

Modeling of the Tensile Strength of Immiscible Binary Polymer Blends Considering the Effects of Polymer/Polymer Interface and Morphological Variation

Sharifzadeh Esmail

Cite this article as:

Sharifzadeh Esmail. Modeling of the Tensile Strength of Immiscible Binary Polymer Blends Considering the Effects of Polymer/Polymer Interface and Morphological Variation[J]. *Chinese J. Polym. Sci.*, 2019, 37(11): 1176-1182. doi: 10.1007/s10118-019-2274-4

View online: <https://doi.org/10.1007/s10118-019-2274-4>

Articles you may be interested in

[Modeling of the Mechanical Properties of Blend Based Polymer Nanocomposites Considering the Effects of Janus Nanoparticles on Polymer/Polymer Interface](#)

Chinese J. Polym. Sci. 2019, 37(2): 164 <https://doi.org/10.1007/s10118-019-2178-3>

[RELATION BETWEEN INTERFACIAL WAVE INSTABILITY AND TENSILE STRENGTH OF TWO-LAYER POLYMER STRUCTURE: AN EXPERIMENTAL STUDY](#)

Chinese J. Polym. Sci. 2010, 28(6): 887 <https://doi.org/10.1007/s10118-010-9176-9>

[Studies on Droplet Size Distributions during Coalescence in Immiscible Polymer Blends Filled with Silica Nanoparticles](#)

Chinese J. Polym. Sci. 2014, 32(3): 255 <https://doi.org/10.1007/s10118-014-1411-3>

[反应性杂化纳米粒子的合成及其对不相容共混物的增容研究](#)

Synthesis of Reactive Hybrid Nanoparticles and Their Compatibilization Effects on Immiscible Polymer Blends

高分子学报. 2017(2): 334 <https://doi.org/10.11777/j.issn1000-3304.2017.16288>

[聚合物熔体冷却阶段结晶行为的多尺度模拟](#)

MULTISCALE MODELING AND SIMULATION OF POLYMER CRYSTALLIZATION IN COOLING STAGE

高分子学报. 2010(6): 653 <https://doi.org/10.3724/SP.J.1105.2010.09202>

[高浓度不相容聚合物共混体系相形态与流变性能研究](#)

PHASE MORPHOLOGY AND RHEOLOGICAL PROPERTIES OF CONCENTRATED IMMISCIBLE POLYMER BLENDS

高分子学报. 2010(1): 114 <https://doi.org/10.3724/SP.J.1105.2010.00114>

Modeling of the Tensile Strength of Immiscible Binary Polymer Blends Considering the Effects of Polymer/Polymer Interface and Morphological Variation

Esmail Sharifzadeh*

Polymer Research Center, Faculty of Petroleum and Chemical Engineering, Razi University, Kermanshah, Iran

Abstract In this work, a unique model is proposed for predicting the tensile strength of binary polymer blends considering the effects of polymer/polymer interface and the morphological variation of the system. The modeling was performed based on the combination of analytical and artificial neural network (ANN) modeling methods. For the analytical part, Kolarik's model was developed in accordance with the system requirements and ANN was simultaneously involved in order to interpret some effective model parameters using the tensile test result of an actual sample (*e.g.* the yield strength and thickness of the interface, *etc.*). Furthermore, the model accuracy was evaluated by comparing the tensile test results of differently prepared iPP/PA and PS/PMMA blend samples and also some other data from literature with the model predictions. It was revealed that the designed ANN perfectly elevates the capability of the analytical section in order to predict the tensile strength of binary polymer blends with different compositions (prediction error < 10%).

Keywords Polymer blends; Tensile strength; Analytical modeling; Artificial neural networks; Polymer/polymer interface

Citation: Sharifzadeh, E. Modeling of the tensile strength of immiscible binary polymer blends considering the effects of polymer/polymer interface and morphological variation. *Chinese J. Polym. Sci.* 2019, 37, 1176–1182.

INTRODUCTION

Improving the mechanical and physical properties of an individual polymer phase *via* blending processes and adding nanoparticles is quite well-known.^[1–10] However, the presence of many involving parameters in such processes (*e.g.* random orientation of nanoparticles, properties of the polymer/nanoparticle interphase, properties of the polymer/polymer interface, blend morphology, *etc.*) makes it quite challenging to control the properties of the final products.^[11] As a result, mathematical modeling has always been considered as an efficient method for reducing unnecessary costs.^[12–16] Focusing on the immiscible binary polymer blends, there are many practical models which are capable of providing reliable predictions for the mechanical properties (*e.g.* tensile modulus).^[16–19] The improvement of modeling procedures during the last decades has been increasingly elevating the prediction accuracy and reliability due to the consideration of more detailed parameters in the models.^[16–19] Accordingly, for polymer blends, the morphological variation (as a result of the increasing content of the dispersed phase) and the formation of polymer/polymer interface can be considered as the most effective parameters in determination of the mechanical/physical properties.^[20,21] The morphological variation

can be simply categorized into four intervals using percolation thresholds and phase inversion point, but it is essential to consider the polymer/polymer interface as a specified part of the main geometrical structure of the model.^[5,17,20] This makes it possible to evaluate the performance of the interface as well as its effects on the reaction of other model parts against the exerted stress. Similarly, the crucial role of the polymer/polymer interface in analytically modeling the tensile modulus of certain blend-based polymer nanocomposites has been previously proven.^[5]

Modeling of the tensile strength of polymer blends is a quite challenging field of research where the drastic effects of polymer/polymer interface on the final predicted results are inevitable. Considering either strong or weak reaction of the interface against the exerted stress in a binary polymer blend, there are some well-known models which are capable of predicting the final tensile strength.^[16,22–26] However, the deficient consideration about the effects of morphological variation in such models raises the question that how the exerted stress acts on polymer/polymer interface while it is also changing with the blend composition.

