

Deformation Mechanism of Polypropylene Composites Filled with Magnesium Hydroxide

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ABSTRACT: The fracture behavior and deformation mechanism of polypropylene (PP) composites filled with magnesium hydroxide [Mg(OH)₂] were investigated. The incorporation of Mg(OH)₂ particles into the PP matrix led to an increase in Young's modulus and a significant reduction in the tensile yield strength and elongation at break. Surface modification on filler particles with stearic acid could reduce the interfacial adhesion between the filler and PP matrix and improve the stress transferability. The deformation mechanism of the Mg(OH)₂/PP composites depended on the interfacial adhesion and the deformability of ligaments between microvoids caused by debonding. The deformability

of the ligaments could be significantly improved by surface modification on the particle surface. The dependence of the deformation behavior of the Mg(OH)₂/PP composites on the filler content was in accordance with percolation theory. The agglomeration of microvoids and fibrillation of ligaments in the PP composites with excessive filler content indicated the weak resistance of the polymer matrix to crack propagation and premature fracture in a brittle manner. © 2005 Wiley Periodicals, Inc. *J Appl Polym Sci* 97: 1922–1930, 2005

Key words: composites; morphology; poly(propylene) (PP)

INTRODUCTION

Polypropylene (PP) has been widely used in different application fields because of its good mechanical and physical properties. However, its poor flame resistance hinders its practical applications in some fields. Magnesium hydroxide [Mg(OH)₂] is a good flame retardant because of its high decomposition temperature and smoke suppressibility, and it is widely used in polyolefin composites. The disadvantage of Mg(OH)₂, however, is the high levels (>60% w/w) required to achieve the desired flame-retardant effect.¹ Therefore, the improvement in flame retardancy is accompanied by a dramatic deterioration in the composite mechanical properties.

The mechanical properties of polymer composites depend largely on interfacial interactions between the filler and matrix and the dispersion of the filler in the matrix. Poor interfacial adhesion is regarded as the most important factor responsible for deterioration in the mechanical properties of polymer composites, and many studies have been focused on the improvement of composite interfacial adhesion. Weak or strong adhesion can be reached by virtue of different modifications on the filler or matrix and results in different mechanical properties of composites. Jancar and

Kucera^{2,3} studied PP composites filled with Mg(OH)₂ having strong or weak adhesion and concluded that strong adhesion led to increased modulus and tensile strength, whereas zero adhesion caused decreased tensile strength and increased elongation at break. Hornsby and Watson⁴ studied the influence of filler surface modification on the mechanical properties of PP composites containing Mg(OH)₂. Fatty acid derivatives were found to be the most effective surface treatment for toughness improvement, which was accompanied by declines in flexural and tensile properties. Liu and Naumam⁵ suggested that weak adhesion is sufficient for high toughness when there is a soft interlayer between the filler particles and polymer matrix. The dispersion of the filler particles in the polymer matrix also plays an important role in determining the mechanical properties of highly filled composites, which depends on the interaction between the particles.⁶ The interaction of the particles is mainly determined by the size and distribution of the filler particles. Generally, large filler particles induce premature cracks in the polymer matrix, whereas small particles lead to aggregation of the filler particles. Those are both detrimental to the mechanical properties of composites. However, the dispersion of particles in the polymer matrix is quite good in many cases.⁴

According to the criterion of the brittle–ductile transition proposed by Wu,^{7,8} in rubber-toughened nylon 6,6, the interparticle matrix ligament thickness has to

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be smaller than a certain critical value for enhanced toughness. Therefore, mineral fillers can be predicted as novel toughening modifiers in a polymer matrix to overcome the main drawback of sacrificing the elastic properties of the rubber-toughened polymer. The only purpose of blending rubber particles into a matrix is to cavitate at the beginning of deformation and to loosen the constraint to the deforming matrix ligaments between particles. Thus, soft rubber particles can be replaced by rigid particles according to the concept of percolation theory.^{9,10} Wilbrink et al.¹¹ investigated the toughenability of nylon-6 with calcium carbonate (CaCO_3) filler particles. The poor Izod impact strength of the nylon-6/ CaCO_3 composites was attributed to a combination of the high-strain-rate-induced elevation of the plastic resistance and the development of a triaxial tensile stress due to incomplete separation of the rigid particles from the matrix in the crucial early stages of the impending matrix ligament stretching process. The work of Bartzczak et al.¹² demonstrated that the same volume fraction of CaCO_3 filler as that of rubber particles could achieve the same level of toughness increase in polyethylene (PE). The results indicated that the significant increase in toughness was due to the debonding of CaCO_3 filler particles from the PE matrix at the yield of the matrix, which permitted the matrix ligaments to undergo large and unhindered plastic extensions. The normal states of adhesion of the CaCO_3 particles to the matrix satisfied the debonding requirements in all instances. Fu and coworkers^{13–16} also reported an impressive toughening effect in PE/ CaCO_3 composites at a certain ligament thickness. However, few reports have concerned considerable improvements in the toughness of PP composites filled with $\text{Mg}(\text{OH})_2$.

