#### **Review Article**

Haotian Guan, Hui Xiao, Sihui Ouyang\*, Aitao Tang, Xianhua Chen, Jun Tan\*, Bo Feng, Jia She\*, Kaihong Zheng, and Fusheng Pan

# A review of the design, processes, and properties of Mg-based composites

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Abstract: Magnesium-based composites are promising materials that can achieve higher strength, modulus, stiffness, and wear resistance by using metals, ceramics, and nanoscale carbon-based materials as reinforcements. In the last few decades, high-performance magnesiumbased composites with excellent interfacial bonding and uniformly distributed reinforcements have been successfully synthesized using different techniques. The yield strength, Young's modulus, and elongation of SiC nanoparticle-reinforced Mg composites reached ~710 MPa, ~86 GPa, and ~50%, respectively, which are the highest reported values for Mg-based composites. The present work summarizes the commonly used reinforcements of magnesium composites, particularly nano-reinforcements. The fabrication processes, mechanical properties, reinforcement dispersion, strengthening mechanisms, and interface optimization of these composites are introduced, and the factors affecting these properties are explained. Finally, the scope of future research in this field is discussed.

**Keywords:** magnesium-based composite, nanoscale reinforcement, mechanical properties, interfacial bonding

#### 1 Introduction

In recent years, new energy vehicles and the aerospace industry have been facing a constant requirement to improve energy efficiency through the weight reduction of their components. Owing to their low density and high strength-to-weight ratios, light metals have thus attracted considerable global attention. Magnesium (Mg) and magnesium alloys are the lightest engineering structural materials [1–6] and have been extensively considered as good weightreducing components because of their low density, high specific strength, good thermal conductivity, castability, machinability, and high damping capacities [7-11]. After decades of development, the conventional mechanical properties of Mg alloys, such as strength and plasticity, have been significantly improved [12-15]. The ultimate tensile strength (UTS) of Mg alloys has reached 500 MPa, and its elongation (EL) has reached more than 50%.

However, as structural materials, their inherently low elastic modulus and strength restrict their extensive utilization in critical engineering applications [16,17]. For this reason, many researchers have used traditional strengthening methods (solid-solution strengthening, precipitation strengthening, fine-grain strengthening, and work hardening) to improve the performance of Mg alloys for many years. Mg alloys are mainly composed of hexagonal close-packed (hcp) α phases (α-Mg). A limited number of slip systems can be offered by the hcp structure, which limits the role of plastic deformation and work hardening methods in the strengthening of Mg alloys [18]. Solid-solution strengthening and fine-grain strengthening by adding alloying elements (mainly rare earth elements) are commonly used to improve the mechanical properties of traditional Mg alloys [19,20], but the increased content of alloying elements greatly affects

Haotian Guan, Hui Xiao: College of Materials Science and Engineering, Chongqing University, Chongqing 400044, China Aitao Tang, Xianhua Chen, Fusheng Pan: College of Materials Science and Engineering, Chongqing University, Chongqing 400044, China; National Engineering Research Center for Magnesium Alloys, Chongqing University, Chongqing 400044, China

**Bo Feng, Kaihong Zheng:** Institute of New Materials, Guangdong Academy of Sciences, Guangzhou 510650, China

<sup>\*</sup> Corresponding author: Sihui Ouyang, College of Materials Science and Engineering, Chongqing University, Chongqing 400044, China; National Engineering Research Center for Magnesium Alloys, Chongqing University, Chongqing 400044, China,

e-mail: sihuiouyang@cqu.edu.cn

<sup>\*</sup> Corresponding author: Jun Tan, College of Materials Science and Engineering, Chongqing University, Chongqing 400044, China; National Engineering Research Center for Magnesium Alloys, Chongqing University, Chongqing 400044, China, e-mail: jun.tan@cqu.edu.cn

<sup>\*</sup> Corresponding author: Jia She, College of Materials Science and Engineering, Chongqing University, Chongqing 400044, China; National Engineering Research Center for Magnesium Alloys, Chongqing University, Chongqing 400044, China, e-mail: jiashe@foxmail.com

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the weight and cost of Mg alloys. According to the rule of mixtures [21], the elastic modulus of a multiphase alloy is determined by the elastic modulus and volume fraction of all constituent phases. The elastic modulus of pure Mg (~45.0 GPa) [22] and each secondary phase in conventional commercial Mg alloys are both low [23,24]. This results in the modulus of these commercial Mg alloys being generally between 35.0 and 45.0 GPa [25,26]. Therefore, these conventional research ideas are of limited use in improving the performance of Mg alloys. New research directions are thus urgently needed to improve the comprehensive mechanical properties of Mg alloys.

These intrinsic drawbacks can be circumvented by the addition of reinforcements to create Mg-based composites [27-30]. With respect to material design, such reinforcements combine the best properties of the Mg matrix, and they provide significant improvement in mechanical properties [31–33]. These improvements include high specific strength, high elastic modulus, good wear resistance, and excellent high-temperature properties [34-39]. The fabrication processes of Mg matrix composites can be categorized into liquid-phase methods [40-44], solid-phase methods [45-50], and other methods such as in situ synthesis [51,52]. A crucial task of Mg-based composites is to ensure that the load is evenly distributed between the matrix and the reinforcements [53,54]. Hence, the reinforcements require good physical and chemical compatibilities, load-bearing capacity, wettability, and interface reactions with the matrix [55,56]. The strengthening mechanism of the reinforcements mainly acts as a forcebearing body. When external stress is applied, the load is efficiently shifted from the softer Mg matrix to the harder reinforcements [57,58]. The strengthening and hardening effects of the reinforcements can be achieved through effective stress transfer at the interfaces between the matrix and reinforcements. The refinement and dispersion of the reinforcements can not only achieve the effect of grain refinement during the fabrication process but can also hinder the movement of dislocations and grain boundary slip during the loading and deformation process, thereby improving the strength of the composites [59-63]. At present, the widely recognized strengthening mechanisms of Mg-based composites include fine-grain strengthening, Orowan strengthening, load transfer effect, and thermal mismatch strengthening [64-68]. The Orowan strengthening mechanism originates from interactions with nanoparticles, dislocations, and grain boundaries. The better the interfacial bonding, the better the load transfer effect. Similarly, the smaller the grain size, the better the enhancement effect. Therefore, uniform distribution of the reinforcements, good interfacial bonding, and fine Mg matrix grains are prerequisites

that need to be addressed to achieve Mg-based composites with excellent mechanical properties. Current research hotspots include the influence of the type, volume fraction, and size of the reinforcements, reasonable fabrication processes, and interface optimization treatment on the microstructure and mechanical properties of Mg-based composites. Hence, this study reviews the selection of different types of reinforcements (especially nano-reinforcements), fabrication processes, mechanical properties, and interface optimization treatments that have been reported in recent studies in the field of Mg-based composites. The current problems, as well as directions for future development of Mg-based composites, are offered from the authors' perspective.