Accordingly, in this study, a new modeling procedure as a combination of analytical and neural network (ANN) methods was used in order to predict the tensile strength of binary polymer blends. The analytical modeling was performed based on developing Kolarik's model so as to reach a comprehensive geometrical structure in which the morphological variation and polymer/polymer interface region were simul-

* Corresponding author: E-mail E_sharifzadeh@sut.ac.ir

Received March 5, 2019; Accepted April 11, 2019; Published online June 4, 2019

taneously considered.^[5,16,19] Furthermore, the designed ANN was appended to the procedure in order to interpret some effective model parameters using the tensile test result of an actual sample. It is noteworthy to mention that in a few studies, ANNs have been limitedly used to interpret the mechanical properties (mainly tensile modulus) of some specific polymer blends.^[27–30]

The accuracy of the model was evaluated based on comparing the tensile test result of different prepared PA/iPP and PS/PMMA blend samples (with different compositions) with the model predictions. It was revealed that the coupling of analytical and ANN modeling methods works perfectly and the prediction error is less than 10%.

MODELING

Considering an equivalent box model (EBM) corresponding to the structure of a binary polymer blend, Kolarik *et al.* proposed the following equation for predicting the tensile strength of the system (S_{yb}):^[19]

$$S_{ub} = (\phi_{1p}S_{u1} + \phi_{2p}S_{u2}) + A\phi_s S_{u1} \quad (1)$$

where $S_{u1} < S_{u2}$, ϕ_{jp} denotes the volume fractions of the parallel section of phase 1 (or 2) in the EBM model, ϕ_s is the total volume fraction of the series section of the EBM model, S_{uj} is the tensile strength of phase 1 (or 2) and A is the extent of interfacial debonding. It is clear that if parameter $A = 1$, the tensile strength of series section of phase 1 (as the weaker phase) affects S_{ub} while $A = 0$ means interface failure and therefore it is just the parallel section which reacts against the exerted stress. Considering a polymer/polymer interface region, Kolarik's COS model can be reformed as a new geometrical structure in which the morphological variation is also involved (Fig. 1).^[16,19]

The presence of both phases in the interface region makes it crucial to consider its role in determining the final tensile strength of the system.^[31] Accordingly, Eq. (1) can be reformed based on the geometry of the model structure at different intervals and the phase inversion point.

I. Phase Inversion Point (ϕ_{PI}) (Fig. 2)

$$S_{ub|PI} = (\phi_{1p}S_{u1} + \phi_{2p}S_{u2} + \phi_{In,p}S_{ui}) + A(S_{u1} + S_{ui})\phi_s \quad (2)$$

$$\square \begin{cases} \phi_{2s} = 2\pi r_p^2 - 2(16 - \sqrt{128})r_p^3; \\ \phi_{1s} = 4(R - R^2) - (3\pi R^2 - 2(16 - \sqrt{128})R^3); \\ \phi_{In,s} = 2\pi R^2 - 2(16 - \sqrt{128})R^3 - \phi_{2s}; \\ \phi_s = \phi_{2s} + \phi_{1s} + \phi_{In,s}; \\ \phi_{In,p} = \pi(R^2 - r_p^2); \\ \phi_{2p} = \pi r_p^2; \\ \phi_{1p} = 1 - (\phi_{2s} + \phi_{1s} + \phi_{In,s} + \phi_{2p} + \phi_{In,p}); \end{cases}$$

where S_{ui} denotes the tensile strength of the interface region.

It is considered that a very small percentage of each polymer phase (y_p) participates in the polymer/polymer interface region.^[5,32,33] Therefore,

$$\bar{\phi}'_2 = \bar{\phi}_2 \times y_p \quad (3)$$

$$\bar{\phi}'_1 = \bar{\phi}_1 \times y_p \quad (4)$$

where $\bar{\phi}'_1$ and $\bar{\phi}'_2$ are the volume fractions of the participant portions of Phase (I) and Phase (II) in the interface region, respectively. $\bar{\phi}_1$ and $\bar{\phi}_2$ denote the actual (experimental) volume fractions of Phase (I) and Phase (II), respectively. It is now possible to calculate parameter r_p using ϕ_{2s} and ϕ_{2p} as follows:

$$\begin{cases} f(r_p) = \phi_{PI}(1 - y_p) - 3\pi r_p^2 + 2(16 + \sqrt{128})r_p^3 = 0 \\ f(r_p) \rightarrow r_p \end{cases} \quad (5)$$

where ϕ_{PI} is the volume fraction of Phase (II) corresponding to the phase inversion point. Similarly, parameter R can be calculated using Eq. (6):

$$\begin{cases} f(R, r_p) = \phi_{In,s} + \phi_{In,p} - (\bar{\phi}'_1 + \bar{\phi}'_2) = 0 \\ f(R, r_p) \rightarrow R \end{cases} \quad (6)$$