The debonding of filler particles appears to be the dominant factor influencing both the stress-strain and volumetric behavior of particulate-filled composites. Kwon et al.¹⁷ studied the formation of air holes between CaCO_3 and a resin matrix at various draw ratios. The air holes were created by dewetting in the interface between the filler particles and polymer matrix and enlarged up to 300% elongation by dewetting of the calcite. However, further enlargement was attributed to the breakdown of a superimposed fibril structure and the merging effect of preformed air holes of various sizes upon further stretching. Velasco et al.¹⁸ studied the fracture behavior of an injection-molded $\text{Mg}(\text{OH})_2$ -filled PP copolymer at a low strain rate as a function of the $\text{Mg}(\text{OH})_2$ content. The main crack propagation mechanism identified in these materials was ductile tearing, which was initiated by microvoid nucleation at the interfacial sites because of the weak particle-matrix interface.

In this study, we investigated the micromechanical deformation mechanism of $\text{Mg}(\text{OH})_2$ /PP composites at various filler loadings. The influence of the filler

particle surface modification with stearic acid (SA) on the deformation mechanism of the composites was studied with respect to an unmodified filler. In addition, the tensile properties and impact strength of the composites were estimated as functions of the filler volume fraction.

EXPERIMENTAL

Materials

A PP homopolymer (ExxonMobil PP 1304E1, ExxonMobil Chemical) with a melt flow index of 11 g/10 min (2.16 kg at 230°C) was used as a matrix polymer. $\text{Mg}(\text{OH})_2$ (H5, Martinswerke Magnifin, Bergheim, Germany) with an average particle size of 1.25–1.45 μm and a specific surface area of 4.0–6.0 m^2/g without any surface treatment was used as a filler. SA was used as a surface modifier without further purification.

Blending and specimen preparation

PP matrix and $\text{Mg}(\text{OH})_2$, uncoated or coated with SA (1 wt %), were mixed with a Berstoff ZE25A corotating twin-screw extruder (length/diameter = 41, diameter = 25 mm; Hannover, Germany) with a temperature profile of 180/190/200/200/200/200/190/200°C and a rotating speed of 250 rpm; the melting temperature was about 220°C. Various volume contents of $\text{Mg}(\text{OH})_2$ were used, ranging from 0 to 31%. The extrudates were pelletized and dried in a vacuum oven at 80°C for 18 h for specimen preparation.

The test specimens were injection-molded with an injection-molding machine. The temperatures of the barrel and mold were set at 220 and 40°C, respectively. Before the mechanical tests, the specimens were dried at 23°C *in vacuo* for 18 h.

Mechanical properties

The tensile properties were determined with an Instron 4465 tensile machine (Instron Corp., Canton, MA) according to ASTM D 638 at a crosshead speed of 50 mm/min and a testing temperature of 20°C. The dimension of the dumbbell bar followed ASTM D 638M-93 type M-II. Thus, the gauge length, width, and thickness of the dumbbell tensile bars were 25, 6, and 2 mm, respectively.

Notched Izod impact specimens had a 2-mm-deep, 45° V-shape notch with a tip radius of 0.25 mm on rectangular bars (80 × 10 × 4 mm^3) according to ISO 178. The test was performed with a Ray-Ran impact test (Ray-Ran Test Equipment, Ltd., Nuneaton, United Kingdom) according to ISO 180 with an impact velocity of 3.5 m/s and a pendulum weight of 0.818 kg at a

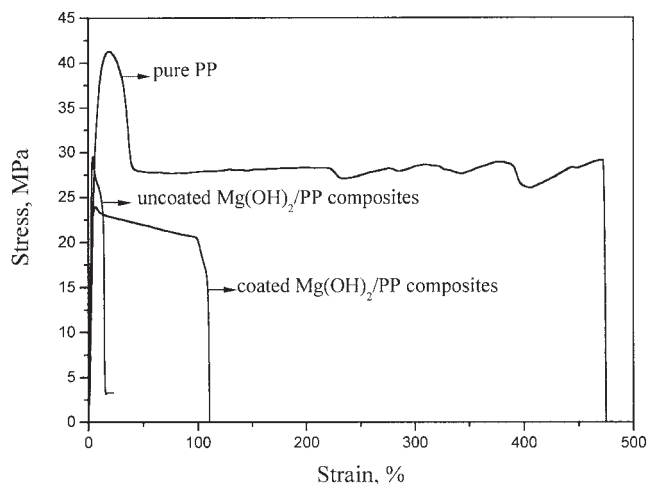


Figure 1 Load-displacement curves of pure PP, uncoated $\text{Mg}(\text{OH})_2/\text{PP}$ composites, and coated $\text{Mg}(\text{OH})_2/\text{PP}$ composites ($\phi_f = 0.31$).

testing temperature of 23°C. The test results were the average data of five samples.

Morphology observations

The morphology and deformation mechanism of the $\text{Mg}(\text{OH})_2/\text{PP}$ composites were studied by scanning electron microscopy (SEM) analysis of the fracture zone. The surfaces for filler dispersion observations were obtained from notched impact specimens that were immersed in liquid nitrogen and fractured under high-speed impact. The fracture surfaces for the deformation mechanism study were prepared by notched Izod impact tests at 23°C. All the surfaces were then gold-coated and observed with a Hitachi S-2150 scanning electron microscope (Santa Clara, CA).

