Shear Compaction Processing of SiC Nanoparticles Reinforced Magnesium Composites Directly from Magnesium Chips

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Abstract

In this work, a novel synthesis technique, shear compaction processing (SCP), is developed to produce magnesium (Mg) based composite from Mg machining chips and nano-sized silicon carbide (SiC) particulates reinforcement. During the process, Mg machining chips are first loaded into the container and slightly compacted, and then SiC nanoparticles are added on top of the compacted Mg chips. A scroll-faced rotating tool with a designated diameter is plunged into the Mg chips at a selected spindle rotation speed and feed rate. Due to the huge amounts of heat generation, the softened materials are compressed and synthesized to form a nanocomposite disc eventually. The results show that the SCP process is a feasible solution for producing a void-free bulk nanocomposite material directly from Mg chips and SiC nanoparticles in a single step. The microstructural analysis with optical microscopy and scanning electron microscopy clearly shows the significant grain refinement in the produced nanocomposite disc. The reinforcement of nanoparticles is uniformly distributed in the mixed domain. Good interfacial bonding is captured between the matrix and SiC nanoparticles. The produced specimen has a higher hardness than the parent material, and also exhibits excellent wear resistance. The significant improvement is attributed to the resulting fine-grained microstructure and homogenous distribution of SiC nanoparticles throughout the matrix material of Mg, due to severe plastic deformation and mixing during the SCP process.

Keywords: *Shear Compaction Processing, SiC, Magnesium, Nanoparticle, Composites*

1. Introduction

Due to the high specific strength, stiffness, and thermal conductivity, magnesium (Mg) and its alloys have been widely used in the automobile, aerospace and electronics industries instead of conventional materials, such as aluminum alloys, steel and plastics [1]. However, applications of Mg alloys are limited due to the poor formability and ductility at room temperature [2]. Also, the low hardness and poor wear resistance of the Mg alloys confine the broader potential applications [3]. These limitations, however, can be avoided by using harder and stiffer ceramic particulates as reinforcements, which can improve material mechanical properties [4]. The Mg-based metal matrix composites (MMCs) produced with silicon carbide (SiC) particles have better microstructures and mechanical properties at both room and elevated temperatures [5].

In general, Mg or Mg alloy based MMCs reinforced with SiC particles have been produced by powder metallurgy [6], stir casting [7], squeeze casting [8], spray deposition [9] and ultrasonic method [10]. Usually, structural defects, such as porosity, particle cluster, interfacial reactions, and oxide inclusions, arise in the aforementioned conventional techniques [9].

In order to overcome these types of pitfalls, a solid-state method of friction stir processing (FSP) has been successfully adopted to fabricate defect free composites [11]. It has been proven that FSP can produce in situ nanocomposites with superior mechanical properties, because fine and stable reinforcements are uniformly distributed in the matrix with good interfacial bonding [12]. In the past decades, many authors have used FSP to produce MMCs successfully, e.g. Cu based composites with SiC particles [13], and Al based composites with reinforcing particles of SiC [11] or alumina (Al_2O_3) [14]. In addition, the fabrication of Mg-based composites with SiC particles has been investigated by many other researchers [15–19].

Asadi et al. [1] experimentally investigated the effects of the particle type and number of process passes in the fabrication of Mg-based nanocomposites by FSP. It was observed that the Al_2O_3 particles were agglomerated in the form of clusters at different locations in the matrix during FSP. These clusters uniformly distributed in the stirred zone. Although the SiC particles were not uniformly distributed in the Mg matrix, they did not stick together to form the clusters as those of Al_2O_3 particles. Compared to the original hardness of as-cast AZ91 alloy, the average hardness of

nanocomposites after two passes were increased from 63 HV to about 115 HV and 105 HV reinforced by SiC and Al₂O₃ particles, respectively. Asadi and his co-workers [20] also found out that equiaxed grains were observed and the grain size was significantly decreased compared to the base material due to the dynamic recrystallization (DRX) resulted from severe plastic deformation and the high process temperature. Lee et al. [18] incorporated nano-sized SiO₂ into AZ61 Mg alloy to form bulk composites with FSP. It has been found that the hardness of the composite was up to 75% higher than that of the base material. Faraji and Asadi [21] identified that the wear resistance was significantly increased for the AZ91/Al₂O₃ composite produced by FSP, and the wear mechanism was changed from severe wear in the as-received AZ91 to mild wear in the FSPed specimens. The results also showed that the higher ratio of the rotational speed to traverse speed was beneficial to achieve better particle distributions.

