the silicon morphology and this modification in morphological change to the generalized faster-cooling outcome formed by the SiCp due to the temperature difference in the melt around its vicinity. The solidification rate is affected by refined eutectic Si-particles. It can be explained based on the surface energy of the Al–Si liquid to the solid interface [20]. The rate of augmentation in the solidification interface depends on a balance between the heat flow rate from the Al–Si liquid to the solid through the interface and the latent heat to be released during solidification. Since the difference between the magnitude of the thermal conductivity of Al alloy matrix and reinforced SiCp and the difference between the magnitudes of the latent heat of fusion of Al–Si liquid to the (SiCp) solid interface are large. So, the Al alloy matrix will go hard much more rapidly than Si. Hence, Al achieved a lead in the course of the solidification of the eutectic and changed the dendritic structure.

4.2. Microhardness

In this experiment investigation, to systematically evaluate the effect of the modified microstructure on the microhardness of LM27 alloy (Al-Si8Cu2Mn0.6)/SiCp composite, consisting of three phases (interface of particle-matrix, particle and matrix) are included. Figure 4(a) displays the microhardness notches marks taken at these phases. The difference in the notches marks size at different phases was observed. Microhardness tests were performed in specific areas, such as the centre of the particle (SiC), matrix (dendrites-a-Al-rich) and interface of particle and matrix (interdendritic regions). As expected, in the cast sample, it has been observed that the hardness trend of the composite is decreasing as we move away from the SiCp to the Al–Si matrix. The microhardness value in the interface (interdendritic regions) is high, due to the presence of the harder phases in the eutectic mixture [24–26]. Fine particles offer more surface area of the reinforcement for a given volume fraction because of less interpretive spacing resulting
in decreased dendrite size, which exhibits good bonding strength with the particle and matrix [26]. The addition of coarse size particles also improves the hardness of the matrix, but the increase is less as compared to fine size particles. Increased hardness of the interface can also be due to the strain energy at the periphery of the particles dispersed in the material. As observed in Figure 4(b), the hardness value increased with the increase in weight fraction and decrease in size of SiCp.

4.3. Wear mechanism

Figures 5–8 show the graphs representing the trend of data collected from the wear tests of all the composites. The wear rate of LM27/SiC3F and LM27/SiC3C composites against sliding distance (3000 m) at different applied loads (ranging from 9.8N to 49N) with a constant sliding velocity at 1.6 m/s are graphically shown in Figure 5(a–b). An increment in the wear rate of the LM27/SiC composites was observed with increasing applied loads from 9.8N to 49N. For LM27/SiC3F composites, the wear rate is lower than for LM27/SiC3C composites. Similarly, the minimum wear rate for 6wt.% to 12wt.% fine SiCp composites were observed as compared to 6wt.% to 12wt.% coarse-sized SiCp composites at applied loads from 9.8N to 49N. Graphically, it can be assumed that the composite shows better wear rate performance with LM27/SiC12F composites. This is due to the modified microstructure, stronger interface bonding between particles to the matrix, stronger oxides and oxygen-containing tribo-layers on the wear surface of the matrix by creating stronger boundaries and improving the wear resistance of composites [27–33]. All the composites showed a similar type of behaviour under different loading conditions as seen in Figures 5–8.

The wear rate of the LM27/SiC3F is marginally improved due to the increase in the interface area of the particle-matrix phase compared to the LM27/SiC3C composites in Figure 5.

Figure 5. Wear rate of composites against sliding distance at different loads for (a) LM27/SiC3F and (b) LM27/SiC3C.

Figure 6. Wear rate of composites against sliding distance at different loads for (a) LM27/SiC6F and (b) LM27/SiC6C.
So, the number of asperities per unit area in LM27/SiC6F composite is higher than that of LM27/SiC6C composite. Because of the greater number of particles per unit area in the LM27/SiC6F composite, the particle distance is shorter if compared to the LM27/SiC6C composite because asperities are generated in the ductile phase. The high contact area of the asperities on the counter surface is responsible for the transfer of stress at the contact points. The wear rate may be due to the plastic deformation of the asperities and enhancement in the disc-to-pin contact area with an increase in applied load from 9.8N to 49N. This generates a high amount of frictional heat between the disc-to-pin contact surface areas. The high frictional heat is responsible for the softening of the pin’s surface.