RESULTS AND DISCUSSION

Tensile properties

The stress-strain curves of the $\text{Mg}(\text{OH})_2/\text{PP}$ composites at a strain rate of 50 mm/min are shown in Figure 1. For comparison, the stress-strain curve of pure PP is also presented. The incorporation of the uncoated filler into the PP matrix leads to a drastic decrease in the tensile yield stress and elongation at break, whereas the surface treatment of the filler particles with SA can increase the elongation at break of the PP composites, with the cost of decreasing tensile yield stress.

The dependence of Young's modulus of the $\text{Mg}(\text{OH})_2/\text{PP}$ composites on the volume fraction of the filler is shown in Figure 2. The relative Young's modulus ($M_{\text{rel}} = M_c/M_m$, where M_c is the composite modulus and M_m is the matrix modulus) is used for

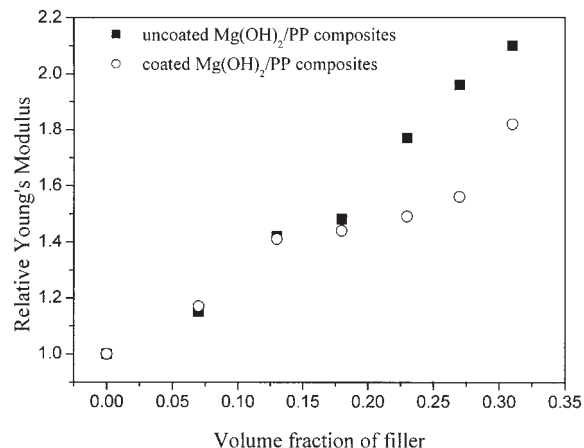


Figure 2 Dependence of the relative Young's modulus of $\text{Mg}(\text{OH})_2/\text{PP}$ composites on the filler volume content.

the description of the dependence of Young's modulus on the filler content. Young's modulus increases with increasing filler content, and this is attributed to the higher modulus and stiffness of inorganic particles. In a weak polymer matrix such as PP, the filler carries a large part of the load and reinforces the polymer^{6,19} at the elastic deformation stage. The influence of the filler content on Young's modulus of the coated $\text{Mg}(\text{OH})_2/\text{PP}$ composites is quite similar to that of the $\text{Mg}(\text{OH})_2/\text{PP}$ composites at a relatively small loading level. However, with the filler content increasing further, the reinforcement of inorganic filler particles on the PP matrix is obviously weakened by the surface treatment of SA on $\text{Mg}(\text{OH})_2$.

The dependence of the tensile yield strength of $\text{Mg}(\text{OH})_2/\text{PP}$ composites on the filler content is shown in Figure 3. For the uncoated $\text{Mg}(\text{OH})_2/\text{PP}$ composites, the tensile yield strength decreases considerably with increasing filler content, whereas for

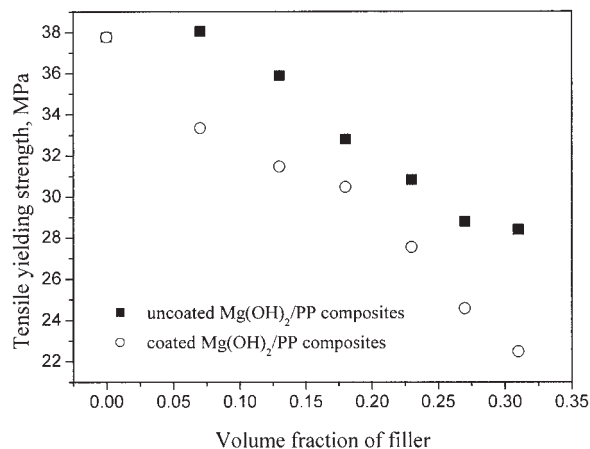


Figure 3 Dependence of the tensile yield strength of $\text{Mg}(\text{OH})_2/\text{PP}$ composites on the filler volume content.

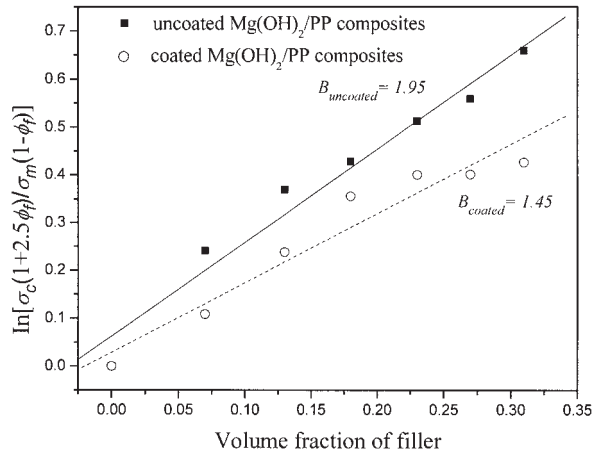


Figure 4 Dependence of the relative tensile strength of Mg(OH)₂/PP composite on the filler volume content.

