Mechanical properties of polymer grafted nanoparticle composites

Marissa Giovino, Julia Pribyl, Brian Benicewicz, Ronald Bucinelli and Linda Schadler

Materials Engineering, Rensselaer Polytechnic Institute, Troy, NY, USA; Chemistry and Biochemistry, University of South Carolina, Columbia, SC, USA; Mechanical Engineering, Union College, Schenectady, NY, USA; Mechanical Engineering, University of Vermont, Burlington, VT, USA

ABSTRACT

Polymer nanocomposites have improved mechanical, optical and thermal properties compared to traditional thermoplastics. To study the impact of grafted brush–matrix interactions on the mechanical properties of nanoparticle filled polymers, uniaxial tensile testing with digital image correlation (DIC) was done on polystyrene (PS) grafted SiO2 in low molecular weight (MW) \( N = P \), medium MW \( (2N = P) \) and high MW \( (N = 2P) \) matrices at different loadings. The low matrix MW composites had higher strength than high MW composites at high loading but lower strength than the pure matrix. The high matrix MW composites had higher toughness at low loadings. SEM images of fracture surfaces revealed particle debonding and plastic void growth in the high toughness samples. The results show that the \( P/N \) ratio of polymer grafted nanoparticle composites is important for controlling the mechanical properties of polymer nanocomposites.

ARTICLE HISTORY

Received 2 December 2018
Accepted 17 December 2018

KEYWORDS

Nanocomposite; toughening; mechanical properties; particle debonding; polymer matrix; plastic void growth; grafted brush; RAFT polymerization

1. Introduction

Polymer nanocomposites often exhibit improved mechanical properties,\(^1\)–\(^4\) optical properties,\(^5\)–\(^6\) thermal stability\(^7\)–\(^9\) and electrical properties\(^10\)–\(^12\) over their polymer counterparts. The filler size, distribution and dispersion can change each of those properties significantly.\(^13\)–\(^16\) It has been shown that for low filler loadings, composites with well dispersed fillers are most likely to exhibit enhanced properties.\(^17\) For example, in well dispersed 40 nm alumina filled PMMA, increases of an order of magnitude in fracture strain were observed.\(^18\) While for clustered composites with 40 nm \( \mathrm{Al}_2\mathrm{O}_3 \) and 90 nm \( \mathrm{Fe}_3\mathrm{O}_4 \) in PMMA and PS,\(^19\) the modulus of the composites was slightly lower than pure polymer in all composites measured, which is opposite to what is predicted by filler reinforcement theories.\(^20\),\(^21\)

Short molecule modification can lead to improved dispersion, and there are many examples of well dispersed nanoparticles increasing the modulus.\(^20\),\(^21\) It is also typical that there is an optimal loading to achieve the highest toughness or...
impact energy. For example, in 16 nm diameter chlorosilane modified particles in polystyrene, the modulus increased with SiO₂ loading. The highest toughness, impact energy and fracture strength were found at 0.75 wt% SiO₂ and SEM images revealed some evidence of particle debonding.

To further improve dispersion, grafted polymer chains have been used. A bimodal population of brushes (one group is densely grafted short brush and the other is sparsely grafted long brush) have been found to be optimal for controlling dispersion. At low loadings, bimodal samples have the largest modulus in both compression and tension compared to monomodal brush modified particles. In fact the bimodal samples have stiffness even higher than predicted by the Guth–Gold or Halpin–Tsai predictions.

