Effect of Particle Size on Impact Strength of Polymer Blends

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SYNOPSIS

The effect of particle size on impact strength of polymer blends with ductile fracture was studied. The results are in agreement with the experiments. © 1995 John Wiley & Sons, Inc.

INTRODUCTION

Rigidity and toughness are two key parameters that determine whether a polymer can be used as an engineering material. Thus the toughening mechanism for a polymer has always been an important issue in polymer material science. Wu[1] first proposed the critical interparticle distance (ID_c) model. A blend will be tough when the ID (Fig. 1) is smaller than a critical value (ID_c), and it will be brittle when ID is greater than ID_c. This model has been further confirmed not only by plastic/rubber blends,[2,3] but also by plastic/rigid particle blends.[4,5] Although the tough–brittle transition for polymer blends is independent of the size and volume fraction of dispersed particles, the impact strength changes with particle size in nylon-66/rubber blends and high density polyethylene (HDPE)/C_aCO_3 blends in the ductile fracture region ($ID < ID_c$). For nylon-66/rubber blends, applying Wu's formula [eq. (6) in this article] to figure 3 in Wu,[3] we find that the impact strength increases with the increase of rubber size. However, for HDPE/C_aCO_3 blends from figure 5 in Fu et al.,[4] we can obviously find that the impact strength decreases with the increase of particle size in the ductile fracture region. These phenomena have not been interpreted until now.

THEORETICAL

Because the elastic moduli of the dispersed phase are different from those of the matrix, when the force is applied on the polymer blend, stress concentration around particles of the dispersed phase will form. Its scope can be described as a stress volume with diameter $S = d + ID$, where d is diameter of dispersed phase, ID is interparticle distance (Fig. 2). According to Wu and Margolina's percolation model, if $ID < ID_c$, the stress volume will yield and propagate. During impact fracture, the polymer blend is tough.[6,7] (ID_c is also called critical matrix ligament thickness.)

Although Wu studied the problem on impact energy dissipation for nylon-66/rubber blends,[8] he did not distinguish between matrix contribution and dispersed phase contribution. In order to study the effect of particle size on impact strength, we write the dissipation energy in the form

$$G = G_s + G_d + G_m \tag{1}$$

where G is the total dissipation energy, G_s the surface energy contribution, G_d the dispersed phase contribution, and G_m the matrix contribution given by

$$G_m = G_{mz} + G_{my} \tag{2}$$

where G_{mz} is the matrix craze contribution and G_{my} the matrix yielding contribution. Based on Wu's results,[8] G_{mz} and G_s are negligible for ductile fracture. Therefore eq. (1) can be expressed

$$G = G_d + G_{my}. \tag{3}$$

G_{my} can be given as

$$G_m \sim W_{my} V_m \tag{4}$$

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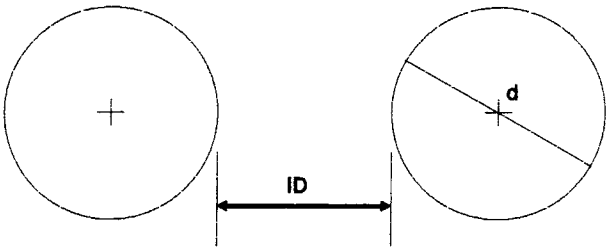

Figure 1 Definition for (surface-to-surface) interparticle distance ID and dispersed phase particle diameter d.

where W_{my} is the yield energy of matrix (per unit volume). V_m, the volume fraction of the matrix in stress volume can be shown as

$$V_m = V_s - V_d \tag{5}$$

where V_s is the volume fraction of stress volume. V_d is the volume fraction of the dispersed phase obtained from Wu's formula

$$\mathrm{ID} = d[(\pi/6V_d)^{1/3} - 1]. \tag{6}$$

So V_d can be expressed as

$$V_d = (1 + \mathrm{ID}/d)^{-3}\pi/6 \tag{7}$$

and

$$V_s = (1 + \mathrm{ID}/d)^3 V_d \tag{8}$$

so

$$V_m = [1 - (1 + \mathrm{ID}/d)^3]\pi/6. \tag{9}$$

G_d can be given by

$$G_d \sim W_d V_d \tag{10}$$

where W_d is the energy dissipated by the dispersed phase (per unit volume) and V_d is the volume fraction of dispersed phase in eq. (7).

Adding eqs. (4)–(10) to eq. (3), we obtain

$$G \sim \pi/6[1 - (1 + \mathrm{ID}/d)^{-3}]W_{my} + W_d(1 + \mathrm{ID}/d)^{-3}\pi/6. \tag{11}$$

Now we discuss the following three special cases.

1. If $W_{my} \gg W_d$, eq. (11) can be shown as

$$G \sim \pi/6[1 - (1 + \mathrm{ID}/d)^{-3}]W_m. \tag{12}$$

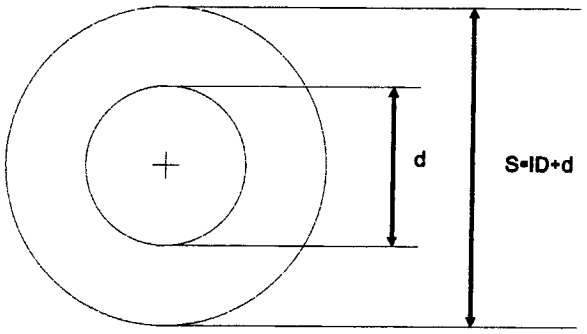

Figure 2 Schematics of stress volume around a dispersed phase particle.

In this case, impact strength will decrease with the increase of particle size, which is shown in Figure 3.