As illustrated in Fig. 2, the thickness of polymer/polymer

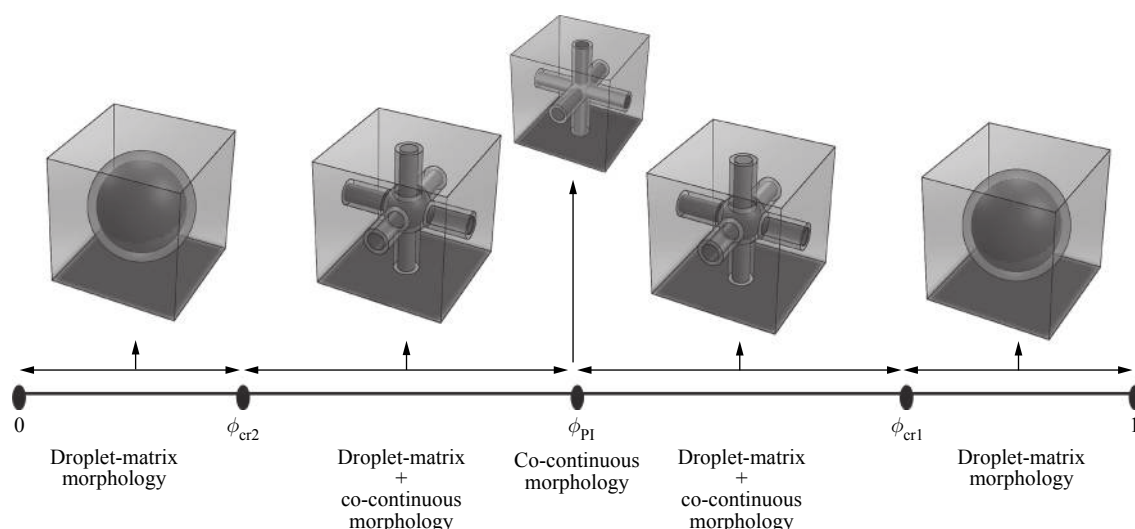


Fig. 1 Geometrical structure of the model corresponding to the actual blend morphology in different intervals plus phase inversion point^[5]

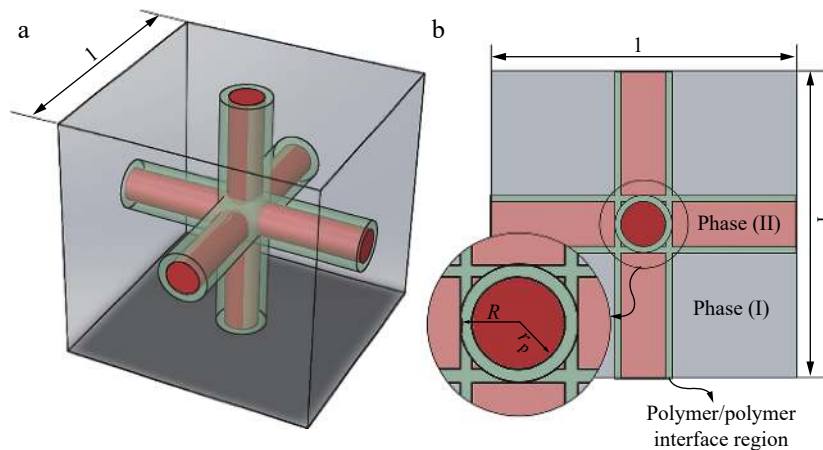


Fig. 2 (a) Geometrical structure of the model at phase inversion point; (b) Side view of the structure (The whole structure is surrounded by a cube of unite side length.^[5])

interface region (T) is equal to $R - r_p$ which is constant and independent of the volume fraction of polymer phases.

II. The Second Interval ($\phi_{cr2} \leq \bar{\phi}_2 < \phi_{PI}$) and the Third Interval ($\phi_{PI} < \bar{\phi}_2 \leq 1 - \phi_{cr1}$) (Fig. 3)

$$S_{ub|2,3} = (\phi_{2p}S_{u2} + \phi_{In,p}S_{ui}) + A(S_{u1} + S_{ui})\phi_s + S_{u1}\phi_c + AS_{ui}\phi_d \quad (7)$$

