# Influence of the mechanical properties of the dispersed phase upon the behaviour of nylon/rubber blends: crosslinking effect

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Rubber-modified polyamides were obtained directly by blending modified poly(styrene-*b*-butadiene-styrene) (SBS) triblock copolymer used as the rubbery component and nylon 6. The rubber was modified with various amounts of maleic anhydride (MA) with the aid of dicumyl peroxide. In this study, the influence of MA concentration on the crosslinking level of the rubber and its effect on the blend properties were investigated. Crosslinking level of MA-modified SBS was measured using isooctane, the calculation being done following the Flory–Rehner equation. Impact testing of the blends revealed a maximum peak in impact strength *versus* rubber crosslinking level. However, the maximum was not so eminent as we expected. Explanations about the impact behaviour of the blends related to toughening mechanisms are presented.

## INTRODUCTION

The commercial importance of toughened polymer blends has rapidly increased in recent years. Impact strength or toughness is a complex mechanical property of a material reflecting the absorption of impact energy during different processes preceding fracture of the specimen: viscoelastic deformation, yielding, voiding, crazing and finally formation and propagation of true cracks leading to fracture. As is well known, the toughness of most thermoplastics can be considerably enhanced by the incorporation of a dispersed rubbery phase. The toughening mechanisms involved are influenced by the properties of the matrix material and by the morphology of the blend[1]. Thus the dominant energy absorbing process is different in different blends. *Figure 1* shows schematic diagrams of some toughening mechanisms. Toughening is typically explained by invoking two major deformation mechanisms, crazing and shear yielding[2], while cavitation and deformation also play an important role. High impact polystyrene (HIPS) is exclusively toughened by crazing and semi-ductile polymers, such as nylons, are toughened by shear yielding. Sometimes they occur simultaneously. Hole formation (cavitation) in rubber particles and deformation of rubber particles are suggested as additional factors contributing to toughening. Crazing and hole formation are dilative processes, while shear yielding and particle deformation occur with constant volume. The mechanism by which a rubber-modified polymer dissipates energy depends on the matrix, rubber content, type of rubber, rubber particle

size and degree of adhesion between the rubber phase and the matrix phase.

Although there have been many intensive studies to explain the toughening effect of the dispersed phase, the subject still has many controversial points[3]. Numerous studies have been carried out to investigate the effects of dispersed phase properties or structures on the toughening of nylons, such as rubber concentration, particle size, mechanical properties of the rubber, intrinsic interparticle distance, degree of functionalization, to name a few [4–10]. However, few studies have been concerned with the crosslinking of the rubber
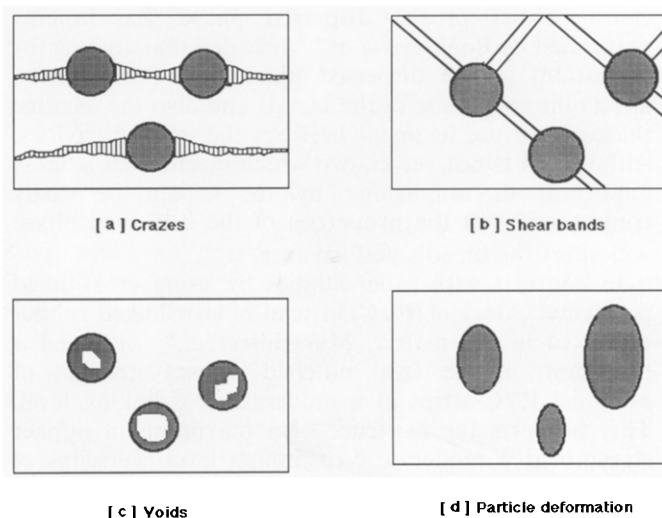


Figure 1   Schematic diagram of toughening mechanisms

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particle. Bucknall[1] states that crosslinking of the rubber particle is desirable since the rubber phase is subjected to very large tensile strains during impact, giving a craze-like structure. For uncrosslinked polymers like poly(butadiene), entanglements of the molecules are unable to prevent rapid flow and fracture in response to an applied stress; at room temperature the rubber is far above its glass transition temperature ($T_g$) and relaxation times are extremely short. A moderate degree of crosslinking allows the rubber to reach high strains by fibrillation and at the same time renders mechanical strength to the fibrils. Little experimental work has been reported for the toughening effect by a crosslinked rubber. Dao[11] found that a highly crosslinked ethylene–propylene–diene terpolymer (EPDM) rubber was slightly more effective than uncrosslinked EPDM. However, as Boggreve and Gaymans[6] pointed out, he did not report on the particle size used in the blends and this limited the validity of his conclusions because particle size strongly influences the impact behaviour of the blend. Crosslinking also has consequences for the rubber viscosity, which affects the blending process and thus the rubber particle size. Although still conjectural, light crosslinking may not seriously affect the blend's performance. According to Boggreve *et al.*, lightly crosslinked EPDM did not seem to have any effect on the impact properties of nylon 6/EPDM blends[7]. Tinker[12] is quoted in Boggreve and Gaymans' report[6] to insist that an optimum crosslinking level in natural rubber (*cis*-1,4-polyisoprene) exists for impact modification of polypropylene. However, he also did not report whether the dispersed rubber particle size was kept constant.

It is well known that the brittle–tough transition depends on particle size. Wu[8] stated recently that a sharp brittle–tough transition occurred at a critical rubber particle size. Although it remains a controversial point whether the sole material parameter controlling the brittle–tough transition of nylon/rubber blends is the interparticle distance (*ID*), there is no doubt that particle size is decisive for toughening. Following Wu's argument, if *ID* is greater than the critical value (*ID*$_c$) the fracture mode is brittle and vice versa. *ID*$_c$ is independent of particle size and rubber volume fraction and is characteristic of a given matrix. However, Wu's explanation is concerned only about the matrix and his proposition does not take into consideration the characteristics of the dispersed phase. Experiments performed by Boggreve *et al.*[4,5] revealed that voiding (or cavitation) in the dispersed phase and delamination affects the toughness of the blends and also the average thickness of the ligament between the created cavities. Although it is not yet known which mechanism is most important in toughening nylons, it can be easily conjectured that the properties of the dispersed phase will affect the blend's performance.

