



Preparation and properties of Ti_p/Mg-RE composites through laser melt injection

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ABSTRACT

Hardness and wear resistance of magnesium alloys should be improved without reducing mechanical properties to realize their widespread application. By adding metal particles into magnesium instead of ceramic particles can reducing the influence caused by large difference of thermophysical and mechanical properties between ceramics and Mg. The existing methods for Mg based composites are time-consuming, cause agglomeration, and not suitable for preparing complex surfaces. In this study, Ti particles were injected into Mg-RE alloy through laser melt injection. Results show that the prepared Ti_p/Mg-RE composites with large surface area have no obvious welding defects in microstructure. Ti particles have good metallurgical bonding with the Mg-RE matrix. Microhardness of the Ti_p/Mg-RE composites is 100% higher than that of the Mg-RE matrix, at a cost of reducing partial mechanical properties. Moreover, the composites showed a stable friction coefficient and running-in stage. The Ti_p/Mg-RE composites show excellent comprehensive properties in a summary.

1. Introduction

Mg and its alloys show excellent physical and chemical properties, such as specific strength, damping capacity, and non-toxicity, and hence are considered as promising materials for application in the automotive, aircraft, aerospace, and biomedical industries [1–5]. However, Mg and its alloys exhibit poor mechanical properties and low wear resistance, which greatly limit their practical applications [6–8]. Over the past few years, considerable progress has been made in improving the properties of Mg alloys through plastic processing and surface treatment [9–12]. Normally, plastic deformation can significantly improve the strength of Mg alloys, while the influence of plastic deformation on the hardness and wear resistance is limited [13,14]. Traditional surface treatment methods, such as the deposition of hydrothermal treatment coatings, are not suitable for improving the hardness and wear resistance of Mg alloys as such coatings form a poor interface with the Mg matrix [15,16]. Thus, achieving good interface between the Mg matrix and surface coating, and improving the hardness and wear resistance of Mg alloys

simultaneously have received significant attention [17–19]. Reinforcing the Mg matrix with various particles is a feasible method to improve the hardness and wear resistance of Mg alloys [20,21]. Ceramic particles, such as oxides, carbides, borides, and nitrides, are often used as reinforcing particles in Mg alloys [22–24]. Some ceramic particles showing good interface relationship with the magnesium alloys can enhance the mechanical properties of the magnesium alloys matrix [25,26]. While some ceramic particles would dramatically reduce the ductility of the Mg matrix owing to their brittleness and lacking good interface relationship [27,28]. Metallic particles possessing high plastic deformation capacity and appropriate hardness, such as Ti, Ni, and Mo, can be used to reinforce the Mg matrix to maintain its ductility [28–31]. Ti particles are appropriate reinforcements for Mg matrix composites, for three reasons. First, Ti and Mg are compatible in terms of the crystal structure as they both possess the hexagonal close-packed (HCP) crystal structure [28]. Second, Ti exhibits better biocompatibility and corrosion resistance than the Mg matrix. This is advantageous for their biomedical applications [32]. Third, Ti metal particles exhibit better plastic deformation ability

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than typical ceramic particles, and hence can render Mg matrix composites suitable for structural applications. Moreover, Ti does not mix with Mg either in the solid or liquid state to form a solid solution or any intermetallic compounds [28]. However, it can interact with the elements in the Mg alloy, such as Zr, Zn and some rare earth (RE) elements, to form metallurgical bonds. Therefore, Ti metal particles evenly distributed into the Mg-RE matrix to form a harder and better wear-resistant barrier is a feasible solution in order to improve the wear resistance of Mg-RE alloys.

Researchers have investigated on Ti particles reinforced Mg matrix (bimetallic) composites prepared using different methods. Lu et al. [33] prepared metal-metal (bimetallic) composites by mechanical ball milling Mg, Al, and Ti powders followed by cold pressing sintering and hot extrusion at 400 °C to obtain Mg10.3% Ti and Mg5% Al10.3% Ti alloy powders. As Ti is not solid soluble in Mg, it inhibits the grain growth, resulting in grain refinement. Umeda et al. [34] prepared Mg composites reinforced with Ti particles using the powder metallurgy method and then extruding the composites. The extruded Ti-reinforced Mg composites showed significantly improved tensile strength and elongation as compared to the pure Mg powder. Braszczynska-Malik et al. [35] prepared AM50 Mg matrix composites reinforced with Ti particles with a size of less than 50 µm using the stir casting method. The as-prepared composites showed finer grains, and hence better mechanical properties than the Mg matrix. Ouyang et al. [32] prepared Ti/Mg metal-metal composites (MMCs) by spark plasma sintering. The effects of the Ti/Mg MMCs on the proliferation and differentiation of Saos-2 cells into the femoral condyle of rats were better than those of pure Ti. Wang et al. [31] used commercial AZ91D as the matrix and the TC4 (Ti-6Al-4V) powder with a particle diameter of 15 µm as the reinforcement phase to prepare deformable TC4 particle-reinforced Mg matrix composites by stir casting followed by extrusion. The deformable TC4 particles promoted the nucleation of dynamic recrystallization during the extrusion to improve the mechanical properties of the composites. Dinaharan et al. [28,29] prepared Ti particle-reinforced hot-rolled AZ31 Mg plates through friction stir processing (FSP). The addition of Ti particles resulted in an increase in the ultimate tensile strength (UTS) of the AZ31 matrix from 226 (0% Ti particles by volume) to 283 MPa (21% Ti particles by volume).

The powder metallurgy method generally yields small samples and requires a time-consuming ball milling pre-treatment. On the other hand, it is difficult to control the powder distribution in the stir casting and FSP methods. In addition, powder agglomeration is observed in these methods. Laser processing with powder feeding is a flexible method to control the powder distribution. Recently, the laser melt injection method has been proposed to prepare various composites, such as Ti matrix, Fe matrix, and Al bronze matrix composites [36–39]. The laser melt injection method is similar to the laser cladding process. The biggest difference between the two laser process methods is that the particles are injected into the ‘tail’ part of the molten pool in the laser melt injection method to avoid the melting of the particles [40,41]. The laser melt injection process only melts the matrix, while the laser cladding process melts both the matrix and injected particles. The application of the laser melt injection method for preparing Mg-based alloys has rarely been reported in the literature. Moreover, the reports on the use of metals as the reinforcement phase to obtain MMCs through laser melt injection are also rare. This is mainly because Mg alloys show poor weldability because of their physical and chemical characteristics [11,22,42,43]: low density, low melting point, high thermal conductivity and coefficient of thermal expansion, strong chemical activity, easy oxidation, etc. Therefore, pores and other defects are often observed in Mg alloys prepared using the laser method.