the coated Mg(OH)₂/PP composites, the reduction of the tensile yield strength with increasing filler content is more significant because of the surface treatment of the particle surfaces with SA. The decrease in the tensile yield strength can be regarded as the result of the reduction of the effective load-bearing cross section and the weak interfacial adhesion between the polymer matrix and filler. The latter can be estimated by a semiempirical correlation developed by Pukanzsky and coworkers^{20,21} and based on the composition dependence of the tensile yield strength in heterogeneous polymer systems:

$$\sigma_c = \sigma_m \frac{1 - \phi_f}{1 + 2.5\phi_f} \exp(B_y \phi_f) \quad (1)$$

where σ_c and σ_m are the yield stresses of the composites and polymer matrix, respectively; ϕ_f is the volume fraction of the filler; and B is an empirical parameter related to the stress transfer and proportional to the interfacial adhesion. A linear plot of $\ln[\sigma_c(1 + 2.5\phi_f)/\sigma_m(1 - \phi_f)]$ against the filler content is shown in Figure 4, and parameter B of the composites is calculated as the slope of a linear curve. The result demonstrates that the interfacial adhesion of the coated Mg(OH)₂/PP composites is obviously weaker than that of the uncoated Mg(OH)₂/PP composites, and this is consistent with other reports.^{22,23} The surface treatment with SA decreases the surface free energy of the filler drastically and thus decreases the interfacial tension and the reversible work of adhesion. The reduction of the interaction between the matrix and particles leads to decreases in the modulus and strength, as shown in Figures 2 and 3.

It is well known that the existence of inorganic particles changes the stress distribution in the polymer matrix under an external load and that the particles act as stress concentrators because of the discrepancy of

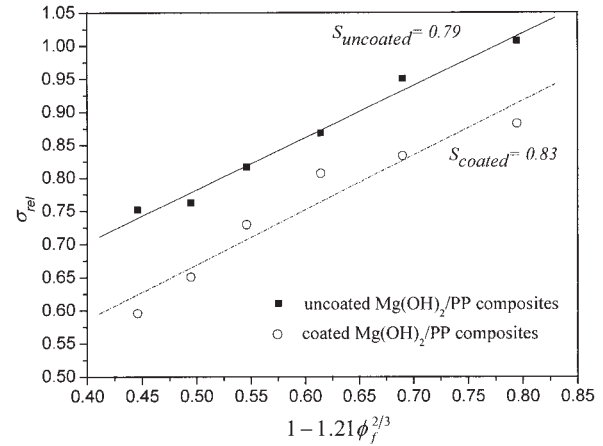


Figure 5 Dependence of the relative yield stress (σ_{rel}) of Mg(OH)₂/PP composites on the filler volume content.

the modulus and the deformability between the filler and polymer. This is detrimental to stress transferability and subsequently the deformation process because of the nondeformability of inorganic particles. The degree of stress concentration of Mg(OH)₂/PP composites can also be estimated by a modified form of the equation based on the composition dependence of the tensile yield strength developed by Nicolais and Narkis²⁴ as follows:

$$\sigma_c/\sigma_m = (1 - 1.21\phi_f^{2/3})S \quad (2)$$

where S is the stress concentration factor and varies between 0.2 and 1.0 according to finite element analysis. Parameter S of the composites is determined from a linear plot of the relative yield stress ($\sigma_{rel} = \sigma_c/\sigma_m$) against the filler content according to eq. (2), as shown in Figure 5. The result suggests that the surface treatment of filler particles with SA can reduce

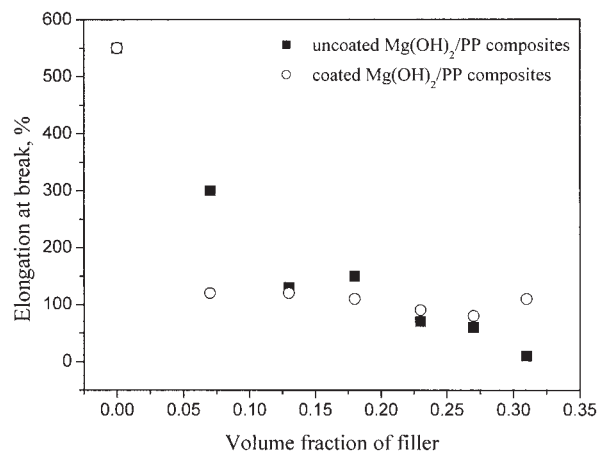


Figure 6 Dependence of the elongation at break of Mg(OH)₂/PP composites on the filler volume content.

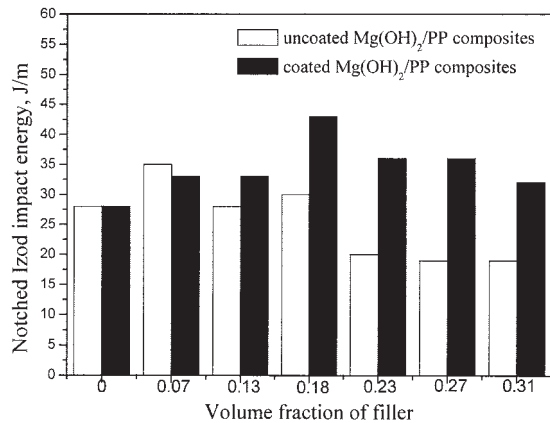


Figure 7 Dependence of the notched Izod impact strength of Mg(OH)₂/PP composites on the filler volume content.

the stress concentration of the coated Mg(OH)₂/PP composites under an external load in comparison with the uncoated Mg(OH)₂/PP composites. The decrease in the degree of stress concentration indicates the improvement, by the surface treatment, of the stress transferability of the matrix surrounding filler particles, which can be regarded as the result of the reduction in the interfacial adhesion between the filler and matrix.