2. If $W_{my} \ll W_d$, eq. (11) can be expressed as

$$G \sim W_d(1 + \mathrm{ID}/d)^{-3}\pi/6. \tag{13}$$

In this case, impact strength will increase with the increase of particle size shown in Figure 4.

3. If $W_{my} = W_d$, eq. (11) can be given by

$$G \sim \pi/6. \tag{14}$$

In this case, impact strength will not change with particle size.

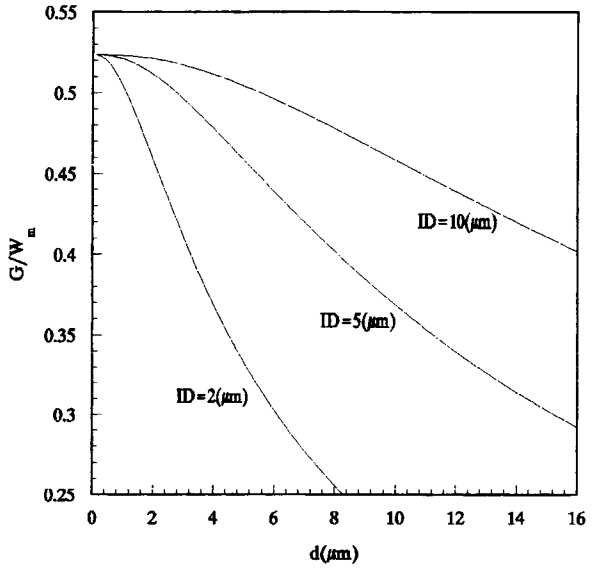

Figure 3 Calculation results of dependence of impact strength on particle size for $W_{my} \gg W_d$.

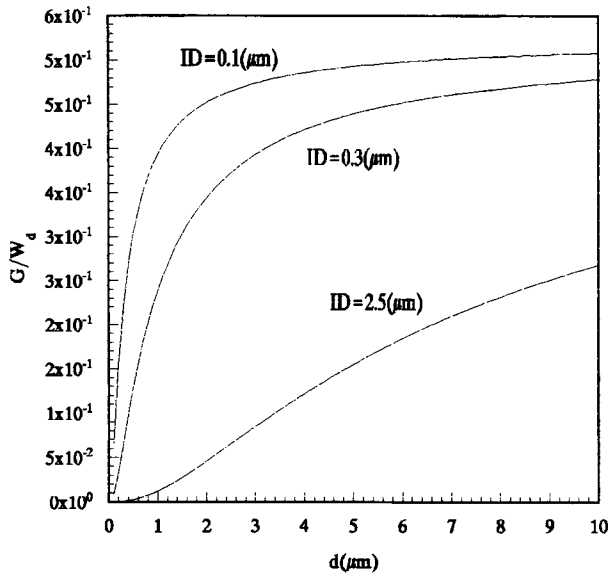

Figure 4 Calculation results of dependence of impact strength on particle size for $W_{my} \ll W_d$.

For the general case, if $W_{my} > W_d$, impact strength will decrease with the increase of particle size; if $W_{my} < W_d$, impact strength will increase with the increase of particle size.

For HDPE/C_aCO_3 blends, although the stress volumes will yield during ductile fracture, it is hard for the C_aCO_3 particle to be deformed during impact fracture because the moduli of C_aCO_3 is much higher than that of HDPE. The energy dissipated by the C_aCO_3 particle is almost zero. So the inequality of $W_{my} \gg W_d$ is tenable. According to eq. (12), impact strength will decrease with the increase of C_aCO_3 particle size for HDPE/C_aCO_3 blends. This is in agreement with the experimental results.

For nylon-66/rubber blends with ductile fracture, although no rubber particles are visible on the fracture surfaces,[1,8] there are more and more experimental results confirming that the rubber particle will be much deformed. And finally cavitation of the rubber particles will occur followed by shear yielding of the matrix[9-11] during impact fracture. According to Figure 2, we proposed that cavitation of rubber exists in every yielding stress volume. Although we do not know the energy dissipated by cavitation of rubber (per unit volume) and the yield energy of nylon-66 (per unit volume) in the impact test, we know the yield energy of nylon-66 is about 2.0 MJ/m^3 and the fracture energy of rubber is more than 16 MJ/m^3 in the tensile test. So the inequality of $W_{my} < W_d$ should be tenable in the impact test. For nylon-66/rubber blends, ID_c = 0.30 μm and d changed from 0.32 to 1.7 μm for the ductile fracture. We can see that it is in this region that impact strength increased rapidly with particle size d in Figure 4.

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REFERENCES

1. S. Wu, *Polymer*, **26**, 1855 (1985).
2. R. J. M. Borggreve, R. J. Gaymans, J. Schuijer, and J. F. Ingen Housz, *Polymer*, **28**, 1489 (1987).
3. S. Wu, *J. Appl. Polym. Sci.*, **35**, 549 (1988).
4. Q. Fu, G. Wang, and J. Shen, *J. Appl. Polym. Sci.*, **49**, 673 (1993).
5. Q. Fu and J. Shen, *J. Appl. Polym. Sci.*, **49**, 1985 (1993).
6. A. Margolina and S. Wu, *Polymer*, **29**, 2170 (1988).
7. S. Wu, *Polymer*, **31**, 972 (1990).
8. S. Wu, *J. Polym. Sci., Polym. Phys. Ed.*, **21**, 699 (1983).
9. R. J. M. Borggreve, R. J. Gaymans, and H. M. Eichenwald, *Polymer*, **30**, 78 (1989).
10. D. S. Parker, H. J. Sue, J. Huang, and A. F. Yee, *Polymer*, **31**, 2267 (1990).
11. D. D. Huang and B. A. Wood, *Polym. Prepr.*, **33**, 629 (1992).

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