The wear rate of the soft pin samples increases due to the penetration of some hard asperities into the softening of the pin’s surface [30]. If the increase applied load from 9.8N to 49N, the plastic deformation nature of the material increased during sliding wear due to the micro-crack nucleation tendency of the subsurface as well as deformation and rupture of asperities [12]. In this case, these asperities are removed either by subsurface or by deforming the surface. But in composites where the inter-particle distance is an important variable, controlling the wear performance of otherwise material removal during wear tests at high loads.

Figure 6(a–b) shows that the LM27/SiC6F composite exhibit better wear behaviour than the LM27/SiC6C composite at all applied loads. The wear rate of the composites decreases when the SiCp content is increased from 6 to 9 wt%. A further decrease in the wear rate is observed with different particle sizes as shown in Figure 6(a–b). SiCp has been reinforced to act as a barrier for the greater dislocation density to provide greater load-bearing capacity to the composite material. During the run-in-wear phase, the high wear rate of the LM27/SiC6F composites was observed based...
on the grinding of the asperity on the disc-to-pin contact surface area during the wear test. Figure 7(a–b) depicts the wear behaviour of the composites LM27/SiC9F and LM27/SiC9C at different applied loads. Figure 7(a–b) shows that the wear rate of LM27/SiC9F composites is lower than that of LM27/SiC9C composites. The LM27/SiC9F composite contains harder phases with a larger dominance in the eutectic mixture surface area in the matrix than the LM27/SiC9C composite. The presence of fine-sized particles with a large dominance in the eutectic mixing surface area in the matrix enhances the material’s overall hardness as well as the material’s wear resistance [32]. The protection provided by the harsh abrasive of the composite to the matrix in the initial stages reduces the wear resistance. The protection provided by the hard abrasive of the composite to the matrix in the initial stages reduces the wear resistance. After some sliding distance, an increase in wear resistance in the steady-state is also observed [12–14]. At this stage, the contact area of the asperities increases due to the decrease in the height of the asperities.

The continuous reduction in the wear rate with 9 to 12wt.% SiCp reinforced in the matrix is displayed in Figure 8(a–b). At this stage, the reduction in the wear rate is controlled by a large number of hard particles that successfully transfer the load from the matrix to the particle. However, the wear rate of the LM27/SiC12F composite (Figure 8(a)) during the run-in-wear condition is less sloping than that of the LM27/SiC12C composite (Figure 8(b)). This indicates that the wear resistance of the composites depends on the plastic deformation rate and the breakdown of the asperities Therefore, the LM27/SiC12F composite has shown a lower plastic deformation rate and breaking of asperities than the LM27/SiC12C composite. However, with increasing the applied load from 9.8N to 49N for fine and coarse sized 12wt.% SiCp reinforced composites, the wear rate of the composites increased significantly. The mild to severe wear nature of the composites was observed under high load (49N) conditions in Figure 8(a–b).

During high applied load (49N), sharp edge SiCp (fine shape) can also be inserted into a soft Al–Si matrix affected by frictional heat without any trouble in comparison to coarse shaped SiCp. The interesting point that is the rapid transition in severe running-in wear to mild steady wear nature is observed after that of sliding distance 1500 m. The increased load on the pin surface favours the build-up of wear debris from the worn surface and this worn debris comes into contact with the pin-on-disc in the next rotation during sliding and forms a MML generates. The thickness of the MML is affected by the amount of frictional heat in the contact area. In the oxygen-rich environment, oxides and other alloy elements of materials are come to contact with the frictional heat and form a tribo-oxide-layer. This tribo-oxide-layer protects the wear track surface of the LM27/SiC composite and reduces the wear rate. Similar nature of wear behaviour has also been reported by Kumar et al. [12]. A continuous increase in the wear rate of all composites is observed with increasing load from 9.8N to 49N.

Figure 9 shows a schematic representation of the wear mechanism involved between composite and counter body during the wear testing. The formation of a void was mostly identified by the plastic flow of the matrix around the reinforced one. A void formed very easily around the reinforced particles, but crack propagation was gradual [14]. The depth and void size depend on the size of the reinforced particles, the applied load, and the initiate of void nucleation.