In this study, a novel synthesis technique of shear compaction processing (SCP) is proposed as another energy-efficient process to directly convert Mg chips and SiC nanoparticles into Mg/SiC nanocomposite. It follows the principle of the friction stir processing and friction stir extrusion (FSE), in both of which a rotating tool causes local softening of material through heat generated by friction and plastic deformation [22]. FSE is a novel extrusion method for the fabrication of metal wires from chips. It was invented and patented in 1993 by The Welding Institute [23]. In this process, extra energy consumptions are not required for re-melting which is necessary for the conventional synthesis processes. Recently, Tang and Reynolds investigated the synthesis process of AA2050 and AA2195 aluminum chips by FSE [24]. They observed that the fabricated wire microstructure consisted of fully equiaxed, recrystallized grains, and the hardness was generally homogeneous across the wire transverse cross-section. Abdi-Behnagh et al. [25–27] found that the rotational speed of the tool had an important role to achieve defect free wires in the FSE process of aluminum alloy AA7227 and pure Mg chips.

In this work, the SCP method is tested experimentally on pure Mg chips and SiC nanoparticles with various rotational speeds, aiming at producing fully-consolidated composite discs. The microstructural and mechanical properties of the fabricated discs are also investigated with different microscopic analysis and mechanical tests.

2. Principles of Shear Compaction Processing

The principle of shear compaction processing is illustrated in Fig. 1. As shown in Fig. 1a, the experimental set-up for the SCP process consists of two major components: a container with a cylindrical central hole and a stirring tool. The container has a backing plate mounted with three tension bolts. This container works as a die and is fixed on the table of a computer numerical control (CNC) vertical milling machine during the process. The stirring tool has a scrolled face which can facilitate the stirring action of the tool. In the first step of the process, the machining chips are filled into the container, and then are slightly compressed along the feed direction at room temperature. After the compaction of chips, the SiC particles are applied on top of the compacted chips (shown in Fig. 1a). Then, a rotational stirring tool is plunged into the container with a specific feed rate and rotational speed. A huge amount of friction-induced heat is generated due to the rotation and translation along the feed direction, which softens and fully consolidates the chips in the cavity. The scrolled face of the tool enhances the mixing of chips and SiC particles. Based on the amount of loaded chips, a designated plunge stroke of the stirring tool was firstly determined, which also defined the final thickness of the fabricated composite disc. As shown in Fig. 1b, the process is to be terminated when the stirring tool reached the plunge stroke.

Fig. 1 Schematics of SCP experiment: (a) before plunging; (b) completely compressed.

3. Experimental Procedure

The materials used in this work were clean and dry Mg chips produced by a planning machine of pure Mg ingot without using any lubricants. The chemical composition of pure Mg ingot based on mass percentage was Mg-Al < 0.01-Cu 0.006-Mn 0.03-Zn 0.005-Ca 0.005 and Sn < 0.002. The average grain size was about 1 mm in the as-received pure Mg ingot, while it was refined to 14 μm after machining process, as observed in Fig. 2. The reinforcement was SiC particles supplied by Nanosany Corporation^{TM}. The transmission electron microscopy (TEM) image and specifications of the SiC particles are shown in Fig. 3 and Table 1, respectively. For all the experiments, a constant mass fraction of 25% for the SiC nano-particles was applied into the Mg substrate.

Fig. 2 Mg chips and microstructures. (a) Mg machined chips; (b) Optical micrograph of the base Mg ingot material; (c) Optical micrograph of the machined chips.

Fig. 3 TEM micrograph of SiC particles

Table 1. Specifications of the SiC nanoparticles.

Material	Purity	APS (nm)	Color	Bulk Density (g/cm ³)	True Density (g/cm ³)	Morphology
Silicon Carbide Nanopowder (SiC, Beta)	99+%	45-65	Grayish white	0.05	3.216	Cubic

The process was performed on a CNC vertical milling machine. The major components were made from H13 tool steel. To avoid sticking of Mg, all the components were treated with the case hardening and subsequent nitriding processes. The stirring tool was 20 mm in diameter. In the container, the cavity was 50 mm in depth and 20.6 mm in diameter, which had a clearance fit of 0.3 mm between the tool and the container wall. The experimental conditions are listed in Table 2. A constant axial feed rate of 20 mm/min was chosen; and three clockwise rotational speeds were tested at 250, 355, and 500 rpm, respectively. The plunge stroke for the stirring tool was set to 45 mm, aiming at 5 mm thickness discs.