In contrast with other studies by using crosslinked poly(vinyl chloride) (PVC) instead of crosslinked rubber dispersed in the matrix, Marshall *et al.*[13] observed a maximum in the Izod notched impact strength of extruded PVC strips at a moderate crosslinking level. This supports the existence of a maximum in impact strength at a moderate crosslinking level. Terselius *et al.*[14] performed instrumented falling weight impact tests on unnotched PVC pipe specimens and verified that the maximum in falling weight impact strength at moderate

crosslinking level was a true material property. According to their results, the most efficient resistance to crack initiation was observed in the specimen of moderate crosslinking level and could be attributed to enhanced post-yield deformation, presumably due to yielding and voiding in weak, interparticular regions of reduced enganglement density in the PVC.

The objective of this paper is to investigate experimentally a possible relationship between the toughening mechanism in nylon/rubber blends and the crosslinking level of the rubber phase. We used nylon 6 as a matrix material and blends were made with lightly crosslinked poly(styrene-*b*-butadiene-*b*-styrene). Level of crosslinking, which was determined during the process of rubber functionalization by reaction with maleic anhydride (MA), was measured and its effect on the mechanical properties of the blends was investigated.

## EXPERIMENTAL

### Materials

The materials used in this study are blends of nylon 6 (KN111, $M_n = 20\,000\,\text{g mol}^{-1}$, $\eta_{rel} = 2.45$ in 96% $H_2SO_4$) obtained from Kolon Co. and SBS rubber (Kraton D 1102, $M_n = 69\,000\,\text{g mol}^{-1}$, styrene/butadiene = 28/72 wt%) supplied by Shell. SBS was grafted with various amounts of MA (Kanto Chemical Co.), using dicumyl peroxide (DCP) as an initiator for the grafting reaction. All chemical reagents used in this work were of analytical grade and were used as received.

### Rubber modification

The functionalization of SBS with MA was carried out in a 42 mm Brabender twin screw extruder (AEV 651) at a fixed rotation speed of 30 rev min$^{-1}$. SBS was dried overnight in a vacuum oven at 100°C. The initiator concentration was set as 0.3 wt%; MA concentration was varied from 0.25 to 2 wt%. After removing free MA by washing with methanol (48 h at room temperature) and drying overnight in a vacuum oven at 100°C, the amount of MA grafted on to the rubber was measured with potentiometric titration and *FT*i.r. spectroscopy.

### I.r. spectroscopy

The i.r. spectrum of the reactive rubber was taken on compression moulded film using an Alpha Centauri spectrophotometer (Mathson Instrument).

### Titration

SBS-*g*-MA was washed in methanol for 48 h to remove free MA. Then it was dissolved in tetrahydrofuran (THF)/methanol (90/10). By titrating with 0.1 N NaOH solution using phenolphthalein as an indicator, the amount of MA grafted onto the rubber was determined[4–6].

### Rheometry

Rheological properties of nylon 6 and rubbers were measured using a Rheometrics Dynamic Spectrometer (RDS7700, Rheometrics Co., USA) on which a 2.5 mm diameter parallel plate, plate gap of 1.2 mm, was mounted. Frequency range was set as 0.1–500 rad s$^{-1}$ and strain was 10%. Temperature was set at 240°C. Samples were dried overnight in a vacuum oven at 100°C before measurement, which was done under a nitrogen atmosphere.
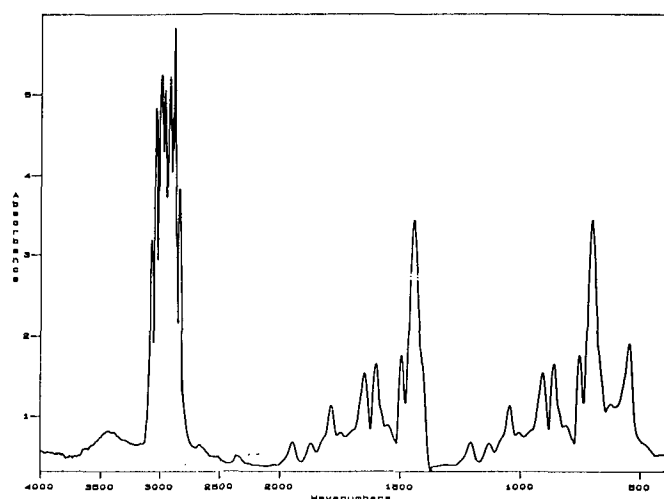
**Figure 2** I.r. spectrum of m-SBS rubber before blending

*Blend preparation*

A number of blends were prepared by compounding nylon 6 with modified SBS (m-SBS) in a Brabender twin screw extruder with screw speed of 30 rev min$^{-1}$. The rubber fraction was varied from 0 to 30 wt%. The extrusion temperature was set as 200/240/240°C (hopper/melting zone/die). To prepare the samples for impact testing, the blends were injection moulded using an injection moulder (PM-30M, Hillard Co.). Specimens of the blends were prepared following ASTM D256-84.

*Crosslinking level measurement*

Crosslinking level, defined as the inverse of molecular weight between crosslinks, was measured as follows. When a crosslinked polymer is placed in a suitable solvent, the polymer imbibes the solvent and undergoes swelling to an extent determined by the nature of the polymer and the solvent. At equilibrium, the swelling by a good solvent is given by the following Flory–Rehner equation[15]:

$$v = \frac{1}{M_c} = -\frac{V_r + \chi V_r^2 + \ln(1 - V_r)}{d_r V_0(V_r^{1/3} - 2V_r/f)} \quad (1)$$

where $v$ is the effective number of moles of crosslinked chains per g of polymer, $M_c$ is the molecular weight between crosslinks, $V_r$ is the volume fraction of polymer in the swollen mass, $V_0$ is the molar volume of the solvent, $\chi$ is the polymer–solvent interaction parameter, $d_r$ is the density of the polymer, and $f$ is the functionality of the crosslinks. The volume fraction of the polymer in a swollen mass can be calculated by equations (2) and (3):

$$V_r = 1/(1 + Q) \quad (2)$$

$$Q = \frac{\text{wt of solvent in gel}}{\text{wt of gel}} \times \frac{d_r}{d_s} \quad (3)$$

where $d_s$ is the density of the solvent.

In this study, isooctane was used as the solvent imbibed in the polymer. The weight change of the rubber in isooctane was recorded with time and the equilibrium value was obtained. The interaction parameter $\chi$ between isooctane and SBS was obtained using the experimental relationship of Bishop and Davison[16]:

$$\chi = 0.406 + 0.522V_r \quad (4)$$

*Mechanical properties*

Notched Izod impact strength was measured according to ASTM D256-84 using a cryogenic impact tester (Yasuda Seiki 195-LFR, owned by Cheil Industries Co.). Measurements were done at temperatures of $-40$, $-20$, 0 and 20°C. Tensile properties were determined on an Instron 4202 tester with a computerized data acquisition system at a crosshead speed of 10 mm min$^{-1}$. Specimens for tensile testing were prepared at room temperature according to ASTM D638M-84.

