Electrical properties of Sb$_2$O$_3$-modified BiScO$_3$–PbTiO$_3$-based piezoelectric ceramics

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Compared with pure Pb-based perovskite ferroelectric materials, BiMeO$_3$–PbTiO$_3$ (Me = Sc$^{3+}$, In$^{3+}$, and Yb$^{3+}$) systems have remarkable advantages in their Curie temperatures. As a member of this group, the BiScO$_3$–PbTiO$_3$ (BS–PT) solid solution has drawn considerable attention from scientists for its high Curie temperature and excellent piezoelectric coefficient. However, BS–PT ceramics still have some shortcomings, such as high dielectric loss and low mechanical quality factor, which make them unsuitable for high-temperature applications. Herein, we report the effect of the addition of complex ions on the electrical properties of BS–PT ceramics. Sb$_2$O$_3$-doped 0.36BiScO$_3$-0.64PbTi$_{0.97}$Fe$_{0.03}$O$_3$ + 1 mol% MnO$_2$ (BS–PTFMn + x% Sb$_2$O$_3$) ceramics were fabricated and their electrical properties were studied. BS–PTFMn + 0.75% Sb$_2$O$_3$ had an optimal piezoelectric coefficient, exhibiting $d_{33}^{\text{FMn}} = 442$ pm V$^{-1}$, which indicates that Sb$_2$O$_3$ doping can improve the piezoelectric properties of the BS–PT ceramics, exhibiting a "soft" effect of Sb$_2$O$_3$ doping. In addition, the thermal depolarization temperature ($T_d$) of BS–PTFMn + 0.75% Sb$_2$O$_3$ ceramics remained above 300 °C, such as 325 °C for BS–PTFMn + 0.75% Sb$_2$O$_3$. It was concluded that the piezoelectric properties of BS–PT ceramics were enhanced by the addition of Sb$_2$O$_3$.

1. Introduction

High-temperature piezoelectric materials have received unprecedented attention due to the urgent demand for high-temperature piezoelectric sensors and actuators in aerospace, automobile, oil and gas exploration fields. $^3$–$^5$ The commercial lead–zirconium–titanium (PZT) piezoelectric ceramics have been studied for decades and are widely used in piezoelectric sensors and transducers due to their outstanding piezoelectric properties, such as $d_{33} = 285$ pC N$^{-1}$ for PZT-4, $d_{33} = 375$ pC N$^{-1}$ for PZT-5A. $^4$ Nevertheless, their curie temperatures, $T_C$, are below 380 °C. $^5$ It is generally believed that the safety usage temperature range cannot exceed half of the Curie temperature in bulk PZT-based piezoelectric materials because of the depolarization caused by the thermal activation. Therefore, the traditional PZT-based piezoelectric ceramics cannot be used in the environments above 200 °C. $^5$ This is far from enough for today's industrial demands. Therefore, the development of new high-temperature-stable piezoelectric materials is an extremely urgent need.

In 2001, Eitel Richard first reported Bi(Me)O$_3$–PbTiO$_3$ (Me = Sc, In, Yb, Mg/Ti, Fe and Ga) systems, which have a perovskite structure and high Curie temperature. $^7$–$^{12}$ Among them, BiScO$_3$–PbTiO$_3$ (BS–PT) ceramics have drawn considerable attention because of their large piezoelectric coefficient ($d_{33} = 460$ pC N$^{-1}$) and high Curie temperature ($T_C = 450$ °C) near the morphotropic phase boundary (MPB), and have a good application prospect at high temperatures. $^{13,14}$ However, the thermal degradation of BS–PT ceramics occurs at 250 °C, which limits their applications at high temperatures. $^{13,14}$ Moreover, BS–PT ceramics have a high dielectric loss, resulting in high energy consumption and heat dissipation in the use of high-power electronic devices. $^{17}$