The interactions between grafted brush and matrix polymer are also important in determining properties. In samples without any matrix added, the toughest samples had the highest brush molecular weight (and the most entanglements). Not surprisingly, in samples with matrix added, the largest strain to failure was also for samples with the greatest degree of entanglement. These experimental observations are supported by computational work. Self-consistent field theory (SCFT) has shown that when the matrix and particles are in the wetting regime, the matrix chains penetrate the grafted corona. To quantify wettability, the interfacial tension of matrix and grafted brush, γ_{b/h} was used. It was found that γ_{b/h} > 0 for all P (degree of polymerization of matrix) > N (degree of polymerization of grafted chain) ratios, while for P < N, γ_{b/h} was too small to cause film dewetting. To further understand that interaction, the penetration width, w, of the grafted chain has also been determined computationally. Matrix penetration into the brush is largest for brushes with a graft density in the semi-dilute regime. And as P/N increases, w decreases. Further, if the brush-matrix interactions are repulsive, interaction decreases and leads to a lower yield strength than pure matrix. For attractive filler-matrix interactions, computation predicts that the yield strength will be the same as the matrix. For grafted chains, larger than 5 Mₐ, the craze failure mechanism is predicted to be bond breaking, whereas for matrix chains which have lower tension (since both ends aren’t fixed) the transition to bond breaking occurs at 20 Mₐ. This experimental and computational work suggest that the fracture behavior of polymer nanocomposites depends on the brush–matrix interactions which are controlled by the molecular weight and graft density of the brush.

The challenge experimentally has been that it is difficult to separately alter the matrix-brush interaction for a fixed dispersion. Bimodal grafted brushes allow independent modification of the interaction while maintaining good dispersion. Thus, we have developed a materials system of bimodal polystyrene grafted SiO₂ nanoparticles dispersed in polystyrene. The two sets of composites have P/N ratios of 1 and 2. We find that at specific P/N ratios and loadings, the mechanical properties are altered resulting in significantly increased strain to failure.

2. Experimental

2.1. SiO₂ 8k 110k synthesis

Bimodal brush particles were synthesized using a sequential reversible addition-fragmentation chain transfer (RAFT) polymerization similar to a process described previously. To a solution of colloidal silica particles (Nissan Chemicals Inc., 30 wt% dispersion in MIBK, SiO₂ density = 2.2 g/mL) diluted with THF, added 3-aminopropylmethyloxysilane and a trace amount of octyldimethylmethoxysilane to maintain dispersibility through this synthetic step. This mixture was heated at 65 °C for 4 h under an inert (N₂) atmosphere. The surface-anchored amine groups were then reacted with 2-mercaptothiazoline-activated 4-cyanopentanoic acid dithiobenzoate (CPDB). The graft density of these covalently bound chain transfer agents was determined by comparing a UV–Vis spectrum of a grafted particle sample dispersed in THF to a calibration curve constructed from known amounts of free CPDB in solution. The surface polymerization of styrene monomer was performed at 65 °C. The polystyrene-grafted particles were precipitated in hexane and recovered by centrifugation. The chains from a small sample of polystyrene-grafted particles were cleaved using hydrofluoric acid (HF), and the chain length and dispersity were analyzed by gel-permeation chromatography (GPC). The remainder of the sample was redispersed in THF. A large excess of azobis(isobutyronitrile) (AIBN) was used to cleave the RAFT agent to prevent further chain growth.

The particles, which have been grafted with the short brush of polystyrene, were refunctionalized with CPDB as described above. The graft density of the second population of chain transfer agent was determined by normalizing the % silica in the sample based on TGA weight loss due to grafted polystyrene, and this corrected mass was correlated with the UV–Vis spectrum of the sample to determine the density of RAFT agents on the silica surface. The polymerization of this second population of polystyrene chains was conducted as described.
above. After the polystyrene-grafted particles were precipitated in hexane, a small sample was taken and the chains cleaved with HF. The molecular weight and dispersity of both populations of cleaved chains was analyzed by GPC. The bimodal polymer grafted particles were redispersed in THF, and the second population of RAFT agent was cleaved using a large excess of AIBN before further use.

The bimodal polystyrene grafted SiO2 is referred to as SiO2 8k 110k in this study. Two populations of polymer brushes were grafted to the surface: a short polystyrene brush with MW = 8 kg/mol and a graft density = 0.25 chains/nm², and long polystyrene brush with MW = 110 kg/mol and a graft density = 0.07 chains/nm².

2.2. SiO2–PS composites

SiO2 8k 110k in THF was solution mixed with homopolymer matrix (MW = 100 kg/mol or MW = 200 kg/mol, from Polymer Source). The samples were mixed with a probe sonicator (Sonics and Materials Vibrcell VCX 750 W unit) for 1 min at 40% amplitude using the pulsing setting of 2 s on and 0.5 s off. The samples were solution cast into aluminum boats at 90 °C. Composites were annealed for 2 days at 120 °C in a vacuum oven. Samples were hot pressed at 190 °C and slow cooled.