*Scanning electron microscopy (SEM) and transmission electron microscopy (TEM)*

Fractured surfaces of the samples after impact testing were furnished with a gold layer and examined in a Hitachi scanning electron microscope (S-2500 and S-510). Rubber etching was done using THF. Ultra thin sections of injection moulded specimens, stained with OsO$_4$, were examined by TEM (Jeol 2000EX2, owned by Cheil Industries Co.). The particle size distribution was obtained from the photomicrographs[15].

## RESULTS AND DISCUSSION

*Functionalization of SBS rubber*

The functionalization of SBS was carried out with different amounts of MA, at a constant DCP content (0.3 wt%). The i.r. spectrum shows the presence of the carbonyl stretching vibration of the anhydride group at 1785 cm$^{-1}$ which indicates functionalization of SBS (*Figure 2*). By relating this peak to other peaks such as the C–H stretching vibration of the CH group at 715 cm$^{-1}$, we could determine MA content. MA content was also determined by titration (*Figure 3*). The amount of grafted MA increased with the amount of MA added during functionalization, in agreement with the i.r. results. MA grafting efficiency was about 10% in all three cases presented in *Figure 3*, i.e. for MA additions of 0.5, 1 and 1.5 wt% (hereafter referred to as 0.5 wt% MA m-SBS, 1 wt% MA m-SBS and 1.5 wt% MA m-SBS, respectively).

*Rheological properties*

The melt viscosities of nylon 6 and modified rubbers are plotted as functions of shear rate at 240°C in *Figure 4*. The materials exhibit the shear thinning behaviour
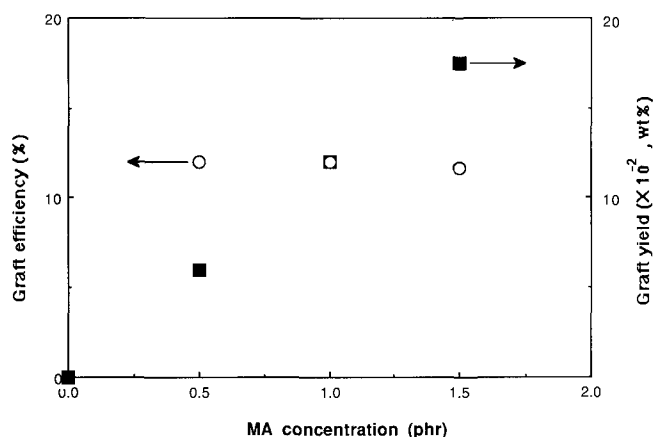


**Figure 3** Graft efficiency and MA concentration grafted on the rubber (graft yield) *versus* amount of MA added
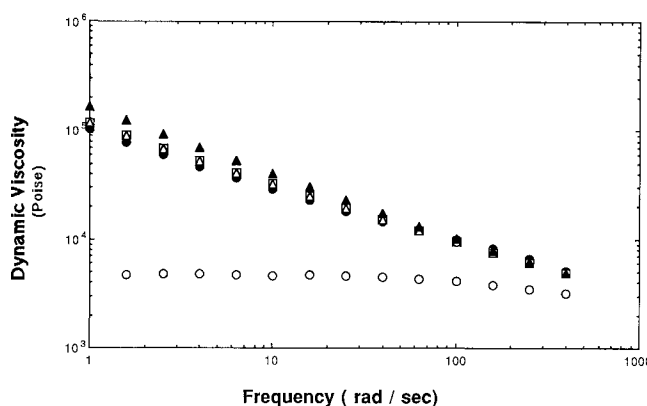
**Figure 4**  Viscosity *versus* shear rate at 240°C: (○) nylon 6; (●) SBS; (□) 0.5 wt% MA m-SBS; (△) 1 wt% MA m-SBS; (▲) 1.5 wt% MA m-SBS

typical of polymers, but the viscosity reduction is rapid for the rubbers whereas it is very slow for nylon 6.

*Figure 5* gives the shear moduli of the rubbers and nylon 6 as functions of shear rate. Different MA concentrations in the rubber seem to have only a slight effect on G' and η'. The shear viscosity and modulus of the 1.5 wt% MA m-SBS added were higher than those of the other rubbers. This appears to be related to the effect of MA on crosslinking, rubber viscosity and modulus both increasing with the degree of crosslinking. Experimental results shown below in *Figure 10* confirm this explanation.

*Blending*

The proportion of rubber in the blend was varied from 0 to 30 wt%. As described in an earlier study[6], MA-modified polymers react with nylon 6 during the blending process in the extruder. The use of MA as a coupling agent strongly improves dispersion of the SBS rubber in nylon 6 because nylon-grafted SBS reduces the interfacial tension between the matrix and the dispersed phase. Generally it might be expected that more MA-grafted dispersed phase wille be smaller in particle size. However, this is not always true. Non-uniform MA grafting and the low efficiency of the nylon graft reaction do not guarantee a linear relationship between the rubber particle size and the amount of added MA. This fact is supported by the experimental results of Boggreve *et al.*[7], who found that the concentration of the coupling agent has hardly any influence on the dispersion process due to low reaction efficiency. Later it is shown that MA grafting does not only proceed by simple addition to the double bond but also by some other reactions.

The particle size distribution was obtained from SEM micrographs. In this study, the so-called weight average particle size is used:

$$d_w = \sum_i n_i d_i^2 / n_i d_i \qquad (5)$$

where $n_i$ is the number of particles whose size is $d_i$. *Figure 6a* shows the blend of nylon 6 and unfunctionalized SBS (80/20 wt%). Very large (> 10 μm) rubber particles are observed in the blend with unmodified rubber but rubber particle size changes dramatically with the addition of m-SBS, see *Figures 6b* to *6d*. Unfunctionalized rubber is immiscible with nylon 6, producing a large domain in the matrix due to poor dispersion and large

interfacial tension. *Figure 7* shows the morphology of the fractured surfaces when the amount of 1 wt% MA m-SBS was varied from 10 to 20 to 30 wt% of the blends. The weight average particle size is gradually reduced (0.98 μm > 0.83 μm > 0.75 μm); however, because of the relatively mild mixing conditions and the short residence time in the extruder, the particle size was not markedly changed.