### 2 Metallic-reinforced Mg-based composites

The final properties of Mg-based composites are controlled by many factors, such as the fabrication process, matrix size/constitution, volume fraction and morphology of reinforcements, and secondary processing. Among these, the selection of reinforcements compatible with the Mg matrix remains one of the most critical factors for realizing the best properties of the resultant composite. Metallic reinforcements have good wettability and selfextensibility, which can collectively enhance the modulus and plasticity of the Mg matrix. In recent years, various metallic-reinforced Mg-based composites (mainly based on pure Mg), such as Cu/Mg [69,70], Ni/Mg [71,72], Mo/Mg [73], Ti/Mg [74–76], (Cu, Ti)/Mg [69], TC4/Mg [77], and Mn/Mg alloys [78], have been developed. These metallic reinforcements have higher melting points than the Mg matrix, and they can become nucleation centers and obtain fine equiaxed grains during melting. The solid solubility of the metallic reinforcements in the Mg matrix is limited. It can not only maintain a certain number of reinforcements but also form secondary phases.

The mechanical properties of some metallic-reinforced Mg-based composites are listed in Table 1. As shown in the table, these metallic-reinforced Mg-based composites exhibit excellent ductility owing to the good wettability between the Mg matrix and the metallic reinforcements, the generation of secondary phases, and the toughness of the metallic reinforcements. Therefore, metallic-reinforced Mg-based composites have the dual functions of particle strengthening and second-phase strengthening. The simultaneous addition of multiple metallic reinforcements may improve the strengthening effect.

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**Table 1:** Mechanical properties of metallic-reinforced Mg-based composites

Materials	Fabrication process (es)	Ten	Tensile properties	Se	ŭ	Compressive properties	roperties	Hardness (HV)	Hardness (HV) Young's modulus (GPa)
		YS (MPa)	UTS (MPa)	EL (%)	CYS (MPa)	UCS (MPa)	YS (MPa) UTS (MPa) EL (%) CYS (MPa) UCS (MPa) Failure strain (%)		
Pure Mg [79]	PM	125 ± 15	172 ± 12	5.8 ± 0.9	ı	ı	ı	38.6 ± 1.5	I
AZ31/WE43 [80]	PD + CE	296 ± 3	368 ± 9	16	243 ± 5	$419 \pm 12$	1	1	I
3.0 wt% Cu/Mg [70]	PM + MS	$237 \pm 24$	286 ± 8	$5.4 \pm 1.2$	1	ı	ı	57 ± 1	59.7 ± 6.7
17.18 wt% Cu/Mg [81]	DMD + HE	355 ± 8	358 ± 7	$2.2\pm0.9$	1	ı	I	ı	$47 \pm 1$
14.46 wt% Ni/Mg [72]	DMD + HE	$370 \pm 12$	389 ± 5	$3.1\pm0.1$	1	1	1	74 ± 2	$47 \pm 1$
3.6 wt% Mo/Mg [73]	DMD + HE	$123.3 \pm 3.5$	$123.3 \pm 3.5  197.6 \pm 5.9$	$9.0 \pm 2.1$				$63.5 \pm 7.3$	$43.0 \pm 1.6$
5.6 wt% Ti/Mg [75]	DMD + HE	$163 \pm 12$	248 ± 9	$11.1\pm1.4$	ı	ı	I	ı	I
(3.0  wt%  Cu + 5.6  wt%  Ti)/Mg [69]	DMD	$201 \pm 7$	$265\pm11$	$7.5\pm0.5 126\pm8$	126 ± 8	<b>380</b> ± 6	$19.1 \pm 2.9$	91 ± 3	I
1.88 wt% Mn/Mg-1Sn [78]	WI	233	321	17.5	198	1	1	ı	1

1/5: yield strength, UTS: ultimate tensile strength, EL: elongation, CYS: compressive yield strength, UCS: ultimate compressive strength, PM: powder metallurgy, PD: pressing diffusion, CE: co-extrusion, MS: microwave sintering, DMD: disintegrated melt deposition, HE: hot extrusion, IM: induction melting. Seetharaman *et al.* [69] added 5.6% Ti + 3% Cu (wt%) to Mg powder using the mechanical ball-milling method. The significantly enhanced mechanical properties of the composites are a result of grain refinement and Orowan strengthening mechanisms. The formation of a  $Mg_2Cu$  eutectic phase significantly refined the matrix grains. The  $Ti_3Cu$  phase was also conducive to the improvement of strength and plasticity.

It is necessary for metallic reinforcements to have limited solubility in the Mg matrix so that the reinforcements can partially dissolve into the matrix while retaining a certain number of reinforcement particles inside the matrix. In one study, micron-size Mo particles were added to pure Mg [73] because Mo is immiscible with Mg [82]. Minimal porosity and good interfacial integrity were obtained by combining disintegrated melt deposition (DMD) and subsequent hot extrusion (HE) methods, thus improving the hardness, elastic modulus, and ductility of the micron-sized Mo/Mg composites. Powder metallurgy (PM) is the most popular and widely used method for preparing Mg composite and does not require specific/difficult process steps. Wong and Gupta [70] used the PM method combined with microwaveassisted two-directional sintering to successfully prepare Mg-based composites reinforced with different nano-Cu particle contents. They pointed out that because of the presence of a continuous network structure composed of nano-Cu particles and the distribution of a Mg<sub>2</sub>Cu intermetallic phase inside the grains, the elastic modulus, yield strength (YS), and UTS of 3Cu/Mg (wt%) composites were 40, 82, and 51% higher, respectively, than those of pure Mg. Coextrusion is a process in which two or more materials are concurrently extruded into a single composite billet. This is a new method for manufacturing Mg-based composites with excellent ductility and high strength [80,83,84]. As shown in Figure 1, Zhao et al. [80] prepared a bimetal composite rod composed of a softer AZ31 sleeve and a harder WE43 core. This was achieved by a special process that combines hot-pressing diffusion and co-extrusion. Compared with AZ31, AZ31/WE43 composite rods, which had a gradient of both composition and microstructure, had excellent interfacial bonding owing to higher compression and tensile YS.