Table 2. Experimental conditions

Experiment	Plunge rate (mm/min)	Rotational speed (rpm)
Test-1	20	250
Test-2	20	355
Test-3	20	500

After SCP, the fabricated composite discs were sectioned following standard metallographic techniques. The disc top surfaces were ground against 600-grit silicon carbide paper and cleaned with alcohol. For microstructural observations, the samples were etched using acetic-picral (5mL acetic acid, 6 g picric acid, 10mL water and 100 mL ethanol) at room temperature. The microstructure was mainly examined through OLYMPUS PMG3 optical microscopy (OM). Image analysis software was used to calculate the volume fraction of SiC particles from SEM micrographs. The chemical composition of local areas on the disc cross sections was analyzed by VEGA II TESCAN-XMU scanning electron microscope (SEM) equipped with an energy-dispersive X-ray spectroscopy (EDS) analysis system. The phases presented on the disc cross sections were studied using a

Philips PANalytical X'Pert X-Ray Diffraction (XRD) System with 45 kV and Cu-K α radiation. The diffraction data were collected over a range of 20°-110°. The phases were determined using the X'pert Highscore software.

Vickers microhardness tests were performed on the disc cross section in two directions, radial and through the thickness. Microhardness test was carried out by applying a 50-grf load for 5 sec. The wear behavior of the produced sample was also evaluated with a Pin-on-Disc wear tester with a disc made of hardened En-32 steel (13 mm in diameter). The dry wear tests were carried out on the composite disc (pin) top surface at room temperature with a normal load of 10 N, a sliding velocity of 0.2 m/sec, track radius of 7.9 mm, for a total sliding distance of 500 m. The worn surfaces after the wear tests were examined using SEM.

4. Results and Discussions

4.1. Composite Disc Form Integrity

Composite discs were produced with all the experimental conditions. The quality of the discs was examined on the cross section of composite discs produced in each experimental condition. Fig. 4 shows the SEM images of cross sections of the composite discs produced by SCP at rotational speeds of 250, 355 and 500 rpm (Test 1-3), respectively. The right half of the composite disc cross sections are shown in Fig. 4a-b for the specimens produced in Tests 1 and 2, while the whole cross section of the composite disc is shown in Fig. 4c for Test-3. For all specimens, there are generally two domains on the cross section: a bright one and a dark one. The bright domain is the center top domain enclosed by a curved boundary, which is just the synthesized nanocomposite. The dark domain is beyond the boundary, which consists of the consolidated pure Mg chips. The different colors in the gray scale between the two major domains are attributed to the distribution and concentration of SiC particles, i.e. the domain with SiC is brighter, while it is darker with Mg only. This is because SiC particles have greater atomic weight than pure Mg. As a result, more backscattered electrons can be emitted to form the brighter domain on the SEM image.

Fig. 4 SEM images of SCP samples for test conditions of: a) 250 rpm; (b) 355 rpm; (c) 500 rpm.

As shown in Fig. 4a, for the specimen produced in Test-1, many macro- and micro-cracks were observed in the dark domain at the bottom and right boundary, where the major fiction occurred during the SCP process. Chip residuals can even be seen on the cross-section near the disc top surface. Thus, synthesis of chips was incomplete at the spindle speed of 250 rpm. In addition, bandarea can be

seen within the bright domain as shown in Fig. 4a, while the brightest band is suggested to be a rich SiC particle area. The banded area indicates the distribution of SiC particles were not uniform, which is not ideal. By increasing the rotational speed to 355 rpm, there were still cracks on the cross section as shown in Fig. 4b, but they were fewer and smaller than those of the specimen produced with 250 rpm. In addition, there was no chip residual at all, and the composite domain was deeper and wider in Fig. 4b. The banded area cannot be seen anymore, but there was still a domain along the boundary with color in between the bright composite domain and the dark Mg only domain. Nevertheless, the synthesis of chips was improved by the increase of rotational speed, and the distribution of SiC particles in the composite domain was much more uniform. Fig. 4c shows the cross section of a disc produced in Test-3. Cracks can hardly be seen on the cross-section. No chip residual can be seen in the composite domain, and the color in the composite domain was very uniform. The cross-sectioned view clearly shows that a void-free disc was directly formed from Mg chips and SiC nanoparticles in a single step at tool rotational speed of 500 rpm. By comparing the SEM images of the samples, a higher ratio of tool rotational speed to vertical axial feedrate was beneficial to improve the consolidation of Mg chips, synthesis of the composite. It can also promote the deeper and wider dispersion of the SiC particles. These changes were mostly attributed to the greater heat generation with a higher ratio, which is similar to the findings observed in Ref. [21].