2.3. Transmission electron microscopy (TEM)

Composite samples were embedded in epoxy. Samples were cut into 60 nm sections using an RMC PowerTome XL microtome. The sections were floated onto water and transferred to copper grids. The sections were imaged using a JEOL 2011 TEM at an accelerating voltage of 200 kV. The images were converted to binary format using program ImageJ. For each composite 30 images were taken and binarized. To measure cluster size an algorithm developed by the Brinson and Chen groups at Northwestern University was used.31–33

2.4. Differential scanning calorimetry (DSC)

Composite samples were weighed into aluminum pans for determination of glass transition temperature ($T_g$). The sample and an empty reference pan were loaded into a DSC-Q100. The sample chamber was filled with N2 (g). Composites were heated at 10 °C/min to 160 °C and cooled to 50 °C at a rate of 10 °C/min. Three heating and cooling cycles were run. The glass transition temperature ($T_g$) was taken as the inflection point of the step.

2.5. Thermal gravimetric analysis (TGA)

For SiO2–PS composites, a TGA-Q50 was used to determine the filler loading. The sample was loaded into an alumina crucible. The sample chamber was filled with N2 (g). The sample was heated from 25 to 800 °C at a rate of 20 °C/min.

2.6. Tensile testing with digital image correlation (DIC)

Eight different samples were made: PS matrix 100k, PS matrix 200k, 0.7 vol.% SiO2 in 100k PS, 1.3 vol.% SiO2 in 100k PS, 5.5 vol.% SiO2 in 100k PS, 0.5 vol.% SiO2 in 200k PS, 1.3 vol.% SiO2 in 200k PS, and 5.5 vol.% SiO2 in 200k PS. For each sample 5 specimens were made. Samples were hot pressed at 190 °C using a Carver hot press into dog bone samples with dimensions specified by ASTM D638 type V.

The PS and SiO2–PS composites were prepared for 3D DIC measurements by applying a random speckle pattern.34 The pattern was applied in two steps: first a uniform white base coat (white enamel paint (Model Master FS37875) with some air brush thinner) was spray painted. Next a random black speckle pattern (Model Master FS37038) was applied. The samples and setup are shown in Figure 1. Samples were tested in uniaxial tension using an electro-mechanical load frame (load cell capacity 1 kN) at room temperature. A preload of 5...
N–7 N was applied to each sample. The samples were strained at a strain rate of 1 mm/min. The calibrated field of view was 15 mm width × 10 mm height × 5 mm depth of field. Images were captured using 2 CCD cameras (2448 × 2050 pixel resolution) with Schneider UNIFOC 2.8/50 lenses. The load data was taken from the load cell. The ARAMIS software by GOM was used to correlate the images and calculate strains. Virtual extensometers were placed on images to compute strain. A virtual extensometer approximately 7 mm in length with a precision of 0.001 mm was used to calculate axial strain.

2.7. Scanning electron microscopy (SEM) Fractography

The SiO$_2$–PS composites and PS matrices after tensile testing were preserved. The fracture surfaces were sputter coated with 60% Au/40% Pd using a Hummer V sputter coater. The metal layer was 0.5 nm thick. Samples were then imaged with FEI Versa 3D Dual Beam SEM at accelerating voltage of 3 kV.

3. Results and discussion

The materials system details are shown in Table 1. The polymer grafted nanoparticles in 100k PS are predicted to have matrix penetration into the grafted brush since $N$~$P$, however the grafted chains in 200k PS are expected to have more collapsed chains or a smaller amount of matrix penetration since $N$~$P$.\textsuperscript{28,29}

To evaluate the dispersion state of SiO$_2$–PS composites, TEM was used. The images were binarized and 30 images were taken per composite. An algorithm was applied to the binarized images to extract average cluster radius ($R_c$).\textsuperscript{31–33} Representative TEM images for the SiO$_2$–PS composites are shown in Figure 2. Each image clearly shows that the particles are singly dispersed and well distributed throughout the micrograph. To quantify the dispersion, $R_c$ for each sample was calculated. For each composite sample, $R_c=7$ nm as expected for well dispersed polymer nanocomposites with 14 nm diameter particles. Thus, any differences in properties cannot be attributed to the particle distribution.

