Enhancement of Dielectric Performance of Polymer Composites via Constructing BaTiO$_3$–Poly(dopamine)–Ag Nanoparticles through Mussel-Inspired Surface Functionalization

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ABSTRACT: We demonstrate the synthesis of a novel core–satellite-structured BaTiO$_3$–poly(dopamine)–silver (BT–PDA–Ag) nanoparticle for improving dielectric properties of nitrile-butadiene rubber (NBR) nanocomposites. The BT–PDA–Ag nanoparticles are synthesized by dopamine oxidative polymerization and electroless plating of silver. The Ag nanoparticles decorated on the BT nanoparticles enhanced the dielectric constant of NBR nanocomposites due to the increased conductivity of the filler/matrix interlayer and nanocapacitor structure. In addition, the incorporation of the BT nanoparticles prevented the continuous connection of Ag nanoparticles and suppressed the formation of a conductive path in the NBR matrix. Moreover, the ultrasmall Ag nanoparticles trapped the carriers by Coulomb blockade and quantum confinement effects, which results in low dielectric loss and electrical conductivity of nanocomposites. The proposed method with simplicity and scalability can be adapted to process high-dielectric polymer nanocomposites.

1. INTRODUCTION

Recently, with the tendency of miniaturization and multifunctionalization of electronics, flexible polymer-based materials with tunable properties have garnered significant research focus.1−3 Particularly, the polymer nanocomposites with high dielectric constant, low dielectric loss, high electric breakdown strength, and high energy density are highly desirable for high-speed integrated circuits.4 A large number of approaches have been adopted to attain this goal. The most commonly used method is adding either high-dielectric-constant ceramic fillers, such as BaTiO$_3$,5,6 and TiO$_2$,7,8 or conductive fillers, such as graphene,9,10 carbon nanotubes,11,12 and metallic nanoparticles,13,14 into the polymeric matrix to synthesize two-phase polymer-based nanocomposites. However, the high-dielectric-constant ceramic/polymer nanocomposites suffer from the high concentration of rigid ceramic particles, which induces agglomeration of nanoparticles and results in poor mechanical properties and reduced flexibility.15,16 On the other hand, the incorporation of conductive nanoparticles increases the dielectric constant by several orders of magnitude when the filler concentration reaches the percolation threshold.17 However, the poor compatibility between conductive particles and the polymeric matrix results in aggregation and formation of numerous local conductive paths, leading to a high dielectric loss and a low electric breakdown strength. This hinders the dielectric application of these nanocomposites in electronic devices.18

To overcome these problems, a number of studies have been carried out to prevent the aggregation and form a homogeneous dispersion of nanoparticles in the polymeric matrix.19,20 One efficient strategy is the surface functionalization of nanoparticles in the polymer nanocomposites, which improves the compatibility and homogeneous dispersion. The surface functionalization can minimize the distortion degree of the local field and decrease the amount of microstructural defects, such as voids and flaws at the filler/matrix interface.21 On the basis of this strategy, various surface-modification agents, such as the coupling agents,22 surfactants,23,24 and phosphoric acids,25,26 have been employed. Another way of improving nanoparticle dispersion is the “graft to” and “graft from” method, which directly connects the nano-fillers to the molecular chain of the polymeric matrix by chemical bonding. Although these techniques improve the dispersion of nanoparticles to a certain extent, they also have some practical limitations, such as tedious procedure, complex instrumentation, and use of toxic solvents.27

Recently, some research groups have proposed decorating conductive nanoparticles on high-dielectric-constant ceramic nanoparticles can avoid the agglomeration without compromising the insulative properties of the film.28,29 The increased conductivity of the interlayer between the ceramic nano-
particles and polymeric matrix by conductive nanoparticles enhances the space charge polarization and polarization reverse speed, leading to the enhanced dielectric constant. In this work, we demonstrate the synthesis of core−satellite-structured BaTiO$_3$−poly(dopamine)−Ag (BT−PDA−Ag) nanoparticles by dopamine oxidative polymerization and electroless plating of silver. The nitrile-butadiene rubber (NBR) with a relatively high dielectric constant (>10) and a large number of polar acrylonitrile groups was used as a matrix. The incorporation of BT nanoparticles prevented the continuous connection of Ag nanoparticles and suppressed the formation of a conductive path in the NBR matrix. Moreover, the ultrasmall Ag nanoparticles trapped the carriers by Coulomb blockade and quantum confinement effects, which resulted in a low dielectric loss and electrical conductivity. This research proposes a simple and effective route to fabricate high-dielectric materials.