Previously, numerous studies were involved in controlling the reduction of dielectric loss, improving the electrical properties and thermal stability by the addition of a third element, $^{16}$–$^{20}$ such as Pb(Mg$_{1/3}$Nb$_{2/3}$)O$_3$ (PMN), BaTiO$_3$ (BT), and Pb(Mn$_{1/3}$Sb$_{2/3}$)O$_3$ (PMS). It is reported that the defects and oxygen vacancies generated by the multi-valence elements, such as Mn and Fe, can diffuse to the domain boundaries and reduce the total energy of the ceramics, leading to the pinning behavior of the motion of 180° domain and effective reduction of the dielectric loss. $^{17,21}$ Chen et al. reported that 0.36BiScO$_3$–0.64Pb(Ti$_{0.97}$Fe$_{0.03}$)O$_3$ + 1 mol% MnO$_2$ ceramics exhibited a high Curie temperature ($T_C = 492$ °C) and low dielectric loss ($\tan \delta = 0.006$). $^{21}$ Unfortunately, its $d_{33}$ significantly deteriorated ($d_{33} = 238$ pC N$^{-1}$) and the dielectric constant significantly
decreased ($\varepsilon_r = 815$). Mixed Mn and Sb elements are usually used to modify commercial PZT-based piezoelectric ceramics because the mixed Mn and Sb can synchronously enhance $k_p$, $Q_m$, $\varepsilon_r$, and $d_{33}$. Therefore, in this study, Sb$_2$O$_3$ was added to 0.36BiScO$_3$–0.64Pb(Ti$_{0.97}$Fe$_{0.03}$)O$_3$ + 1 mol% MnO$_2$ ceramics. The effects of Sb$_2$O$_3$ content doping on the phase structure, dielectric, piezoelectric, ferroelectric and electromechanical coupling properties of 0.36BiScO$_3$–0.64Pb(Ti$_{0.97}$Fe$_{0.03}$)O$_3$ + 1 mol% MnO$_2$ ceramics were investigated.

2. Experimental procedure

2.1 Sample preparation

In this study, 0.36BiScO$_3$–0.64Pb(Ti$_{0.97}$Fe$_{0.03}$)O$_3$ + 1 mol% MnO$_2$ + $x$ mol% Sb$_2$O$_3$ ($x = 0, 0.25, 0.5, 0.75, 1$, and $1.5$) (BS–PTFMn + $x\%$ Sb$_2$O$_3$) were prepared by a conventional solid-state method. The oxide powders, Bi$_2$O$_3$ (99.9%), Sc$_2$O$_3$ (99.99%), PbO (99.9%), TiO$_2$ (99.5%), Sb$_2$O$_3$ (99.5%), Fe$_2$O$_3$ (99.5%), and MnO$_2$ (99.5%), were accurately weighed and mixed according to the desired stoichiometric compositions. In addition, 3 wt% excess Bi$_2$O$_3$ was added to compensate for the volatilization of bismuth at high temperatures. The weighed powder was put into an agate grinding tank with an agate ball and anhydrous ethanol as the ball-mill medium for the ball milling. The ball milling was used with a planetary ball mill for 24 hours. After ball-milling, the mixed powder was dried, pressed into blocks and placed in a crucible, and calcined at 750 °C for 4 h. The calcination of blocks can reduce the distance of the powders, which is conducive to matter diffusion, the uniformity of a reaction and the reduction in the calcination temperature and time. Then, small amounts of MnO$_2$ and Sb$_2$O$_3$ were added to the calcined mixture and ground in alcohol for 24 hours. After drying, the calcined powders were granulated with 5% polyvinyl alcohol binder (PVA) and then pressed into pellets of 10 mm in diameter and 1 mm in thickness under 200 MPa. The green pellets were heated at 550 °C for 4 h to remove PVA. Furthermore, the green pellets were placed in an Al$_2$O$_3$ crucible and sintered at 1000–1100 °C for 3 h, while embedded in the calcined powders for preventing the vaporization of Bi and Pb.