4.2. Chemical Characterization

The chemical composition of the composite was characterized on the cross section of a composite disc produced in Test-3 by X-ray elemental compositional imaging and XRD. Fig. 5 shows an SEM image of the nanocomposite domain and the X-ray mapping images for the distribution in the same area of three elements of Mg, Si, and C. The results show that these elements were mixed uniformly in the composite domain. The process is found to be capable of recycling of Mg chips directly, and synthesizing them into a nanocomposite disc with SiC nanoparticles in a single step.

Fig. 5: SEM image of the nanocomposite domain (a) on the cross section of the disc produced in Test-3 and the X-ray mapping images at the same area for (b) Mg- K_{α} , (c) Si- K_{α} , and (d) C- K_{α} .

Fig. 6 shows the XRD patterns of the original SiC powders and Mg/SiC composite. As shown in Fig. 6a, all the peaks were for SiC, so the purity of SiC powders was verified. From Fig. 6b, the presence of SiC in the composite was apparent. Special attentions must be given to any reaction products at the interface between SiC particles and the Mg matrix. For instance, Li et al. [28] synthesized AZ91 Mg alloy reinforced with SiC particulates using a powder metallurgy technique and mechanical alloying process, and found that Mg_2Si appeared in the Mg/SiC composites and can cause embrittlement. Mg_2Si was also formed in the Mg/SiC composites from a high-intensity ultrasonic cavitation-assisted casting experiment of AZ91 Mg alloy and nano-sized SiC particles by Lan et al. [10]. The formation of Mg_2Si was believed to originate from the reaction between the SiO_2 on the

surface of SiC and the Mg matrix. The same chemical compound was also found in the Mg based nano-composites fabricated by friction stir processing of AZ61 Mg alloy and SiO₂ particles[29]. For pure Mg reinforced by SiC nanoparticles, Ferkel and Mordike[30] suspected the formation of Mg₂Si led to the reduction in hardness. In this work, the XRD analysis did not show any trace of interfacial reaction compound between the ceramic particles and Mg matrix, which can be considered to be a merit of the SCP technique over other traditional processes.

Fig. 6 XRD analysis: a) SiC powders; b) Mg/SiC composite in Test-3.

4.3. Microstructural Analysis

The evolution of the grain structure was examined for the Mg/SiC nanocomposite after SCP and compared to the as-received material. Optical micrographs in Fig. 7 show the microstructure at different regions of the consolidated Mg/SiC nanocomposite as labeled on the disc cross section. The

Domains a, b, and c are the center of the nanocomposite, the transition between nanocomposite and non-composite domain, and the non-composite domain, respectively. The nanocomposite domain was surrounded by the non-composite, which consisted of fully consolidated Mg chips and recrystallized Mg grains without any particles.

As shown in Fig. 7a, very fine equiaxed grains were observed in nanocomposite domain. The uniformly distributed dark dots are SiC particles. A particle-reinforced microstructure can be seen without any obvious grain boundaries or grain structures. There is a noticeable interface between the composite and non-composite domain as shown in Fig. 7b, as a result of the sudden drop in SiC amount from the composite to thenon-composite domain. In the non-composite side of the interface, sheared Mg grains can be seen along the tangential direction of the interface. In Fig. 7c, clear grain boundaries and fully-recrystallized grain structures were observed with some twinning laminae in a few grains. The average recrystallized grain size is 18 μm for this region, which is much greater than the grain size in the composite domain as shown in Fig. 7a. By comparing the different microstructures from the composite to thenon-composite domain, the change of the microstructure and the grain size are suggested to be mainly attributed to the existence of ceramic particles.

Fig. 7 Microstructure appearance of the disc sample: a) OM at the center of nanocomposite domain; b) OM of the transition zone; c) OM in the non-composite domain.

