Studies on polypropylene composites filled with talc particles

Part I *Mechanical properties*

S. N. MAITI, K. K. SHARMA
Centre for Material Science and Technology, Indian Institute of Technology, New Delhi 110016, India

Tensile and impact properties of talc-filled isotactic polypropylene composites are investigated at 0–60 wt% filler contents. Tensile modulus registered an increase whereas tensile yield strength and strain-at-break decreased with increasing filler content. Mechanical restraint imposed by the talc particles on the molecular mobility or deformability of polypropylene explained the increase in modulus and decrease in strain-at-break while decrease in tensile yield strength was attributed to decreased crystallinity and formation of stress concentration points around the filler particles. Izod impact strength decreased with increased talc content. Surface modification of talc with a titanate coupling agent LICA 38 enhanced the filler–polymer interaction, further modifying the composite properties consequent upon significant decrease in the stress concentration. Scanning electron microscopic studies revealed better dispersion of surface-modified filler particles in the polymer matrix.

1. Introduction

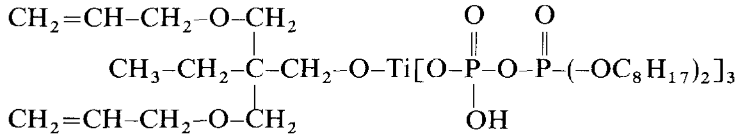
Various additives are often incorporated into polymeric matrices to modify the physical, mechanical, rheological as well as thermal properties in order to suit a wide range of applications [1–6]. The major constituents of these additives are inorganic particulate fillers [4, 7–10] such as talc, limestone, silicates, glass beads, ceramics, etc. The bulk properties of these two-phase composites are largely governed by the interphase [4, 11] between the dispersed phase and the continuous phase. The inclusion–polymer interaction can be enhanced to a varying extent by treating the filler surface with a suitable coupling agent which, in turn, imparts optimum properties [4, 5, 12]. In particular, titanate-based coupling agents have been reported [6, 12, 13] to modify the filler surface quantitatively, rendering the polymer composites easily processable due to improved filler dispersion in the polymer through enhanced wettability of the modified filler surface by the former.

Mechanical properties such as tensile modulus, tensile strength, breaking strain and Izod impact strength of isotactic polypropylene (i-PP)/talc composites with and without a titanate coupling agent are reported. Tensile properties are compared with theoretical predictions to assess the degree of discontinuity generated in terms of stress concentration factors. Izod impact strength data have been analysed on the basis of crystallinity of i-PP. The state of dispersion of talc in i-PP has been examined by scanning electron microscopic (SEM) studies.

2. Experimental procedure

2.1. Materials

The polymer used was isotactic polypropylene (i-PP), Koylene M0030 (MFI 10, density 0.89 g cm^{-3}). Commercial grade talc (magnesium silicate $MgSiO_3$) was obtained from the local market. The particle size and density of talc was < 0.5 µm and 2.69 g cm^{-3}, respectively. Surface treatment of talc was performed by a titanate coupling agent neopentyl (diallyl) oxy, tri(dioctyl) pyrophosphato titanate (LICA 38) of the following chemical formula [6]



$$\begin{array}{l} CH_2{=}CH{-}CH_2{-}O{-}CH_2 \\ \qquad CH_3{-}CH_2{-}C{-}CH_2{-}O{-}Ti[O{-}\underset{OH}{\overset{\overset{O}{\|}}{P}}{-}O{-}\overset{\overset{O}{\|}}{P}{-}({-}OC_8H_{17})_2]_3 \\ CH_2{=}CH{-}CH_2{-}O{-}CH_2 \end{array}$$

2.2. Surface treatment of talc

Talc was added with vigorous stirring to a moisture-free solution of LICA 38 (0.4 wt % on the basis of talc) in toluene. The slurry was stirred for 3 h using a magnetic stirrer. Toluene was expelled by heating the coated filler for 8 h at 393 K followed by vacuum drying at 383 K for 2 h. The hydrophobicity acquired by the fillers was confirmed by floating test on water. This technique of filler surface treatment was reported

0022-2461 © *1992 Chapman & Hall*

[14, 15] to give a uniform coating of the coupling agent.

2.3. Mixing and moulding

Thin sheets of the composites containing 0–34 vol % (0–60 wt %) talc were prepared by homogenizing vacuum-dried i-PP and talc for 10 min on a two-roll mill at 433 ± 2 K. Several plies of these sheets were compression moulded into sheets of thickness 1 and 3 mm at 513 ± 2 K and 10.35 MPa pressure for 5 min. The mould assembly was cooled under the same pressure at a cooling rate of 20 K min^{-1}. Dumb-bell shaped tensile specimens were machined from 1 mm thick sheets while notched bar Izod impact samples were machined from sheets 3 mm thick. Unfilled i-PP was also passed through identical mixing and moulding conditions to eliminate any degradation effect of prolonged heating of the polyolefin.

2.4. Measurements

Tensile properties were measured on an Instron universal testing instrument (Model 1121) at an extension rate 20% (initial crosshead separation 5 cm and crosshead speed 10 mm min^{-1}) following ASTM D638 test procedure. An FIE instrument (Model IT-0.42) was used to evaluate the Izod impact strength values of the notched samples according to ASTM D256 test procedure. At each composite composition at least five samples were tested and the average results reported. All tests were carried out at ambient temperature 303 ± 2 K.

Cryogenically fractured surfaces were scanned by a Cambridge Stereoscan (Model S4-10) to study the dispersion of talc in i-PP and the effect of surface treatment on the morphology of these two-phase composites. The fracture surfaces were sputter coated with gold prior to scanning.