*Impact properties and crosslinking level*

Notched Izod impact strength measured for different samples is presented in *Figure 8* as a function of temperature. As a result of coupling of the phases and the change in morphology, impact strength increases with the amount of m-SBS. The blend containing 30 wt% m-SBS rubber showed very tough behaviour, one order of magnitude higher than that of pure nylon 6. The general behaviour followed was that of a typical nylon toughened by a modified rubber.

Unexpectedly, however, we observed one interesting point in the 80/20 nylon 6/m-SBS blends: modification of the SBS with 0.5, 1 and 1.5 wt% MA resulted in almost the same average particle size in the blends (0.82, 0.83 and 0.82 μm, respectively). Their notched Izod impact strength values were 11.7, 13.8 and 12.5 kJ m$^{-2}$, respectively, at 20°C. Despite repeated measurements and excluding experimental errors, the higher notched Izod impact strength for the nylon 6/1 wt% MA m-SBS blend was reproduced. Since this and the blends with 0.5 wt% MA and 1.5 wt% MA were processed identically and had the same rubber particle size, they should have the same impact strength. *Figure 9* shows histograms of particle size distribution in these blends. Since they are quite similar, the only explanation we could imagine for the difference in strengths was the difference of modified rubber properties.

When we added more than 2 wt% MA to SBS, the extrudate was totally cracked like a powder, showing typical crosslinked rubber behaviour. Also when only DCP was added to the rubber, the same phenomenon occurred, owing to the DCP initiating crosslinking. Depending upon the reaction conditions and the presence of free-radical initiators, MA reacts with SBS via four different basic reactions with the products shown in *Figure 10*. Crosslinking occurs through intermolecular reactions. The more MA is added, the greater the addition to double bonds within the SBS
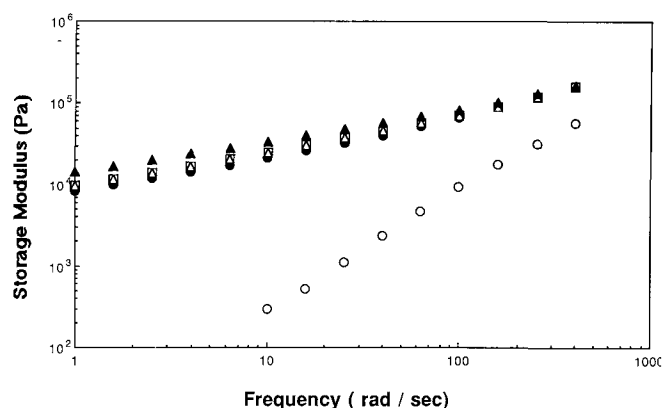


**Figure 5**  Shear modulus *versus* shear rate at 240°C: (○) nylon 6; (●) SBS; (□) 0.5 wt% MA m-SBS; (△) 1 wt% MA m-SBS; (▲) 1.5 wt% MA m-SBS
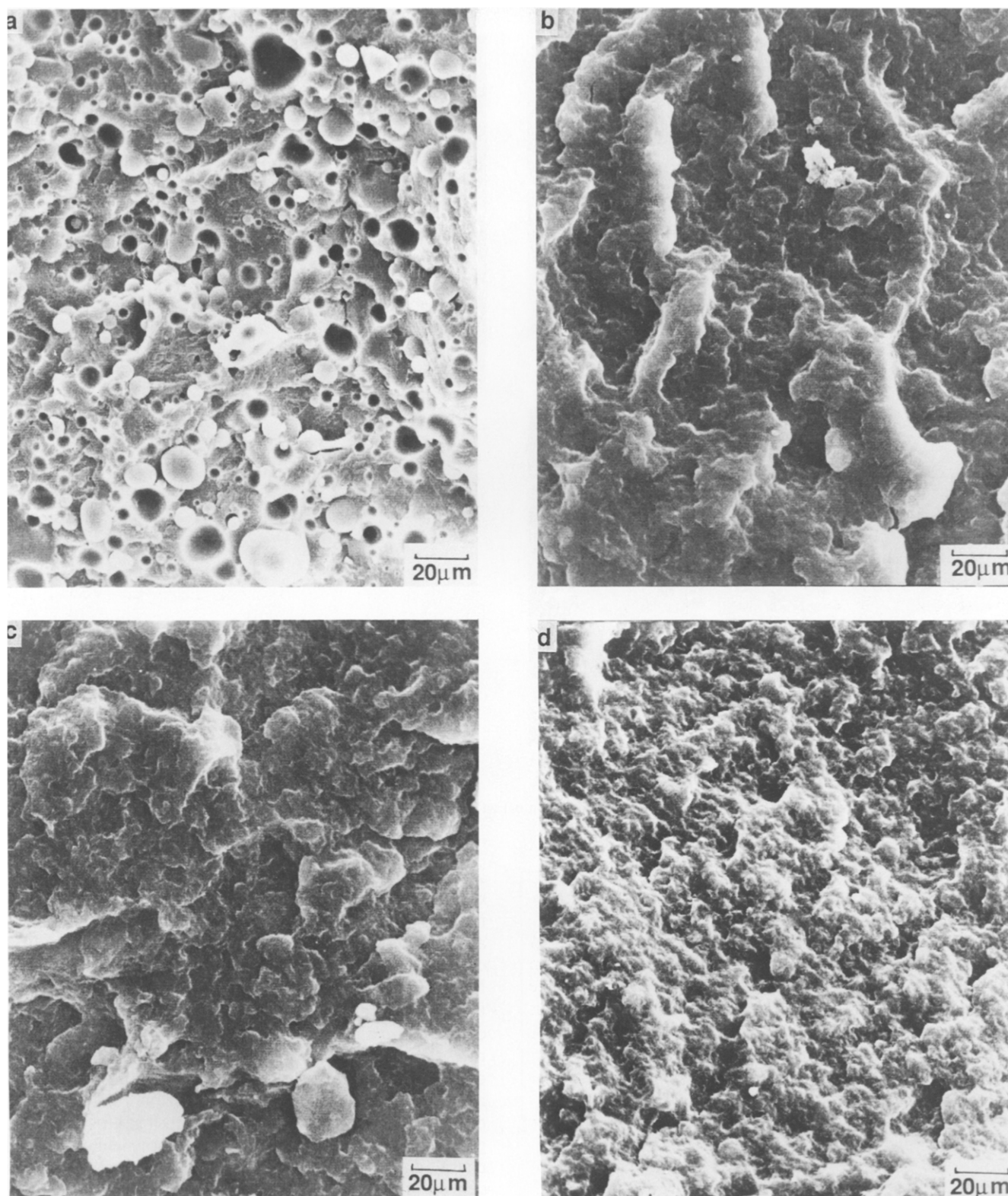
**Figure 6** Scanning electron micrographs of fracture surfaces of 80/20 wt% nylon 6/SBS (or m-SBS) blends: (a) SBS; (b) 0.5 wt% MA m-SBS; (c) 1 wt% MA m-SBS; (d) 1.5 wt% MA m-SBS

chains; on the other hand, intermolecular addition is also increased. Crosslinking level is shown in *Figure 11* as a function of the amount of MA added. A minimum in crosslinking level is obtained under the impulse of two crosslinking paths, one by peroxide and the other by MA intermolecular addition. There might be some physical crosslinking by the styrene block in the SBS rubber. We assume that it is the same in all the rubbers.