The dependence of the elongation at break of the Mg(OH)₂/PP composites on the filler content is shown in Figure 6. For the uncoated Mg(OH)₂/PP composites, as the result of the poor interfacial adhesion between the matrix and filler and the limited deformability of the polymer matrix, which is related to the decreased mobility of the adsorbed chains on the filler particle surface, the elongation at break decreases dramatically with increasing filler content. However, for the coated Mg(OH)₂/PP composites, no significantly change in the elongation at break is observed with increasing filler content after a abrupt decrease of the composite at a small loading level (ϕ_f

= 0.07) of the filler. The elongation at break remains at a high level when ϕ_f is up to 0.31 in the coated Mg(OH)₂/PP composites. This is probably due to an improvement in the matrix deformability caused by SA under tensile stress, which is discussed in detail later.

Impact strength

The dependence of the notched Izod impact strength on the filler content is shown in Figure 7. For the uncoated Mg(OH)₂/PP composites, the impact strength first increases with increasing filler content. However, when the filler content reaches a high level, the impact strength of the uncoated Mg(OH)₂/PP composites is lower than that of pure PP. For the coated Mg(OH)₂/PP composites, the effect of the filler content on the toughness is similar to that of the uncoated Mg(OH)₂/PP composites. The improvement of the coated filler with respect to the impact resistance is more remarkable, especially at a high level of filler loading. The maximum value of the impact strength appears at the filler content of 18 vol %. With increasing filler content, the impact strength of the coated Mg(OH)₂/PP composites decreases, but it is still higher than that of pure PP. All the specimens of the Mg(OH)₂/PP composites with various filler contents are fractured in a brittle manner, and slightly plastic deformation is observed on the fracture surfaces, even for the composites with higher impact strength.

Morphology analysis

Figure 8 shows the microstructure of the uncoated Mg(OH)₂/PP (31/69 v/v) composites, and the cryogenic fracture surface was etched with hydrochloride (HCl). The filler particles were completely dissolved after immersion in the acid solution for 1 h. This observation suggests that the filler particles are well

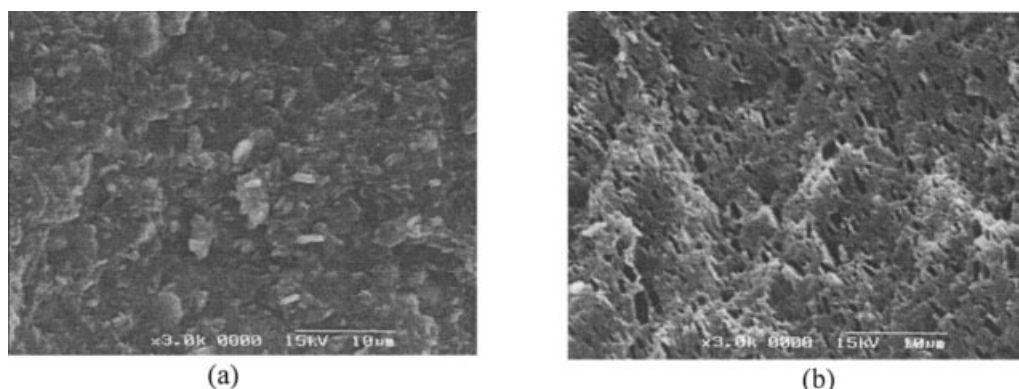


Figure 8 SEM micrographs of the cryogenic fracture surfaces of uncoated Mg(OH)₂/PP composites: (a) Mg(OH)₂/PP ($\phi_f = 0.31$) and (b) Mg(OH)₂/PP ($\phi_f = 0.31$) etched with 1N HCl for 1 h.