4.4. Coefficient of friction (CoF)

Since the wear rates of all developed composites show similar types of wear behaviour with different values under different operating conditions. To avoid
repetition and confusion with similar data, in this section we have used two contact pressures 0.125 and 0.624 MPa to study the COF for only four composites with 3wt.% and 12 wt.% fine and coarse-sized reinforcements. A fluctuated COF was observed along with the sliding distance (Figure 10). Figure 10 (a) shows the graph of the average COF for the LM27/SiC3F and LM27/SiC3C during the sliding process for a sliding distance of 3000 m. The COF of LM27/SiC3F composites varied inside the range of 0.31–0.33, in the event of the contact pressure of 0.125 MPa. The lower value of COF is due to less interlocking on the sample surface and counter steel disc. From the graph of the LM27/SiC3C composite, it is observed that the COF varies within the range of 0.41–0.43. Although the COF of LM27/SiC3C composites increases to about 0.42 at a contact pressure of 0.125 MPa, it peaks at 0.43 over the next few seconds.

The sudden increase in COF of LM27/SiC3C composites may be due to the crushing of coarse-sized SiCp against counter steel discs at near about 2250–2700 sliding distance (metres). For most tests operating at a contact pressure of 0.125 MPa, the COF remained nearly constant at 0.31. The COF for the LM27/SiC12F composite (Figure 10(b)) was found to be lower than that of the LM27/SiC12C composite. The foremost reason for this was the presence of finer SiC particulates in the LM27/SiC12F composite. The distance from the particle to particle depends on the amount and size of particles that reason less friction at the wear pin surface and the counter steel disc surface. It was described that the COF of the LM27/SiC12C composite was around 0.6 at 0.624 MPa. Although the COF also had some fluctuations up to a full run of 3000 m sliding distance. At contact pressure, the COF the LM27/SiC12F composite increased to 0.53, but most of the time it remained between 0.50 and 0.53.

Figure 11(a–b) shows the graphs of COF for the LM27/SiC3F, LM27/SiC3C, LM27/SiC12F and LM27/
SiC$_{12}$C composites during the sliding process under the contact pressure of 0.624 Mpa. From the graph of the LM27/SiC$_{3}$F composite (Figure 11(a)), it was observed that the COF varied within the range of 0.38–0.44. Under the contact pressure of 0.624 MPa, the value of COF reached almost 0.82 from 1 to 500-metre sliding distance. The fluctuated COF value of 0.8 is during the 3000 m sliding distance run. However, the COF of the LM27/SiC$_{3}$C composite reached a contact pressure of about 0.6 under 0.624 MPa. This is caused by plastic deformations and then easily transfers the stresses at the contact points [34]. For most of the tests operating under contact pressure 0.624 MPa, the COF of the LM27/SiC$_{12}$F and LM27/SiC$_{12}$C composites remained of similar behaviour (Figure 11(b)). However, for 0.624 MPa, the LM27/SiC$_{12}$C composite remained almost constant with an increase in the average COF sliding distance for the coarse SiC$_{p}$ composites.

It was very clear that the load factor has a great effect on the COF because increasing the load also increases the COF. Both LM27/SiC$_{12}$F and LM27/SiC$_{12}$C composites indicated a decrease in COF, with sliding distances ranging from 500 to 3000 m. However, composites with finer SiC particulate exhibited lower COF than composites with coarse SiC particles. For under contact pressure 0.624 MPa, the average COF for any AMC remained almost constant with an increase in sliding distance from 750 to 3000 m. The initial increase in average COF with an increase in frictional heat was attributed to the growth of the oxide film on the surface of the pin. The continuous sliding leads to an increase in the interface temperature between counter surfaces. At these high interface temperatures, atmospheric oxygen causes oxidation of the matrix and disc material causing various oxides to form viz. Al$_2$O$_3$, SiO$_2$ and Fe$_3$O$_4$. These oxide layers help to prevent metal-to-metal contact because these particles have a tendency to attach to the pin surface, increasing wear resistance and lowering the COF value [34]. However, an increase in frictional heat softens the material thus reducing the stability of the oxide film. This exposed fresh material to the contact surface and increased COF values with a further increase in frictional heat. Due to the high thermal stability, low CTE and high stiffness, SiC$_{p}$ behaves like an anti-friction material, and therefore, reduces the COF of the LM27/SiC$_{p}$ composite.