The thermal state of polymer nanocomposites is also important when considering the mechanical properties. Specifically, for amorphous polymers the $T_g$ is very important. The $T_g$ values are shown in Table 2 below. The 100k PS composites and matrix $T_g$ values are within error of each other. The thermal state does not change with particle loading for 100k samples. For the 200k composites, the largest variation is less than 2 degrees. Thus, any significant difference in mechanical properties is not due to changes in $T_g$.

Figure 3a shows the modulus data vs filler volume fraction data (the core SiO$_2$ volume fraction).

| Material | Matrix MW (kg/mol) | SiO$_2$ loading | Surface modification |
|----------|------------------|----------------|---------------------|
| 100k PS  | 100              | 0.7 vol%       | PS: 8 kg/mol, 0.25 chains/nm$^2$ PS: 110 kg/mol, 0.07 chains/nm$^2$ |
| 100k_1.3 | 100              | 1.3 vol%       | PS: 8 kg/mol, 0.25 chains/nm$^2$ PS: 110 kg/mol, 0.07 chains/nm$^2$ |
| 100k_5.5 | 100              | 5.5 vol%       | PS: 8 kg/mol, 0.25 chains/nm$^2$ PS: 110 kg/mol, 0.07 chains/nm$^2$ |
| 200k PS  | 200              | –              | –                   |
| 200k_0.5 | 200              | 0.5 vol%       | PS: 8 kg/mol, 0.25 chains/nm$^2$ PS: 110 kg/mol, 0.07 chains/nm$^2$ |
| 200k_1.3 | 200              | 1.3 vol%       | PS: 8 kg/mol, 0.25 chains/nm$^2$ PS: 110 kg/mol, 0.07 chains/nm$^2$ |
| 200k_5.5 | 200              | 5.5 vol%       | PS: 8 kg/mol, 0.25 chains/nm$^2$ PS: 110 kg/mol, 0.07 chains/nm$^2$ |

Figure 2. TEM images of SiO$_2$–PS composites, all scale bars 200 nm. (a) Image of 100k_1.3, (b) Image of 100k_0.7, (c) Image of 200k_1.3, and (d) Image of 200k_0.5.
The prediction from filler reinforcement theory by Guth–Gold is shown for comparison. The 100k and 200k composites have very similar values for modulus at constant loading and follow the Guth–Gold prediction closely. There are a few deviations from Guth–Gold at very low and high loadings.

The strength data vs core SiO2 volume fraction is shown in Figure 3b. The strength of the 100k MW matrix composites does not change significantly with loading, while the 200k matrix composites show a decrease in strength at higher loading. The 100k_0.7 composite failed mostly outside the gauge section so the data point is from only 2 specimens. Clearly in these systems, the impact of the matrix–brush interaction is not large enough to change the maximum load significantly at low loadings. At higher loadings, when the brushes have the potential to overlap, the strength begins to decrease. A plot of the interparticle distance is shown in Figure 4. Here interparticle distance, \(d_{c-c}\), includes the core SiO\(_2\) particle and the grafted brush. The grafted brush height was calculated as radius of gyration of the long brush (\(h = 9.1\) nm). This has been shown to be a good approximation of brush height according to SAXS and SANS measurements.35 It is clear from Figure 4 that at a high particle loading the interparticle distance is very small. This small distance is likely the cause of the strength decrease.