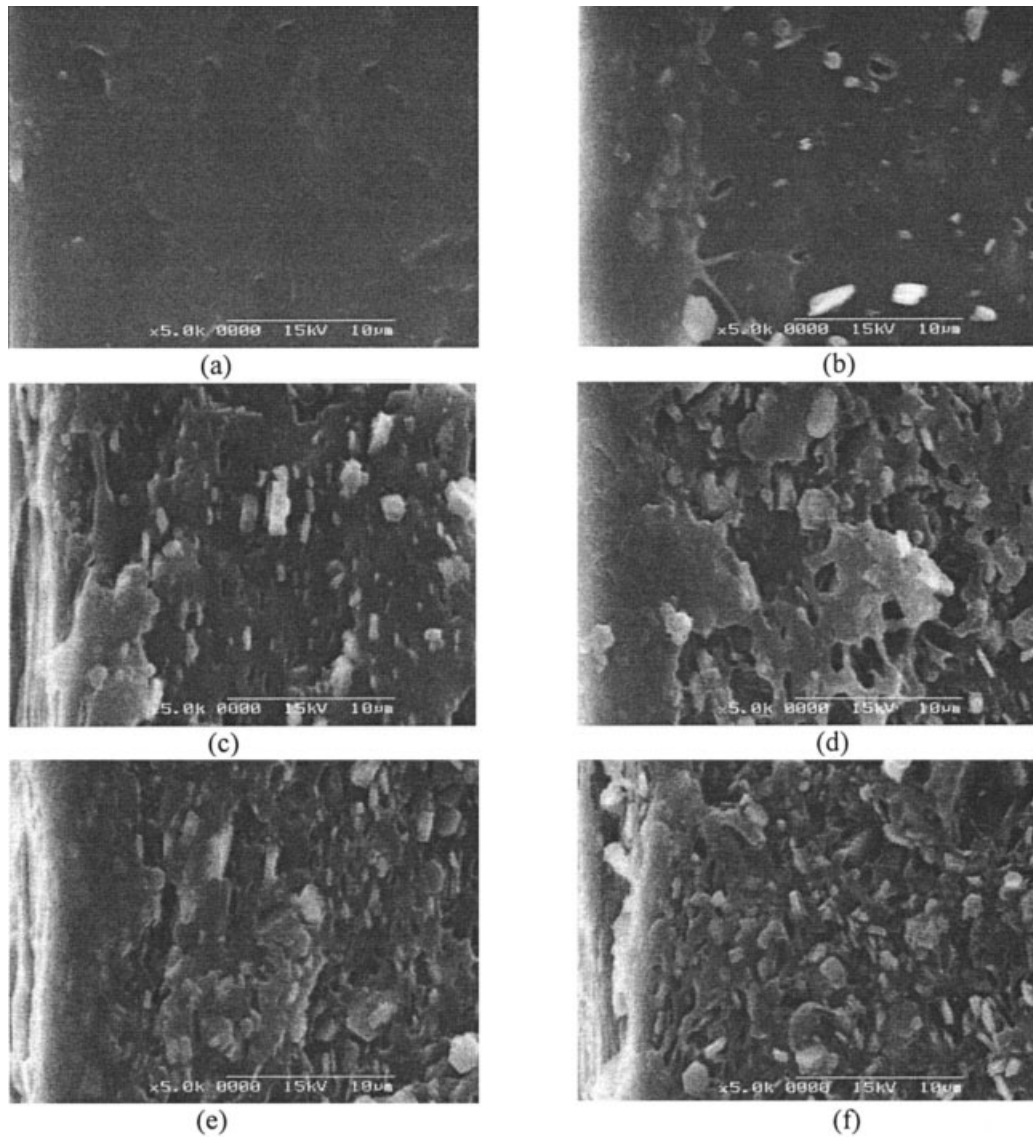


Figure 9 SEM micrographs of the impact fracture surfaces on the zone ahead of the notched crack tip of uncoated $\text{Mg}(\text{OH})_2/\text{PP}$ composites with various filler contents: (a) $\phi_f = 0$ (pure PP), (b) $\phi_f = 0.07$, (c) $\phi_f = 0.13$, (d) $\phi_f = 0.18$, (e) $\phi_f = 0.23$, and (f) $\phi_f = 0.31$ (the notched crack tip is on the left edge of the micrograph).

dispersed in all the composites, and few signs of an agglomeration structure have been observed in the fracture surface.

The observation of the impact fracture surface is focused on the zone close to the notched crack tip on the fracture surface. Figure 9 shows SEM micrographs of local impact fracture surfaces of the uncoated $\text{Mg}(\text{OH})_2/\text{PP}$ composites with various filler contents. There is a layer ahead of the crack tip that does not show any deformation such as debonding or shear yielding. The existence of a nondeformation layer suggests complete relaxation of the polymer matrix, which might be due to the adiabatic deformation and temperature rising under high-test-speed conditions.²⁵ No significant plastic deformation can be observed in the fracture surface of

pure PP [Fig. 9(a)], and this is due to the limited molecular chain mobility restrained by high stereoregularity. For the uncoated $\text{Mg}(\text{OH})_2/\text{PP}$ composites with low filler contents [Fig. 9(b,c)], the fracture surfaces are similar to that of pure PP, and the debonding of particles from the polymer matrix has little effect on the morphology of the composites. With the filler content increasing further, excessive debonding of filler particles from the matrix occurs in the PP matrix, and this is followed by the formation of microvoids. However, no obvious sign of shear yielding can be observed in the ligaments between microvoids even in the composites with the highest filler content [Fig. 9(f)], and the microvoids retain the same size as the filler particles. This might be attributed to the weak interfacial adhesion be-