Notched Izod impact strength *versus* crosslinking density is presented in *Figure 12*. (In this figure are shown two other points for which rubbers containing 0.25 and 0.75 wt% MA were used. They were controlled to have almost the same particle size as rubbers with 0.5, 1 and 1.5 wt% MA added.) The blend of nylon 6/1 wt% MA m-SBS had a maximum strength. Crosslinking affects the elastomer's modulus and viscosity, which can change the
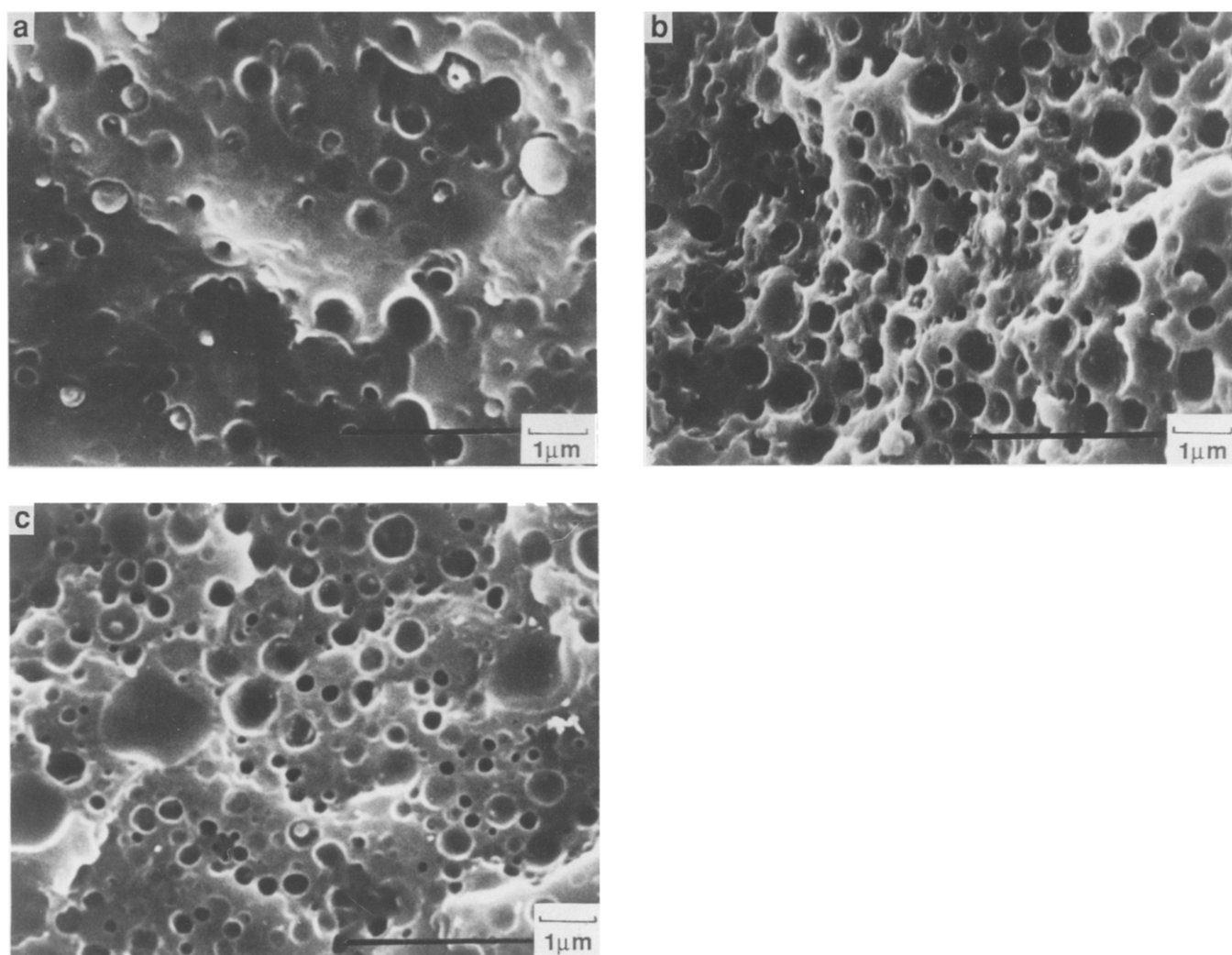
**Figure 7**  Scanning electron micrographs of fracture surfaces of nylon 6/1 wt% MA m-SBS blends after etching: (a) 90/10 wt%; (b) 80/20 wt%; (c) 70/30 wt%
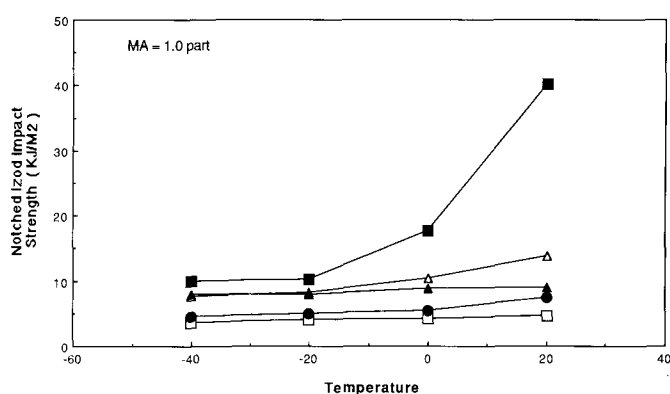