In this study, Ti particles with an appropriate size were distributed on the surface of a Mg-RE matrix to improve its hardness and wear resistance. The laser melt injection processing method was used to prepare the Ti_p/Mg-RE composites to achieve good metallurgical bonding interface. The stability of the molten pool was monitored in real

time to analyze the metallurgical process during the laser melt injection of the Ti_p/Mg-RE composite. The phases and microstructures, of the as-prepared Ti_p/Mg-RE composites were investigated. Moreover, the microhardness and wear resistance of the Mg-RE alloy and Ti_p/Mg-RE composites were evaluated.

2. Experimental procedures

Mg-3.0Gd-2.7Zn-0.4Zr-0.1Mn (GZKM-1 alloy, wt.%) prepared via semi-continuous casting was used as the Mg-RE metal matrix, and its typical microstructure and properties were consistent with those observed in our previous study [14]. GZKM-1 specimens with a height of 5 mm and different surface dimensions were used as the substrates.

Subsequently, the metal-metal Ti_p/Mg-RE composites were prepared using the laser melt injection process. The laser melt injection experiment was carried out using a 4 kW IPG YSL-4000 fiber laser. The laser beam with a diameter of 3 mm was focused on the substrate surface through optical systems. Ti particles entered the four channels with the carrier argon gas blown out from a double-cylinder powder feeder. Then, the Ti particles in the four channels converged to the coaxial position of the laser beam before entering the molten pool. The movement of the laser melt injection system was controlled by a KR C4 robot from the KUKA Company. The schematic of the laser melt injection experiment is shown in Fig. 1. Spherical Ti particles obtained by atomization with a diameter of 40–100 µm and an average size of around 75 µm were used. In order to truly present the Ti particles used in this study, several optical micrographs (OM) of Ti particles were spliced, as shown in Fig. 2(a). The scanning electron microscopy (SEM) images of the Ti particles are shown in Fig. 2(b-c), and their size distribution is shown in Fig. 2(d). The amount of Ti particles in the channel was controlled by adjusting the carrier gas velocity. The coaxial shielding argon gas was supplied simultaneously at a flow rate of 10 L/min to prevent the oxidation of the molten pool. To investigate the possible metallurgical mechanism during the laser injection process, an IDT Motion pro Y4 high-speed camera was utilized to monitor the instantaneous molten pool.

Transverse cross-sections of the laser melt injection samples were cut, ground, and polished. Microstructures of the samples were analyzed using optical microscopy and a scanning electron microscope equipped with a backscattering scanning electron probe. X-ray diffraction (XRD) with Cu Kα radiation was used to identify the phases in the coatings. To further characterize the microstructures of the composites, electron backscatter diffraction (EBSD) (Gemini SEM300 in FEI Sirion 200 SEM) was used. Prior to the EBSD measurements, the samples were subjected to vibration polishing and ion polishing. In addition, the hardness values of the cross-sections of the Ti_p/Mg-RE composites were measured using a Vickers microhardness tester with a load of 1000 g and a holding time of 12 s. Tensile experiments are carried out on a universal material-testing machine (Zwick Z020, Germany). The friction and wear testing of the matrix and composites were conducted with a friction pair of Si₃N₄ at a low linear reciprocating speed of 10 mm/s, friction load of 15 N, and friction time of 10 min on a high-speed and high-temperature friction and wear testing machine (UMT-Tribolab, Bruker). The worn surfaces of the samples were examined using confocal laser scanning microscopy (Super Depth of Field Microscope, VHX-1000C) and SEM (FEI Sirion 200).

3. Results

3.1. Ti_p/Mg-RE composites prepared by laser melt injection

The Ti_p/Mg-RE composites prepared under optimum laser conditions using the laser melt injection method. Considering the energy accumulation during the multi-pass process, it can be stated that the laser spot offset a certain distance with decreasing laser power after each pass. When the substrate with the appropriate length, width, and thickness

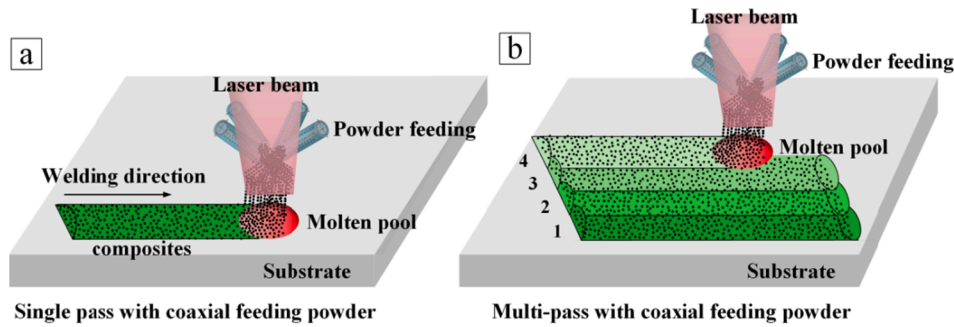


Fig. 1. Schematic of the laser melt injection experiments with coaxial nozzles: (a) single pass and (b) multi-pass.