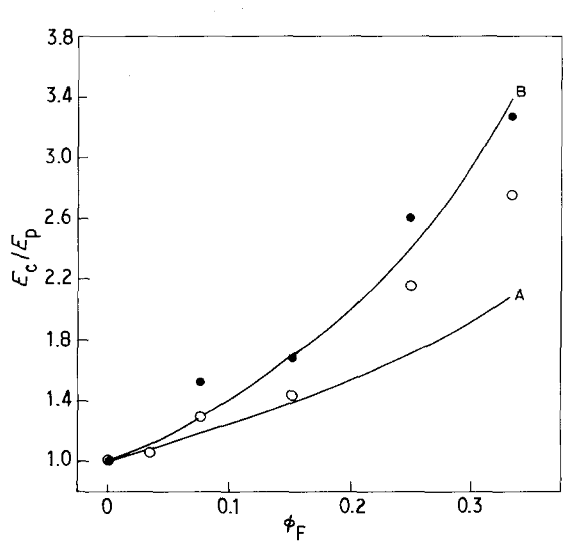


Figure 1 Plot of relative tensile modulus, E_c/E_f, of (○) i-PP/talc and (●) i-PP/talc/LICA 38 composites, as functions of talc volume fraction, ϕ_F. Curves A and B represent the Kerner and Guth–Smallwood equations, respectively.

3. Results and discussion

3.1. Tensile properties

Tensile properties of i-PP/talc composites were evaluated from the stress–strain curves (not shown) and are presented in Figs 1–8 as variations of the ratio of the property of the composite (subscript c) to that of the unfilled polymer (subscript p) versus volume fraction, ϕ_F, of talc.

The variation of relative tensile modulus E_c/E_p of the composites against ϕ_F, Fig. 1, shows that the modulus increases significantly with ϕ_F, the value at maximum ϕ_F (33.3 vol %) being ~ 2.8 times that of unfilled i-PP. Upon treatment of talc with LICA 38 the modulus increases further. The modulus data are compared with some of the predictive models for two-phase composites in Fig. 1. Curve A represents the Kerner equation [16, 17] (Equation 1), for filled polymer composites with Poisson ratio, ν_p, for i-PP taken as 0.35, while Curve B exhibits Guth–Smallwood equation [11, 18, 19] (Equation 2)

$$E_c/E_p = 1 + \left(\frac{15(1-\nu_p)}{8-10\nu_p}\right)\left(\frac{\phi_F}{1-\phi_F}\right) \quad (1)$$

$$E_c/E_p = 1 + 2.5\phi_F + 14.1\phi_F^2 \quad (2)$$

For composites with untreated talc, the modulus

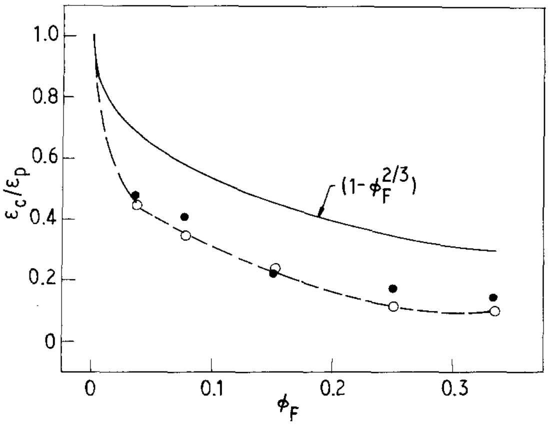


Figure 2 Variation of relative breaking strain, $\varepsilon_c/\varepsilon_p$, of (○) i-PP/talc and (●) i-PP/talc/LICA 38 composites versus ϕ_F. (---) Average of experimental data, (——) predicted behaviour according to Equation 3.

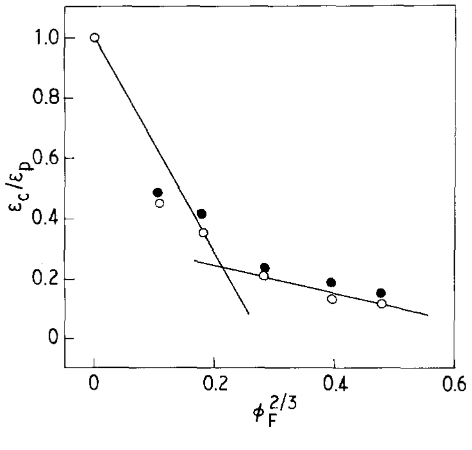


Figure 3 Variation of $\varepsilon_c/\varepsilon_p$ of (○) i-PP/talc and (●) i-PP/talc/LICA 38 composites against $\phi_F^{2/3}$.

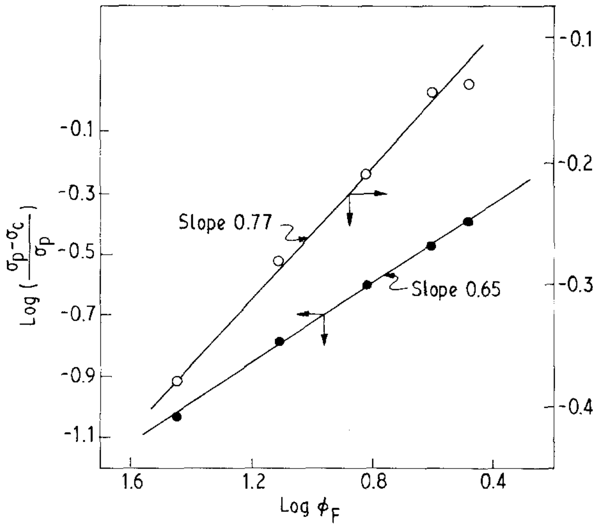


Figure 4 Plot of $\log[(\sigma_p - \sigma_c)/\sigma_p]$, of (○) i-PP/talc and (●) i-PP/talc/LICA 38 composites versus log ϕ_F.

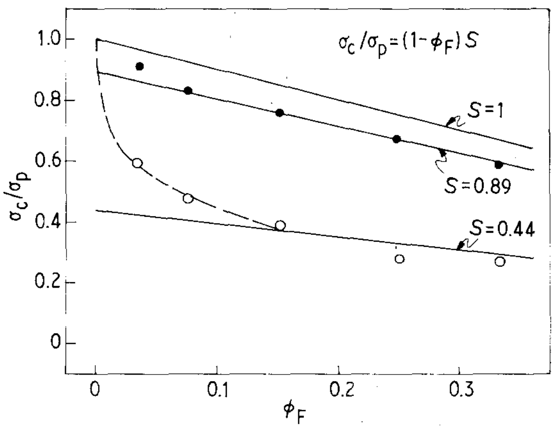


Figure 5 Relative tensile stress, σ_c/σ_p, of (○) i-PP/talc and (●) i-PP/talc/LICA 38 composites versus ϕ_F. (---) Average of data, (——) predicted behaviour according to Equation 5 with S values indicated.