Metallic reinforcements can become nucleation centers when the Mg matrix is solidified, thus allowing the matrix to obtain fine equiaxed grains [85]. In addition, metallic reinforcements exhibit good wettability and ductility. Therefore, Mg-based composites reinforced by metallic reinforcements have good interface bonding. Some metallic reinforcements can also form secondary phases with the matrix to enhance interfacial bonding [86,87]. The strength of the composite often reaches a maximum value for certain

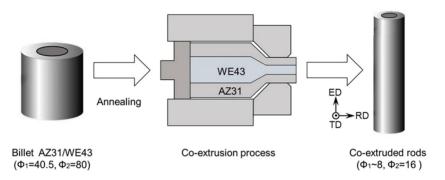


Figure 1: Schematic of the AZ31/WE43 bimetallic composite rod prepared by hot-press diffusion and co-extrusion [80].

content of metallic reinforcement [88,89]. Hence, the added content of metallic reinforcements needs to be better designed; otherwise, the reinforcements and the matrix may form compounds, which deteriorates the mechanical properties of the matrix. The decrease in the ductility of metallic-reinforced Mg-based composites is mainly attributed to the presence of a relatively high volume fraction of brittle intermetallic compounds at the interfaces and inside the matrix. These act as crack nucleation points, resulting in greatly reduced properties.

At present, research on metallic-reinforced Mg-based composites is at its initial stage and has mainly focused on pure Mg as the matrix. However, a few studies have been conducted on metallic-reinforced composites based on Mg alloys. In addition, the density of metallic reinforcements is much higher than that of the Mg matrix, and thus a high content of metallic reinforcement will negatively affect the weight of metallic-reinforced Mg-based composites. Methods to obtain uniformly dispersed nanometallic particle-reinforced Mg-based composites therefore need to be further explored. Solving the problems of interface diffusion, particle distribution uniformity, content ratio, and effective low-density metallic reinforcement is still a difficult problem in the development of metallic-reinforced Mg-based composites, and continuous experimental exploration is required.

# 3 Ceramic-reinforced Mg-based composites

Ceramic reinforcement has several advantages such as high hardness (1,200–3,700 HV) and modulus (300–600 GPa), low thermal expansion coefficient (3–8  $10^{-6}$  K<sup>-1</sup>), and high chemical stability [57–63,90,91]. The commonly used ceramic-based reinforcement materials mainly include carbides

(SiC,  $B_4C$ , and TiC), nitrides ( $Si_3N_4$  and AlN), borides ( $TiB_2$  and TiB), and metal oxides ( $Al_2O_3$  and MgO). The size and morphology of the ceramic reinforcements have a significant influence on the properties of Mg-based composites. Stress concentration occurs at the sharp corners of ceramic reinforcements with an irregular external morphology. Ceramic reinforcements can refine the Mg matrix grains, produce high-density dislocations, and effectively block dislocations, which can significantly enhance the strength of the matrix.

However, as the volume of the micron-ceramic reinforcements increases, significant plastic weakening of the composites will occur owing to the stress concentrations occurring near the particle aggregates and interface decohesion [92-94]. Moreover, as shown in Figure 2, the easy initiation and propagation of cracks in the micronceramic reinforcements or at the interfaces are a result of micron-ceramic-reinforced Mg-based composite failure [95–97]. Therefore, subsequent secondary processes are often used to refine the micron-ceramic-reinforced Mgbased composite grains to improve their microstructures [98]. In addition, the residual stress of ceramic reinforcements with regular and uniform shapes is very small [95,99]. The expected properties can be obtained using nano-ceramic reinforcements owing to their excellent properties (see Table 2). As illustrated in Figure 3, nanoceramic particles can be distributed at the grain boundaries and/or inside the grains, thus resulting in remarkable pinning of grain boundary movement and a fine-grain strengthening effect [103,104]. Hence, the strength and ductility of ceramic-reinforced Mg-based composites can be comprehensively improved [79,105]. More specifically, the strengthening effect of nano-ceramic reinforcements with a small volume fraction can be similar to or even better than that of micron-ceramic-reinforced Mg-based composites with similar or higher volume fractions [100,106]. For example, Hassan and Gupta [107] showed that the YS, UTS, and EL of Mg-based composites containing 1 vol% nanosized

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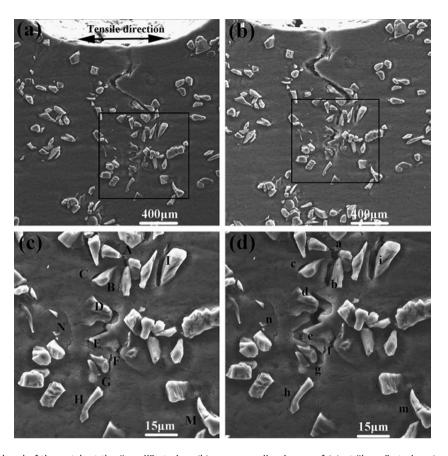


Figure 2: (a) Cracks ahead of the notch at the "small" strains; (b) corresponding image of (a) at "large" strains. (c and d) Further magnification of the rectangular regions in (a) and (b), respectively [95].

SiC particles ( $SiC_p$ ) were higher than those of an AZ91 alloy reinforced with a higher content (10 vol%) of micron- $SiC_p$ .

Similarly, mechanical ball milling is an economical method for producing nano-ceramic-reinforced Mg-based composites. Nano-ceramic particles can also be synthesized in situ by milling precursor elements. Hwang et al. [108] synthesized nano-TiC particle-reinforced Mg-based composites by the mechanical ball milling of Mg, Ti, and C element powders. The tensile strength, toughness, and ductility of the nanocomposite can be improved using nano-TiC<sub>n</sub>, which can inhibit grain boundary sliding. Furthermore, nanoceramic reinforcements can be added to Mg-based composites together with other types of reinforcements (hybrid reinforcements), thus providing greater strengthening efficiency than individual nano-ceramic reinforcements [109,110]. Compared with the AZ91 alloy, the hardness and tensile strength of MgO-Al<sub>2</sub>O<sub>3</sub>-MgAl<sub>2</sub>O<sub>4</sub>-reinforced Mg-based composites increased by 64% (97 HV) and 43% (325 MPa) [111,112], respectively. Tun et al. [113] prepared nano-Al<sub>2</sub>O<sub>3</sub> ceramic particles and Cu metal hybrid-reinforced Mg-based composites by PM combined with microwave sintering (MS) and HE. The significant increase in YS and UTS can be attributed to the grain refinement of Cu and its secondary