$$\square \left\{ \begin{array}{l} \phi_{2p} = \pi r_{23}^2; \\ \phi_{In,p} = \pi(R^2 - r_{23}^2); \\ \phi_{2s} = 2 \left(\pi r_{23}^2 - \left[\frac{4}{3} \pi (R_s + T)^3 - 2\pi h \left((R_s + T)^2 - \frac{(R_s \cos(\arcsin(\frac{r_{23}}{R_s + T})))}{3} - r_{23} \right) \right] \right); \\ \Delta \left\{ \begin{array}{l} \phi_{1s} = 4 \left(2(r_{23} + T)(0.5 - h) - \frac{\phi_{2s}}{4} - \frac{\pi h^2}{3} (3(R_s + T) - h) \right) \\ \text{where } h = (R_s + T) \left(1 - \cos \left(\arcsin \left(\frac{r_{23} + T}{R_s + T} \right) \right) \right) \end{array} \right. \\ \phi_{In,s} = 2 \left(\pi(r_{23} + T)^2 - \left(2\pi h_1(R_s + T)^2 - \frac{h_1^2}{3} - (r_{23} + T) + 2\pi h_1(R_s + T)^2 - \frac{h_2^2}{3} - r_{23} \right) \right) \\ \Delta \left\{ \begin{array}{l} \text{where } \left\{ \begin{array}{l} h_1 = (R_s + T) \cos \left(\arcsin \left(\frac{r_{23} + T}{R_s + T} \right) \right) \\ h_2 = (R_s + T) \cos \left(\arcsin \left(\frac{r_{23}}{R_s + T} \right) \right) \end{array} \right. \end{array} \right. \\ \phi_s = \phi_{2s} + \phi_{1s} + \phi_{In,s}; \\ \phi_{2d} = \frac{4}{3} \pi (R_s + T)^2 - (r_{23} + T)^2; \\ \phi_{1c} = 1 - (\phi_{2d} + \phi_{In,s} + \phi_{1s} + \phi_{2s} + \phi_{In,p} + \phi_{2p}); \end{array} \right.$$

where, r_{23} and R_s are the model parameters at 2nd and 3rd intervals (Fig. 3b).

Parameter r_{23} can be calculated using r_p (Eq. 5) as follows:

$$r_{23} = r_p \times R(D); \quad R(D) = \frac{\bar{\phi}_2 - (\bar{\phi}'_2 + \phi_{cr2})}{\phi_{PI}(1 - y_p) - \phi_{cr2}} \quad (8)$$

where $R(D)$ is a developed form of the parameter used by Wang *et al.* in order to make a correlation between their proposed model structures for the 2nd/3rd interval and phase inversion point.^[15,20] Considering constant T (Eq. 6), it is possible to determine R_s using Equations (ϕ_{2p}), (ϕ_{2s}), and (ϕ_{2d}):

$$\left\{ \begin{array}{l} f(R_s, r_{23}) = \bar{\phi}_2 - \bar{\phi}'_2 - (\phi_{2p} + \phi_{2s} + \phi_{2d}) = 0 \\ f(R_s, r_{23}) \rightarrow R_s \end{array} \right. \quad (9)$$

III. The First Interval ($0 \leq \bar{\phi}_2 \leq \phi_{cr2}$) and the Fourth Interval ($1 - \phi_{cr1} < \bar{\phi}_2 \leq 1$) (Fig. 4)

$$S_{ub|1,4} = S_{u1}\phi_c + AS_{ui}\phi_d \quad (10)$$

$$\square \left\{ \begin{array}{l} \phi_{In.} = \frac{4}{3} \pi (R_s + T)^3 - R_s^3 \\ \phi_d = \phi_{In.} + \left(\frac{4}{3} \pi R_s^3 \right) (1 - y_p) \\ \phi_c = \left(1 - \left(\frac{4}{3} \pi R_s^3 - \phi_{In.} \right) \right) (1 - y_p) \end{array} \right.$$

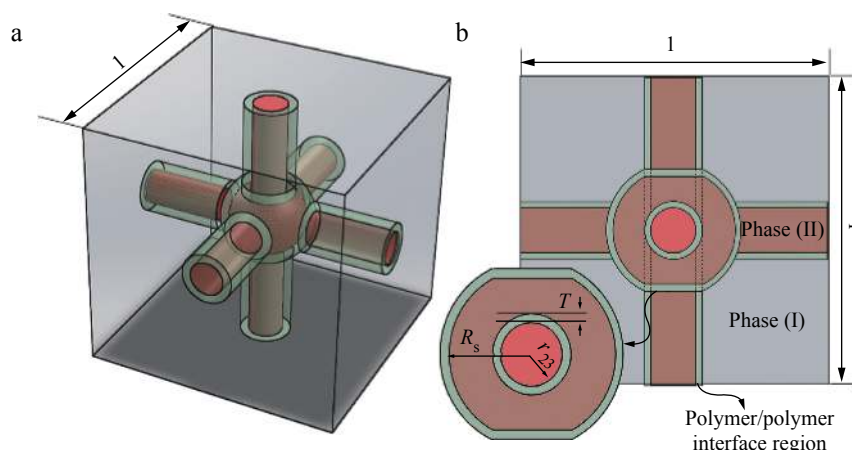


Fig. 3 (a) Geometrical structure of the model in the second and the third intervals; (b) Side view of the structure (The whole structure is surrounded by a cube of unite side length.^[5])