In the present work, grain growth restriction was attributed to the presence of SiC particulates with the coupled effects of: (i) capability of ceramic particles to nucleate Mg grains during recrystallization, and (ii) confined growth of recrystallized Mg grains as a result of pinning effect by reinforcement. In MMCs, the basic principles of the ability of ceramic particulates have been reported by many other researchers on the new recrystallized grains nucleation and the confinement of grain growth [31,32]. During hot deformation processes, dynamic recrystallization usually occurs and causes the formation of very small grains in the thermo-mechanically processed workpiece material. As the onsite of DRX, original grains are broken, and a large number of low-angle misoriented grain boundaries are created in the severe plastic deformed material, where highly favored places for recrystallization nucleation are produced. Subsequently, low angle boundaries transform to high angle ones and the nucleation of new grains takes place at preferential sites. The fine nuclei of grains start to

grow, and a microstructure composed of fine equiaxed grains is created. Furthermore, fine-grain microstructure has a lot of grain boundaries and consequently a large stored energy. So, it is inherently susceptible to grain growth during the high-temperature deformation. By introducing another phase in the form of ceramic particles, significantly more nucleation sites are available for DRX, and dispersion of ceramic particles also helps to confine the grain growth of huge amounts of new recrystallized grains, i.e. the pinning effect.

The grain size resulting from the pinning effect in a composite can be determined by the volume fraction of particles and the particle size. This relationship can be represented by the Zener relation [33]:

$$d_s = \frac{4\alpha d_p}{3V_p} \quad (1)$$

where d_s is the grain size, V_p is the volume fraction of nano-sized particles, d_p is the mean particle diameter, and α is a material constant. From this equation, denser distribution, as well as the smaller size of SiC particles, confine the grain growth. Nano-sized particles are particularly favorable for grain boundary pinning, and hence can result in smaller grain sizes in nanocomposites [34].

To illustrate the ultra-fine grain structure in the composite domain, higher magnification SEM at this domain was captured and is shown in Fig. 8. Fig. 8 shows the significant grain refinement and relatively uniform distribution of SiC particles in the matrix. The grain size was decreased to less than one micron in the composite, while it was about 14 μm in the machined chips before SCP. The result of grain refinement achieved by SCP is comparable to other fabrication methods of Mg-based nanocomposites. For instance, a refined grain size to 0.5–2 μm was observed in the Mg/SiC nanocomposite produced through four-pass friction stir processing [29].

Fig. 8SEM image of particle distribution in the composites domain (15 kx).

Moreover, as seen in Fig. 9, sporadic small amounts of local clusters were observed. This can be attributed to the imperfect stirring action of the tool. For better distribution of reinforcement, further work is required on redesigning of the stirring tool or selection of the process conditions, etc. Nevertheless, the cluster size found in this study was satisfactory compared to the particle size. The cluster chemical composition was characterized by EDS. Fig. 10 presents the results of EDS analysis

at the region marked by a cross in Fig. 9. It was suggested that there was no partial oxidation for SiC particles, which has been observed in other fabrication techniques of Mg-based MMCs[10]. There were a high peak and higher fraction of Mg in the EDS result for SiC clusters. This is because the detection zone determined by EDS beam was bigger than the average size of SiC clusters. Thus, the EDS peaks for SiC clusters inevitably included compositional information of Mg matrix near particles.

Fig. 9 High magnification SEM image of SiC particle clusters in the composite (100 kx).

Fig. 10 EDS composition analysis at SiC particle cluster as labeled by the cross in Fig. 9.

The volume fraction of the SiC nano-particles in the specimen cross-section was evaluated using image processing software and was found to be around 8%, 11% and 18% for 250, 355 and 500 rpm specimens, respectively. The low volume fraction of SiC particles for the specimens produced from Tests 1-2 was mainly caused by the loss of SiC particles from the sample preparation procedures. In order to illustrate the composite structure produced by SCP better, more SEM images of interfacial bonding between matrix and SiC particles are presented in Fig. 11. The pores and good bonding between the SiC particles and Mg matrix are marked in these images. As shown in Fig. 11a, the results show that poor bonding was found between the matrix and particles in Test-2 with a rotational speed of 355 rpm. In addition, SiC particles are surrounded by a relatively large amount of pores. However, the interfacial bonding between the matrix and particles was significantly improved by increasing the rotational speed to 500 rpm in Test-3 as shown in Fig. 11b. This obviously indicated that the pores between the Mg matrix and the SiC particles were gradually diminished upon increasing the ratio of rotational speed to vertical axial feedrate. The other reason for this phenomenon could be that more uniform dispersion of SiC particles by increasing the rotational speed prevented the debonding induced by the clustered nano-particles[13].