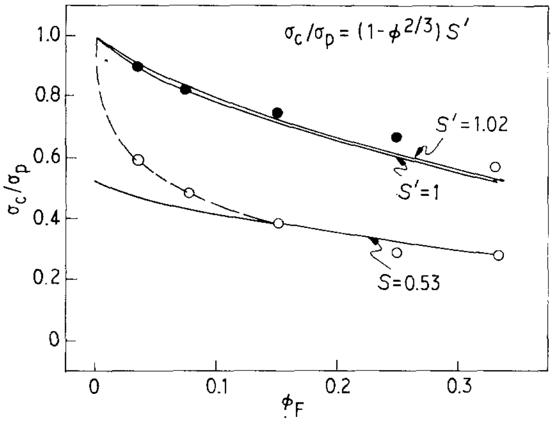


Figure 6 Variation of σ_c/σ_p of (○) i-PP/talc and (●) i-PP/talc/LICA 38 composites versus ϕ_F. (---) Average of data, (——) Equation 6 with S' values indicated.

data showed a good fit with Kerner model up to ~ 15 vol% talc; with further increase in ϕ_F the data increase more towards Curve B. For composites with surface-treated talc the modulus data showed a closer fit with Curve B. Thus simple predictive equations can be used to estimate the Young's modulus of talc-filled i-PP systems similar to other works [20].

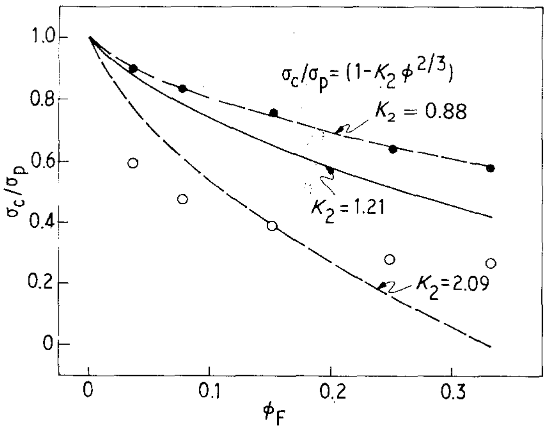


Figure 7 Plot of σ_c/σ_p of (○) i-PP/talc and (●) i-PP/talc/LICA 38 composites against ϕ_F. (——) Predicted behaviour according to Equation 7 with $K_2 = 1.21$, (---) the same with K_2 values indicated.

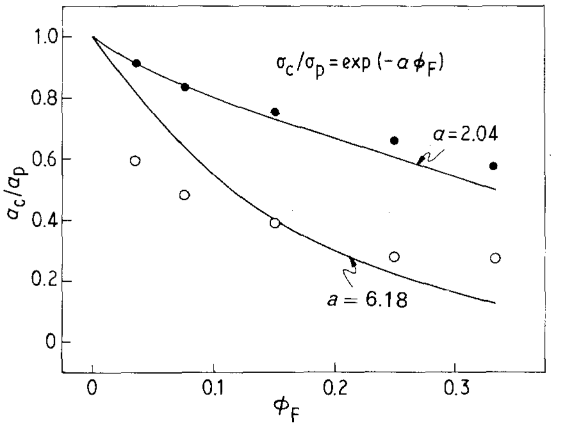


Figure 8 Variation of σ_c/σ_p of (○) i-PP/talc and (●) i-PP/talc/LICA 38 composites versus ϕ_F. The curves are predicted behaviours according to Equation 8 with a values indicated.

The increase in modulus in talc-filled composites indicates an increase in rigidity of i-PP through imposition of restrictions in the mobility and deformability of i-PP molecules by introduction of a mechanical restraint [11] by the filler. It has been pointed out that the degree of this mechanical restraint is dependent on the particulate spacing and the properties of the particle and the matrix. Although chemical-type adhesive bond formation between talc (untreated) and i-PP seems a rare possibility, this mechanical restraint may be a result of the enhanced surface interaction between the two phases in view of the large surface area of the dispersed phase talc (< 0.5 μm particle size). Other works [21–23] also reported modulus increase in polymers in the presence of rigid particulate fillers despite the absence of strong interaction with the matrix. Hindrance in polymer diffusion in the presence of solid particles was shown by Baumgarter and Muthukumar [24].

Upon surface treatment of talc the modulus increases further above the values of i-PP/talc (untreated) systems, in particular at $\phi_F > 3.6\%$. This

TABLE I Values of per cent crystallinity of i-PP component in i-PP/talc and i-PP/talc/LICA 38 composites estimated from DSC studies

Sample composition (vol % filler)	Crystallinity (%) DSC method[a] (ΔH, J g^{-1})	
	i-PP/talc	i-PP/talc/LICA 38
0	70.0	70.0
3.57	62.2	57.3
7.69	55.0	50.1
15.21	45.1	43.2
24.99	36.5	32.6
33.32	24.3	23.2

[a] Heat of crystallization, ΔH, is proportional to crystallinity (%) [25]. Data were taken from Reference 26.

TABLE II Values of polymer–filler interaction parameter, K_1, (Equation 4) in i-PP/talc and i-PP/talc/LICA 38 composites

Talc (vol %)	K_1	
	i-PP/talc	i-PP/talc/LICA 38
0	–	–
3.57	5.074	4.767
7.69	3.598	3.228
15.21	2.785	2.692
24.99	2.185	2.059
33.32	1.839	1.769
Mean value[a]	2.601	2.437

[a] Owing to data variation, the mean values were estimated excluding the value at $\phi_F = 3.57\%$.

implies enhanced interaction of the filler and the matrix possibly through chemical bonding [6] in addition to the physical interaction. Although crystallinity of i-PP showed a decrease with increase in ϕ_F, Table I, this increase in interaction more than compensates for the decrease in crystallinity, contrary to other reports [27].