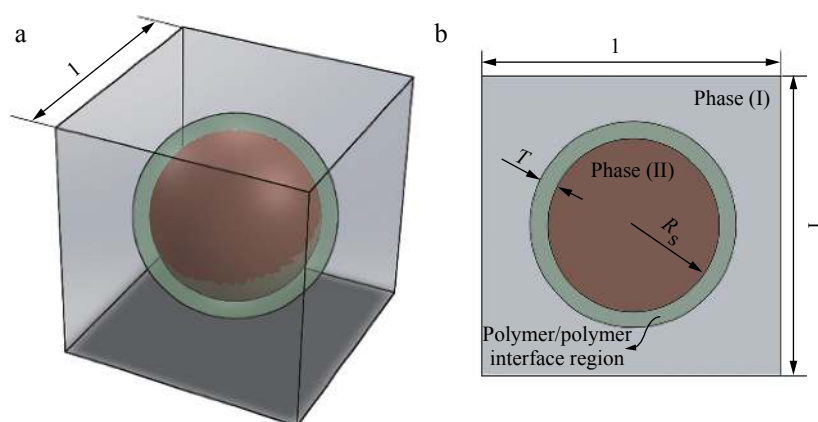


Fig. 4 (a) Geometrical structure of the model in the first and the fourth intervals; (b) Side view of the structure (The whole structure is surrounded by a cube of unite side length.^[5])

The model parameter R_s can be determined using Eq. (11):

$$\bar{\phi}_2 - \bar{\phi}'_2 = \frac{4}{3}\pi R_s^3 \quad (11)$$

where, $\bar{\phi}_2$ denotes the actual (experimental) volume fractions of Phases (II),

In the next step, the designed ANN can be applied in order to interpret parameters “ A ”, “ S_{ui} ”, and “ y_p ” using an actual blend samples in the 1st (or 4th) interval. Based on the findings of Kolarik *et al.*, parameter A can be considered as a constant value between 0 and 1. Furthermore, it is possi-

ble to consider $S_{u2} - S_{u1} \leq 2S_{ui} \leq S_{u2} + S_{u1}$ and $0 \leq y_p \leq 0.5$ (which is completely reliable based on our previously reported findings).^[5,16]

Fig. 5 represents the schematics of the designed ANN. Considering error < 5%, the network trains itself in order to produce the best result with the least error, which may require the network to be trained multiple times. At the final step, the results (S_{yi} , y_p , and A) can be tested using Eq. (10) in order to define the best set with the least error. It is important to mention that, the designed ANN should not be used as a predictive tool since it just interprets parameters

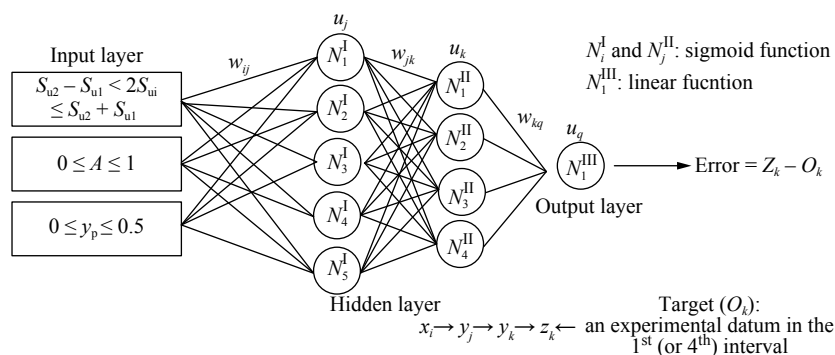


Fig. 5 Schematic of the structure of the designed ANN (If |error| > 5%, the training procedure should be repeated.)

“ A ”, “ S_{ui} ” and “ y_p ” using the tensile test result of a specific sample.

EXPERIMENTAL

Materials

The applied polymers for preparing blend samples were polyamide (PA, Akulon F223D, DSM), isotactic polypropylene (*i*PP, Polynar Petrochemical Co., Tabriz, Iran), polystyrene (PS, 1160 GPPS, Tabriz Petrochemical Co.), and

poly(methyl methacrylate) (PMMA, EG920, LG). To avoid the effect of moisture, all the materials were dried in a vacuum oven at 80 °C for 12 h prior to melt mixing. All the blends with 10 vol%, 20 vol%, 30 vol%, 40 vol%, 50 vol%, 70 vol%, and 90 vol% as the total volume fraction of dispersed phase were prepared in an internal mixer (Brabender Plasticorder W50EHT) with a rotor speed of 60 r/min at 230 and 200 °C (for *i*PP/PA and PS/PMMA, respectively) for 8 min. The obtained compounds were compression molded into suitable pieces for tensile tests. Molding was carried out

Table 1 Yield strength (MPa) of different *i*PP/PA and PS/PMMA blend samples

Sample *	Volume fraction of dispersed phase								
	0	0.1	0.2	0.3	0.4	0.5	0.7	0.9	1
PS/PMMA	43.8	42.6	47.3	46.4	43.24	51.93	57.47	59.11	62.4
<i>i</i> PP/PA	31.8	34.1	37.8	39.2	47.3	52.1	70.7	78.2	84.7

* The latter item (PMMA, PA) refers to the dispersed phase in each sample.