phase. HE is an effective way to improve the microstructure defects of casting [114,115]. For a SiC<sub>p</sub>/AZ91 composite prepared by the stirring casting method, the subsequent HE can form grain deformation zones and high energy density dislocations and ultimately promote dynamic recrystallization to refine the matrix grains [98.116]. Subsequent HE can also allow for uniform distribution of SiC nanoparticles within the matrix and a simultaneous increase in YS, UTS, and failure strain [117]. In fact, the surface properties of Mg alloys significantly affect their service life in certain application scenarios. Surface Mg-based composites are the best examples of such materials, which contain evenly distributed reinforcements at the surface, without affecting the chemical composition and structure of the internal substrate [118-120]. To modify the surface properties of Mg alloys, friction stir processing (FSP), a solid-phase method, is advantageous for preparing surface composites on Mg alloys [121]. Grain refinement, improved hardness, wear resistance, and mechanical behavior are common observations in all Mg-based composites produced by FSP [50,122].

Improving the strength of Mg and Mg alloys at high temperatures is a challenge [123]. However, on the one hand, the fine grains obtained by deformation are easily

Table 2: Mechanical properties of ceramic-reinforced Mg-based composites

Materials	Fabrication process (es)	Te	Tensile properties	es	כי	Compressive properties	roperties	Hardness (HV)	Hardness (HV) Young's modulus (GPa)
		YS (MPa)	YS (MPa) UTS (MPa) EL (%)	EL (%)	CYS (MPa)	UCS (MPa)	CYS (MPa) UCS (MPa) Failure Strain (%)		
14 vol% 60 nm SiC/Mg <sub>2</sub> Zn [99] LSU + EV + HPT	LSU + EV + HPT	710 ± 35 ~900	006~	~50	ı	ı	I	~145	86 ± 5
10 vol% 25 µm SiC/Mg [100]	PM + MS	$140 \pm 2$	$165 \pm 2$	$1.5\pm0.8$	I	I	I	$44.3 \pm 0.5$	I
1 vol% 50 nm SiC/Mg [100]	PM + MS	$157 \pm 22$	$203 \pm 22$	$7.6\pm1.5$	ı	ı	I	$43.2 \pm 2.0$	I
1.1 vol% 1 $\mu$ m Al <sub>2</sub> O <sub>3</sub> /Mg [100]	PM + MS	$209 \pm 1$	242 ± 3	$3.5 \pm 0.3$	ı	ı	I	$58.8 \pm 0.5$	I
1.1 vol% 50 nm $Al_2O_3/Mg$ [10]	PM + MS	$175 \pm 3$	246 ± 3	$14.0\pm2.4$	I	I	I	$65.9 \pm 0.9$	I
Pure Mg [101]	PM + MS + HE	$109 \pm 09$	$161\pm16$	$8.9 \pm 1.1$	93 ± 2	$246 \pm 23$	$19.4 \pm 0.8$	ı	I
0.66 vol% 50 nm B <sub>4</sub> C/Mg [101]	PM + MS + HE	$120 \pm 05$	$164\pm06$	$10.0\pm0.3$	$100 \pm 5$	$335\pm11$	$11.8\pm1.8$	ı	I
$1.11 \text{ vol}\% 50 \text{ nm } B_4 \text{C/Mg } [101]$	PM + MS + HE	$82 \pm 11$	$119\pm17$	$5.5 \pm 1.2$	$105 \pm 3$	$331 \pm 10$	$13.3 \pm 1.4$	ı	I
Pure Mg [102]	PM + MS + HE	9 <del>+</del> 96	137 ± 9	$6.0 \pm 3.0$	$51 \pm 9$	$268\pm16$	$18.9\pm1.6$	41 ± 3	I
0.2 vol% 75 nm AlN/Mg [102]	PM + MS + HE	$102 \pm 6$	159 ± 8	$11.0 \pm 2.2$	$65 \pm 1$	$284 \pm 12$	$19.3 \pm 1.5$	49 <del>±</del> 3	I
0.4 vol% 75 nm AlN/Mg [102]	PM + MS + HE	$120 \pm 1$	164 ± 3	$8.4\pm0.9$	72 ± 5	$314 \pm 20$	$17.5 \pm 0.6$	55 ± 3	I
0.8 vol% 75 nm AlN/Mg [102]	PM + MS + HE	129 ± 5	176 ± 3	$6.3 \pm 0.4$	71 ± 3	$307 \pm 17$	$18.3 \pm 2.3$	53 ± 8	1

1/5: yield strength; UTS: ultimate tensile strength; EL: elongation; CYS: compressive yield strength; UCS: ultimate compressive strength; LSU: liquid-state ultrasonic; EV: evaporation; HPT: high-pressure torsion; BM: ball milling; HPS: hot-pressing sintering; PM: powder metallurgy; MS: microwave sintering; HE: hot extrusion.

coarsened; on the other hand, the precipitates obtained after heat treatment tend to dissolve, leading to the loss of strengthening at high temperatures [96,124]. Generally, the addition of nano-ceramic reinforcements to pure Mg or Mg alloys is regarded as one of the most effective methods owing to their high-temperature properties and high thermal stability [125]. For example, Ganguly and Mondal [126] found that the presence of an Al<sub>2</sub>Ca phase network and SiC nanoparticles increased dislocation accumulation and dislocation tangling, resulting in nanocomposites with excellent creep resistance prior to AZ91 alloy. Labib et al. [127] reported that the average grain size of SiC<sub>p</sub>/Mg composites remained almost unchanged after the addition of SiC, whereas their dimensional stability was improved because of the reduction in the coefficient of thermal expansion. During deformation at elevated temperatures, the enhanced high-temperature properties of the composites were mainly attributed to the coefficient of thermal expansion mismatch strengthening, Orowan strengthening, and grain boundary strengthening induced by nano-ceramic particles [128–130].

Although the strengthening effect of the nano-ceramic reinforcements is excellent, homogeneous dispersion is difficult because agglomeration readily occurs. Favorable performance can only be obtained when the reinforcements are evenly distributed and are not clustered in the Mg matrix. For nano-ceramic-reinforced Mg-based composites prepared solely by the traditional stirring casting method, ultrasonic-assisted dispersion can enhance the surface energy of the reinforcements, reduce the surface energy of the melt, improve wettability, provide homogeneous dispersion of the nanoparticles, and significantly improve the mechanical properties of the nanocomposites [87,131]. For example, Chen et al. [99] used an ultrasonicassisted liquid-phase and evaporation method to prepare a SiC nanoparticle-reinforced Mg composite, and the semicoherent combination of SiC nanoparticles and the matrix improved the bonding force and stability of the interfaces. The high-density SiC nanoparticles were uniformly distributed in the matrix, thus greatly improving the strength  $(710 \pm 35 \,\mathrm{MPa})$  and plasticity (50%) of the nanocomposite (Figure 4a). Moreover, the Mg<sub>2</sub>Zn (14 vol% SiC) specimen showed the highest specific strength and the highest specific modulus among all the reported data collected from micropillars of different metals and alloys (Figure 4b).