The failure strain data as a function of SiO\(_2\) volume fraction is shown in Figure 3c. There is one surprising data point. For the 200k_0.5 composite there is a very large strain to failure. This is the same loading where the decrease in modulus was observed. 100k composites at the same loading tended to fail in the grips, but the few samples tested did not show the increase in strain to failure. The large failure strain and high strength suggest that the 200k_0.5 composite is tougher than all the other samples. Strain energy density, the area under the stress–strain curve, was used as a measure of toughness. The 200k_0.5 composites had twice the strain energy density compared to the matrix. One possible reason for this increased strain energy density is residual solvent. Residual solvent can act as a plasticizer and toughener.36 All samples were processed in the same way so it is highly unlikely that the 200k_0.5 composite has residual solvent. Furthermore, if there was residual solvent a decrease in \(T_g\) would be seen. In Table 2, the \(T_g\) of 200k_0.5 is similar to all other samples. To better understand this large strain energy density, the fracture surfaces of tensile testing samples were imaged with SEM.

The fracture surfaces of the 200k SiO\(_2\)--PS composites were imaged with SEM and are shown in Figure 5. Representative images for low and high magnification are shown for each sample. The low magnification images for each 200k sample are very similar. The high magnification images have significant differences. In Figure 5b the matrix high magnification image is shown for reference, the fracture surface is very smooth with minimal features. The high magnification images for the 200k_1.3 and 200k_5.5 composites are very similar to each other and slightly different than the matrix image (Figure 5f and h). The SEM images for 200k_1.3 and 200k_5.5 have some small pores which could be from particle debonding. The average pore size for Figure 5f is 46 nm diameter and for Figure 5h it is 41 nm diameter. These voids are roughly 3 times

![Figure 3.](image-url) Bulk mechanical properties data for polymer nanocomposites. Data points with asterisk only had 2 samples break in gauge length. (a) Modulus vs. particle loading for polymer nanocomposites, (b) Failure strength vs. particle loading for polymer composites, and (c) failure strain vs. particle loading for polymer nanocomposites.
larger than the SiO₂ diameter. The SEM images for 200k_0.5 have very large pores (Figure 5d), much larger than the particle size, 2R = 14 nm, likely there was particle debonding followed by plastic void growth. If we look more closely at the fracture surface of composite 200k_0.5 we note that there are two regions (Figure 6). There are more voids and larger voids in region 1 than region 2. Region 1 appears to have a rough fracture surface with a large amount of deformation. Region 1 is likely the slow crack growth region, providing more time for void growth, and region 2 is the fast crack growth region.

To further understand the large fracture strain observed for composite 200k_0.5 several toughening mechanisms were considered. Common toughening mechanisms for particle filled composites are: crack pinning, crack deflection, shear banding and plastic void growth.\(^\text{37-40}\) In crack pinning the particles act as obstacles that the crack will bow around. The crack bowing increases the crack length, which, increases toughness. Our obstacles are 14 nm SiO₂ nanoparticles. In most cases of crack pinning the obstacles were larger than the crack tip opening displacement:\(^\text{31-44}\)

\[
\delta_{tc} = \frac{K_{tc}^2}{E\sigma_y(1-\nu^2)} \quad (1)
\]

Figure 4. Interparticle distance vs particle loading. The particle is defined as core inorganic SiO₂ and grafted PS brush.

Figure 5. SEM images of fractured surfaces (a–b) Images of 200k PS, (c–d) Images of 200k_0.5, (e–f) Images of 200k_1.3, and (g–h) Images of 200k_5.5.
where $\delta_{tc}$ is crack opening displacement, $K_{1c}$ is fracture toughness, $E$ is Young’s modulus, $\sigma_y$ is yield stress and $\nu$ is Poisson’s ratio. Using the values in Table 3, $\delta_{tc} = 6.1 \, \mu m$, this is more than 2 orders of magnitude larger than our particle diameter. It is highly unlikely that the particles are crack pinning obstacles. Next, we consider crack deflection as a toughening mechanism. In crack deflection, the crack is tilted and twisted around the particles, which, causes an increase in the fracture surface area and crack growth occurs in mixed mode fracture (mode I and mode II).\textsuperscript{39,45} The particles are likely too small to cause this effect. According to Evans et al., the particles need to be larger than the plastic zone radius in order for this mechanism to apply.\textsuperscript{45,46} To calculate plastic zone radius $r_y$ the following equation was used:

$$r_y = \frac{1}{6\pi} \left( \frac{K_{1c}}{\sigma_y^2} \right)^{1/3}$$  \hspace{1cm} (2)

where $r_y = 24.5 \, \mu m$. The particle diameter is a factor of 1000 smaller than this indicating that crack deflection is not likely for these fillers. Shear banding is another common toughening mechanism for particle filled composites. There was no obvious necking of the sample or evidence of 45° bands in the DIC data or fracture surface. Suggesting there is no significant shear banding in these samples.