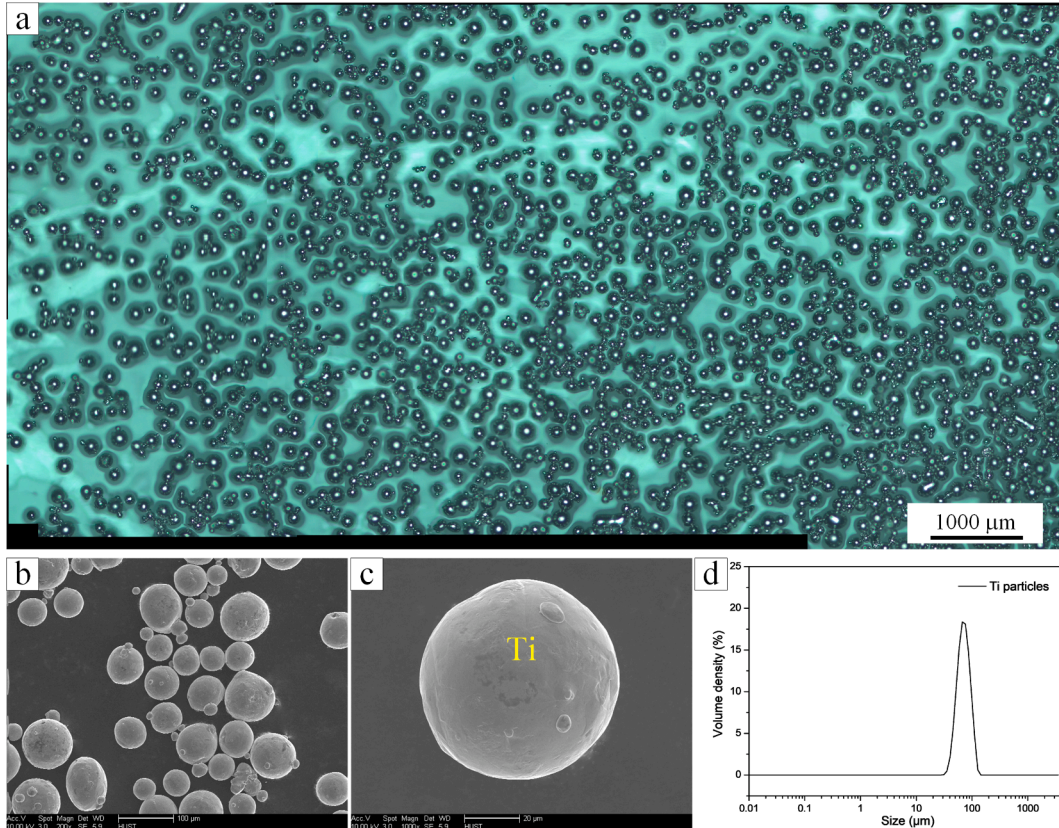


Fig. 2. Ti powder: OM (a), SEM image (b) low-magnification and high-magnification (c), and (d) size distribution of Ti particles.

was used, a Mg-RE composite layer with large surface area (over 30 cm² after 21 passes), silver white surface, good formability, and no significant deformation was obtained through laser melt injection with the laser spot offset by 2.0 mm per pass and a scanning speed (welding speed) of 0.9 m/min. In addition, the laser power decreased from 2800 to 2400 W, as shown in Fig. 3(a). The laser power decreased with a decrease in the laser scanning speed. Fig. 3(b) shows the cross-section of the weld bead obtained after six overlapping passes at a scanning speed of 0.6 m/min and a laser power of 2500–2300 W. The weld did not show any significant defects.

The cross-sections of Ti_p/Mg-RE are shown in Fig. 4. The Ti particles maintained their initial shape in the melt Mg-RE pool, as shown in Fig. 4(a). The volume fraction of the Ti powder was affected by the powder feeding rate, laser power, scanning rate, and other factors at a given powder size. The volume fraction of the Ti powder was approximately 80% at the powder feeding rate of 9.1 g/min, laser power of 2500 W, and scanning speed of 1.2 m/min. The Ti_p/Mg-RE composites had a small heat-affected zone (HAZ), as shown in Fig. 4(a,b). Columnar crystals

appeared in the vertical direction of the fusion line as well as the region between the Ti particles in the composites, marked as rectangular area in Fig. 4(b,d). This is because some energy was consumed during the injection of the Ti powder into the molten pool, which increased the supercooling degree.

The Ti_p/Mg-RE composites mainly consisted of α-Mg and β-(MgZn)₃Gd compounds in the matrix as well as Ti particles, as shown in Fig. 5(a). It can be seen that α-Ti showed good wettability in the Mg-RE alloy as both of them had the HCP structure. No significant reaction layer was observed between the Ti particles and α-Mg, as shown in the SEM images in Fig. 5(a-c). This observation is consistent with the Mg-Ti binary phase diagram. The β-(MgZn)₃Gd compounds showed good affinity to the Ti particles, as indicated by the element distribution at the interfaces between the Ti particles and surrounding compounds (Fig. 5(e,f)). This indicates that the interfaces between the Ti particles and β-(MgZn)₃Gd compounds could be controlled by the laser melt injection process, which has adjustable parameters.

To further investigate the interface between the Ti particles and Mg-

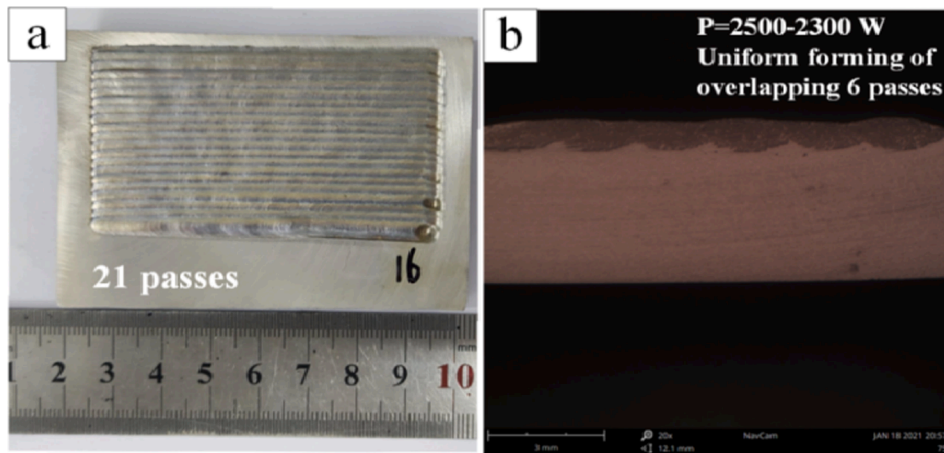


Fig. 3. Ti_P /Mg-RE composites prepared by multi-pass laser melt injection: (a) surface morphology and (b) cross-section.

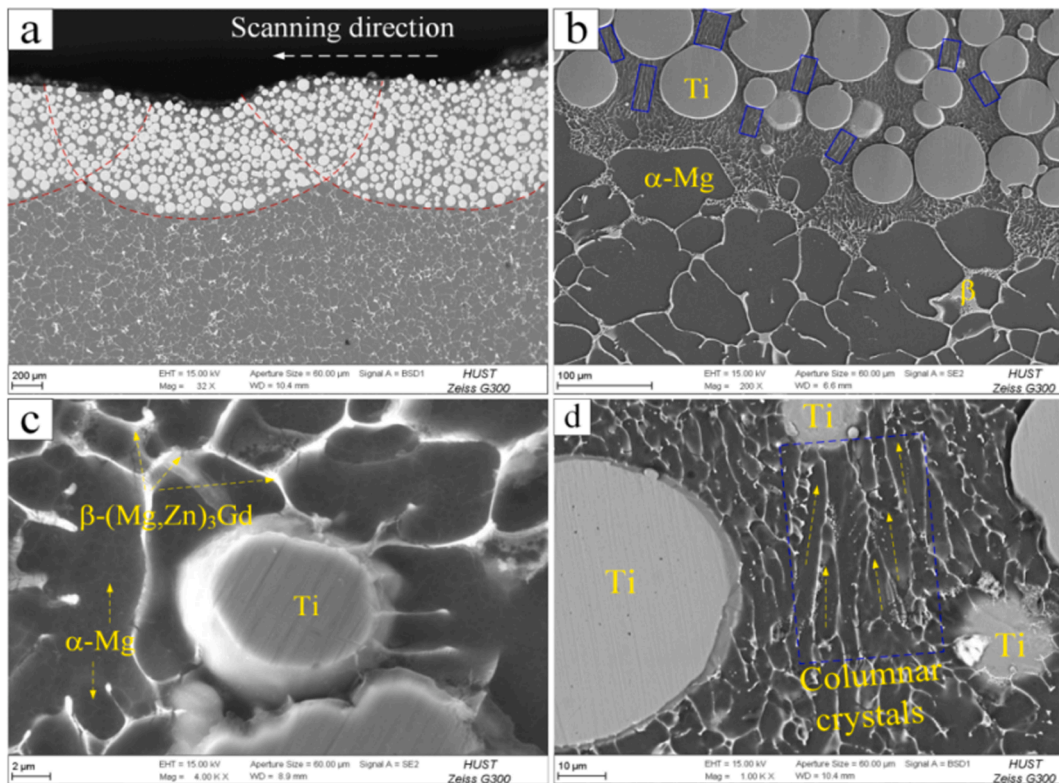


Fig. 4. Cross-sections of the Ti_P /Mg-RE composites: (a) multi-pass, (b) bottom region, (c) typical characteristics of the interface, and (d) columnar crystals between Ti particles.