Poor interfacial bonding between the nano-ceramic reinforcements and the matrix is another reason for the performance degradation. Interface optimization is therefore used to improve the wettability and interfacial bonding between the nano-ceramic reinforcements and the matrix. Several strategies, including adding alloying elements to

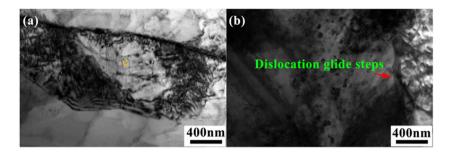


Figure 3: (a) Transmission electron microscopy (TEM) image of the dislocation cell labeled with orange "c." (b) Dislocation glide steps marked with a red arrow in the MgO/Mg nanocomposite [104].

the Mg matrix, surface modification of the reinforcements, and optimization of processing technologies, have been adopted to address the aforementioned problem. Among them, adding appropriate alloying elements to the matrix can effectively improve the wettability and interface bonding between the molten matrix and the nano-ceramic reinforcements [46]. By introducing a small amount of Ti (which has a higher melting point than pure Mg) to the B<sub>4</sub>C ceramic preform, homogeneously distributed B<sub>4</sub>C particles within the matrix were obtained. This is because Ti acts as an infiltration inducer to reduce the surface tension and solid–liquid interfacial tension of the molten Mg. In addition, the addition of metallic particles of similar size can promote the uniform distribution of ceramic nanoparticles [132].

Although ceramic-reinforced Mg-based composites show superior advantages, they still have many short-comings that limit their large-scale application. In addition to the poor interfacial bonding mentioned above, the reinforcement may react with the matrix, resulting in interface instability [133,134]. Therefore, ceramic-reinforced Mg-based composites need further research in the following directions in the future: (1) in terms of the fabrication process, the process flow, parameters, and equipment should be optimized on the basis of the original process, and fabrication processes with low cost

and high efficiency should be developed; (2) adding hybrid ceramic reinforcements may obtain ceramic-reinforced Mg-based composites with comprehensive properties. Individual ceramic particle-reinforced Mg-based composites may improve the stiffness and strength of the composites, but they can also reduce the ductility and toughness. Hybrid ceramic reinforcement-reinforced Mg-based composites can maintain the advantages of each ceramic reinforcement and provide multiphase synergistic strengthening; and (3) it is necessary to deeply study the influence of the interface between the reinforcement and the matrix on the properties of Mg-based composites, explore the interface action mechanism, and seek an effective way to solve the interface problem.

# 4 Carbon-reinforced Mg-based composites

Nanosized reinforcements can improve the strength of Mg composites while maintaining their plasticity. Nanocarbon reinforcements mainly comprise carbon nanotubes (CNTs) [135] and graphene nanosheets (GNSs) [136]. Graphene consists of only a single-atom-thick sheet of sp<sup>2</sup>-

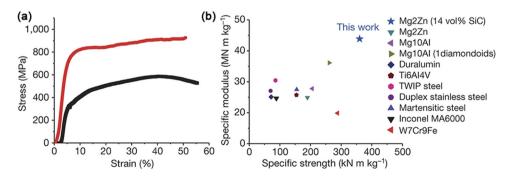


Figure 4: (a) Engineering stress-strain curves of Mg specimens without (black) and with (red) SiC nanoparticles. (b) Relationship between specific modulus and specific YS of Mg<sub>2</sub>Zn (14 vol% SiC) in comparison with the microcolumn test results of other metals and alloys [99].

hybridized carbon atom in a honeycomb lattice. Graphene building blocks can be used to construct single-walled and multiwalled CNTs [137,138]. CNTs are produced by rolling layers of graphene. The walls of CNTs (1–2 nm in diameter) are composed of carbon atoms bonded to each other with strong C-C bonds. Both CNTs and GNSs have low density  $(\sim 1 \text{ g cm}^{-3})$ , ultra-high strength (up to  $\sim 100 \text{ GPa}$ ), high Young's modulus (~1 TPa), excellent mechanical properties, and good thermal stability [57-63,139-143]. These qualities make them ideal reinforcements for high-performance Mg-based composites. Therefore, Mg-based composites reinforced with nanocarbon reinforcements (CNTs and GNSs) have a lower density and higher specific strength than those with ceramic-based and metallic particle reinforcements [144].

Many studies have found that the strength of Mg-based composites reinforced by CNTs and/or carbon nanosheets (CNSs) increased by as much as 40% and that the EL rate exceeded 100% [145-149]. Table 3 lists the mechanical properties of nanocarbon-reinforced Mg-based composites. Ding et al. [151] synthesized CNT-reinforced Mg-based nanocomposites using PW. The CNTs were uniformly dispersed in the matrix, and the nanocomposite exhibited higher strength and reasonable ductility. Its compressive strength and YS reached 504 and 454 MPa, respectively, values that are much higher than those of similar materials reported in other studies. In addition, a novel processing method that combined liquid-state ultrasonic and solid-state stirring was used to prepare bulk nanocomposites [154]. The obtained GNS-reinforced Mg-based nanocomposite showed uniform dispersion of the GNSs and significantly enhanced hardness. Rashad et al. [152,153,155-157] studied GNSs and/ or CNT-reinforced Mg-based nanocomposites using semipowder metallurgy (SPM) followed by HE. For (10 wt% Ti + 0.18 wt% GNSs)/Mg composites, the addition of GNSs enhanced the YS, UTS, and ductility of the matrix. The increased ductility of GNS-reinforced Mg-based nanocomposites has rarely been reported. Rashad et al. [153] also studied the synergistic effect of CNTs and graphene on the mechanical properties of pure Mg. Under the same ratio of using either only CNTs, only graphene, or a combination of CNTs and graphene, CNTs exhibited the best strengthening effect, whereas the CNTs + graphene hybrid reinforcements exhibited the best ductility.