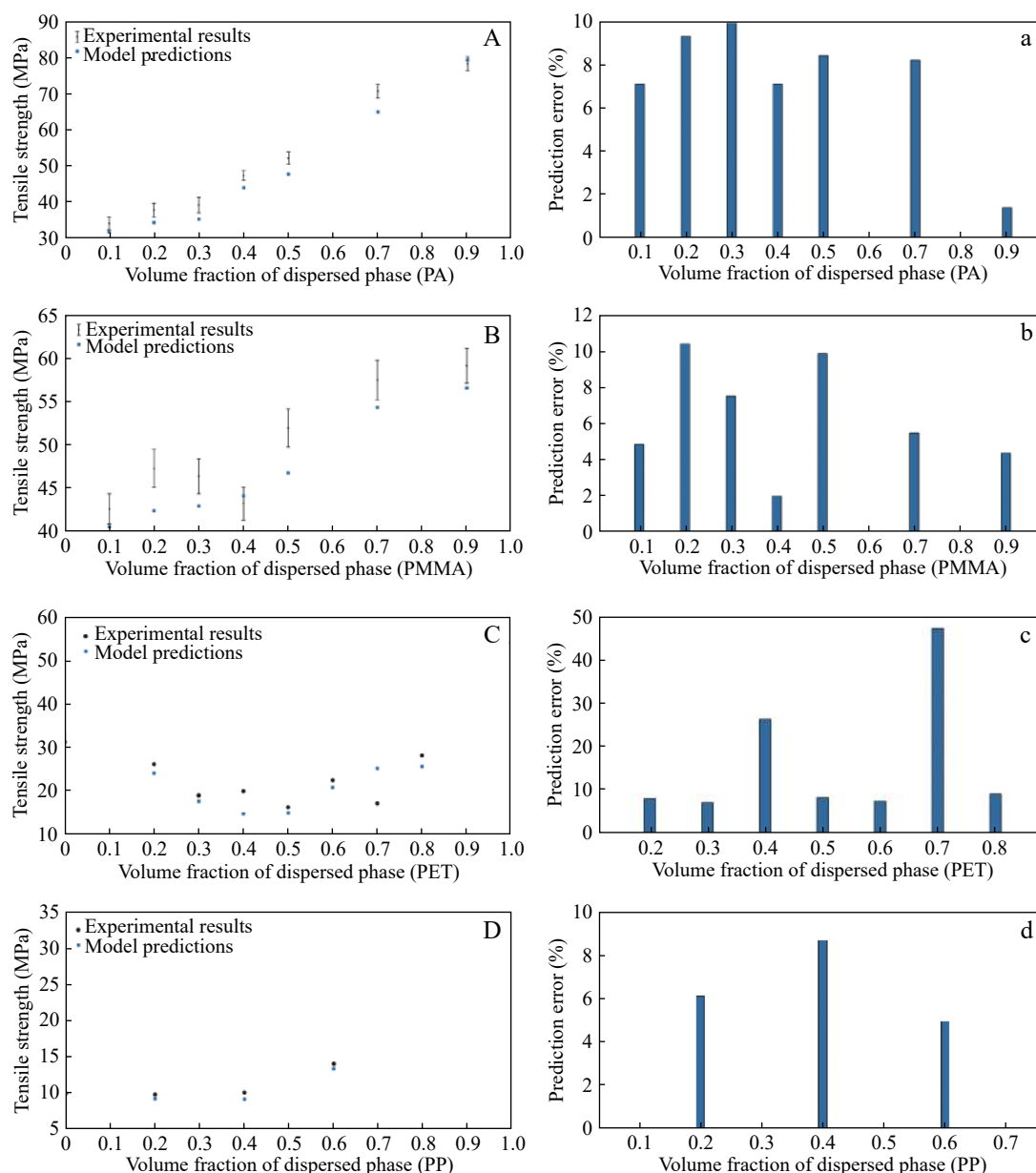


Fig. 6 The comparison of model predictions with the experimental data of (A) *i*PP/PA, (B) PS/PMMA, (C) PP/PET, and (D) LDPE/PP blend samples with the corresponding prediction error represented in (a), (b), (c), and (d), respectively^[5,34,35]

at 230 and 200 °C (for *i*PP/PA and PS/PMMA, respectively) followed by slow water-cooling under 10 MPa.

Mechanical Properties

Tensile tests were conducted on a Zwick/Roell tensile testing machine (Z 010) at a fixed crosshead speed of 5 mm/min at room temperature according to ISO 527. At least five specimens were tested for each composition and the results were averaged. Table 1 lists the yield strength (S_u) of the tested samples.

RESULTS AND DISCUSSION

Fig. 6 compares the experimental results with the model predictions. It should be mentioned that the immiscibility of the constituents of all blend samples is previously proven.^[11,34–36] Clearly, the combination of analytical and ANN methods was completely successful in approximating the tensile strength of different binary polymer blend systems (error < 10%). However, the presence of dispersed phase in the bulk of continuous phase (in immiscible blends) may cause stress concentration points that negatively affect the tensile strength of the system.^[37,38] This is the reason why in some cases the experimental data and/or the model predictions of a blend sample give lower values compared to the tensile strength of the neat continuous phases. It should be mentioned that the high value of prediction error for some of the PP/PET blend samples (Figs. 6c) is most likely due to the inaccuracy of the corresponding experimental data.

