

The Damage Evaluation of Rigid Particle Filled Polymer

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ABSTRACT

In this paper, the rigid particle filled polymer is studied in the hope to understand the real damage mechanisms. Two damage parameters were introduced and measured. One is the macro-damage of the materials calculated from the modulus measured, another is micro-damage describing the interfacial debonding or the percentage of the particle debonded from the matrix. The damage rate of the macro damage decreases, while the micro damage increases with the applied stress.

1. INTRODUCTION

The rigid particle filled polymer has received much attention since the first study was made in 1984[1]. Much work was focused on the brittle-ductile transition of such materials [2-4]. The rigid particle can greatly improve the strength and modulus of polymers, and also in some cases the toughness if suitable parameters such as the size, volume fraction of particles and the adhesive strength of the interface etc. are well designed [5,6]. The reinforcing and toughening mechanisms are related to the damage evolution process in the materials. The voids induced by the interfacial debonding for particle filled polymers contribute to a great extent to the damage. If the interfacial adhesion is weak, the interfacial crack will be created early and then developed into void with the continuous application of the load. In this case, the void growth and coalescence will play an important role in the damage evolution of the materials. However, if the interface adhesion is too strong to be debonded, the matrix crazing will take place and predominate in the damage evolution process, the toughening effect can not be achieved. The growth and coalescence of the voids induced by interfacial debonding was studied theoretically [7,8]. For particle filled polymer, the interfacial debonding represents one of the main factors controlling the damage. Much often, the

modulus change is chosen as a parameter to describe the damage level. By theoretical analysis, the modulus change due to the void formation and growth could be predicted [9]. Also by experiments, the modulus variation can be measured as a function of applied stress or strain.

In this paper, an experimental study was carried out on the damage evolution in a particle filled polymer. Both macro cyclic tensile tests and *in situ* tensile tests are carried out. The percentage of particles debonded is measured as a function of applied strain. The variation of Young's modulus is also measured.

2. MATERIALS AND EXPERIMENTAL STUDY

Both pure high density polyethylene (HDPE) and glass bead (GB) filled HDPE are used as the studied materials. In the manufacture process of GB/HDPE composites, the glass beads are pretreated with 10% of γ -aminopropyltriethoxysilane. The mixture of HDPE, HDPE-g-MA and glass beads with weight ratio of 984.54:71.82:143.64 grams is put into the chamber of a high speed mixer, and then blended in a twin-screw extruder. The pelleted extrudate is injection molded into tensile samples. A strong chemical bonding at the interface is formed according to the chemical analysis. The volume fraction of the glass beads is 5%. The micrograph in Fig.1 shows the particle distribution in the matrix. The size distribution of the particles is shown in Fig.2 with the maximum concentration at about 22 μm in diameter.

Both monotonic and cyclic tensile tests are carried out on a MTS810. The geometry of tensile samples refers to the standards ASTM D638, type M. The gage length is 50mm and the strain rate is $3 \times 10^{-5}/\text{s}$. *In situ* tensile tests are undertaken under a scanning electronic microscope (SEM) with the effective sample section of $4 \times 3 \times 1 \text{mm}^3$. During the *in situ* tensile tests, a continuous observation of the damage evolution process is made.

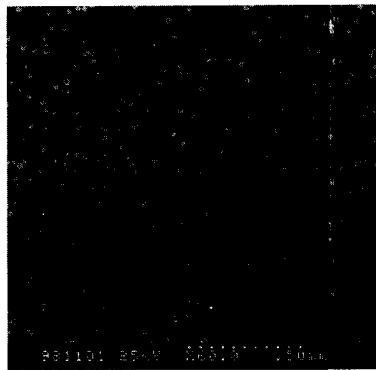


Fig.1. The particle distribution

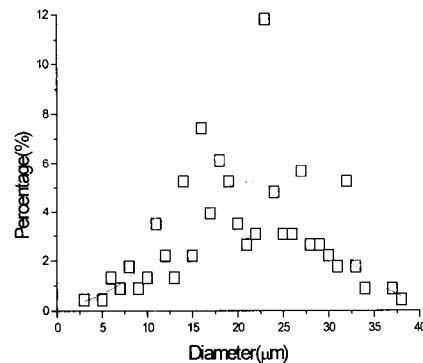


Fig.2. Size distribution of the particles

3. RESULTS AND DISCUSSION

3.1 Macro-behavior under tension

Fig.3 gives the monotonic tensile σ - ϵ curves of GB/HDPE and HDPE. It can be seen that GB/HDPE composite has better mechanical properties than pure HDPE, reflected by 7.4% of

increase in maximum tensile stress and 15.2% in Young's modulus. This result proves the fact that rigid particles can improve the mechanical properties of polymers. Here, the studied materials are too ductile to be ruptured. So the final strain is chosen to be 25 %.