RE matrix, EBSD analysis was carried out. Some of the grains in the Mg-RE matrix showed good orientation with the gains in the Ti particles, especially those in the basal plane, as shown in Fig. 6. This indicates that a good interface relationship and metallurgical bonding existed between the Ti particles and Mg-RE matrix.

3.2. XRD analysis

XRD analysis was carried out to analyze phases in the Ti_P /Mg-RE composites comparing with the Mg-RE matrix. Fig. 7(a) shows the XRD patterns of the GZKM-1 alloy and Ti_P /Mg-RE composites over the 2θ range of $5-85^\circ$. Fig. 7(b) shows the magnified XRD patterns over the 2θ range of $31-42^\circ$. The GZKM-1 alloy mainly consisted of the α -Mg matrix and β -(MgZn)₃Gd compounds, whose diffraction peaks corresponded to

the Mg_3Gd -type phase (ICCD, PDF#65-0040). The detailed analysis is provided in our previous report [14]. The Ti_P /Mg-RE composites consisted of the α -Mg phase and β -(MgZn)₃Gd compounds in the matrix alloy and α -Ti as the reinforcing phase. The Ti particles maintained their HCP lattice structure during the laser melt injection process (ICCD, PDF#65-9266).

3.3. Mechanical properties of the Ti_P /Mg-RE composites

Fig. 8(a) shows the microhardness results of the cross-sections of the Ti_P /Mg-RE composites at different testing positions: upper position in the bead (marked as line 'A'), middle position in the bead (marked as line 'B'), bottom position in the bead (marked as line 'C'), HAZ position in the bead (marked as line 'D'), and matrix position (marked line as

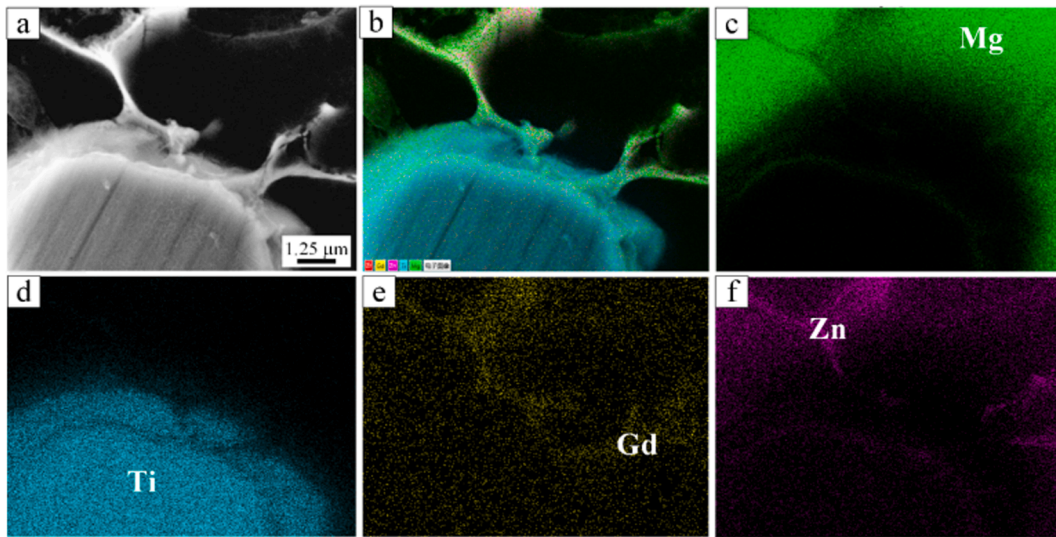


Fig. 5. SEM and EDS analysis of the $Ti_p/Mg-RE$ composites: (a) SEM image, (b) synthesis graph of element distribution of Mg, Ti, Gd and Zn, and (c–f) element distribution at the interface between Ti particles and surrounding β -compounds.

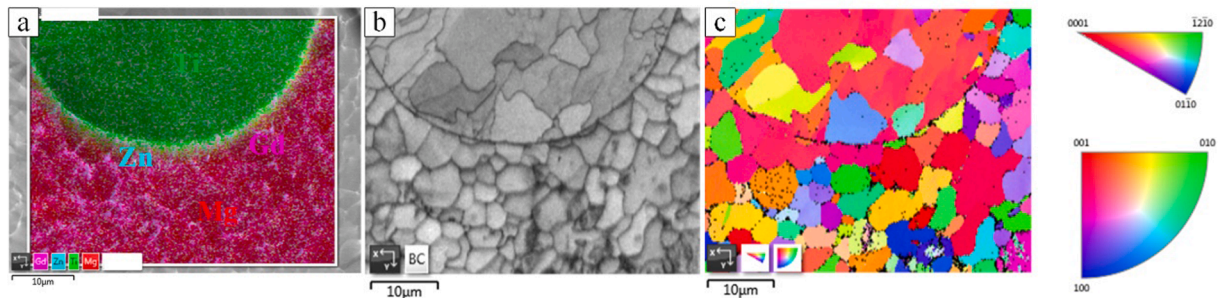


Fig. 6. EBSD analysis of the $Ti_p/Mg-RE$ composites: (a) element distribution, (b) EBSD results, and (c) inverse pole figures of the $Ti_p/Mg-RE$ composites.