As mentioned before, it is assumed that y_p (%) of each polymer phase participates in forming the interface and accordingly the interface relating parameters y_p , A , and S_{ui} were defined using the designed ANN (Table 2). As clearly shown, parameters A and S_{ui} of *i*PP/PA blend are greater than those of PS/PMMA while *i*PP/PA has smaller y_p . In the case of PP/PET blend samples (Fig. 6c), the presence of the dispersed phase weakens the systems (formation of stress concentration points), which, based on the model results, is due to the very low value of parameter A (Table 2).

Table 2 The result of ANN for *i*PP/PA, PS/PMMA, PP/PET, and LDPE/PP blend samples^[34,35]

Blend	y_b (%)	A	S_{ui} (MPa)
PS/PMMA	0.34	0.214	49.7
<i>i</i> PP/PA	0.18	0.535	56.17
PP/PET	0.03	0.037	41.1
LDPE/PP	0.12	0.108	23.89

CONCLUSIONS

A combination of analytical and ANN modeling methods was used to predict the tensile strength of binary polymer blends. The analytical section was developed based on Kolarik's model, emphasizing on drastic effects of the interface on system reaction against the exerted stress. On the other hand, an ANN was designed to interpret parameters A , y_b , and S_{ui} as the effective model parameters, using the tensile test result of an actual sample. As reported, the model predictions were very close to the tensile test results (as an

advantage of applying ANN) and final prediction errors were completely acceptable (error < 10%) (Fig. 6).

REFERENCES

- Gomari, S.; Ghasemi, I.; Karrabi, M.; Azizi, H. Organoclay localization in polyamide 6/ethylene-butene copolymer grafted maleic anhydride blends: The effect of different types of organoclay. *J. Polym. Res.* **2011**, *19*, 9769.
- Fathi, A.; Lee, S.; Breen, A.; Shirazi, A. N.; Valtchev, P.; Dehghani, F. Enhancing the mechanical properties and physical stability of biomimetic polymer hydrogels for micro-patterning and tissue engineering applications. *Eur. Polym. J.* **2014**, *59*, 161–170.
- Robeson, L. M. *Polymer blends: A comprehensive review*. Hanser, **2007**.
- Ezzati, P.; Ghasemi, I.; Karrabi, M.; Azizi, H.; Fortelny, I. Preparation of porous PLLA/PCL blend by a combination of PEO phase and NaCl particulate leaching in PLLA/PCL/PEO/NaCl blend. *Iran Polym. J.* **2014**, *23*, 757–766.
- Sharifzadeh, E. Modeling of the mechanical properties of blend based polymer nanocomposites considering the effects of janus nanoparticles on polymer/polymer interface. *Chinese J. Polym. Sci.* **2019**, *37*, 164–177.
- Minaei-Zaim, M.; Ghasemi, I.; Karrabi, M.; Azizi, H. Effect of injection molding parameters on properties of cross-linked low-density polyethylene/ethylene vinyl acetate/organoclay nanocomposite foams. *Iran Polym. J.* **2012**, *21*, 537–546.
- Shonaike, G. O.; Simon, G. P. *Polymer Blends and Alloys*. Taylor & Francis, **1999**.
- Dasari, A.; Yu, Z. Z.; Mai, Y. W. *Polymer nanocomposites: Towards multi-functionality*. Springer London, **2016**.
- Gupta, R. K.; Kennel, E.; Kim, K. J. *Polymer nanocomposites handbook*. CRC Press, **2009**.
- Utracki, L. A. *Commercial polymer blends*. Springer US, **2013**.
- Sharifzadeh, E.; Salami-Kalajahi, M.; Hosseini, M. S.; Aghjeh, M. K. R. Synthesis of silica Janus nanoparticles by buoyancy effect-induced desymmetrization process and their placement at the PS/PMMA interface. *Colloid. Polym. Sci.* **2017**, *295*, 25–36.
- Chen, S.; Sarafbidabad, M.; Zare, Y.; Rhee, K. Y. Estimation of the tensile modulus of polymer carbon nanotube nanocomposites containing filler networks and interphase regions by development of the Kolarik model. *RSC Adv.* **2018**, *8*, 23825–23834.
- Zare, Y. A Two-step method based on micromechanical models to predict the Young's modulus of polymer nanocomposites. *Macromol. Mater. Eng.* **2016**, *301*, 846–852.
- Zare, Y.; Rhee, K. Y. Development of a model for electrical conductivity of polymer/graphene nanocomposites assuming interphase and tunneling regions in conductive networks. *Ind. Eng. Chem. Res.* **2017**, *56*, 9107–9115.
- Sharifzadeh, E.; Ghasemi, I.; Karrabi, M.; Azizi, H. A new approach in modeling of mechanical properties of binary phase polymeric blends. *Iran Polym. J.* **2014**, *23*, 525–530.
- Kolarik, J. Three-dimensional models for predicting the modulus and yield strength of polymer blends, foams, and particulate composites. *Polym. Compos.* **1997**, *18*, 433–441.
- Sharifzadeh, E.; Ghasemi, I.; Safajou-Jahankhanemlou, M. Modulus prediction of binary phase polymeric blends using symmetrical approximation systems as a new approach. *Iran Polym. J.* **2015**, *24*, 735–746.
- Ji, X. L.; Jing, J. K.; Jiang, W.; Jiang, B. Z. Tensile modulus of polymer nanocomposites. *Polym. Eng. Sci.* **2002**, *42*, 983–993.
- Kolarik, J. Simultaneous prediction of the modulus and yield strength of binary polymer blends. *Polym. Eng. Sci.* **1996**, *36*, 2518–2524.
- Wang, J. F.; Carson, J. K.; North, M. F.; Cleland, D. J. A knot-