The curves in Fig.4 represent the cyclic process of HDPE and GB/HDPE until 20% of final applied strain. After each load-unload cycle, the slope of unloading σ - ϵ curves is measured. In Fig.5 the modulus change is given as a function of applied maximum stress at each cycle. The GB/HDPE has the same type of modulus variation with HDPE. From the fact that GB volume fraction is low, the mechanical properties of GB/HDPE are mainly controlled by the matrix. Therefore, the mechanical behavior of two materials are similar.

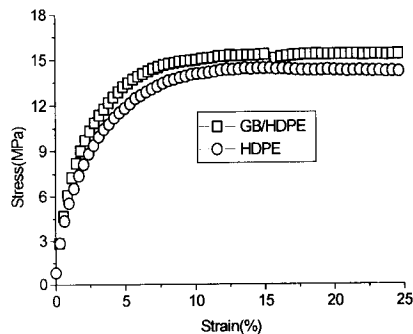


Fig.3. Monotonic tensile curves

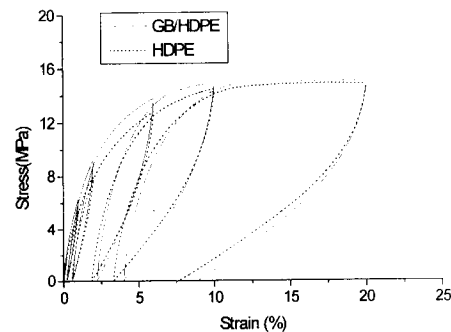


Fig.4. Cyclic tensile curves

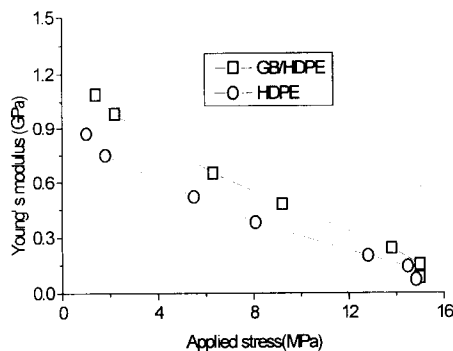


Fig.5. Young's modulus versus σ

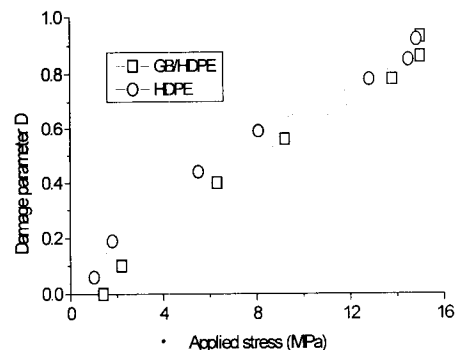


Fig.6. Damage parameter D versus σ

It is noted that with the increasing of applied stress, the difference of Young's modulus between two materials becomes less and less. Owing to the low volume fraction of GB, the voids induced by the interfacial debonding contribute in a limited extent to the modulus. So, the damage in the materials is caused mainly by the matrix microcracking. The interfacial debonding occurred under high external stress. The GBs can still resist the matrix contraction after the interfacial debonding, but the damage in the matrix predominates.

If we use the damage concept, the damage degree of the materials can be expressed by the variation of Young's modulus: $D = 1 - E/E_0$, here, D called damage parameter, E_0 and E are the Young's modulus of intact and damaged materials respectively. The damage degree D as a function of the applied stress is shown in Fig.6. According to D values, under a given stress, HDPE is more damaged than GB/HDPE. At high stress, the damage of two materials is almost same.

3.2-*In situ* tensile tests

In order to observe the damage evolution process for GB/HDPE, the *in situ* tensile tests are carried out under SEM. During the test, the interfacial debonding is observed as shown in Fig.7. The interfacial debonding is initiated at two poles of the GB and the interfacial crack grows into void with the continuous loading. The interface at the top and bottom of GB seems to be in contact. The interfacial debonding represents one of the main damage mechanisms for GB/HDPE. $D^* = V_{\text{m}}/N$ is defined as the damage parameter related to the interface debonding. n and N are the number of the GBs debonded and the total number of the GB in the materials respectively. It is supposed that a GBs debonded can be considered as a void which has the same volume as the GB. In reality, the GB debonded can still resist the transverse constriction of the matrix, so D^* should be considered as the upper limite of the damage induced by the interfacial debonding.