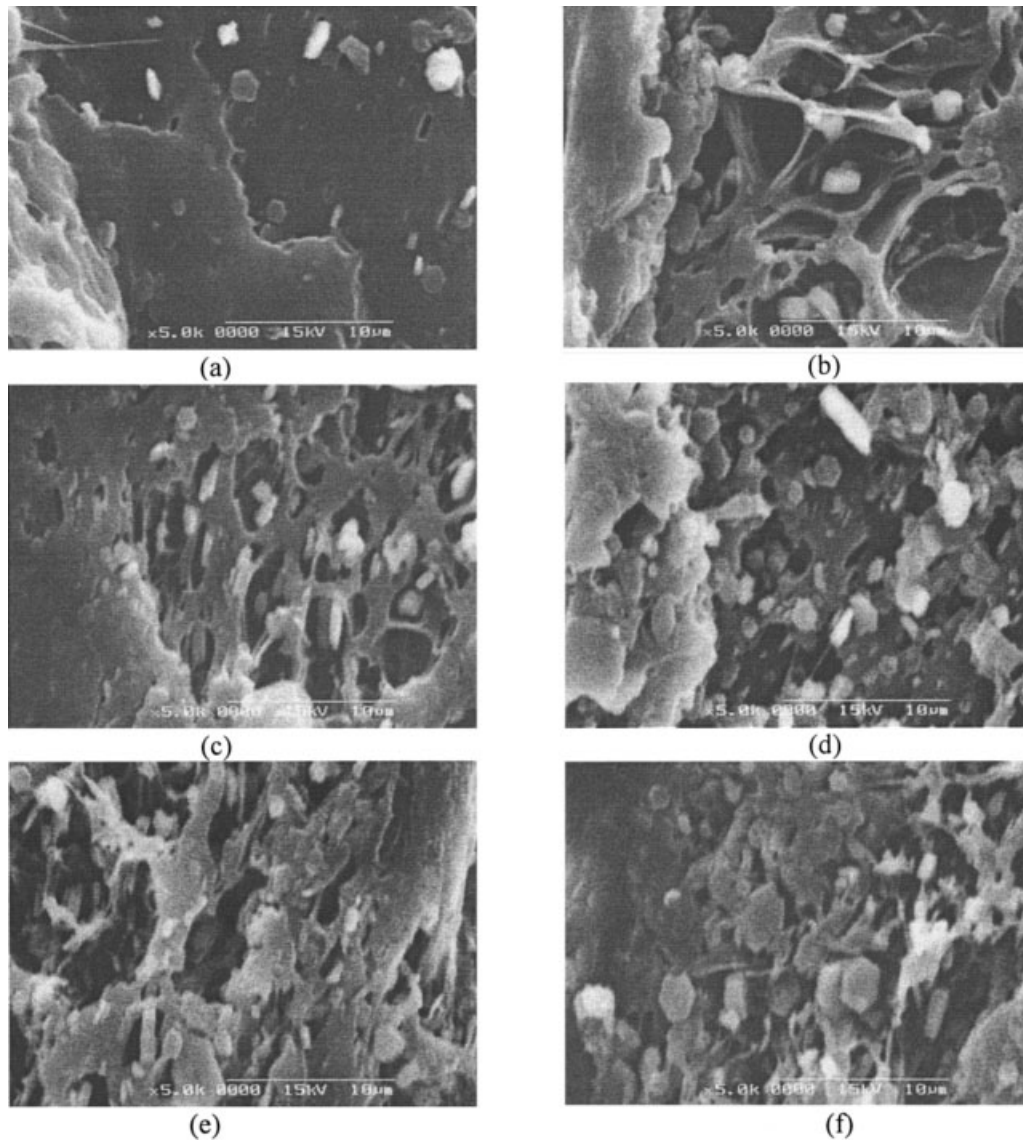


Figure 10 SEM micrographs of the impact fracture surfaces on the zone ahead of the notched crack tip of coated $\text{Mg}(\text{OH})_2/\text{PP}$ composites with various filler contents: (a) $\phi_f = 0.07$, (b) $\phi_f = 0.13$, (c) $\phi_f = 0.18$, (d) $\phi_f = 0.23$, (e) $\phi_f = 0.27$, and (f) $\phi_f = 0.31$ (the notched crack tip is on the left edge of the micrograph).

tween the filler and matrix or the poor chain mobility of the ligaments between microvoids.

SEM micrographs of local impact fracture surfaces of the coated $\text{Mg}(\text{OH})_2/\text{PP}$ composites with various filler contents are shown in Figure 10. The coated $\text{Mg}(\text{OH})_2/\text{PP}$ composites with low filler contents [Fig. 10(a)] show a similar fracture morphology with respect to the uncoated $\text{Mg}(\text{OH})_2/\text{PP}$ composites. Once the filler content reaches a certain level, such as $\phi_f = 0.13$ [Fig. 10(b)], significant plastic deformation occurs on the ligaments between microvoids caused by the debonding of particles, and the size of the microvoids dramatically increases; this is quite different from the situation for the uncoated $\text{Mg}(\text{OH})_2/\text{PP}$ composites with the same filler content. The transition of the deformation mechanism from debonding/craze

damage to debonding/shear-yielding deformation is attributed to an improvement of the deformability of the ligaments surrounding filler particles caused by surface modification with SA on the filler particle surface. However, with the filler content continuously increasing, the initiation and propagation of voids become unstable, and the agglomeration of microvoids and fibril-like structures is observed on the fracture surfaces [Fig. 10(f)] and leads to fracture in a brittle manner.

The results are in accordance with the concept of critical interparticle distance and percolation theory by Wu.^{7,8} When the filler content is relatively small, the stress field around a single particle can be only slightly affected by other particles, and the fracture behavior of the composites is dominated by the prop-

erties of the polymer matrix. Thus, the $\text{Mg}(\text{OH})_2/\text{PP}$ composites with low filler contents deform in a brittle fracture manner because of the high stereoregularity of the matrix. However, when the particles are sufficiently close together, with the filler content increasing, the stress fields of the particles begin to interact with one another, and this is beneficial for the matrix yielding and transition of deformation mechanisms. The transition is thought to occur in the composites with filler contents of 0.07–0.13, and a large amount of plastic deformation takes place in the ligaments between the microvoids in the coated $\text{Mg}(\text{OH})_2/\text{PP}$ composites with the filler content beyond the critical value. The transition of the deformation mechanism from craze damage to shear-yielding deformation does not lead to a transition of the fracture manner from brittle to ductile. This is attributed to the excessive filler loading level and the weak interfacial adhesion between the filler and matrix. Moreover, the thickness of the ligaments also strongly influences the termination of microvoid propagation. From the perspective of energy, the development of microvoids mainly depends on energy dissipation during plastic deformation, which is directly related to volumetric dilatation in plastic deformation fields.²⁶ As shown in Figure 10(b), a sufficiently large ligament thickness can successfully prevent microvoids from developing into premature cracks before the complete yielding of the matrix. With increasing filler content, the thickness of the ligaments decreases, and this results in a reduction of the energy dissipated during plastic deformation. On the other hand, excessive development of microvoids leads to agglomeration of voids and fibrillation of ligaments, followed by crack propagation and premature fracture in a brittle manner. The results suggest weak resistance of the polymer matrix in preventing crack propagation and premature deformation, especially in the PP composites with high filler contents.