Strain-at-break ($\varepsilon_c/\varepsilon_p$) showed a decrease with increase in ϕ_F, Fig. 2, the decrease being quite significant at low filler levels, e.g. 5%, and slow with further increase in ϕ_F. With surface treatment of talc this declining trend is maintained, the values are, however, marginally higher than the previous composites. Nielsen's predictive model [11, 17] (Equation 3) for strain assuming perfect adhesion between the filler and the matrix

$$\varepsilon_c/\varepsilon_p = 1 - \phi_F^{1/3} \quad (3)$$

exhibit higher values than both types of composite. The decrease in breaking elongation of the polymer by the addition of talc implies an interference by the latter with the mobility/deformability of the former through imposition of mechanical restraint by way of physical interaction, as stated earlier. In the surface-treated talc-filled composites, the coupling agent probably functions as a plasticizer/lubricating agent [12, 28] also which accounts for the marginally lesser decrease of $\varepsilon_c/\varepsilon_p$.

Breaking strain decrease to lower values than the Nielsen model (Equation 3) indicates a dependence of the parameter on area fraction $\phi_F^{2/3}$ of the filler. A plot of $\varepsilon_c/\varepsilon_p$ against $\phi_F^{2/3}$ shows that the elongation is quite sensitive to filler area fraction, particularly up to $\phi_F^{2/3} = 0.21$, the sensitivity is lowered at higher $\phi_F^{2/3}$ values, Fig. 3. Filler surface modification does not alter this general trend of sensitivity of the strain. Equation 4 was derived following the method introduced by Mitsuishi *et al.* [29] to estimate the effect of filler–polymer interaction on elongation

$$\varepsilon_c/\varepsilon_p = (1 - K_1\phi_F^{2/3}) \quad (4)$$

where the interaction parameter, K_1, is a constant value which depends on filler size and the modification of fillers. Values of K_1 obtained from Equation 4 are given in Table II. The mean value of K_1 determined for untreated talc-filled composites was 2.60, which was surprisingly very close to a reported value [29] for i-PP/CaCO$_3$ systems. The interaction parameter decreases marginally for the treated talc-filled composites revealing a plasticizing/lubricating effect [12, 28] of the coupling agent. Therefore, the observed effect of filler content and surface-modified filler on breaking elongation of i-PP/talc composites can be explained similar to other works [27, 29].

In particulate-filled polymer composites [11, 17] introduction of discontinuity at the filler–polymer interface is very often a problem which gives rise to poor stress transfer creating, in turn, a weak structure. The composite structure is further weakened due to breakdown of filler agglomeration and dewetting of the polymer at the interface which create stress concentration centres. On the contrary, if continuity in the structure is maintained and interfacial adhesion is promoted, the composite may show enhanced physical/mechanical properties. Analysis of weakness, if any, in the structure is attempted in the subsequent section.

3.2. Analysis of tensile strength data

Weakness in the composite structure introduced by the inclusions through creation of a discontinuity of stress transfer (i.e. stress concentration points) was attempted to be estimated by comparing the tensile strength data with some theoretical predictive models (Equations 5–8) described elsewhere [13, 30]. In these equations, σ_c and σ_p

$$\sigma_c/\sigma_p = (1 - \phi_F)S \quad (5)$$

$$\sigma_c/\sigma_p = (1 - \phi_F^{2/3})S' \quad (6)$$

$$\sigma_c/\sigma_p = (1 - K_2\phi_F^{2/3}) \quad (7)$$

$$\sigma_c/\sigma_p = \exp(-a\phi_F) \quad (8)$$

represent the tensile strength of the composite and the polymer, respectively. Equations 5–7 are power-law models based on the relationship of area fraction and volume fraction of the dispersed phase [31, 32] and the composite structure formed is stated to be no adhesion type. The parameter S in the first power-law model (Equation 5) describes weakness in the structure created through stress concentration at the

TABLE III Values of stress concentration parameters S (Equation 5), S' (Equation 6), a (Equation 8), a' (Equation 9), and adhesion parameter, K_2 (Equation 7) in i-PP/talc and i-PP/talc/LICA 38 composites

Talc (vol %)	S	S'	K_2	a	a'
0	–	–	–	–	–
3.57	0.607	0.656	3.827	15.011	4.943
	(0.863)	(1.016)	(0.866)	(2.767)	(0.911)
7.69	0.520	0.587	2.873	9.525	4.054
	(0.826)	(1.018)	(0.918)	(2.294)	(0.976)
15.21	0.454	0.539	2.159	6.272	3.350
	(0.759)	(1.049)	(0.875)	(1.865)	(1.006)
24.99	0.379	0.472	1.802	5.026	3.166
	(0.671)	(1.104)	(0.841)	(1.623)	(1.023)
33.32	0.405	0.520	1.519	3.926	2.723
	(0.597)	(1.100)	(0.891)	(1.672)	(1.163)
Mean value[a]	0.439	0.529	2.088	6.187	3.323
	(0.895)	(1.027)	(0.878)	(2.048)	(1.016)

[a] Owing to data scatter, mean values were estimated excluding some data points: for S, S', K_2, a and a' at $\phi_F = 3.57\%$. Values in parentheses are for i-PP/talc/LICA 38 composites. Excluded data points for the mean values are: for S' at $\phi_F = 24.99\%$ and 33.32%.

inclusion–matrix interphase, analogous to S' Nielsen's model [33] (Equation 6). Unity in the value of S (or S') means a "no stress concentration effect", whereas the lower the value the "greater the stress concentration effect" (or poorer the adhesion). In Equation 7, the adhesion between the polymer and the dispersed phase [34] is described by the so called weightage factor, K_2, the value of which is determined by the details of the model [20]. A value of $K_2 = 1.1$ describes dense hexagonal packing in the plane of highest density; $K_2 = 1.21$ represents the extreme case of poor adhesion [31] with spherical fillers for the minimum cross-section between spherical particles [35]; and $K_2 = 1$ stands for strain consideration [33]. In general, the lower the value of K_2 than 1.21 the better the adhesion [36, 37] and $K_2 = 0$ represents the upper limit with the unfilled polymer. The porosity model (Equation 8) considers the inclusions to be analogous to pores/voids in nonpolymeric materials such as metals and ceramics [38] and in polymer blends [31] and composites [39]. The pores are assumed not to have any influence on the mechanical properties of the composites due to non-adhesion at the interphase boundary. Stress concentration was shown to be described [31] by the parameter a; the higher the value of a the higher the stress concentration.