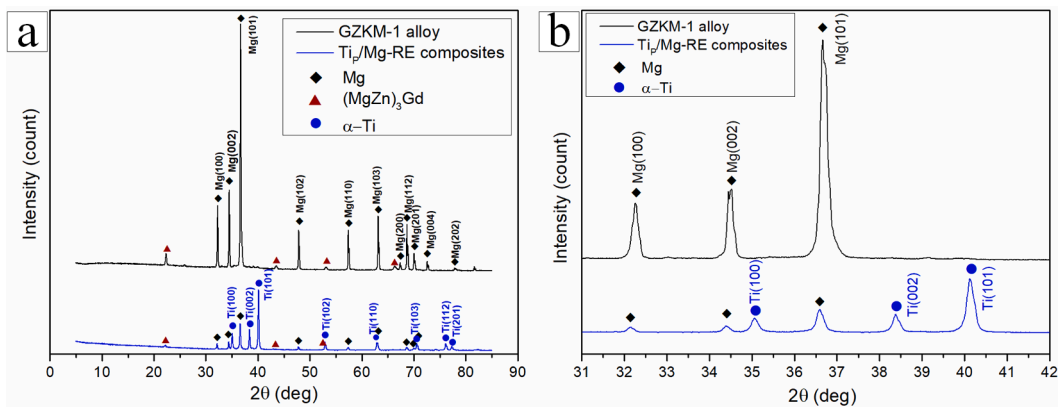


Fig. 7. XRD analysis of the GZKM-1 alloy and $Ti_p/Mg-RE$ composites: (a) $2\theta = 5-85^\circ$, and (b) magnified XRD pattern for $2\theta = 31-42^\circ$.

lines 'E' and 'F'). The average and standard deviation of the values obtained at the five testing points on each line are also marked in the corresponding position. Fig. 8(b) shows the microhardness results of the five testing points on each line marked in Fig. 8(a). As can be observed from the figure, the composite area showed significantly higher microhardness than the matrix because of the presence of its high Ti particle volume fraction. The upper and middle positions showed the same microhardness (approximately 145 HV). This indicates that the distribution of the Ti particles in the bead was relatively uniform. The bottom position corresponded to the transition area with lesser Ti particles.

Thus, its microhardness changed remarkably. The microhardness values of the different points on the horizontal line 'C' at the bottom of the bead were significantly different. This can be attributed to the presence of few Ti particles in some local areas near the bottom. Nevertheless, the microhardness at the bottom position was approximately 132 HV, which is significantly higher than those in the HAZ and matrix. As there were no Ti particles in the HAZ, its microhardness decreased to approximately 70 HV, which is the same level as that of the matrix area. In summary, the microhardness of the Mg-RE matrix increased by approximately 100% after the incorporation of Ti particles via laser melt injection to

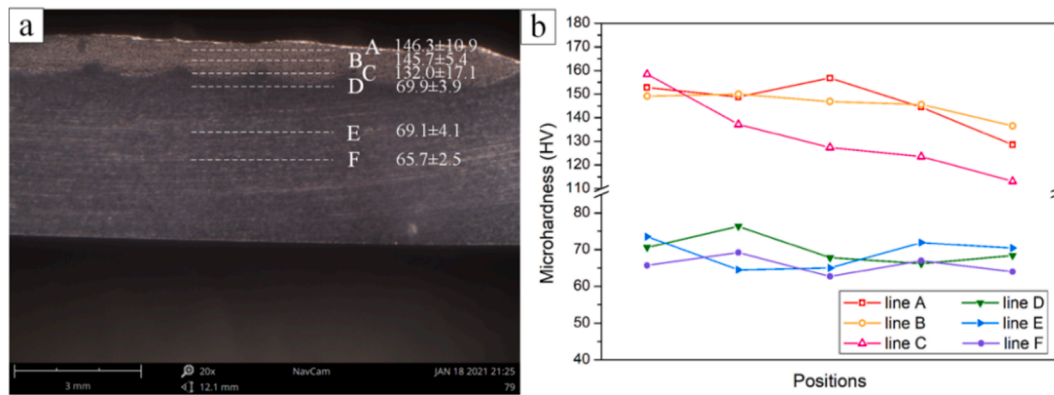


Fig. 8. Microhardness of the $\text{Ti}_p/\text{Mg-RE}$ composites: (a) testing positions in the laser melt injected sample, and (b) the microhardness measurement results.

prepare the $\text{Ti}_p/\text{Mg-RE}$ composites.

Fig. 9 shows the typical tensile engineering stress-strain curves of the $\text{Ti}_p/\text{Mg-RE}$ composites and Mg-RE matrix. The average value of three times UTS of the $\text{Ti}_p/\text{Mg-RE}$ composites is around 160 MPa, which is 10% lower than that of the matrix. The average value of three times elongation of the $\text{Ti}_p/\text{Mg-RE}$ composites is 4.99%, which is 4% lower than that of the matrix. In summary, the tensile properties of the $\text{Ti}_p/\text{Mg-RE}$ composites decreased slightly while the hardness increased significantly, as compared to the Mg-RE matrix.

To further investigate the reason behind the decrease in the tensile strength of the $\text{Ti}_p/\text{Mg-RE}$ composites, their fracture characteristics were evaluated. Under the action of force, some Ti particles are came off from the Mg-RE matrix and these may appear as particle-pits after the Ti particles have fallen off or as exposed Ti particles on the surface, as shown in Fig. 10(a-b). Despite the low elongation of the $\text{Ti}_p/\text{Mg-RE}$ composites, the Mg-RE alloy around the Ti particles caused certain plastic deformation during the tensile process, leaving behind dimples, as shown in Fig. 10(c). In particular, some residual Mg on the Ti particles also produced dimples, which indicates that good interfacial bonding between these Ti particles and the Mg-RE matrix, as shown in Fig. 10 (d-f). Wu [44] et.al reported that in the studied interface models, Ti (0001) has a mismatch of 4.1% with its 9-atom cell matching 7-atom Mg (0001) cell. Baldi [45] et.al stated that although large lattice mismatch between Mg and Ti, the as-deposited Pd-capped Mg/Ti metallic multilayers show good structural coherence. The reason is that Mg-Ti alloys have a small degree of chemical short-range order in the atomic distribution. The researches show that Ti and Mg can have a good interface

bonding. The main reason for the formation of these dimples is the grain refinement of the Mg-RE matrix during the laser melt injection process, which induced the fine grain strengthening effect during the tensile experiments. Another reason for the low elongation of the $\text{Ti}_p/\text{Mg-RE}$ composites is that the $\beta\text{-(MgZn)}_3\text{Gd}$ compounds in the Mg-RE matrix were a kind of brittle phase, which decomposed first during the tensile process, as shown in Fig. 10(f).