2. RESULTS AND DISCUSSION

The schematic illustration of the fabrication of the core−satellite-structured BT−PDA−Ag nanoparticles is given in Figure 1. First, the BT nanoparticles were treated with dopamine to form a functional PDA layer, followed by the decoration of Ag nanoparticles on the PDA shell via in situ reduction of AgNO$_3$. To date, the dopamine self-polymerization mechanism is debatable and the research community believes that the catechol groups oxidize into quinone at weak alkaline pH and form a key intermediate 5,6-dihydroxyindole (DHI) after a series of oxidation, cyclization, and rearrangement reactions. The DHI further oxidizes into 5,6-indolequinone, which undergoes further polymerization to form PDA. Furthermore, the concurrent presence of the catechol and amine groups immobilizes the silver nanoparticles onto the BT−PDA surface. It is worth mentioning that...
The elemental surface composition of the BT, BT−PDA, and BT−PDA−Ag nanoparticles was investigated by X-ray photoelectron spectroscopy (XPS), and results are presented in Figure 2. Figure 2 exhibits that the three types of nanoparticles demonstrate peaks of C 1s, O 1s, and Ba 3d, but the N 1s peak (binding energies (BEs) ≈ 400 eV) is observed only in BT−PDA and BT−PDA−Ag nanoparticles, attributed to the existence of the nitrogen component in PDA. The N 1s spectra can be deconvoluted into two peaks at 399.5 and 398.5 eV, which can be assigned to pyridinic N species (p-N, amine groups) and graphitic N species (g-N, indole groups), respectively. As we all know, the samples will come in contact with air inevitably while saving, transporting, and analyzing; therefore, the C 1s signal in pristine BT particles mainly comes from the contaminant.37 The C 1s region in pristine BT nanoparticles can be fitted into two peaks located at 284.6 and 286.4 eV, which are assigned to C−C species and C−O species, respectively. Two additional peaks with BEs of 285.5 and 288.5 eV in BT−PDA nanoparticles can be attributed to the C−N and O−C=O species introduced by PDA, respectively (Figure 2b). These results demonstrate that...

Figure 3. XRD spectrum of the BT, BT−PDA, and BT−PDA−Ag nanoparticles.

Figure 4. HR-TEM images of (a) BT, (b) BT−PDA, and (c) BT−PDA−Ag nanoparticles.
PDA has been successfully coated on the surface of BT nanoparticles. In addition, a strong peak of Ag (BEs ≈ 370 eV) can be seen in Figure 2c. The Ag 3d spectrum can be deduced from two peak components at BEs of 368.0 and 374.0 eV, for Ag 3d_{5/2} and Ag 3d_{3/2} of the metallic Ag, respectively. No peak corresponding to the silver oxide species is found in the XPS spectrum, which indicates the successful reduction of Ag⁺ to Ag⁰, using glucose as a reducing agent in the presence of PDA.

**Figure 3** shows the X-ray diffraction (XRD) spectrum of BT, BT−PDA, and BT−PDA−Ag nanoparticles. As shown in Figure 3, the XRD pattern of BT nanoparticles exhibits the typical cubic lattice system. After functionalization with PDA, the XRD patterns of BT and BT−PDA nanoparticles exhibit no difference, indicating that PDA has no effect on the crystal structure of BT nanoparticles and the PDA layer is amorphous. The BT−PDA−Ag nanoparticles show four sharp characteristic peaks of silver (2θ = 38.2, 44.4, 64.6, and 77.4°), corresponding to the (1 1 1), (2 0 0), (2 2 0), and (3 1 1) planes of face-centered cubic silver, respectively. The XRD pattern is consistent with the JCPDS card no. 4-783.

**Figure 4** shows the high-resolution transmission electron microscopy (HR-TEM) images of BT, BT−PDA, and BT−PDA−Ag nanoparticles. On comparing with pristine BT nanoparticles that exhibit a smooth surface (shown in Figure 4a), the BT−PDA nanoparticles display a distinct rough PDA layer with a thickness of 2 nm, which is coated on the surface of BT nanoparticles (shown in Figure 4b). From Figure 4c, it can be seen that some silver nanoparticles (with an average diameter of 10 nm) without a continuous silver layer are dotted on the PDA layer.