The uniform distribution of the nanocarbon reinforcements is the premise underlying the excellent mechanical properties of nanocarbon-reinforced Mg-based composites. As shown in Figure 5a and b, the uniform distribution and continuous CNTs or graphene-reinforced composites have mechanical properties that exceed those of carbon-fiber-reinforced composites [158]. The best bearing capacity achieved thus far was achieved by designing continuous CNT or graphene preforms in the composite design. As shown in Figure 5c, Kinloch et al. [158] summarized four dispersion methods based on the combination of CNTs/GNSs and functional groups. Owing to the second-phase strengthening effect of the high modulus of the nanocarbon reinforcements, Mg-based nanocomposites usually have a higher hardness than the matrix, which limits local deformation during the indentation process [159]. The high elastic modulus of the nanocarbon reinforcements indicates a high load-bearing capacity and prevents damage to the worn surface. Once the load is removed, the reinforcements are restored to their original shape. Therefore, the wear resistance of Mg-based composites with highmodulus reinforcements can be further improved [160].

CNTs may cause agglomeration because of the weak van der Waals forces between the carbon atoms. Similarly, owing to the strong van der Waals force and the  $\pi$ - $\pi$  attractive force, CNSs are prone to aggregation [154]. Therefore, the uniform distribution of these nanocarbon reinforcements in the Mg matrix is a major problem. Compared with adding reinforcements in the matrix directly, the in situ method has the advantages of good interfacial wettability and strong interfacial bonding with the matrix [161] and finer size and uniform distribution of in situ synthetic reinforcements [162,163]. Recently, a preparation process where chemical predispersion was combined with stirring casting was used to achieve good dispersion of CNSs in the Mg matrix [31]. This process effectively solved the problem of dispersing CNSs using the stirring casting method and prepared a CNS-reinforced Mg-based composite with high strength while maintaining plasticity.

The ball-milling method can also significantly improve the aggregation of CNSs. GNSs showed uniform distribution in the matrix with no significant graphene aggregation with increasing ball-milling time [164,165]. Good interfacial bonding can transfer the load from the soft matrix to the nanocarbon reinforcements to improve the strength of the composites. In addition, the noncoherent interfaces between CNTs and the matrix lead to poor interfacial bonding [166,167], and the natural nonwettability of CNTs in molten Mg leads to inhomogeneous dispersion of the CNTs [168,169]. Therefore, the mechanical properties of CNTs/Mg nanocomposites are much lower than predicted [170], which largely limits the application of CNTs in Mg-based composites [171]. To overcome these problems, some improved methods have been applied, such as rapid MS and spark plasma sintering technologies [172-174]. However, it is difficult to avoid the agglomeration of CNTs in the matrix. DMD and ultrasonicassisted extrusion can prevent the agglomeration of CNTs [175–177], but these methods usually introduce structural

Table 3: Mechanical properties of carbon-reinforced Mg-based composites

Materials	Fabrication	ī	Tensile properties	S		Compressive properties	perties	Hardness (HV)	Young's
	process (es)	YS (MPa)	UTS (MPa)	EL (%)	CYS (MPa)	UCS (MPa)	Failure strain (%)		modulus (GPa)
AZ91 [150]	SPM + HE	168 ± 5.0	$215 \pm 6.0$	7.0 ± 0.2	I	ı	I	72.4 ± 2.0	ı
1 wt% CNT/AZ91 [150]	SPM + HE	$173 \pm 4.0$	$228\pm5.0$	$8.6\pm0.1$	ı	ı	I	$79.2 \pm 2.0$	ı
2 wt% CNT/AZ91 [150]	SPM + HE	$197 \pm 4.5$	$263 \pm 5.5$	$8.7 \pm 0.2$	ı	ı	I	$87.1\pm1.5$	ı
3 wt% CNT/AZ91 [150]	SPM + HE	$250 \pm 3.8$	$301 \pm 4.5$	$\textbf{9.4} \pm \textbf{0.1}$	I	I	I	$94.1 \pm 2.0$	ı
4 wt% CNT/AZ91 [150]	SPM + HE	$187 \pm 3.5$	$248 \pm 3.9$	$8.5\pm0.1$	ı	I	I	$84.3 \pm 1.6$	ı
Ni-CNT/Mg [151]	PM + BM + HPS	454	504	10.5	ı	ı	I	I	ı
1 wt% (MgO-CNT)/AZ91 [150]	SPM + HE	$190 \pm 3.6$	$260 \pm 4.2$	$7.6 \pm 0.1$	I	I	I	$80.2\pm1.5$	ı
2 wt% (MgO-CNT)/AZ91 [150]	SPM + HE	$210 \pm 5.0$	$294 \pm 6.0$	$8.2 \pm 0.2$	ı	I	I	$89.5\pm1.0$	ı
3 wt% (MgO-CNT)/AZ91 [150]	SPM + HE	284 ± 4.6	$331 \pm 5.0$	$8.6\pm0.1$	1	1	I	$96.4 \pm 1.2$	1
4 wt% (MgO-CNT)/AZ91 [150]	SPM + HE	$206 \pm 3.7$	$272 \pm 4.8$	$8.0\pm0.1$	I	I	I	$86.5\pm1.2$	ı
Mg-1 wt% Al-1 wt% Sn [152]	SPM + HE	$161\pm04$	$236 \pm 5.1$	$16.7\pm0.3$	ı	1	ı	I	ı
Mg-1 wt% Al-1 wt% Sn-0.18	SPM + HE	$208 \pm 5.3$	$269 \pm 03$	$10.9 \pm 3.4$	I	1	ı	I	1
wt% GNSs [152]									
Pure Mg [152]	SPM + HE	$104 \pm 4$	$164 \pm 5$	$6.2 \pm 0.2$	136 ± 3	286 ± 6	$12 \pm 0.2$	46 ± 2	$7.0 \pm 0.3$
Mg-1 wt% Al [152]	SPM + HE	155 ± 3	202 ± 3	$6.9 \pm 0.5$	$100 \pm 2$	377 ± 8	$18\pm0.5$	50 ± 4	$12.8\pm0.4$
Mg-1 wt% Al-0.60 wt% GNSs [152]	SPM + HE	204 ± 9	265 ± 8	4.0 ± 0.6	230 ± 5	407 ± 3	13 ± 0.3	<b>63</b> ± 2	$17.2 \pm 0.1$
Mg-1 wt% Al-0.60 wt% CNTs [152]	SPM + HE	210 ± 5	287 ± 4	$10 \pm 0.3$	237 ± 4	425 ± 5	$12.6\pm0.2$	<b>61</b> ± <b>5</b>	15.7 ± 0.3
Mg-1 wt% Al-0.60 wt% [1:5] (CNT + GNSs) [153]	SPM + HE	185 ± 4	234 ± 3	$16.4\pm0.5$	167 ± 6	397 ± 3	15 ± 0.4	56 ± 3	$15.0\pm0.2$

YS: yield strength; UTS: ultimate tensile strength; EL: elongation; CYS: compressive yield strength; UCS: ultimate compressive strength; SPM: semipowder metallurgy; HE: hot extrusion; PM: powder metallurgy; BM: ball milling; HPS: hot-pressing sintering; CNT: carbon nanotube; CNS: carbon nanosheet/nanoplate.