**Figure 8**  Notched Izod impact strength as a function of temperature for (□) nylon 6 and nylon 6/1 wt% MA m-SBS blends: (●) 90/10 wt%; (△) 80/20 wt%; (■) 70/30 wt%; (▲) 80/10/10 wt% (nylon 6/m-SBS/SBS)

dispersion of the particle and its deformation. Gent[17] states that tensile strength is usually found to pass through a sharp maximum as the degree of crosslinking is increased from zero. This maximum is primarily due to changes in intrinsic strength. It is generally known that introduction of high crosslink density into polymer chains decreases the fatigue performance. The crack propagation rate for crosslinked polystyrene is higher than that for uncrosslinked polystyrene at the same stress

intensity factor. Hertzberg and Manson[18] conclude that there is a strong correlation between molecular chain mobility and fatigue resistance: those polymers that offer a larger energy dissipation during deformation (low crosslinking level) exhibit better fatigue behaviour. In this sense, highly crosslinked polymers possessing a tightly constrained structure do not allow easy particle deformation and lead to accelerated fatigue crack growth rates. However, a recent study of the effect of crosslinking on the fatigue resistance of polystyrene[19] shows a maximum in fatigue crack propagation resistance at a particular, optimum degree of crosslinking. Uncrosslinked glassy polymers which tend to craze show some localized energy dissipation. With an increasing, but still very low degree of crosslinking, the deformation mode in such films tends to change from crazing to crazing plus shear deformation. The presence of shear deformation restricts the growth of crazes and this may result in an improvement of fracture properties. However, further increases in the degree of crosslinking provide additional constraints on chain mobility which restrict localized fracture properties. This is for the simple matrix, not for toughened plastics. The situation is a little different for toughened ductile polymers.

*Toughening mechanism*

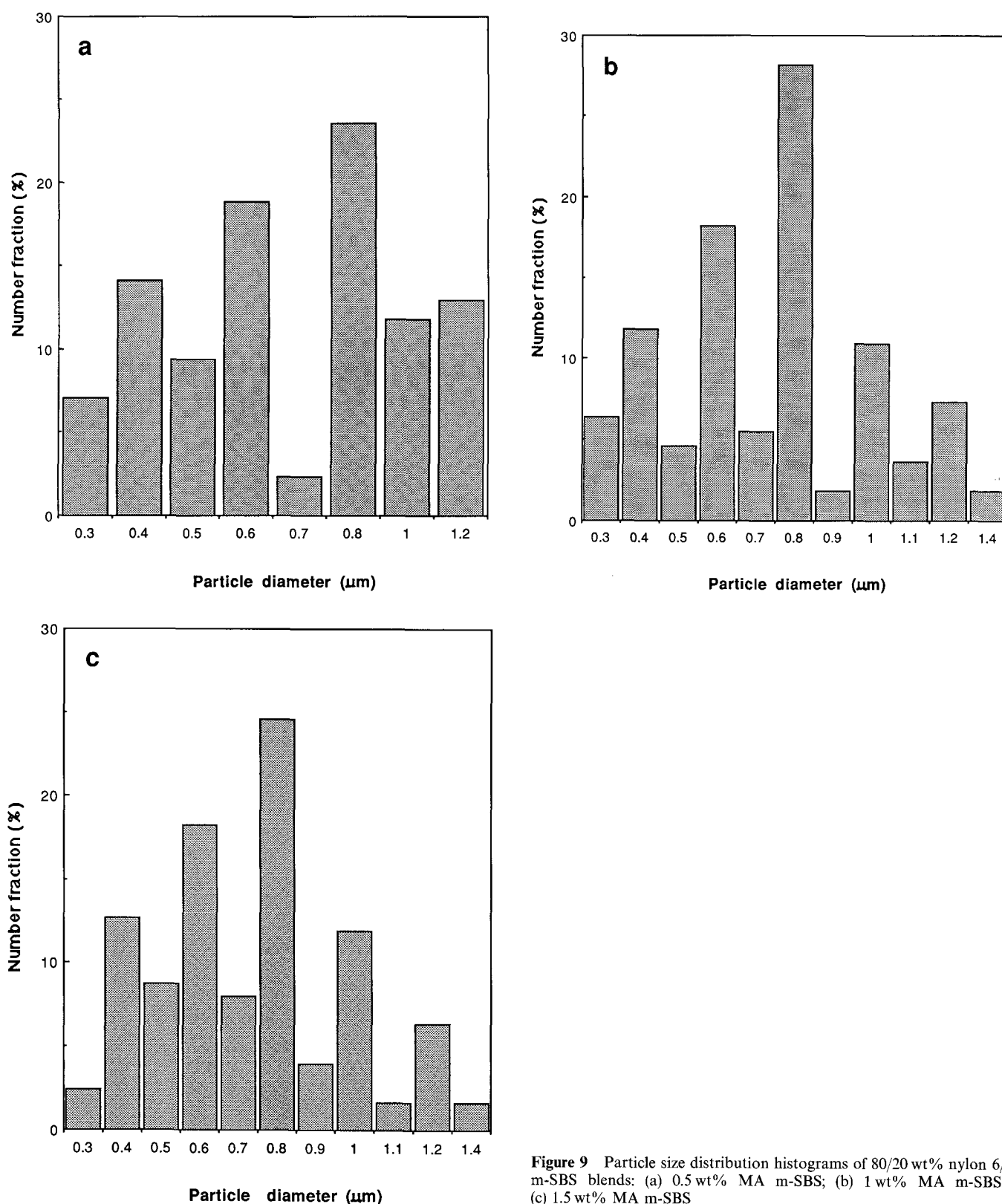In tensile testing, the maximum strength obtained is

**Figure 9** Particle size distribution histograms of 80/20 wt% nylon 6/ m-SBS blends: (a) 0.5 wt% MA m-SBS; (b) 1 wt% MA m-SBS; (c) 1.5 wt% MA m-SBS

two or three times higher at the optimum crosslinking level[17]. The maximum impact strength value, however, was not so eminent in our experiment. This can be understood by considering the toughening mechanism of the nylon 6/rubber blends. For semi-ductile polymers such as nylon 6, since the shear initiation stress $\sigma_{sh}$ is lower than the craze initiation stress $\sigma_{cr}$, shear yielding is the main deformation mechanism by which a large energy absorption takes place. On the other hand, void