The final mechanism to consider is plastic void growth. This mechanism starts with particle debonding from the matrix and is followed by growth of voids. The void growth is the dominant energy dissipation mechanism. To model this the following equations were used:\textsuperscript{37}

$$\Delta G_{comp} = \Delta G_m + \Delta G_v$$  \hspace{1cm} (3)

$$\Delta G_v = \left(1 - \frac{U_m^3}{3V_f} \right) (V_e - V_f) \sigma_y r_y k_{vm}$$  \hspace{1cm} (4)

where $\Delta G_{comp}$ is the strain energy release rate of composite, $\Delta G_m$ is the strain energy release rate of the matrix, $\Delta G_v$ is the energy dissipated from plastic void growth, $\mu_m$ is von Mises pressure sensitivity, $V_e$ is void volume fraction, $V_f$ is filler volume fraction, $\sigma_y$ is compressive yield stress, and $K_{vm}$ is the von Mises stress concentration factor. The properties used in this calculation are shown in Table 3.\textsuperscript{38,47,48} Instead of strain energy release rate, we used the strain energy density, area under stress–strain curve, for composite ($U_c$) and matrix ($U_m$), $U_c/U_m = 2.6$. We assumed that this strain energy density ratio is equal to the strain energy release rate ratio: $\Delta G_{comp}/\Delta G_m$ ratio. A maximum $\Delta G_v$ was calculated by assuming all the particles debonded and knowing the average void diameter is 50 nm (Figure 6c). The calculation found $\Delta G_v/\Delta G_m > 1.6$ indicating 100% particle debonding is an overestimate. Partial particle debonding followed by plastic void growth is the best explanation for the high strain energy density seen in the 200k$_0.5$ composite.

4. Conclusions
The effect of grafted chain conformation on bulk mechanical properties of SiO$_2$–PS composites was

![Figure 6. SEM images of fracture surface for composite 200k$_0.5$. (a) Low magnification image of fracture surface with region 1 and 2 labeled, (b) high magnification image of region 2, and (c) high magnification image of region 1.](image-url)

Table 3. Properties used in evaluation of toughening mechanisms.

| Property          | Reference | Property          | References |
|-------------------|-----------|-------------------|------------|
| $E$               | 3.14 GPa  | $\mu_m$          | 0.25       |
| $\nu$             | 0.33      | $K_{vm}$         | 2.22       |
| $\sigma_y$        | 41.9 MPa  | $a_y$            | 70 MPa     |
| $K_{1c}$          | 0.9 MPa/\sqrt{m} | $r_y$          | 24.5 \, \mu m |

The – in reference column indicates it was measured experimentally.
tested using tensile testing with DIC. Two composite systems were used: one with $N \sim P$ and one with $N > P$. For $N \sim P$ the grafted chains are expected to be swollen and increase the composite strength, whereas when $N < P$ the grafted chains are expected to be collapsed and facilitate particle pull-out. The composites with $N \sim P$ had roughly constant strength with particle loading and even higher strength than $200k_5.5$. The $N < P$ composite had high fracture strength and strain at low loadings, which suggests higher toughness. Toughness estimated with strain energy density for composite 200k_0.5 was twice as large as 200k matrix. SEM images of fracture surfaces suggested that the large strain energy density was due to particle debonding followed by plastic void growth. The high magnification images showed two zones: 1 a slow crack growth zone containing a lot of deformation and large voids and 2 a fast crack growth zone with less deformation and smaller voids. The bulk mechanical properties of polymer nanocomposites were systematically studied, it was found that composite $P/N$ ratio can be tuned to increase strength or strain energy density.

Acknowledgments

MG acknowledges funding from the NSF under cooperative agreement EEC-0812056, New York State under NYSTAR contract C090145 and the materials engineering department at Rensselaer Polytechnic Institute.