A plot of variation of log $[(\sigma_p - \sigma_c)/\sigma_p]$ versus log ϕ_F is shown in Fig. 4 in order to assess the applicability of first or two-thirds power-law to i-PP/talc system. From the slope of the plot the power-law exponent can be estimated using Equation 5 or 6 when the value of S (or S') is unity. For both types of composite the slope was closer to the two-thirds power-law than the first power-law, the values being 0.77 and 0.65, respectively. Better applicability of first power-law than the fractional power-law was shown in the analysis of tensile strength [40] and yield stress [41, 42] of composites. Other works on filled polymer composites [11, 37] showed the applicability of the two-thirds power-law. The present analysis, however, will be further modified on inclusion of the stress concentration and adhesion parameters (Equations 5–8).

Table III presents the values of S, S', K_2 and a at each individual talc concentration obtained from the experimental tensile stress data and the first and fractional power-law models, the Nicolais and Narkis model and the porosity model (Equations 5–8). Stress concentration values are seen to be less than unity according to the first and two-thirds power-law models, and greater than unity according to the Nicolais–Narkis model and porosity model, both types indicating significant weakness or discontinuity in the composite structure. Interestingly the values of the above parameters were modified to very close to unity upon surface treatment of talc, clearly indicating enhanced interaction of the two phases and drastically reducing the stress concentrations in the process.

The tensile stress data were compared with the predictive models in Figs 5–8. In the first power-law relation (Equation 5) the data with untreated talc up to $\phi_F = 7.7\%$ lie within two theoretical lines corresponding to $S = 1$ and 0.44 and beyond this ϕ_F, the data showed a reasonably good fit with the curve with $S = 0.44$, Fig. 5. The indication is that from the state of no stress concentration the composite system quickly changes to a state of high degree of stress concentration upon addition of talc. Upon surface treatment of talc the tensile stress values showed significant enhancement at all filler levels and the data showed very good agreement with the curve with $S = 0.89$ implying a low degree of stress concentration.

The Nielsen model (Equation 6) also exhibited the occurrence of significant stress concentration in the untreated talc-filled composites, Fig. 6, where the data at $\phi_F > 7.7\%$ show a good fit with the curve with $S' = 0.53$. According to this model, stress concentration decreased to almost zero in the surface-treated talc-filled composites, the value of S' being 1.02.

The Nicolais and Narkis model (Equation 7) shows higher values than the data with untreated filler, Fig. 7, indicating the absence of adhesion in these composites. The data lie scattered around a curve with the adhesion parameter $K_2 = 2.09$ which is higher than the value $K_2 = 1.21$ for no adhesion type composites with spherical fillers [31]. This data deviation may be attributed to the difference in shape of the plate-like talc particles from spherical, as will be shown later (Fig. 11). However, the absence of adhesion between talc and i-PP is quite possible in view of the lack of any chemical interaction between the two phases similar to other composite system [27, 43]. Surface treatment brought about a better degree of adhesion with a K_2 value of 0.88, analogous to other works [36, 37].

According to the porosity model (Equation 8) the data in i-PP/talc (untreated) composites did not show good agreement with the average a value of 6.18. However, the individual data points do exhibit the significant extent of stress concentration which is reduced appreciably on surface treatment of talc ($a = 2.04$), Table III, Fig. 8. A better fit of the data for both types of composite was noted when filler area fraction was used in place of filler volume fraction, according to Equation 9

$$\sigma_c/\sigma_p = \exp(-a'\phi_F^{2/3}) \tag{9}$$

the average a' values being 3.32 and 1.01, respectively, Fig. 9, Table III. Thus, in this model filler area fraction explains the data better than the filler volume fraction. The general trend of the stress concentration, however, does not change.

The decrease in stress concentration (or increase in adhesion) in the i-PP/talc (treated) composites may be attributed to increased filler–polymer interaction, possibly through chemical bond formation [6]. This observation stands at variance with the results in i-PP/$CaCO_3$ composites on surface treatment where adhesion was reduced due to a predominating lubricating/plasticizing role of the coupling agent [27].

It may be noted that although i-PP is a polyolefin with very low degree of interchain attraction forces, the mechanical strength of the polymer is derived from its crystallinity [5]. Hindrance to fitting of the chains into a crystal structure by the inclusion of interfering external agents such as fillers and additives may eventually decrease the strength properties unless it is compensated for by some other means, e.g. enhanced polymer–additive interaction. The role of filler–additive interphase thus is very important in these composite systems [11].

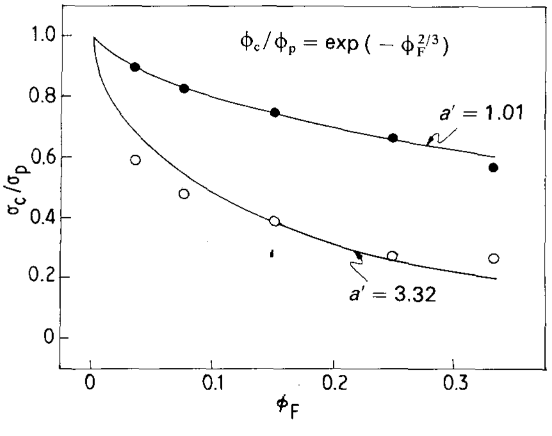


Figure 9 Plot of σ_c/σ_p of (○) i-PP/talc and (●) i-PP/talc/LICA 38 composites against ϕ_F. The curves are theoretical predictions according to Equation 9 with a' values indicated.

In the i-PP/talc (untreated) composite systems, tensile strength and strain-at-break decreased and tensile modulus increased with increasing filler concentrations. Although with very fine talc particles (diameter < 0.5 µm) some degree of surface interaction was possible, the presence of filler particles brought about a decrease in crystallinity, Table I. While the former effect tends to increase the strength properties, the latter would have the opposite effect so that the resultant of these two opposing factors will determine the strength properties in these systems. In the measurements, enhancement was exhibited by the modulus, a property estimated in general at low solid displacement which may not be sensitive to micro flaws, if any, in the structure. On the contrary, tensile strength is determined at relatively high deformations where any weakness may be magnified reducing in the process the stress transfer. Thus the resultant effect in these composites was a decrease in tensile strength despite a degree of physical interaction of i-PP and talc. Formation of an increased extent of amorphous polymer in the presence of talc (Table I) would also function in reducing the tensile strength. The decrease in the breaking strain of i-PP may be attributed to the immobilization of the polymer chains due to the mechanical restraint imposed by the filler particles. A similar observation in tensile properties has been reported [27, 44].