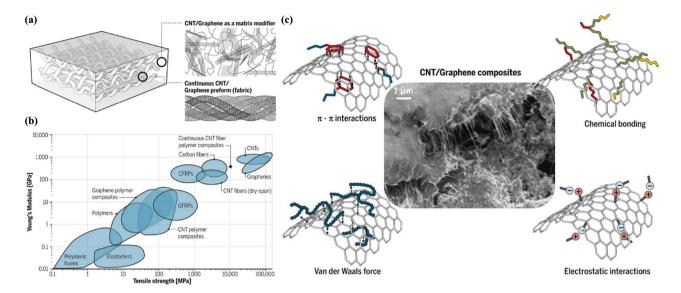


Figure 5: (a) Schematic of a composite that consists of the uniform distribution and continuous CNTs or graphene as reinforcements. (b) Ashby plot of Young's modulus and tensile strength of composites containing nano-reinforcements [158]. (c) Schematic diagram of four dispersion methods for the interaction between CNTs or GNSs and a polymer matrix [158].

defects in the CNTs. However, the in situ synthesis of CNTs can be used to solve this problem. Using an in situ reaction method, CNT-reinforced Mg-based nanocomposites with a small matrix grain size, uniform CNT dispersion, and good interface bonding can be prepared. For example, Sun et al. [178] used a Co catalyst to synthesize uniformly dispersed CNT/Mg nanocomposite powders in situ. Using chemical vapor deposition (at 480°C) combined with a Co catalyst, CNTs with uniform morphology and high purity were synthesized on Mg powder. Yuan et al. [179] improved the mechanical properties of a GNS/AZ91 alloy using in situ-synthesized MgO nanoparticles, which significantly improved the interfacial bonding between the GNSs and α-Mg. This is attributed to the formation of a semicoherent interface, MgO/ $\alpha$ -Mg, and the distortion area interfacial bonding of GNS/MgO, as shown in Figure 6. When 0.5 wt% GNPs was added, the YS and the EL of the nanocomposite increased by 76.2 and 24.3%, respectively, compared to the matrix. The ball milling of the CNT/Mg composite powder enabled CNTs to be embedded in the Mg matrix, thereby forming an ideal combination of CNTs and pure Mg. Hence, the obtained nanocomposite powders produced a Mg-based composite with uniform CNT distribution.

The surface coating of nanocarbons is also one of the most effective ways to improve the wettability and interfacial bonding between the reinforcements and the matrix, such as Ni [180], Al [46], Si [181], and MgO [150]. For example, Yuan *et al.* [150] studied a new method to improve the interfacial bonding strength by coating magnesium oxide (MgO) nanoparticles on the surface of CNTs. The results showed that a nanoscale contact interface and diffused bonding interface formed between the CNTs and

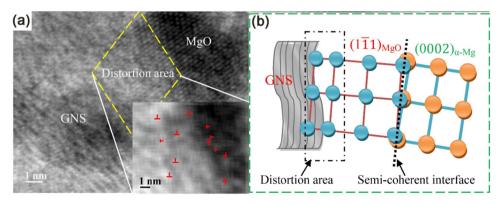


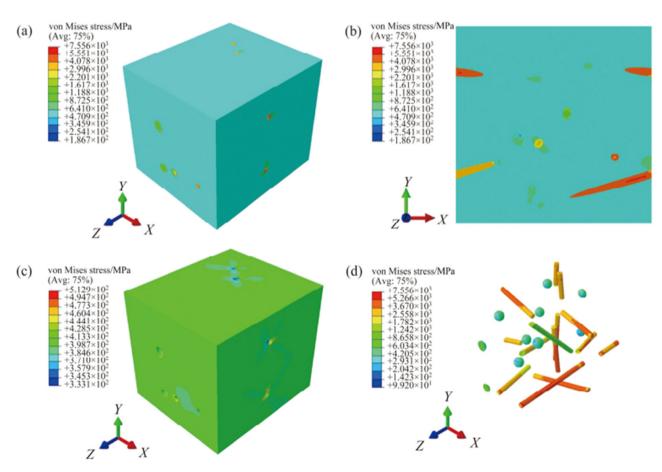
Figure 6: (a) High-magnification image of the distortion area. (b) Schematic illustration of the interface of MgO/ $\alpha$ -Mg and GNS/MgO [179].

MgO in the composite. Consequently, this ensured the effective transfer of the load from the matrix to the CNTs. In recent years, among numerous research methods for Mg alloys, numerical simulations and molecular dynamics simulations have gradually become important [182–186]. As shown in Figure 7, Zhou *et al.* [186] investigated the effects of interfacial properties and hybrid nano-reinforcements on the dynamic mechanical properties of Mg-based composites and hybrid strengthening mechanisms by comparing three-dimensional (3D) simulation results and experimental data. The use of numerical simulation calculations may save more time for optimizing the interfacial bonding and mechanical properties of Mg-based composites in the future, as well as the development of new Mg-based composites.

Nanocarbon-reinforced Mg-based composites have the advantages of low density, high specific strength, and stiffness, which have attracted the attention of scholars worldwide. However, carbon nano-reinforcements also have deficiencies, including agglomeration and uneven dispersion, both of which can cause their performance to be far lower than expected. In addition, the weak interfacial bonding between the nanocarbon reinforcement and the matrix reduces the properties of nanocarbon-reinforced Mg-based composites. Future research should focus on developing new nanocarbon reinforcements, finding appropriate interface optimization methods, and developing reasonable fabrication processes. It is worth mentioning that the combination of experiments and modern computer simulation technology can directly and effectively find the best process plan, thereby avoiding the blindness of experiments.