4.5. Analysis of surface morphology of wear track surfaces

Analysis of surface morphology of worn surfaces along with the debris provides a superior confirmation about the wear mechanism involved. The graph of wear rate (Figure 5–8) indicates that all the manufactured composites display similar nature of wear behaviour with different values at different applied loads. However,
composite with fine size SiCp shows superior wear resistance in comparison to the composites with coarse size SiCp at different applied loads. So, to avoid repeatability and confusion, we have analyzed the worn surfaces of the composites with fine size SiCp tested at high load (49N). At the higher load, worn surface morphology exhibits deep grooves, more damaged areas and a large cavity area as shown in Figure 12(a).

These surface morphologies indicated to higher wear rate. It is observed from the wear track that some debris is welded with parent materials. When in the beginning asperity-to-asperity encounters each other they create large scratches on the pin surface. These asperities become blunt and the abrasive grooves are generated at 49N with the increment in the sliding distance. Some debris and narrow grooves are also clearly observed in Figure 12(b). The oxide layer area is a result of the generation of a MML from the worn surface due to the frictional heat effect [12].

The grooves on the wear surface are generated due to hard asperities on the thermally soften matrix which is shown in Figure 12(c). Due to the formation of the oxide layer, lower damaged area and low depth of grooves reveal the lower wear rate (Figure 8(c)) in comparison to the worn surface of the other composites at higher applied load (49N). The oxide layer along with some narrow grooves is observed in Figure 12(d). This oxide layer work as a self-lubricant in nature, it tries to reduce the surface deterioration of the pin sample [14]. Hence, the addition of more amount of fine SiCp reinforced found to be advantageous in improving the wear resistance of composites.

SEM-EDS images showing the presence of different elements on wear tracks of composites with different amount of 3wt.% and 12wt.% fine size SiCp reinforced at 49N load in Figure 13(a,b), respectively. Some elements such as C, O, Al and Si were detected by EDS analysis in the wear track of 3 wt.% fine size SiCp reinforced composite (Figure 13(a)). The more weight percentage of Al is detected by EDS analysis in the wear track which shows the continuous destruction of ductile material in the form of plastic deformation on the wear track. And carbon and silicon elements are present which shows that SiCp present in the matrix. The maximum wt.% of oxygen, Al and Fe elements confirms that some oxide layer presence on wear track during the oxidative wear mechanism which is shown in Figure 13(b).

EDS analysis of these square cover areas revealed a high oxygen content, which suggests the occurrence of strong oxidation due to relatively high temperatures generated during the sliding process. The oxygen-rich tribolayers (Al2O3/Fe2O3) worked as a self-lubricant, which improves the wear resistance of composite materials. These oxygen-rich tribolayers (Al2O3/Fe2O3) is evidenced. These types of tribolayer materials explained as a self-lubricant by many researchers [13,14].

5. Conclusions

(1) Microstructural refinement of the Si morphology in Al–Si Alloy (LM27) can be altered with the addition of SiCp. The thermal gradient at the
interface is responsible for the nucleation of aci-
form eutectic silicon.
(2) The composite with LM27/SiC12F has less particle
to particle distance in the matrix as compared to
coarse particle so; an increase in capacity to trans-
fer the load from particle to matrix reduces the
wear rate of the composite by increasing the
amount of fine size SiCp.
(3) Microhardness, COF and wear resistance of AMC
showed improvement with an increase in
reinforcement level and concentration of fine
SiCp.
(4) For conditions with under contact pressure of
0.125–0.624 MPa, LM27/SiC12F composites
showed lower COF than other composites.
(5) At the beginning of the wear test, asperity-to-
asperity meets each other and creates large
scratches on the pin surface. In the continuous
sliding process, these asperities become blunt by
breaking the touch point of asperities due to a
higher concentration of shear stress acting on the
softer transferred materials.
(6) At high loads, under the influence of large strain,
air-induced de-cohesion at particle-matrix inter-
faces nucleate the crack network. This phenomen-
ological description leads to the formation of
MML in composite material.
(7) SEM-EDS analysis of wear tracks revealed that the
oxygen-rich tribolayers (Al2O3/Fe2O3/MgAl2O4/
MgFe2O4) worked as a self-lubricant which
improves the wear resistance of fine size SiCp
reinforced composite materials.

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Availability of data and materials
The datasets generated during and/or analysed during
the current study are available on request from the corresponding author.

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