Disclosure statement

No potential conflict of interest was reported by the authors.

Notes on contributors

Marissa Giovino is a materials engineer at the Air Force Research Lab in WPAFB, Ohio. She recently completed her PhD at Rensselaer Polytechnic Institute. Her thesis was focused on the mechanical and viscoelastic properties of polymer nanocomposites.

Julia Pribyl received her PhD from the University of South Carolina where she focused on synthesis of polymer-grafted nanomaterials and foams. She is now a post-doctoral associate at the University of Florida.

Brian C. Benicewicz joined the University of South Carolina, where he holds the SmartState Chair in Polymer Nanocomposites in the Department of Chemistry and Biochemistry. His research focuses on the development of high temperature membranes for fuel cells, reversible addition-fragmentation chain transfer (RAFT) polymerization, and the preparation of multifunctional nanoparticles and polymer nanocomposites.

Ronald B. Bucinell is an associate Professor of Mechanical Engineering at Union College, a Fellow of the American Society of Mechanical Engineers, and a licensed Professional Engineer in the state of New York. His areas of interest are experimental mechanics and composite materials.

Linda Schadler is the Dean of Engineering and Mathematical Sciences at the University of Vermont and professor in the Mechanical Engineering Department. Her research focuses on the dielectric, mechanical, and optical properties of polymer nanocomposites.

References

1. Shah D, Maiti P, Jiang DD, et al. Effect of nanoparticle mobility on toughness of polymer nanocomposites. Adv Mater. 2005;17:525–528.
2. Zhang H, Zhang Z, Friedrich K, et al. Property improvements of in situ epoxy nanocomposites with reduced interparticle distance at high nanosilica content. Acta Mater. 2006;54:1833–1842.
3. Wetzel B, Haupert F, Zhang MQ. Epoxy nanocomposites with high mechanical and tribological performance. Compos Sci Technol. 2003;63:2055–2067.
4. Zhao X, Zhang Q, Chen D, et al. Enhanced mechanical properties of graphene-based polyvinyl alcohol composites. Macromolecules. 2010;43:2357–2363.
5. Dang A, Ojha S, Hui CM, et al. High-transparency polymer nanocomposites enabled by polymer-graft modification of particle fillers. Langmuir. 2014;30:14434–14442.
6. Kim J, Yang H, Green PF. Tailoring the refractive indices of thin film polymer metallic nanoparticle nanocomposites. Langmuir. 2012;28:9735–9741.
7. Zhu J, Uhl FM, Morgan AB, et al. Studies on the mechanism by which the formation of nanocomposites enhances thermal stability. Chem Mater. 2001;13:4649–4654.
8. Wen X, Wang Y, Gong J, et al. Thermal and flammability properties of polypropylene/carbon black nanocomposites. Polym Degrad Stab. 2012;97:793–801.
9. Gilman J. Flammability and thermal stability studies of polymer layered-silicate (clay) nanocomposites. Appl Clay Sci. 1999;15:31–49.
10. Wang J, Dai J, Yarlagadda T. Carbon nanotube-conducting-polymer composite nanowires. Langmuir. 2005;21:9–12.
11. Geng Y, Liu MY, Li J, et al. Effects of surfactant treatment on mechanical and electrical properties of CNT/epoxy nanocomposites. Compos Part A: Appl Sci Manuf. 2008;39:1876–1883.
12. Gómez H, Ram MK, Alvi F, et al. Graphene-conducting polymer nanocomposite as novel electrode for supercapacitors. J Power Sources. 2011;196:4102–4108.
13. Chevigny C, Dalmas F, Di Cola E, et al. Polymer-grafted-nanoparticles nanocomposites: dispersion, grafted chain conformation, and rheological behavior. Macromolecules. 2011;44:122–133.
14. Akcora P, Kumar SK, Moll J, et al. Gel-like mechanical reinforcement in polymer nanocomposite melts. Macromolecules. 2010;43:1003–1010.
15. Moll IF, Akcora P, Rungra A, et al. Mechanical reinforcement in polymer melts filled with polymer grafted nanoparticles. Macromolecules. 2011;44:7473–7477.