The morphology of BT and BT−PDA−Ag nanoparticles was also observed by scanning electron microscopy (SEM), and the result is presented in Figure 5. In Figure 5a, the smooth surface of the pristine BT nanoparticles can be readily observed. After the decoration process, the Ag nanoparticles with an average diameter of about 10 nm were individually distributed on the surface of BT−PDA nanoparticles (shown in Figure 5b). The electronic probe microanalyzer was used to investigate the chemical composition of BT−PDA−Ag nanoparticles, and the results showed that the silver content is about 13.5 wt %. In addition, the electrical conductivity of BT−PDA−Ag nanoparticles determined by Keysight B2902 Precision Source/Measure Units was about 1.5 × 10⁻⁷ S/m.

**Figure 6** presents the freeze-fractured cross-sectional SEM images of the NBR composite films filled with different contents of fillers. Small agglomerated particles were observed in the 10 phr BT/NBR nanocomposite, and the size of the agglomeration increased with the increasing BT nanoparticle content, which led to the formation of the filler network in the 50 phr BT/NBR nanocomposite. This can be attributed to the large difference in surface energy between BT nanoparticles and the NBR matrix. While comparing BT/NBR nanocomposites with BT−PDA−Ag/NBR nanocomposites, it can be easily observed that the later displays more homogenous distribution with limited agglomeration of nanoparticles. In addition, the interface between BT−PDA−Ag nanoparticles and the matrix is indistinguishable, suggesting the enhanced interface compatibility of BT−PDA−Ag/NBR nanocomposites due to the strong adhesion properties of PDA. More importantly, the high-magnification SEM images reveal that the Ag nanoparticles are still tightly stuck to the BT nanoparticles after strong shearing stress during mixing, suggesting strong adhesion between the Ag nanoparticles and the BT core.

The mechanical properties of the NBR composites have been studied, and the stress−strain curves are presented in Figure 7. As shown in Figure 7, the tensile strength of NBR nanocomposites increased with the increasing filler loading content, indicating the reinforcement effect of inorganic nanoparticles. When compared with BT/NBR nanocomposites, the BT−PDA−Ag/NBR nanocomposites displayed a higher tensile strength at the same filler content due to the
uniform dispersion of nanoparticles and strong interfacial interaction. Usually, the loading of inorganic ceramics will decrease the flexibility of the polymeric matrix due to the interfacial defects, such as voids and air bubbles. However, the enhanced interfacial interaction between two phases can restrict the chain slippage and orient the stretching chains to uniformly distribute the external force. The highest tensile strength was achieved for the 50 phr BT−PDA−Ag/NBR nanocomposite and reached up to 2.91 MPa, which is about 1.79 times higher than that of pure NBR. Moreover, the elongation at break of BT−PDA−Ag/NBR nanocomposites was also higher as compared to that of BT/NBR nanocomposites at the same filler loading. The larger elongation at break can be ascribed to the improved interfacial interactions between BT−PDA−Ag nanoparticles and the NBR matrix, which prevents the inevitable elongation losses. However, the NBR composite loading with 10 phr BT or BT−PDA−Ag nanoparticles exhibits a larger elongation at break as compared with the pure NBR, which can be ascribed to the decreased cross-link density due to the interference of BT nanoparticles on the cross-linking process.

The dependence of the dielectric constant of pure NBR, BT/NBR, and BT−PDA−Ag/NBR nanocomposites on frequency is displayed in Figure 8. From Figure 8, we can find that the dielectric constant and dielectric loss tangent of all samples decrease with the increasing frequency, meaning that the dielectric properties are strongly frequency-dependent. This phenomenon is responsible for the dipole polarization of −CN in NBR and interfacial polarization between the filler and matrix, which cannot catch up with the changing electrical field frequency. Compared with that of pure NBR, the dielectric constant of BT/NBR composites increased with an increase in the filler content and can be attributed to the interfacial polarization, namely, the Maxwell−Wagner−Sillars effect. Furthermore, the BT−PDA−Ag/NBR nanocomposites containing a large amount of filler (30 or 50 phr) exhibit a larger dielectric constant at low frequency, when compared to BT/NBR nanocomposites filled with the same filler content. For instance, the 50 phr BT−PDA−Ag/NBR nanocomposite showed the dielectric constant of 21.5 at 1 Hz, whereas the 50 phr BT/NBR nanocomposite has a lower dielectric constant of 13.52 at 1 Hz (shown in Table 1). The higher dielectric constant can be attributed to the introduction of Ag nanoparticles. The Ag nanoparticles enhance the conductivity of the interlayer between the filler and the polymeric matrix, which increases the space charge polarization and polarization.
reverse speed and results in a higher dielectric constant.\cite{3,14} Moreover, the core−satellite-structured BT−PDA−Ag nanoparticles can be considered as a nanocapacitor structure in a polymeric matrix, where two neighboring BT−PDA−Ag nanoparticles act as the electrode and the thin NBR layer acts as the dielectric.\cite{16} Although some nanoparticle clusters might exist in the matrix, there was no obvious influence on dielectric properties of the film. However, it is seen from Table 1 that the dielectric constant at 1 Hz of 10 phr BT/NBR is higher than that of 10 phr BT−PDA−Ag/NBR. The lower dielectric constant of 10 phr BT−PDA−Ag/NBR demonstrating the introduction of a little amount of Ag nanoparticles affected the dielectric constant of NBR composites. This phenomenon can be explained by the Coulomb blockade effect of ultrasmall Ag nanoparticles on the surface of BT, which prevents the electron migration and suppresses the space charge accumulation in the interface, resulting in a lower dielectric constant.\cite{32}