Enhanced interaction between filler and polymer takes place upon surface treatment of the filler with LICA 38 so that modulus values increased further. This interaction more than compensates for the decrease in crystallinity of i-PP, Table I. This is reflected in the tensile strength also, so that although the declining trend is maintained the values are much higher than in the untreated talc-filled composites. The strain values maintained the declining trend; the values are, however, marginally higher. Although the lubricating/plasticizing [12, 28] effect of the coupling agent plays some part, the filler–polymer interaction takes predominance in these composites, possibly through chemical bonding [6], as stated earlier.

3.3. Impact strength

Variation of relative Izod impact strength (I_c/I_p) values with ϕ_F is shown in Fig. 10. The strength decreases with filler content, the decrease beyond $\phi_F = 10\%$ being only marginal so that at the maximum ϕ_F studied (33.3%) the value is 0.6 times that of unfilled i-PP. The decrease in impact strength may be attributed to the immobilization of the matrix by the filler so that the latter fails to deform before failure. In addition, lack of stress transfer at the filler–polymer interphase would also aid the impact failure. This observation is contrary to other reports [27, 45] where an initial increase in the strength was reported. The relative impact strength was modified to a great extent upon surface treatment of the filler. The strength almost remains unaltered up to 7.5% filler, then shows only a small decrease up to 15% filler content; the value, however, decrease rapidly beyond this ϕ_F. This may be explained on the basis of enhanced stress transfer consequent to the increased extent of interaction between the filler and the polymer.

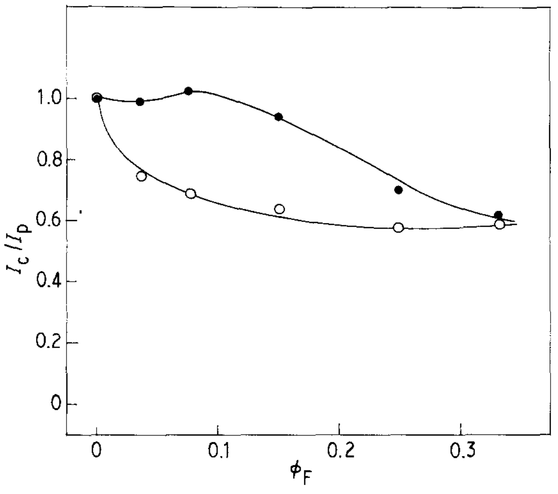


Figure 10 Relative Izod impact strength, I_c/I_p, of (○) i-PP/talc and (●) i-PP/talc/LICA 38 composites versus ϕ_F.

3.4. State of dispersion

Scanning electron micrographs of the i-PP/talc composites are shown in Fig. 11a–g. The dispersion of untreated talc in i-PP is rather poor, as evidenced by the non-uniform fracture surface with uneven voids, Fig. 11b–d. The filler also shows a high degree of agglomeration even at low concentrations. Talc is known to have active sites on its surface which are involved in the formation of a network-like structure [46]. The accicular platy shape of talc would be a favourable factor in the formation of agglomerates. Surface-treated talc exhibits better dispersion with somewhat uniform and smoother fracture surfaces with a larger extent of polymer residues adherent to filler particles in particular at low ϕ_F (i.e. up to ~ 7.7 vol % filler), Fig. 11e, f. This is due probably to enhanced interaction of talc with i-PP. With further increase in ϕ_F, however, the fracture surface becomes