The main reason for decreasing of mechanical properties of the $\text{Ti}_p/\text{Mg-RE}$ composites is that the interface bonding strength between the Mg-RE matrix and the Ti particles is still insufficient. Some interfaces between the Mg-RE matrix and the Ti particles are not bonded enough, which can be proved from the fracture morphology in Fig. 10. The phenomenon of some Ti particles completely came off from the matrix in the fracture indirectly proves that the bonding is poor. Some Ti particles did not come off from the Mg-RE matrix, and the interface bonding of this part was good. It can be inferred from this that during the rapid solidification process of laser processing, the evaporation and burning loss of various elements in the Mg-RE alloy are different, and the elements of the Mg-RE alloy have a certain degree of segregation, which leads to better bonding of Ti particles with the Mg-RE matrix at some positions under the transition action of alloy elements, while weaker bonding of Ti particles with the Mg-RE matrix at other positions. Thus those Ti particles with large size, average size of $75\text{ }\mu\text{m}$ of Ti particles is much larger than that of the grain size in the composites, destroy the integrity of the Mg-RE matrix. Moreover, due to the rapid solidification characteristics of the laser processing, more volume of the brittle $\beta\text{-(MgZn)}_3\text{Gd}$ compounds appear in the composites compared with the amount of the $\beta\text{-(MgZn)}_3\text{Gd}$ compounds in the matrix, which further destroy the integrity of the Mg-RE matrix. In summary, large size of Ti particles with poor bonding strength to the Mg-RE matrix and a large number of the brittle $\beta\text{-(MgZn)}_3\text{Gd}$ compounds damage the integrity of the Mg-RE matrix, resulting in that the refined grains can only play a role in local areas, affecting the effect of fine grain strengthening. Their combined effects have reduced the tensile strength and elongation of the $\text{Ti}_p/\text{Mg-RE}$ composites.

The friction results of the Mg-RE matrix showed that the repeatability of friction testing was not good enough, and the coefficient of friction was unstable indicating the occurrence of an unstable running-in stage under the friction of the Si_3N_4 friction pair, as shown in Fig. 11(a). However, the $\text{Ti}_p/\text{Mg-RE}$ composites showed relatively stable running-in stage and good repeatability in the measurement of the coefficient of friction, as shown in Fig. 11(b). The typical coefficient of friction vs. time curves of the Mg-RE matrix and $\text{Ti}_p/\text{Mg-RE}$ composites are shown in Fig. 11(c). As can be observed from the figure, the coefficient of friction of the $\text{Ti}_p/\text{Mg-RE}$ composites was lower than that of the Mg-RE matrix. This is consistent with the microhardness testing results.

After the friction testing, the worn morphologies of the samples were examined using a depth of field microscope. Fig. 12(a,c) shows the wear marks of the Mg-RE matrix and $\text{Ti}_p/\text{Mg-RE}$ composites. It can be seen

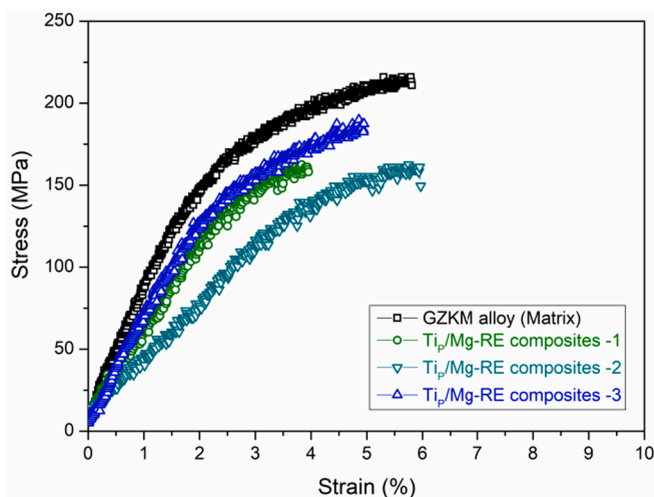


Fig. 9. Tensile stress-strain curves of the $\text{Ti}_p/\text{Mg-RE}$ composites and Mg-RE matrix.

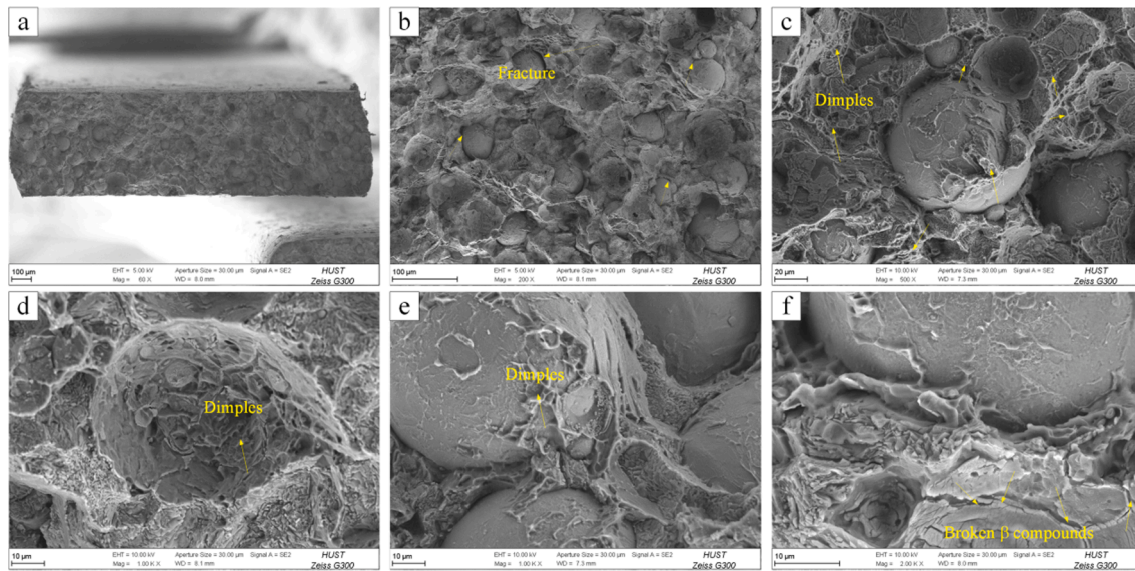


Fig. 10. Fracture characteristics of the $\text{TiP}/\text{Mg-RE}$ composites: (a) overview, (b) typical fracture morphology, (c–e) dimples in the matrix and (f) decomposed β compounds.

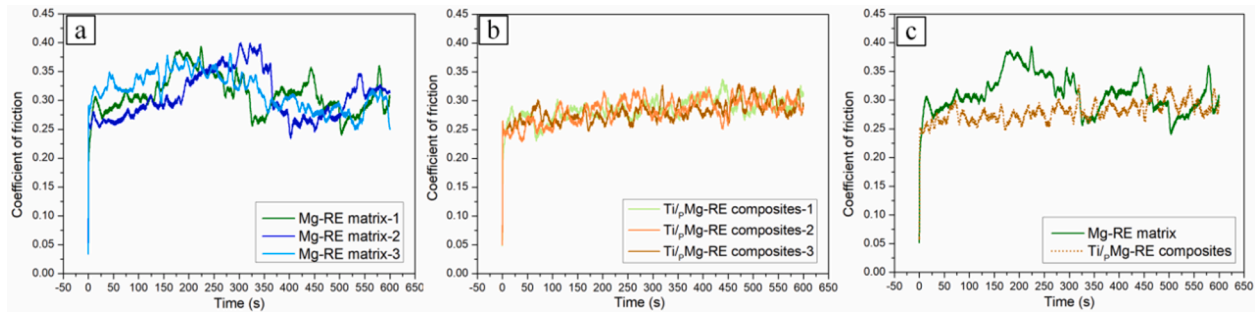


Fig. 11. Coefficients of friction of the Mg-RE matrix and $\text{TiP}/\text{Mg-RE}$ composites: (a) three repeated results of the Mg-RE matrix, (b) three repeated results of the $\text{TiP}/\text{Mg-RE}$ composites, and (c) comparison of the Mg-RE matrix and $\text{TiP}/\text{Mg-RE}$ composites.