Figures 9a,b depict the alternating current (AC) electrical conductivity and dielectric loss tangent of the pure NBR, BT/NBR, and BT−PDA−Ag/NBR nanocomposites, respectively. As expected, the AC electrical conductivity of nanocomposites increases with an increase in the loading content of BT−PDA−Ag nanoparticles (Figure 9a). For instance, the AC electrical conductivity increased from $1.76 \times 10^{-12}$ S/m for pure NBR to $2.34 \times 10^{-11}$ S/m for the 50 phr BT−PDA−Ag/NBR nanocomposite. The small decrease in the electrical conductivity of BT−PDA−Ag/NBR nanocomposites indicates the insulation of the resulting film and confirms that no conducting path has been formed in the nanocomposites. The core−satellite structure suppresses the “percolation” of Ag nanoparticles, resulting in a steady increase in the dielectric constant of NBR composites.

Figure 9b presents the dielectric loss of the nanocomposites and indicates a smaller loss at a higher frequency. It is well known that the interfacial polarizations and ion conductivity are mainly responsible for the dielectric behavior of the material at low frequency, whereas dipole polarization is responsible for the dielectric response at high frequency.\cite{40} In addition, the dielectric loss of nanocomposites increased with the increasing content of BT−PDA−Ag nanoparticles at low frequency, which can be attributed to the increased electrical conductivity.\cite{4} However, the BT−PDA−Ag nanoparticles assist in decreasing the dielectric loss in certain content at high frequency, which can be ascribed to the Coulomb blockade and quantum confinement effects of ultrasmall Ag nanoparticles.\cite{40,41} Finally, the highest dielectric loss was about 1.6 at 1 Hz for the 50 phr BT−PDA−Ag/NBR nanocomposite. The relatively low dielectric loss is achieved and corresponds to blockage of charge carrier movement by insulating NBR chains.\cite{3}

The effect of temperature on the dielectric constant was measured at a constant frequency of 1 Hz, and results are shown in Figure 10a. The dielectric constant varies with the increasing temperature, and the highest dielectric constant was observed at 10 °C. The change in dielectric constant with respect to temperature is due to the competing effect between

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**Figure 7.** Stress−strain curves of the NBR composites filled with different contents of (a) BT and (b) BT−PDA−Ag nanoparticles.

**Figure 8.** Dependence of the dielectric constant of pure NBR and NBR nanocomposites filled with different contents of (a) BT and (b) BT−PDA−Ag nanoparticles on frequency at room temperature.
When the temperature is increased from $-50$ to $10^\circ$C, the interfacial interaction restricts the thermal expansion and the formation of interspace, whereas the thermal expansion is fully exhibited when the temperature is greater than $10^\circ$C. Because of the large difference in thermal expansion coefficient between the BT–PDA–Ag filler and the NBR matrix, the thermal expansion might overcome the effect of polymer mobility on the dielectric constant at high temperature. Therefore, after reaching a maximum value at $10^\circ$C, the dielectric constant decreases with an increase in temperature.

Figure 10b displays the frequency dependence of the dielectric loss tangent for the 50 phr BT–PDA–Ag/NBR nanocomposite at different temperatures. It can be easily observed that the frequency dependence is also affected by the temperature. At high temperature, the change in dielectric constant with respect to frequency is more significant, which can be due to the dominant polar relaxation at high temperature. On the other hand, the anomalous characteristic of the glass-transition temperature ($T_g$) of the polymer dictates the dielectric behavior of nanocomposites at low temperatures.