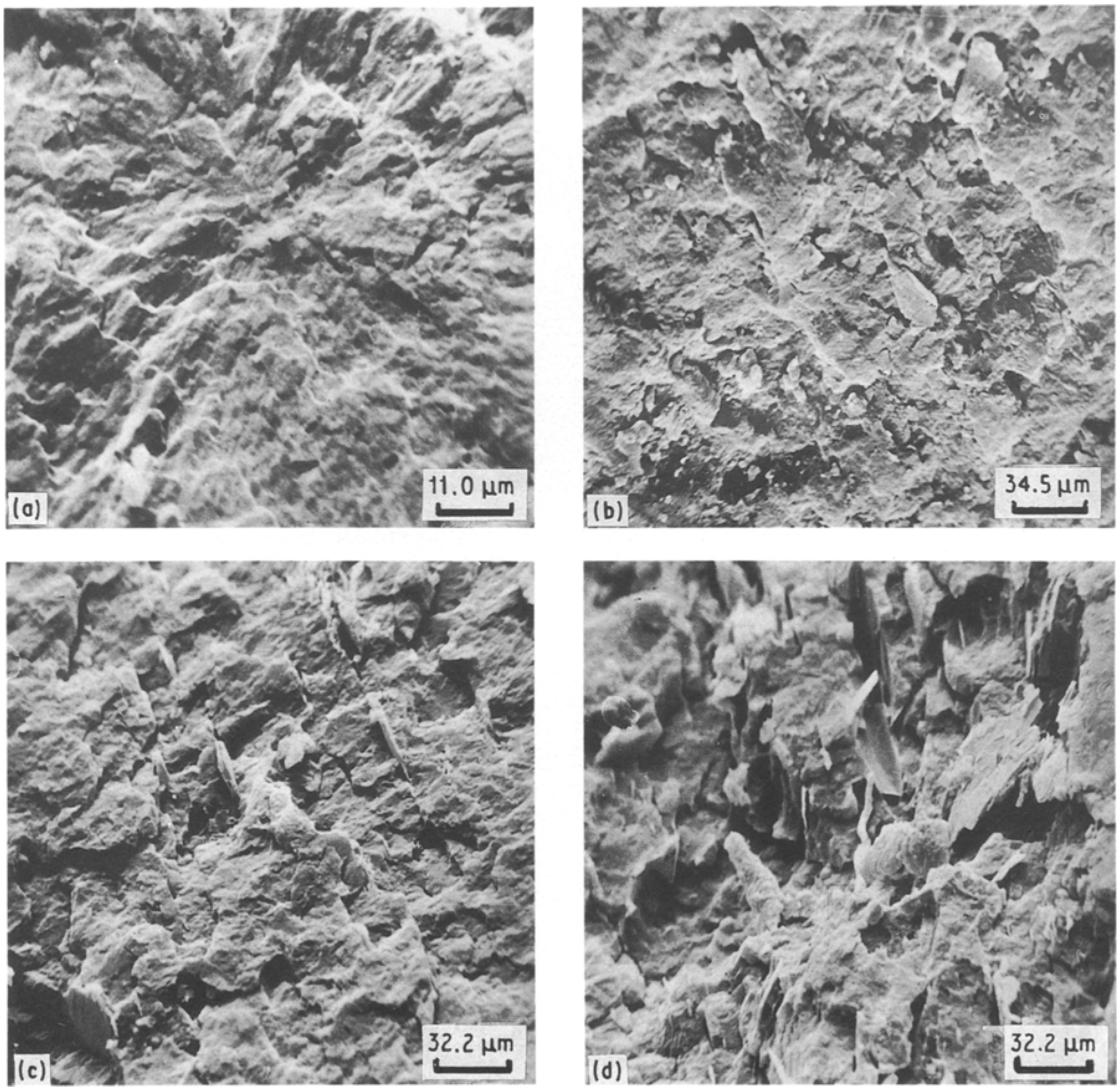
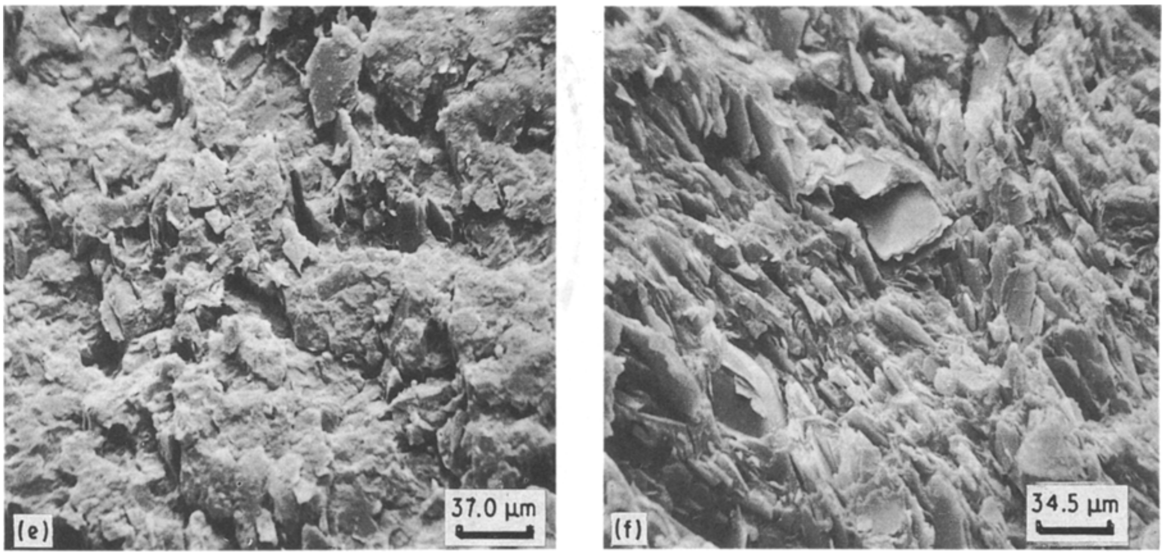


Figure 11 Scanning electron micrographs of (a) i-PP, and i-PP/talc composites at varying ϕ_F: (b) 0.03, (d) 0.07, (f) 0.25. The micrographs of i-PP/talc/LICA 38 composites at corresponding talc levels are shown in (c), (e) and (g), respectively.



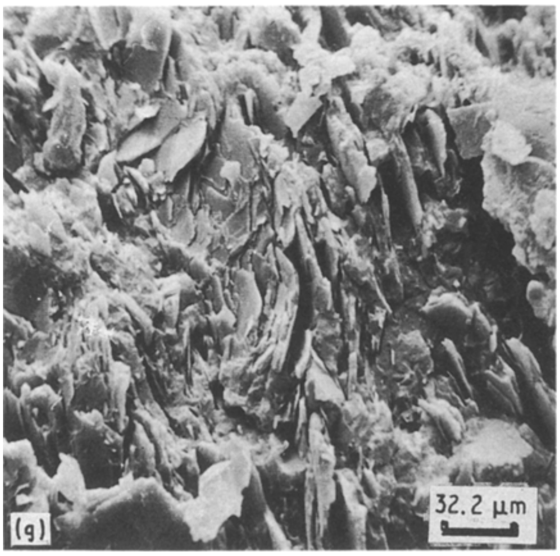


Figure 11 Continued

non-uniform and filler particles become bare, devoid of adherent polymer, Fig. 11g, similar to the untreated talc/i-PP composites. This may be due to dilution of the polymer by the use of a larger volume of filler particles.

4. Conclusion

Addition of talc as a filler into i-PP modifies the properties of the latter to a significant extent. Tensile strength and breaking strain decreased, whereas tensile modulus increased with filler content. Izod impact strength also decreased, the decrease was quite small compared to unfilled i-PP. At the maximum filler concentration of 33.3 vol % (60 wt %) the impact strength was ~ 0.60 times that of unfilled i-PP implying the potential of talc as an economic diluent to i-PP with a possibility of high loading without much trade off on other useful mechanical properties. Stress concentration is introduced at a critical filler level of ~ 3.6 vol %.

Surface treatment of talc with a titanate coupling agent LICA 38 modified the composite properties further. Interphase interaction increased, which in turn increased the tensile modulus over the values with unmodified talc. Although stress concentration was significantly reduced the tensile strength still showed a declining trend due probably to enhanced amorphization of i-PP consequent to this enhanced interaction. Izod impact strength and breaking strain showed a similar trend as in the previous composites; the values were, however, higher.