- ted and interconnected skeleton structural model for predicting Young's modulus of binary phase polymer blends. *Polym. Eng. Sci.* **2010**, *50*, 643–651.
- 21 Sharifzadeh, E.; Ghasemi, I.; Qarebagh, A. N. Modeling of blend-based polymer nanocomposites using a knotted approximation of Young's modulus. *Iran Polym. J.* **2015**, *24*, 1039–1047.
 - 22 Werner, A.; Schmid, F.; Müller, M.; Binder, K. In *Interfaces in immiscible polymer blends: A Monte Carlo simulation approach on the CRAY T3E*, Berlin, Heidelberg, Springer Berlin Heidelberg: Berlin, Heidelberg, **1999**, pp. 176–185.
 - 23 Kar, G. P.; Biswas, S.; Bose, S. Tailoring the interface of an immiscible polymer blend by a mutually miscible homopolymer grafted onto graphene oxide: Outstanding mechanical properties. *Phys. Chem. Chem. Phys.* **2015**, *17*, 1811–1821.
 - 24 Oslanec, R.; Brown, H. R. Entanglement density at the interface between two immiscible polymers. *Macromolecules* **2003**, *36*, 5839–5844.
 - 25 Li, J.; Ma, P. L.; Favis, B. D. The role of the blend interface type on morphology in cocontinuous polymer blends. *Macromolecules* **2002**, *35*, 2005–2016.
 - 26 Pukánszky, B.; Tüdös, F. Miscibility and mechanical properties of polymer blends. *Macromolecular Symposia* **1990**, *38*, 221–231.
 - 27 Yousef, B. F.; Mourad, A. H. I.; Hilal-Alnaqbi, A. Prediction of the mechanical properties of PE/PP blends using artificial neural networks. *Procedia Eng.* **2011**, *10*, 2713–2718.
 - 28 Yousef, B. F.; Mourad, A. H. I.; Hilal-Alnaqbi, A. Modeling of the mechanical behavior of polyethylene/polypropylene blends using artificial neural networks. *Int. J. Adv. Manuf. Technol.* **2013**, *64*, 601–611.
 - 29 Balasubramanian, M.; Paglicawan, M. A.; Zhang, Z. X.; Lee, S. H.; Xin, Z. X.; Kim, J. K. Prediction and optimization of mechanical properties of polypropylene/waste tire powder blends using a hybrid artificial neural network-genetic algorithm (GA-ANN). *J. Thermoplast. Compos. Mater.* **2008**, *21*, 51–69.
 - 30 Roy, N. K.; Potter, W. D.; Landau, D. P. Designing polymer blends using neural networks, genetic algorithms, and markov chains. *Appl. Intell.* **2004**, *20*, 215–229.
 - 31 Werner, A.; Schmid, F.; Müller, M.; Binder, K. In *Interfaces in immiscible polymer blends: A Monte Carlo simulation approach on the CRAY T3E*, High performance computing in science and engineering '98, Krause, E.; Jäger, W., Eds. Springer Berlin Heidelberg, Berlin, Heidelberg, **1999**, pp. 176–185.
 - 32 Dong, B.; Guo, R.; Yan, L. T. Coassembly of janus nanoparticles in asymmetric diblock copolymer scaffolds: Unconventional entropy effect and role of interfacial topology. *Macromolecules* **2014**, *47*, 4369–4379.
 - 33 Chen, P.; Yang, Y.; Dong, B.; Huang, Z.; Zhu, G.; Cao, Y.; Yan, L. T. Polymerization-induced interfacial self-assembly of Janus nanoparticles in block copolymers: Reaction-mediated entropy effects, diffusion dynamics, and tailorable micromechanical behaviors. *Macromolecules* **2017**, *50*, 2078–2091.
 - 34 Teh, J. W. Structure and properties of polyethylene-polypropylene blend. *J. Appl. Polym. Sci.* **1983**, *28*, 605–618.
 - 35 Bataille, P.; Boissé, S.; Schreiber, H. P. Mechanical properties and permeability of polypropylene and poly(ethylene terephthalate) mixtures. *Polym. Eng. Sci.* **1987**, *27*, 622–626.
 - 36 Menyhard, A.; Varga, J.; Liber, Á.; Belina, G. Polymer blends based on the β -modification of polypropylene. *Eur. Polym. J.* **2005**, *41*, 669–677.
 - 37 Bartczak, Z.; Galeski, A. Mechanical properties of polymer blends. In *Polymer Blends Handbook*, Utracki, L. A.; Wilkie, C. A., Eds. Springer Netherlands, Dordrecht, **2014**, pp. 1203–1297.
 - 38 Utracki, L. A.; Wilkie, C. A. *Polymer blends handbook*. Springer Netherlands, **2014**.