Figure 11 shows the direct current (DC) volume resistivity and electric breakdown strength of the pure NBR, BT/NBR, and BT–PDA–Ag/NBR nanocomposites filled with different amounts of filler loading. It is easily observed from Figure 11a that the volume resistivity of BT/NBR nanocomposites does not significantly change with loading of BT nanoparticles. However, the volume resistivity of BT–PDA–Ag/NBR...

**Table 1. Summary of Mechanical Properties and Electromechanical Properties of Pure NBR and NBR Nanocomposites**

| Sample       | Elongation at break (%) | Tensile strength (MPa) | Dielectric constant (1 Hz) | Dielectric loss (1 Hz) | Conductivity S/m (1 Hz) | DC electrical conductivity (Ω cm) | Breakdown strength (kV/mm) | Energy density (kJ/m$^3$) |
|--------------|-------------------------|------------------------|----------------------------|------------------------|-------------------------|-----------------------------------|-------------------------------|--------------------------|
| 0 phr        | 317.3                   | 1.63 ± 0.22            | 11.15                      | 0.20                   | $1.76 \times 10^{-12}$  | $3.67 \times 10^{11}$              | 45 ± 2.0                     | 99.91                    |
| 10 phr BT    | 326.7                   | 1.68 ± 0.17            | 12.28                      | 0.14                   | $1.31 \times 10^{-12}$  | $3.81 \times 10^{11}$              | 47 ± 1.4                     | 120.03                   |
| 30 phr BT    | 280.9                   | 1.75 ± 0.21            | 13.02                      | 0.11                   | $1.07 \times 10^{-12}$  | $4.38 \times 10^{11}$              | 57 ± 2.4                     | 187.19                   |
| 50 phr BT    | 247.7                   | 2.26 ± 0.16            | 13.52                      | 0.05                   | $8.06 \times 10^{-13}$  | $5.13 \times 10^{11}$              | 63 ± 2.5                     | 237.45                   |
| 10 phr BT–PDA–Ag | 326.7           | 1.90 ± 0.17            | 11.76                      | 0.95                   | $9.98 \times 10^{-12}$  | $4.85 \times 10^{10}$              | 43 ± 1.5                     | 96.22                    |
| 30 phr BT–PDA–Ag | 292.3                   | 2.33 ± 0.21            | 16.95                      | 1.32                   | $1.57 \times 10^{-11}$  | $4.11 \times 10^{10}$              | 55 ± 3.1                     | 226.89                   |
| 50 phr BT–PDA–Ag | 264.2                   | 2.91 ± 0.25            | 21.50                      | 1.45                   | $2.34 \times 10^{-11}$  | $3.23 \times 10^{10}$              | 60 ± 2.2                     | 342.50                   |
nanocomposites decreases with an increase in BT–PDA–Ag nanoparticle content, which is consistent with the AC electrical conductivity. In addition, the nanoparticles enhance the electric breakdown strength (shown in Figure 11b) when the filler content increases to 30 phr. This can be explained by the decrease in the mobility of polymer chains when they are tightly bonded to the nanoparticles. The transfer of charge carriers through loose polymer chains not bonded to the nanoparticles is blocked, which results in a higher electrical breakdown strength.\(^43\) For simple linear response dielectric materials, the maximum energy density can simply be defined as \(U_e = 1/2 \varepsilon_r \varepsilon_0 E_b^2\), where \(\varepsilon_r\) is the relative dielectric permittivity (dielectric constant), \(\varepsilon_0\) is the vacuum dielectric permittivity with the value of \(8.85 \times 10^{-12}\) F/m, and \(E_b\) is the dielectric breakdown strength.\(^39\) Compared with BT/NBR nanocomposites, the BT–PDA–Ag/NBR nanocomposites exhibited lower electric breakdown strength, which is attributed to the lower volume resistivity. However, a high energy density was achieved due to the high dielectric constant of BT–PDA–Ag/NBR nanocomposites (Table 1), which is a big advantage for the dielectric materials.

3. CONCLUSIONS

We demonstrate the synthesis of core–satellite-structured BT–PDA–Ag nanoparticles by dopamine oxidative polymerization and electroless plating of silver. The decoration of Ag nanoparticles on the surface of BT nanoparticles resulted in a higher dielectric constant, which can be attributed to the increased conductivity of the filler/matrix interlayer and nanocapacitor structure. Moreover, the BT–PDA–Ag/NBR nanocomposites exhibited low dielectric loss and electrical conductivity, resulting from the introduction of BT nanoparticles, which prevented the continuous connection of Ag nanoparticles and suppressed the formation of a conductive path in the NBR matrix. The proposed method with simplicity and scalability could provide a promising route for improving dielectric properties of polymer nanocomposites.