Fig. 11 SEM images of interfacial bonding between matrix and SiC particles in test conditions of: a) Test-2; b) Test-3.

4.4. Microhardness

SiC particles have two effects on hardness. The first arises from its hard nature, and the other is determined by the resulting grain refinement. The relationship between room-temperature hardness (H) and the average grain size (d) is represented by Hall-Petch equation [35]:

$$H = H_o + kd^{-\frac{1}{2}} \quad (2)$$

where H_o and k are material constants. From Eq. 2, smaller grain size leads to higher hardness. As shown in Fig. 8, the grain size in the nanocomposite domain produced by SCP was less than 1 μm , while the grain size of Mg ingot was about 900 μm . Thus, the room-temperature hardness of the composite disc should be increased drastically from the Mg ingot hardness, due to the linear relationship between room-temperature hardness and the square root of grain size.

Fig. 12 shows the hardness profiles in the horizontal and depth directions of a disc cross-section from the Vickers microindentation measurements. The hardness data for the horizontal profile was obtained at 1 mm below the top surface for all the specimens, while those of the depth profile was obtained at the center of disc specimens. All the hardness values were significantly increased from

that of the Mg ingot (25 HV_{0.05}). The microhardness value was constantly about 80 HV_{0.05} in both directions for the composite disc produced in Test-3 with a rotational speed of 500 rpm. This hardness increase was much higher than the Mg ingot by about 220% and those produced in the other two conditions by about 60%. The maximum hardness achieved in this work was much higher than the value of 55 HV reported by Saravanan and Surappa on fabrication and characterization of pure magnesium-30 vol.% SiC_p particle composite by melt stir technique [7]. Although higher hardness values were reported to range from 60 to 120 HV in Mg/SiC composites by using Mg alloys as matrix [10,11], it is noted that Mg alloys are significantly harder than pure Mg investigated in this work. The microhardness of the as-cast AZ91 alloy was about 63 HV [10], while it was only 25 HV for Mg ingot and limited the hardness of the resultant composites in this study. Relatively low hardness for Test-3 specimen was located in the non-composite domain on both sides of the composite disc, which was still 80% harder than Mg ingot and higher than the hardness value of 43 HV achieved in non-composite by Saravanan and Surappa [7].

The lower hardness of the disks produced in Tests 1-2 were attributed to the formation of cracks and the weaker interfacial bonding between SiC particles and Mg matrix. In addition, it can be caused by the loss of particles during the surface preparation for hardness tests. The hardness values were uniformly distributed inside the composite domain for the composite discs produced from Test 1 and Test 3, while the composite disc produced in Test-2 had completely different hardness values in the horizontal and depth directions, which was attributed to a band of agglomeration of SiC particles occurred between 1.3 and 4 mm in depth from the disc top surface. The significant hardness increase by the SiC particles in Test-3 specimens was a result of the adequate distribution of SiC particles and great bonding strength. It indicated that the rotational speed is an important factor that influences the distribution of ceramic particles in the matrix during SCP.

Fig. 12 Measured hardness profile along: (a) horizontal direction; (b) depth direction

4.5. Wear Behavior

In the MMCs, the fine equiaxed grains and consequently higher hardness positively affect the wear resistance. The increased hardness prevents the material from the severe adhesive wear and consequently decreases the rate of wear. The relationship between the total volume of wear debris produced and room-temperature hardness is represented by Archard's law [36]:

$$Q = KWL/H \quad (3)$$

Where Q is the total volume of wear debris produced, L is the sliding distance, K is the coefficient of wear, W is the normal load, and H is the hardness of the softer material in the two contacting metals. From Eq. 3, the higher the hardness is, the lower the total wear volume would be. This relationship is valid to both the adhesive and abrasive wears[36].