formation relieves the elastic constraint in impact tests that causes extensive plastic flow in the matrix. As explained by Boggreve et al.[7], in a semi-ductile matrix like nylon 6, cavitation within the matrix is prevented by a high entanglement density. The elastic constraint in this case is relieved by internal rubber cavitation. Yielding between the cavities may occur in both ductile and brittle matrices. Whereas cavities usually coalesce quickly and form a fatal crack within a brittle matrix, shear yielding
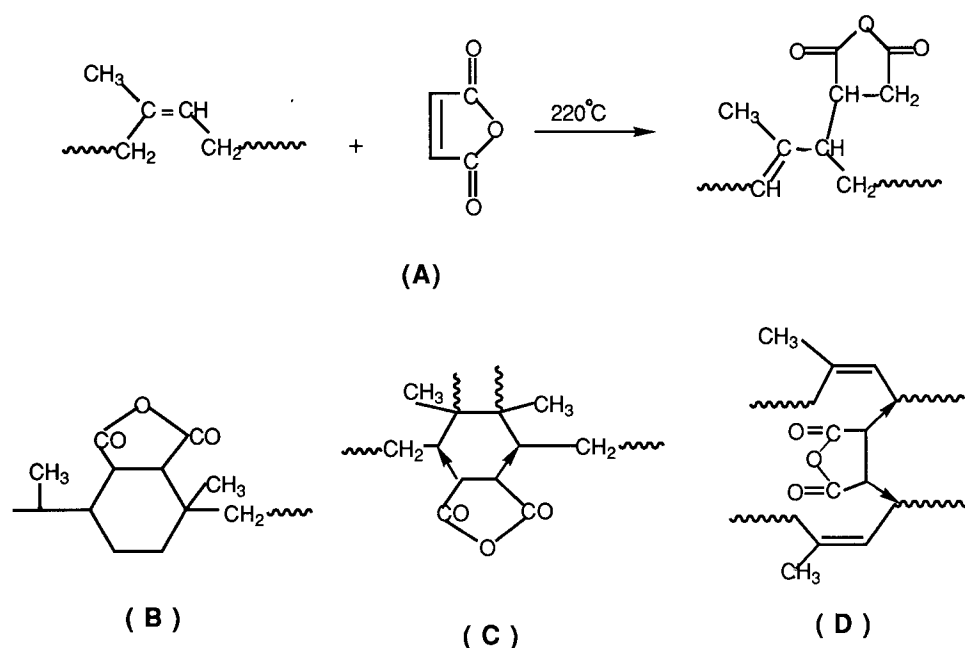
**(A)**



**( B )**　　　**( C )**　　　**( D )**

**Figure 10** Maleic anhydride addition reactions: (a) addition to α-methylenic carbon atom; (b) intramolecular addition to the double bond; (c) intermolecular addition to double bonds in different polymer chains; (d) intermolecular addition to α-methylenic carbon atoms in adjacent chains
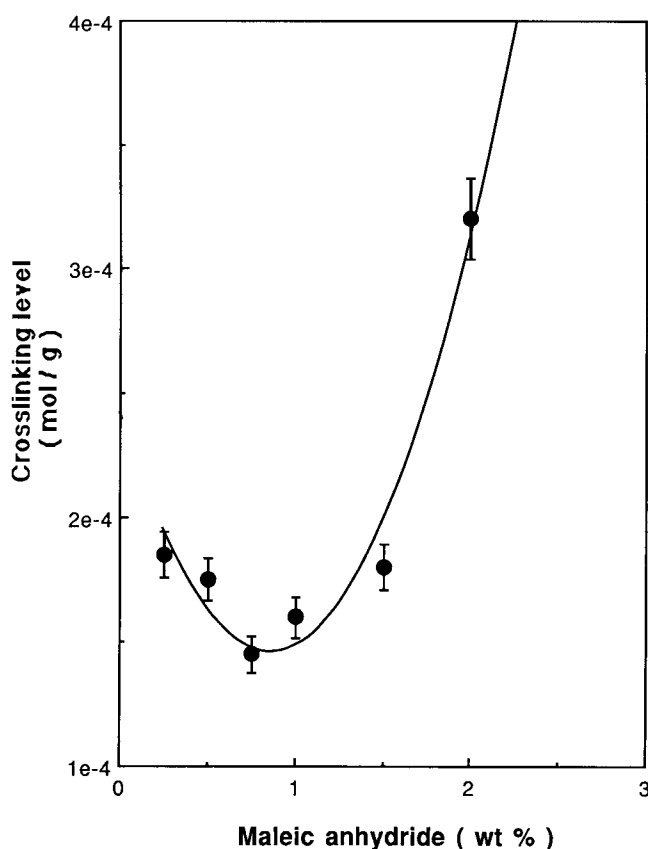


**Figure 11** Crosslinking level *versus* amount of MA added



**Figure 12** Notched Izod impact strength *versus* crosslinking level at room temperature (the last point corresponds to a blend for which 2 wt% MA m-SBS was added; the weight average particle size of this rubber was a little larger than that of the others)