4. METHODS

4.1. Materials. The BT nanoparticles with an average diameter of 100 nm were obtained from Nantong New Electronic Technology Company, China. Dopamine (DA) hydrochloride, tris(hydroxymethyl)aminomethane (Tris), poly(vinylpyrrolidone) (PVP), silver nitrate (AgNO\(_3\)), glucose, ammonia, and dicumyl peroxide (DCP) were obtained from Aladdin Industrial Plant, China. The NBR, with an acrylonitrile content of 28 wt % (DN2850), was received from Zeon Corporation, Japan.

4.2. Dopamine Oxidative Polymerization on BT Nanoparticle Surface. In a typical process, 80 g of BT nanoparticles, 2 g of dopamine hydrochloride, and 1.2 g of Tris were dispersed into 1000 mL of distilled water (pH = 8.5) and stirred for 24 h at room temperature. Then, the obtained nanoparticles were collected by filtration and washed three times using deionized water. The filter material was a commercial medium-speed qualitative filter paper with 100% cotton cellulose and of type 102, which was supplied by Hangzhou Special Paper Industry Co., Ltd. The nanoparticles were dried in a vacuum oven at 60 °C for 12 h and referred as BT–PDA in this article.

4.3. Electroless Plating of Silver Nanoparticles on BT–PDA Surface. The silver plating bath was prepared as follows. The AgNO\(_3\) solution with a concentration of 30 g/L was prepared, and ammonia was slowly dropped into the above solution until it became transparent again. Then, 0.5 wt % PVP was mixed into the above mixture and the reaction mixture was continuously stirred until the complete dissolution of PVP. Then, the BT–PDA nanoparticles were dispersed into the silver plating bath, followed by ultrasonic treatment for 30 min. Glucose with a concentration two times that of AgNO\(_3\) was dropwise added to the above mixture, and the reaction mixture was continuously stirred for 60 min at room temperature. The obtained BT–PDA–Ag nanoparticles were collected by filtration with distilled water and dried in a vacuum oven at 60 °C for 24 h.

4.4. Preparation of NBR Dielectric Composites. First, different amounts (0–50 g) of BT or BT–PDA–Ag nanoparticles, 2 g of DCP, and 100 g of NBR were mixed in a 6 in. two-roll mill. Then, the mixture was poured into a mold and cured at 25 MPa and 160 °C for the optimum curing time (\(t_{90}\)) to obtain cured composites. The \(t_{90}\) was obtained from a GT-M2000-FA rotorless curemeter (Gotech Testing Machines Inc., Taiwan) at 160 °C.

4.5. Characterization Method. X-ray photoelectron spectroscopy (XPS) was carried out using an ESCALAB 250 spectrometer (Thermo Fisher Scientific). The X-ray diffraction (XRD) spectra were measured by a Bruker D8 Focus diffractometer. High-resolution transmission electron microscope (HR-TEM) was performed to observe the morphologies of the particles using Hitachi H9000 at a voltage of 300 kV. Scanning electron microscopy (SEM) was carried out using Nova NanoSEM 430, FEI SEM. An electronic probe

Figure 11. (a) DC volume resistivity and (b) breakdown strength of the NBR nanocomposites filled with different contents of BT and BT–PDA–Ag nanoparticles.
microaccelerometer (EPMA-1600; Shimadzu, Japan) was used with an accelerating voltage of 15 kV. Tensile tests of dumbbell-shaped samples were carried out using a tensile apparatus (Instron 3366, Instron Corporation, American) at a cross-head speed of 50 mm/min, according to ASTM D412. The stress of all samples is engineering stress. The electrical and dielectric responses were measured using a Novocontrol Alpha-A broadband dielectric spectrometer in the frequency range of 10−10Hz at various temperatures. The DC breakdown strength was assessed by an LAS6030P electrical breakdown microanalyzer (EPMA-1600; Shimadzu, Japan).

According to our previous works, the volumetric resistivity considering practical application of dielectric materials, rate of 800 V/s, which was supplied by Boher High Voltage Power Supplies Co. Ltd., China. The samples were performed using circular membrane actuators in the presence of air for considering practical application of dielectric materials, according to our previous works. The volumetric resistivity \( \rho_v \) was calculated using an EST 121 resistivity meter (Beijing Huajinghui, China).

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**Notes**

The authors declare no competing financial interest.

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