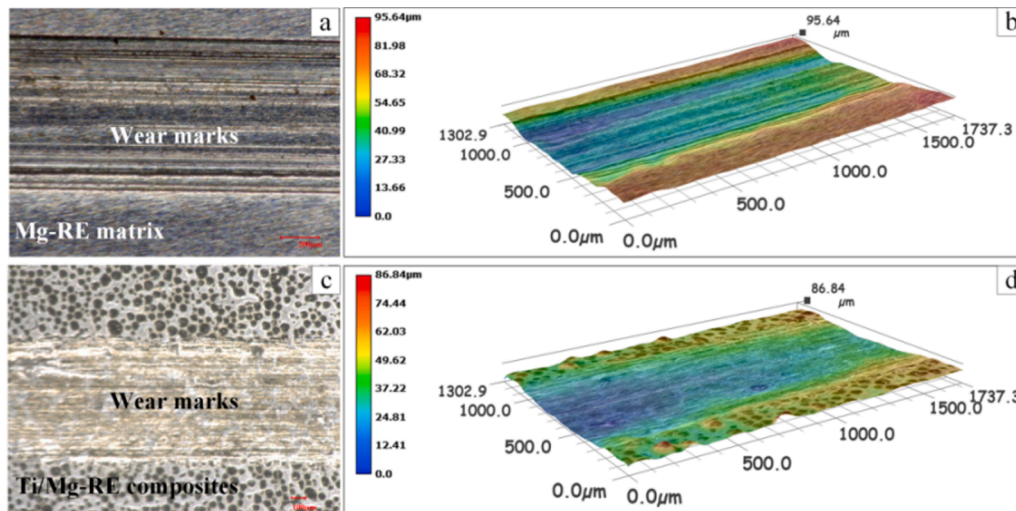


Fig. 12. Morphologies and three-dimensional (3D) measurements of the worn surfaces of the Mg-RE matrix and $\text{TiP}/\text{Mg-RE}$ composites after the friction testing: (a) and (c) macromorphologies of the matrix and composites, (b) and (d) 3D measurements of the worn surfaces of the matrix and composites.

that the Mg-RE matrix showed wider wear marks. The worn area depth of the matrix was higher than that of the composites, as shown in Fig. 12 (b,d). This indicates that the $\text{TiP}/\text{Mg-RE}$ composites showed better

friction resistance than the Mg-RE matrix.

4. Discussion

The molten pool of the Ti_p/Mg-RE composites during the laser melt injection process was monitored using a high-speed camera to observe the dynamics of the plasma and molten pool, as shown in Fig. 13. The laser welding process of the Mg alloy is a dynamic process, which involving melting, evaporation, and combustion, thus forming dynamic plasma and spatters. Fig. 13(a) shows the welding process recorded by the high-speed camera at 0 ms, in which the general protective gas action range, plasma range, molten pool range, and generated spatters in the welding process are shown. The melt flows from the front to the back along the forward direction, and the outer melt flows to the center, forming vortexes under the two combined actions, as shown in Fig. 13(c). As the Ti particles were injected into the tail of the molten pool and did not melt, they moved together with the molten pool. This resulted in the distribution of the Ti particles, which tended to aggregate in the direction of movement after the solidification of the molten pool. The protection of the laser-affected area by the inert argon gas and the matching of the Ti powder-feeding rate with the laser-scanning rate are important control factors that affected the stability of the plasma and molten pool. Considering the inert gas shielding rate, Ti powder feeding rate, laser scanning rate, and laser power, it can be stated that the molten pool moved stably with no significant oscillation, as shown in Fig. 13(b–m). Therefore, after the solidification of the molten pool, no significant welding and microscopic defects were observed in the micromorphology and microstructure of the Ti_p/Mg-RE composites, respectively.

To investigate the reasons of stable coefficient of friction of the Ti_p/Mg-RE composites as compared to those of the Mg-RE matrix, morphologies of their worn surfaces were examined using SEM. Fig. 14(a,d) shows the overview morphologies of the matrix and composites. Significant plough effect was observed in the Mg-RE alloy, as indicated by the rectangles in Fig. 14(b). On the other hand, in the case of the Ti_p/Mg-RE composites, the plough furrows mainly appeared in the matrix area, which was deficient in Ti particles, as shown in Fig. 14(e). The pits formed after the formation of the debris peeled off from the Mg-RE alloy and Ti_p/Mg-RE composites, while the pit area in the Mg-RE alloy seemed larger than that in the composites, as shown in Fig. 14(b, e). Peeling off of the Ti particles from the matrix was hardly observed, indicating that the Ti particles had good bonding with the matrix. Torn sharp corners were observed in both the Mg-RE alloy and Ti_p/Mg-RE composites, as shown in Fig. 14(c,f). The Mg-RE alloy showed larger torn sharp corners. Most of the area on the worn surfaces of both the Mg-RE alloy and Ti_p/Mg-RE composites was plastic extrusion area. The worn surface morphologies of the Mg-RE alloy and Ti_p/Mg-RE composites indicated the occurrence of significant abrasive wear.

The torn sharp corners were formed during the reciprocating friction process due to the repeated shearing stress. The corners could be removed as wear debris at some areas. Fig. 15 shows the schematic of the sliding friction of the Mg-RE alloy and Ti_p/Mg-RE composites: (a) sliding friction process and force analysis of the friction pair: F represents the pressure force in the vertical direction, F' is the sliding traction force acting on the friction pair along the sliding direction, and F'' denotes the squeezing force of the alloy/composites on the friction pair among the friction depth area (marked as 'h' in depth); (b) stress element of the testing materials during the sliding friction process, where σ was caused by the pressure force of the friction pair, σ' was caused by the friction force, and σ'' was caused by the squeezing force of the friction pair; (c) stress in the sliding plane and possible friction mechanism. Under the same shear force, Mg-RE matrix could easily tear the sharp corners as compared to the Ti particles because of its lower strength.

5. Conclusions

In this study, Ti_p/Mg-RE composites were prepared through the laser melt injection process. The microstructures and mechanical properties of the composites were investigated. High-speed camera recording was utilized to evaluate the stability of the laser melt injection process. The conclusions drawn from this study can be summarized as follows.