# 5 Other functional Mg-based composites

As biodegradable implants, Mg-based alloys have unique characteristics that Zn- and Fe-based alloys do not



**Figure 7:** 3D numerical simulations of the dynamical tensile response of hybrid CNT/SiC-reinforced Mg-based composites considering interface cohesion at  $100^{\circ}$ C and strain rate of 1,000 s<sup>-1</sup> [186]. (a) 3D composite, (b) cross section of the composite, (c) Mg matrix, and (d) hybrid reinforcements.

have [55,117,187–189]. Mg-based composites used as biodegradable implants are promising candidates for avoiding the stress shielding problems caused by high Young's modulus metals, such as Zn (~90 GPa) or Fe (~211.4 GPa). This is because pure Mg has Young's modulus (~45 GPa) that is closest to that of natural bone (~5–23 GPa) [190–192]. Therefore, Mg-based composites have superior mechanical and corrosion performance for biomedical applications [193–196].

Mg has a theoretical hydrogen storage capacity of up to 7.6% (mass fraction) and is widely considered a hydrogen storage material. Magnesium hydride/magnesium (MgH<sub>2</sub>/ Mg) composites are potential solid-state hydrogen storage materials. However, the thermodynamic and kinetic properties are far from suitable for practical applications at this stage [197]. Reinforcements can be used as effective catalysts to reduce the reaction enthalpy of Mg-based hydrogen storage materials and significantly improve their hydrogenation/dehydrogenation performance [198–200]. As shown in Figure 8, Jeon et al. [201] reported a type of air-stable composite material that can store high-density hydrogen (up to 6 wt% of Mg, 4 wt% of composites) and has fast kinetics (loading at 200°C for less than 30 min) without using expensive heavy metal catalysts. The composites can enhance the kinetics by reducing the length of the hydrogen diffusion path and reducing the required thickness of the poorly permeable hydride layer formed during the absorption process. In addition, high-energy ball milling can produce hydrogen storage materials with the advantages of

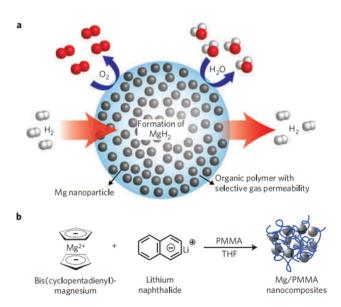


Figure 8: Mg nanocomposites in a gas-barrier polymer matrix.
a) Schematic of hydrogen storage composite material: high-capacity Mg nanocomposites are encapsulated by a selectively gaspermeable polymer. b) Synthetic approach to formation of the Mg nanocomposites [201].

fresh and highly reactive surface nanocrystalline materials and the ability to form nanocomposites [202,203]. Li *et al.* [204] mechanically ball-milled additives (TiO<sub>2</sub>, Ni, and CNTs) and Mg to improve the hydrogen storage performance. The composites can absorb 7.5 wt% hydrogen in 60 s at a hydrogen pressure of  $5.0 \times 10^{-5}$  Pa, and release 6.5 wt% hydrogen in 600 s at a hydrogen pressure of  $5.0 \times 10^{-5}$  Pa at 260°C. The composites exhibited very high absorption, hydrogen capacity, and remarkable dynamic performance during the hydrogenation/dehydrogenation processes.

The micro-galvanic effect is the main corrosion mechanism of reinforcement-reinforced Mg-based composites. This may be attributed to the presence of reinforcements that activate the corrosion of Mg alloys owing to the occurrence of galvanic corrosion. This effect increases with increasing reinforcement content [205]. The addition of reinforcements will also introduce several interfaces, dislocations, twins, compounds, and other structures within the Mg matrix, all of which are closely related to the corrosiveness of Mgbased composites. Nonetheless, surface treatment can significantly improve the corrosion resistance of Mg alloys/ Mg-based composites [196,206-211]. In addition, compared to pure Mg or Mg alloys, some Mg-based composites have better high-temperature properties [99], superior damping [212,213], and improved creep properties [96]. Hence, more research needs to be conducted in these areas so that Mg-based composites can be used in aviation, aerospace, automobiles, civil air-conditioning, and other related fields on a large scale.

### 6 Summary and outlook

This review summarizes the current research progress, challenges, and upcoming explorations of reinforcementreinforced (especially nano-reinforcements) Mg-based composites. Micron-reinforcement-reinforced Mg-based composites, owing to the larger size of the reinforcements, are more likely to contain fracture-inducing defects, thus making particle fracture more common. However, Mg and its alloys reinforced with nano-reinforcements have the advantages of light weight, superior strength, high modulus, and good wear resistance, thus exhibiting improved mechanical properties without significantly reducing the ductility normally associated with the addition of micro-reinforcements, making them an attractive choice for lightweight structural applications. To date, nano-SiC particles or CNTreinforced Mg-based composites have been studied systematically. Their YS, compressive strength, Young's modulus,

and EL have reached ~710 MPa, ~504 MPa, ~86 GPa, and ~50%, respectively, which are the highest reported values for Mg-based composites. To obtain excellent mechanical and physical properties, the microstructure and interface of Mg-based composites need to be considered. It is necessary to understand the influence of reinforcement distribution, orientation, interfacial reaction phases, and matrix microstructure on the properties of Mg-based composites. However, there are three main drawbacks that limit the large-scale application of Mg-based nanocomposites: (a) the agglomeration and inhomogeneous dispersion of these reinforcements, (b) poor interfacial bonding between the reinforcements and the matrix, and (c) most Mg-based composites have poor corrosion resistance and relatively limited high-temperature properties. Although many traditional and novel methods (such as in situ technology) have been employed to disperse the reinforcements into metal matrices, it is very important to use less expensive and simpler methods to produce Mg-based composites without the need for specific technical equipment. Therefore, this remains a significant challenge.

Four aspects should be considered in future work. First, it is necessary to further explore Mg alloy matrix composites. Thus far, most of the work has focused on the pure Mg matrix; however, very few studies have been conducted on Mg alloy matrices. In fact, Mg alloys, rather than pure Mg, are widely used in the industry. Second, effective and inexpensive reinforcements should also be explored in depth. More reasonable sizes and combinations of reinforcements can be studied to improve the comprehensive performance of Mg-based composites. Third, new reinforcement dispersion technology, a less expensive and simpler production technology that is used to disperse the reinforcement into the Mg matrix, should be studied. Reinforced coatings produced by in situ reaction or reinforcement generation should be used to fabricate high-performance composite materials. This can be achieved through proper surface treatment and the development of appropriate coating procedures. This can enable the large-scale production of Mg-based composites. Finally, further discussion on the physical model and the corresponding accuracy is beneficial for the development of new Mg-based composites.

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