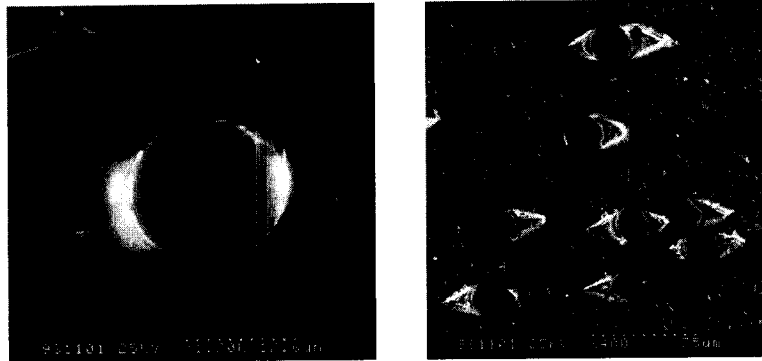


Fig.7. Micrographs showing the interfacial debonding and void formation

The variation of D^* as a function of applied stress is shown in Fig.8. The tendency of this D^* - σ curve is different from that of D - σ curves. According to D^* - σ curve, the damage rate $dD^*/d\sigma$ is more and more great. D^* value is much smaller than D value. In fact, D^* is included in D which is composed of two parts : matrix damage and the interfacial debonding. The matrix damage is mainly the microcrazes created under the stress. With the increasing of GB volume fraction, D^* value will be closer to D value. From D^* - σ and D - σ curves, it is found that the damage rate $dD^*/d\sigma$ increases with the applied strain, while $dD/d\sigma$ decreases. This difference is owing to the different damage mechanisms which are brittle for the interfacial debonding and ductile for the microcrazing of matrix.

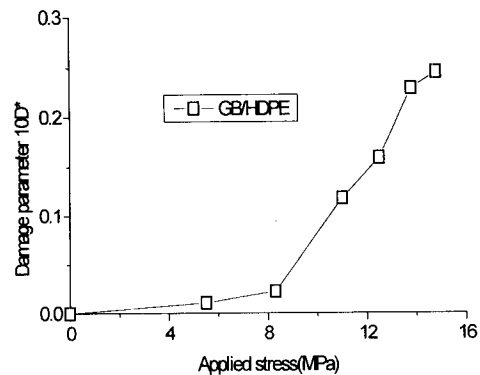


Fig.8. Damage parameters D^* as a function of applied stress

4. CONCLUSIONS

The damage of rigid particle filled polymer materials is studied quantitatively. Two damage parameters are measured. First one D describes the macro-damage evolution of the materials. Second one D^* represents the micro-damage evolution characterized by the interfacial debonding. The damage rate $dD/d\sigma$ decreases, while $dD^*/d\sigma$ increases with the applied stress. This result reflects the different damage mechanisms induced by the interfacial debonding and by the matrix microcrazing.

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REFERENCES

1. T. Kurauchi and T. Ohta, *J. Mater. Sci.*, **19**(1984), P.1699
2. S. Wu, *Polymer*, **26** (1985). P.1855
3. S. Wu and A. Margolina, *Polymer*, **31**(1990), P.971
4. Q. Fu and G. H. Wang, *J. Appl. Polym. Sci.* **49** (1993), P.673
5. K-K. Koo, T. Inoue and K. Miyasaka, *Polymer Engin. Sci.* **25** (1985), P.741
6. J.C. Angola, Y. Fujita, T. Sakai and T.Inoue, *J. Polym. Sci., Polym. Phys.* **26** (1988), P.807
7. Z. P. Huang, J. K. Chen, H. L. Li and Y. Liu, *Proc. IUTAM. Sym. On Rheology of Bodies with Defects*, (Wang R. ed.), Kluwer Academic Publishers, 1998, P.133
8. J. G. Ning and Z. P. Huang, *Progress in Natural Science*, **8**(6)(1998), P.726
9. G. K. Hu, *Int. J. Plasticity*, **12** (1996), P.439

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[2] S. Wu, Polymer, 26 (1985). P.1855

doi:10.1016/0032-3861(85)90015-1

[5] K-K. Koo, T. Inoue and K. Miyasaka, Polymer Engin. Sci. 25 (1985), P.741

doi:10.1002/pen.760251203

[6] J.C. Angola, Y. Fujita, T. Sakai and T.Inoue, J. Polym. Sci., Polym. Phys. 26 (1988), P.807

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