can take place excessively between the cavitated rubber particles in a ductile matrix without fast crack fromation. According to Gent[17], when the rubber block faces are subjected to a uniform triaxial tension, i.e. to a negative hydrostatic pressure $(-P)$, they expand to an equal degree, provided that the rubber is itself undilative. When the expansion is small, it is proportional
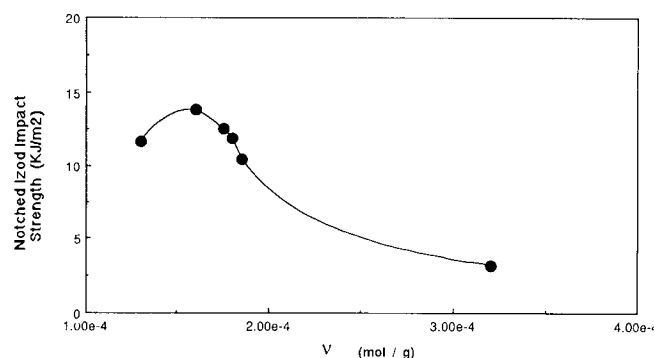
to $-P$ and given by $\lambda = 1 + 3P/4E$ where $\lambda$ is the expansion ratio and $E$ is Young's modulus of the rubber. Rubber is commonly found to undergo internal cavitation at triaxial tensions. The impact strength of rubber-modified nylon 6 with a given matrix structure and specimen geometry depends on the stress at which rubber particle cavitation occurs. Following these interpretations, crosslinking in rubber particles does not contribute to the toughness of nylon 6 but is detrimental. However, rubber particles absorb diffracted energy by deformation. Hence crosslinking takes a dual role in the rubber particle's behaviour. The increase in viscosity and modulus of the rubber (*Figures 4* and *5*) act against easy void formation, as well as enabling it to act against the deformation and thereby absorb more energy. As a result, in spite of the existence of a maximum impact strength for the nylon 6/rubber blends, it is not so eminent as expected. Provided that the toughening mechanism is as described, the maximum peak would be contracted if the impact temperature is low, because of the increased viscosity and modulus of the rubber.
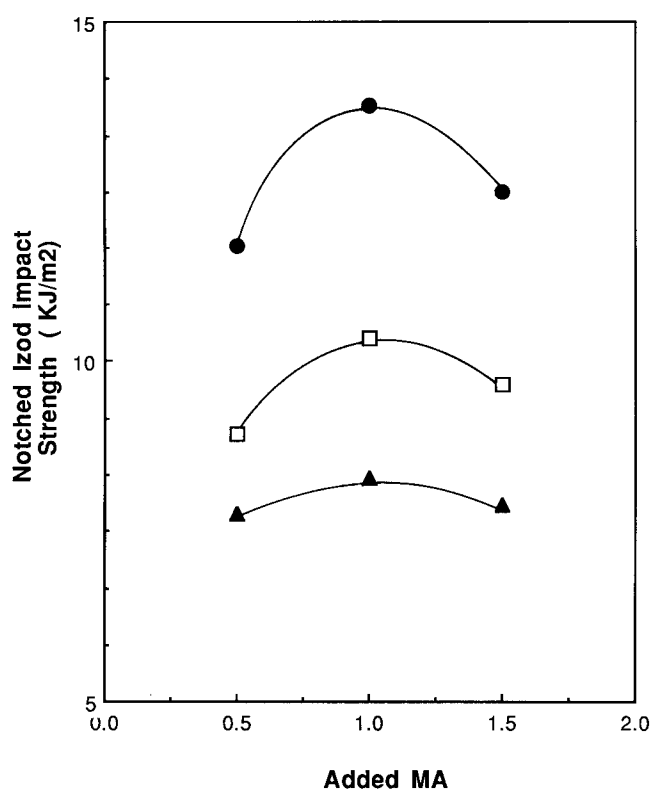
**Figure 13** Notched Izod impact strength *versus* crosslinking level for different temperatures: (●) 20°C; (□) 0°C; (▲) −20°C



**Figure 14** Tensile strength *versus* added MA amount at room temperature

The experimental results illustrate such behaviour, the maximum peak contracting with lower temperature as shown in *Figure 13*. This result implicitly suggests that cavitation is important in nylon 6 toughening. Boggreve *et al.*[4,5] insisted that toughening in the blends was due to rubber cavitation rather than deformation. *Figure 12* also shows this fact. The last point in *Figure 12* pertains to a blend in which the SBS was modified with 2 wt% MA. Its impact strength is much lower than that of the other blends and not much higher than that of the pure nylon 6 matrix. This is because of the high modulus due to the high crosslinking level which suppresses the particle's cavitation, although it helps the particle act against deformation.

Instead of cavitation in the particle, delamination may occur at the particle/matrix interface. In fact, if delamination of the rubber particles does take place, then it would increase with increasing viscosity and modulus of the rubber particles. This can be checked indirectly by tensile strength tests. *Figure 14* shows that a maximum exists in the tensile strength *versus* crosslinking level plot: that this peak is not so eminent supports Boggreve *et al.*'s proposition[4,5] that cavitation rather than delamination occurs in the rubber particle, because in the case of delamination the tensile strength peak should be greater than that of impact strength.

Numerical study by Kikuchi *et al.*[20] on the toughening mechanism in rubber-modified nylon suggests that a tougher material will be achieved when the nylon/rubber interface is designed to prevent debonding. They also suggest that it is unnecessary to take account of cavitation in the rubber particles before matrix fracture. However, a large portion of the energy dissipated is absorbed during cavity formation in the rubber particle by transformation of plain strain to plain stress and it
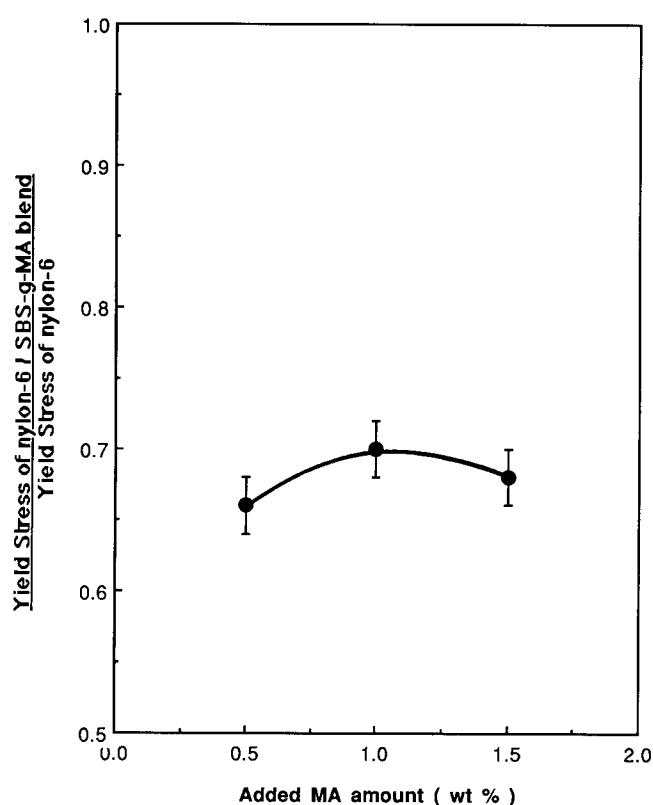
was not checked experimentally that cavitation does not occur in the rubber, even though the authors stated it to be negligible. If cavitation occurs in the rubber as we believe, the problem should be solved as a cavity evolution problem rather than a matrix/void problem. This is a matter of controversy which needs further studies in the future.

## CONCLUSION

Nylon 6 and crosslinked SBS rubber blends were prepared. Impact tests demonstrated the existence of a maximum impact strength at a moderate degree of crosslinking; however, the magnitude of this effect was not as great as that which occurs during tensile testing or impact testing of brittle polymers. This appears to result from a complex mechanism in the dispersed phase. Based on our experimental data, we believe cavitation is important in nylon toughening, and the low peak impact strength can be explained by invoking cavitation and particle deformation as the toughening mechanism in these blends. Even though delamination is also possible, tensile strength data indicate that cavitation in the rubber particles is more plausible for toughened nylon 6. Further work is under way to elucidate the detailed mechanism of toughening in nylon/SBS rubber blends and will be reported in future publications.

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