The wear behaviors of the produced composite discs were evaluated in the dry wear tests. The friction coefficients of Mg ingot and the composite discs were evaluated, and the worn-out surfaces were characterized with SEM images. Fig. 13 represents the variation of friction coefficient with the sliding distance up to 500 m for Mg ingot and SCP specimens. The average friction coefficient of Mg ingot is 0.405, while it was found to be 0.367, 0.358, and 0.294 for the three composite disc produced by SCP with rotational speed of 250, 355, and 500 rpm, respectively. So, the best result was observed in the composite disc produced with rotational speed of 500 rpm.

In the condition of weak bonding as observed in Tests 1-2, the ceramic particles are easily separated from the matrix and results in increasing of contact area between the steel disc and soft Mg substrate. The smaller coefficient of the nanocomposite produced by Test-3 may be attributed to the reduction of actual contact area between the pin and composite surface due to the presence of ceramic particles and also higher hardness values which weakens the cohesion behavior of the contact[37]. On the other hand, the presence of SiC particles in fabricated composites converts the wear mode from a two-body condition to three-body one. In three-body abrasive wear mechanism, since the trapped hard reinforcements particles are not held rigidly, they are free to roll and slide on the specimen surface or penetrate into it causing extra scratches. This increases the rate of wear further by acting as abrader elements[38]. Therefore, the higher number of removed particles in Tests 1-2 compared with Test-3 leads to further material loss and consequently higher wear rate by the three-body abrasion mechanism.

Fig. 13 The friction coefficient as a function of sliding distances: (a) raw Mg ingot surface; (b) top surface of the disc-250 rpm; (c) top surface of the disc-355 rpm; (d) top surface of the disc-500 rpm.

Since the matrix material is much softer than the pin material, the slider could penetrate and cut the surface, causing plastic deformation and loss of material. The abrasive wear is characterized

by the ploughed grooves inside the worn surfaces and generation of wear debris because SiC particles were pulled out from the composite domain during the sliding. Fig. 14 shows the worn-out surfaces of fabricated composites in two magnifications. As shown in Fig. 14, long abrasion grooves and a number of pits are observed which formed parallel to the sliding direction. The pits were found to be deeper and larger for both the Tests 1-2 conditions. In addition, substantial wear debris was observed inside the pits and their surroundings. Fig. 14e and d show the worn-out surface of the composite for Test-3. It shows abrasion grooves parallel to the sliding direction as was observed for Tests 1-2. However, the abrasion grooves were found to be shallower than those of Tests 1-2. In addition, a small number of surface pits, relatively finer than those of the worn-out surface of Tests 1-2, can be observed. These imply relatively lower apparent plastic deformation. In general, the shallow and narrow abrasion grooves and small pits are associated with low rate of material loss. The lower wear rate of the composite disc of Test-3 is attributed to the suitable bonding of SiC particles whose hardness is much greater than that of the substrate and provides protection during wear by strengthening the Mg matrix. This confines deformation and resists the penetration and cutting of the slider into the surface.

a) 250 rpm-100x

b) 250 rpm-600x

c) 355 rpm-100x

d) 355 rpm-600x

e) 500 rpm-100x

f) 500 rpm-600x

Fig. 14 SEM images of the worn-out surfaces of the SCP specimen in all the conditions (Tests 1-3) in different magnifications of 100x (a, c, e) and 600x (b, d, f). The zoom-in SEM images were captured in the area labeled with dotted box in the corresponding lower magnification images on the left.

5. Conclusions

In this study nanocomposite discs were successfully produced by employing a novel shear compaction method. The form integrity, microstructure, chemical composition, microhardness, and

wear behavior of specimens were experimentally evaluated and analyzed. The experimental results show that increasing of the rotational speed to 500 rpm was the effective method to produce defect-free composite discs. A good dispersion of nano-sized SiC particles was achieved in the composite disc produced by the rotational speed of 500 rpm. The microstructural analysis indicated that SiC particles were well distributed in the Mg matrix and good bonding with the Mg matrix was generated. The SCP with SiC particles obviously increased the microhardness of the substrates even in unsound samples. The maximum microhardness value for Mg/SiC composite was about 80 HV, while that of the as-received Mg ingot was about 25 HV. The composites exhibited better wear resistance than Mg ingot due to their superior load bearing capacity, which protects against metal-to-metal contact with the steel counterface during sliding. Furthermore, using a rotational speed of 500 rpm increased the wear resistance significantly and changed the wear mechanism from severe wear in the low rotational speeds to mild wear.

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