A degree of better dispersion into i-PP of surface-modified talc was revealed by SEM morphology studies.

References

1. M. E. J. DEKKERS and D. HEIKENS, *J. Appl. Polym. Sci.* **30** (1985) 2389.
2. S. CACOUTIS, R. T. WOODHAMS and P. G. K. VAN DE POLL, *Polym. Compos.* **9** (1983) 51.
3. J. E. STAMHUIS, *ibid.* **9** (1988) 72.
4. H. KATZ and J. V. MILEWSKI (eds), “Handbook of Fillers and Reinforcement for Plastics” (Van Nostrand, Reinhold, New York, 1978).
5. J. A. BRYDSON, “Plastics Materials”, 5th Edn (Newnes-Butterworths, London, 1989).
6. S. J. MONTE and G. SUGERMAN, “Ken-React Reference Manual” (Bulletin No. KR-1084L, 1985).
7. O. LAGUNA, J. P. VIGO and E. P. COLLAR, *Rev. Plast. Mod.* **53** (1987) 367; (C.F.C.A. 106(24), 197237u 1987).
8. B. SERAFIMOV, *Plast. Kautsch.* **29** (1982) 598 (C.F.C.A. 98(8), 54975p 1982).
9. S. N. MAITI and P. K. MAHAPATRO, *Int. J. Polym. Mater.* **13** (1990) 1.
10. L. CZARNECKI and J. L. WHITE, *J. Appl. Polym. Sci.* **25** (1980) 1217.
11. J. A. MANSON and L. H. SPERLING, “Polymer Blends and Composites” (Plenum, New York, 1976) Ch. 12.
12. C. D. HAN, T. VAN DEN WEGHE, P. SHETE and J. R. HAW, *Polym. Engng Sci.* **21** (1981) 196.
13. S. N. MAITI and P. K. MAHAPATRO, *Polym. Plast. Technol. Engng*, **30**(5&6) (1991) 559.
14. N. K. JHA, A. C. MISRA, N. K. TIWARI and P. BAJAJ, *Polym. Engng Sci.* **25** (1985) 434.
15. J. A. ASHMAN and A. MUDRAK, US Pat. 41 000 075 (1978) Harshaw Chemical Co. (C.F.C.A. 90, 55741f 1979).
16. E. H. KERNER, *Proc. Phys. Soc.* **69B** (1969) 808.
17. L. E. NIELSEN, “Mechanical Properties of Polymers and Composites” (Marcel-Dekker, New York, 1974).

18. V. DOLAKOVA-SVEHLOVA, *J. Macromol. Sci. Phys. B* **21** (1982) 231.
19. E. GUTH and H. SMALLWOOD, *J. Appl. Phys.* **15** (1944) 758.
20. F. RAMSTEINER and R. THEYSON, *Composites* **15** (1984) 121.
21. F. F. LANGE, in "Fracture and Fatigue", "Composite Materials", Vol. 5, edited by L. J. Broutman (Academic Press, New York, 1974) p. 2.
22. G. W. BRASSEL and K. B. WISCHMANN, *J. Mater. Sci.* **9** (1974) 307.
23. L. J. BROUTMAN and R. H. KROCK, "Modern Composite Materials" (Addison-Wiley, Reading, MA 1967).
24. A. BAUMGARTER and M. MUTHUKUMAR, *J. Chem. Phys.* **87** (1987) 3082.
25. A. K. GUPTA, V. B. GUPTA, R. H. PETERS, W. G. HARLAND and J. P. BERRY, *J. Appl. Polym. Sci.* **27** (1982) 4669.
26. K. K. SHARMA, MTech dissertation, IIT, New Delhi (1987).
27. S. N. MAITI and P. K. MAHAPATRO, *J. Appl. Polym. Sci.*, **42** (1991) 3101.
28. C. D. HAN, C. SANDFORD and H. J. YOO, *Polym. Engng Sci.* **18** (1978) 849.
29. K. MITSUISHI, S. KODAMA and H. KAWASAKI, *ibid.* **25** (1985) 1069.
30. A. K. GUPTA and S. N. PURWAR, *J. Appl. Polym. Sci.* **22** (1984) 3513.
31. T. KUNORI and P. H. GEIL, *J. Macromol. Sci. Phys. B* **18** (1980) 135.
32. M. R. PIGGOTT and J. LEIDNER, *J. Appl. Polym. Sci.* **18** (1974) 1619.
33. L. E. NIELSEN, *ibid.* **10** (1966) 97.
34. L. NICOLAIS and M. NARKIS, *Polym. Engng Sci.* **11** (1971) 194.
35. L. NICOLAIS and L. NICODEMO, *Int. J. Polym. Mater.* **4** (1974) 229.
36. L. NICODEMO and L. NICOLAIS, *J. Mater. Sci. Lett.* **2** (1983) 201.
37. S. N. MAITI and R. JEYAKUMAR, *J. Polym. Mater.* 7 (1990) 29.
38. E. M. PASSMORE, R. M. SPRIGGS and T. VASILOS, *J. Amer. Ceram. Soc.* **48** (1965) 1.
39. L. E. NIELSEN, *J. Compos. Mater.* **1** (1967) 100.
40. S. SAHU and L. J. BROUTMAN, *Polym. Engng Sci.* **12** (1972) 91.
41. G. W. BRASSEL and K. B. WISCHMANN, SPE Regional Technical Conference on Advances in Reinforced Thermoplastics, Paper 1 (El Segundo, CA 1972).
42. O. ISHAI and L. J. COHEN, *J. Compos. Mater.* **2** (1968) 302.
43. L. NICOLAIS and L. NICODEMO, *Polym. Engng Sci.* **7** (1967) 1.
44. S. N. MAITI and R. SUBBARAO, *Int. J. Polym. Mater.*, **15** (1991) 1.
45. Y. N. SHARMA, S. F. XAVIER and J. S. ANAND, in "Proceedings of the Seminar on Matrix Resins for Composites", DST, New Delhi (1986) p. 223.
46. M. S. BOIRA and C. E. CHAFFEY, *Polym. Engng Sci.* **10** (1977) 715.

Received 9 January
and accepted 7 June 1991