Deformation mechanism

Inorganic particles are undeformable because of their rigidity, which hinders the toughening effect with respect to elastomer particles. Therefore, inorganic particles can act only as stress concentrators in the polymer matrix and change the stress distribution under an external load. According to the three-stage mechanism of particular composites,²⁷ triaxial stress develops around the particles, and its maximum lies in polar regions of the particles because of the stress concentration and undeformability of inorganic particles. Because of the poor interfacial adhesion, debonding can easily take place in both polar regions of filler particles and leads to the formation of microvoids. With increasing strain, weak shear bands are formed in the matrix between microvoids. Then, the triaxial

tension can be locally released in the ligaments and turn into the state of uniaxial or biaxial stress,²⁸ and this results in the acceleration of the deformation of the ligaments.

No significant plastic deformation can be observed in the polymer matrix between microvoids caused by the debonding of particles in the uncoated $\text{Mg}(\text{OH})_2/\text{PP}$ composites, as shown in Figure 9. Generally, in polymer composites filled with inorganic particles, the preferred absorption of large molecules into the filler particle surface results in the development of an interphase layer, with the structure and properties of the interphase drastically different from those of the matrix.^{29–32} The absorption of backbone bonds of polymer chains in the particle surface leads to considerable parallel orientation to the surface and thus the formation of transcrystallinity.^{33,34} Therefore, a rigid interphase with reduced mobility of adsorbed chains and higher rigidity is formed in the composites, and this leads to earlier crack initiation and crack propagation.^{35,36} The results observed for the uncoated $\text{Mg}(\text{OH})_2/\text{PP}$ composites can be attributed to lower deformability of the polymer chains surrounding the particles caused by the rigid interphase due to preferred absorption. The slight increase in the impact strength of the uncoated $\text{Mg}(\text{OH})_2/\text{PP}$ composites might be attributed to the additional consumption of impact energy caused by the formation of the new surfaces generated during the debonding process.

When particle surfaces are modified with SA, the plastic deformation of the ligaments surrounding the microvoids is remarkable, as shown in Figure 10. As discussed before (Fig. 4), the surface coating with SA results in a significant decrease in the surface free energy of the particles and a reduction in the particle–particle and particle–matrix interactions, which will have considerable influence on the debonding behavior and deformability of the composites. The reduction of the particle surface energy weakens the absorption of the polymer chains surrounding the particle surface and has a positive effect on the chain mobility. The increase in the chain mobility is confirmed by the improvement in the stress transferability, which is based on the model calculation in Figure 5. On account of the improvement of the deformability in the adsorbed polymer chains, the restrained strain can be easily released surrounding microvoids because of the stress state transition from triaxial to uniaxial or biaxial. With increasing strain, the size of the microvoids increases, and this is accompanied by the formation of shear bands in the ligaments and a large amount of plastic deformation. This will result in a considerable improvement in the impact strength.

CONCLUSIONS

The incorporation of $\text{Mg}(\text{OH})_2$ into the PP matrix leads to an increase in Young's modulus and de-

creases in the tensile yield strength and elongation at break of the $\text{Mg}(\text{OH})_2/\text{PP}$ composites. The surface treatment of the particle surface with SA leads to considerable increases in the elongation at break and impact strength of the composites. An analysis of the tensile yield strength has indicated that the surface modification of the filler particle with SA can reduce the interfacial adhesion between the filler and matrix, and this is accompanied by an improvement in the stress transferability.

The deformation behavior of the $\text{Mg}(\text{OH})_2/\text{PP}$ composites depends on the interfacial adhesion between the filler and matrix and the deformability of the matrix ligaments between the filler particles. For the uncoated $\text{Mg}(\text{OH})_2/\text{PP}$ composites, no sign of plastic deformation has been observed in the ligaments between microvoids caused by the debonding of the filler particles. The slight increase of the impact strength might be attributed to a new surface generated during the debonding process. For the coated $\text{Mg}(\text{OH})_2/\text{PP}$ composites, the dependence of the deformation behavior on the filler content is in accordance with percolation theory. The considerable plastic deformation occurring in the fracture surface at a certain filler content reveals the improvement in the deformability of the ligaments by surface modification of the particle surface. However, the transition of the deformation mechanism from debonding/craze damage to debonding/shear-yielding deformation does not lead to a significant improvement in the impact strength. The agglomeration of the microvoids and the fibrillation of the ligaments observed in the coated $\text{Mg}(\text{OH})_2/\text{PP}$ composites with high filler contents indicate the weak resistance of the polymer matrix to crack propagation and premature fracture in a brittle manner before the complete yielding of the PP matrix.

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