- (1) Ti_p/Mg-RE composites with large surface area and no defects in microstructure could be designed using the laser melt injection method under condition of stable molten pool.
- (2) The laser process significantly refined the grains in the Mg-RE matrix of the Ti_p/Mg-RE composites. Ti particles maintained their initial shape in the molten pool and formed good metallurgical bonding with the matrix through alloy elements, as indicated by the SEM, EBSD, and tensile measurement results.
- (3) Microhardness of the Ti_p/Mg-RE composites increased by 100%, while the tensile strength and elongation of the Ti_p/Mg-RE composites decreased a little, as compared to those of the Mg-RE matrix.
- (4) Ti_p/Mg-RE composites showed abrasive wear and had a stable friction coefficient and running-in stage attributing to the good metallurgical interface between Ti particles and the matrix to give play to the advantages of high hardness and strength of Ti particles.

Declaration of Competing Interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence

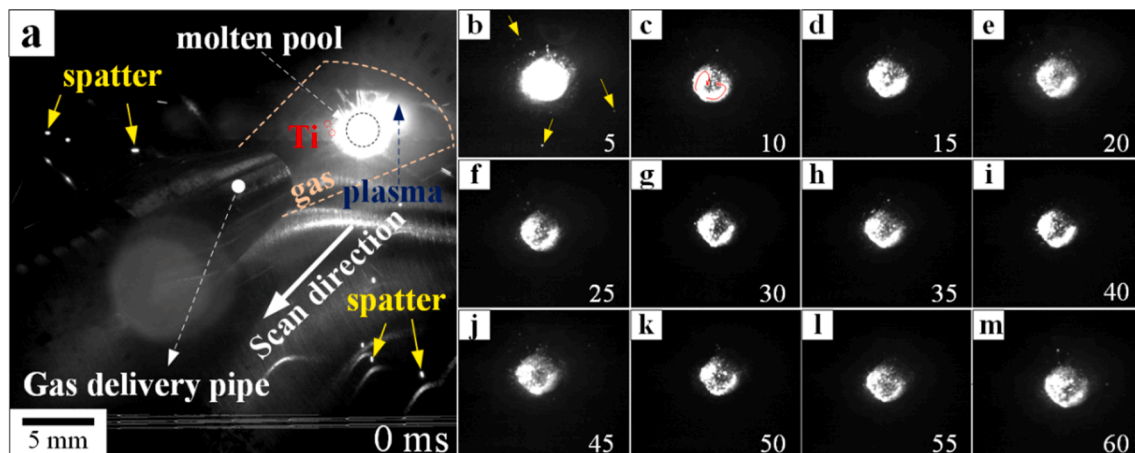


Fig. 13. Molten pools of the Ti_p/Mg-RE composites during the laser melt injection process monitored using a high-speed camera.

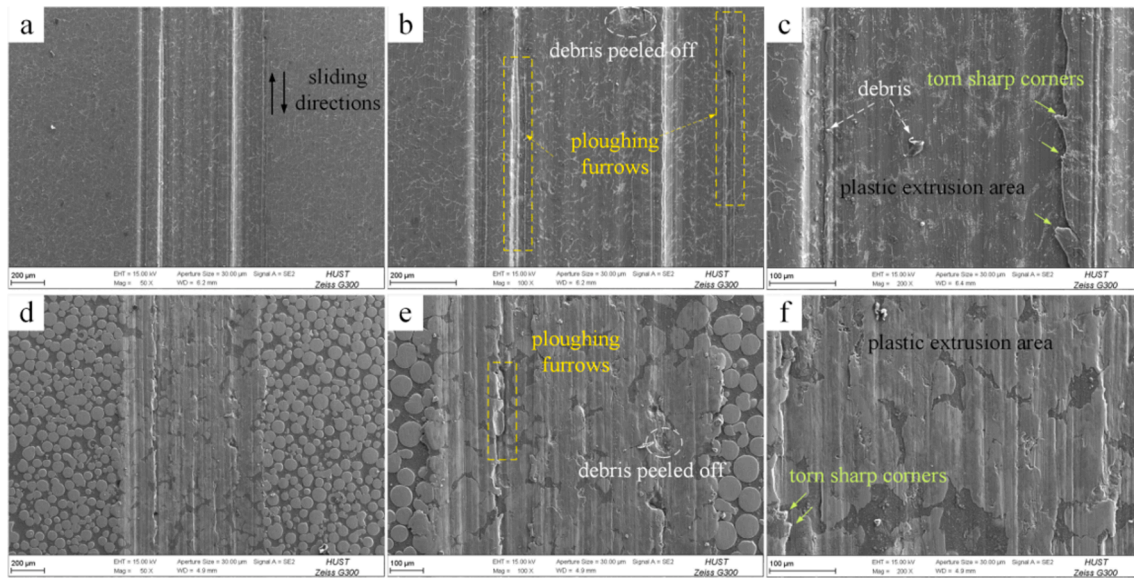


Fig. 14. SEM images showing the worn surface morphologies of the (a-c) Mg-RE alloy and (d-f) TiP/Mg-RE composites.

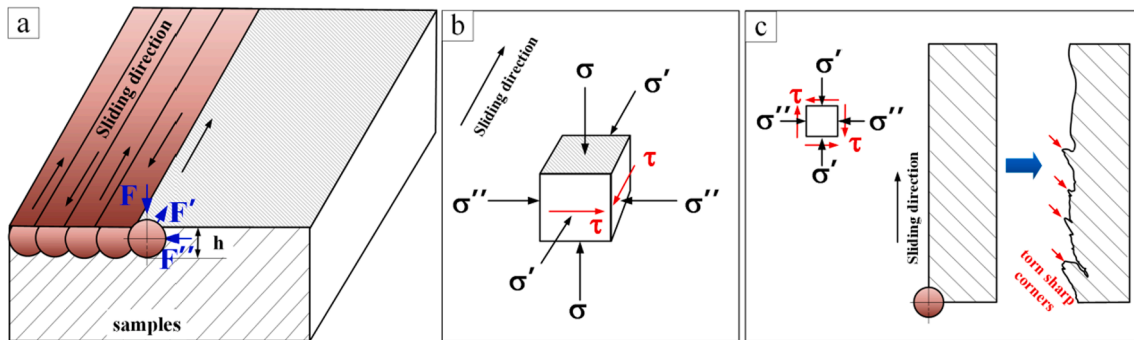


Fig. 15. Schematic of the sliding friction of the Mg-RE alloy and TiP/Mg-RE composites: (a) sliding friction process, (b) stress element during the sliding friction process and (c) stress in the sliding plane and the possible friction mechanism.

the work reported in this paper.

Data availability

Data will be made available on request.

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