2.2 Characterization methods

The crystalline phase of sintered samples was analyzed using an X-ray diffractometer (Miniflex 600, Rigaku, Japan) with Cu-K$_{\alpha 1}$, K$_{\alpha 2}$ mixing radiation, operating at 40 kV and 40 mA, in the 2θ range of 10–60° with a step size of 0.05° and 0.5 s per step. The fractured surface morphologies of the obtained ceramic samples were characterized by scanning electron microscopy (SEM, SU8010, Hitachi, Japan). The sintered samples were polished to the thickness of 0.5 mm, and then coated with a silver paste for electrical characterization. The samples were placed in a silicon oil at 120 °C and poled for 20 min under a DC electric field of 40 kV cm$^{-1}$. The electrical properties of poled samples were tested after placement for 24 h. The piezoelectric coefficients $d_{33}$ were measured using a quasi-static $d_{33}$ meter (Institute of Acoustics, Chinese Academy of Sciences, model ZJ-4AN). The planar electromechanical coupling factor ($k_p$) and the mechanical quality factor ($Q_{m}$) were calculated according to the resonance and anti-resonance methods using an impedance analyzer (E4990A, Agilent Technologies, USA). The dielectric constant ($\varepsilon_r$) and dielectric loss (tan $\delta$) as functions of temperature were also measured using the impedance analyzer. The polarization–electric field ($P$–$E$) hysteresis loops, current–electric field ($I$–$E$) loops and strain–electric field ($S$–$E$) curves were obtained using an aix-ACCT TF2000 analyzer with a frequency of 1 Hz. The resistivity of ceramics was measured using an electrometer/high resistance meter (Keithley 6517B, USA).

3. Results and discussion

3.1 Structure and phase analysis

Fig. 1(a) shows the XRD patterns of sintered ceramic samples with different Sb$_2$O$_3$ contents. All the samples presented a pure perovskite structure without any clear trace of impurity phases, indicating stable perovskite solid solutions. Since the radii of Fe$^{3+}$ ion (0.645 Å for CN = 6) and Mn$^{3+}$ ion (0.645 Å for CN = 6) are similar to Ti$^{4+}$ radius (0.605 Å for CN = 6), Fe$^{3+}$ and Mn$^{3+}$ were likely to replace Ti$^{4+}$ in BO$_6$ octahedral perovskite structures. In addition, though Sb is a multivalent element, more
Sb\(^{5+}\) can be formed in the sintering process. It is expected that Sb\(^{5+}\) will most likely enter into the B-site to substitute Ti\(^{4+}\) because the radius of Sb\(^{5+}\) ion (0.60 Å for CN = 6) is close to that of Ti\(^{4+}\) ion (0.605 Å for CN = 6). Moreover, oxygen vacancies can be formed in Fe\(^{3+}\) and Mn\(^{3+}\) doped BS–PT ceramics due to the substitution of Ti\(^{4+}\) with low ionic valence (Fe\(^{3+}\) as an example: Fe\(_2\)O\(_3\) + 2TiO\(_2\) → 2Fe\(_{\text{ff}}\) + V\(_{\text{O}}\) + 3O\(_3\)).

However, the average grain size rapidly decreased from 10 μm to 2 μm with the increase in Sb\(_2\)O\(_3\) incorporation. The detailed grain size is listed in Table 1. Clearly, grain growth can be controlled by the addition of Sb\(_2\)O\(_3\). The Sb ions could not be completely dissolved into the solid solution as the doping amount of Sb\(_2\)O\(_3\) increased to 1.5% in this study. The extra Sb\(_2\)O\(_3\) diffused into the grain boundary, which prevented the movement of the grain boundary during the sintering process and hindered the growth of grain, leading to the reduction of grain size of Sb-doped samples. Therefore, the average grain size decreased rapidly at this composition. In addition, the Archimedes method was used to calculate the actual density (\(\rho_1\)) of ceramic samples, while the theoretical density (\(\rho_2\)) was calculated based on the XRD data. The calculated results showed that the relative density (\(\rho_r\)) of ceramic samples was 96.5–98.6%, indicating a high density of ceramic samples. The results are listed in Table 1. The density of the ceramic samples reached the maximum at \(x = 0.75\) with a grain size of 5.0 μm. The interlayer and porosity were created when extra Sb\(_2\)O\(_3\) diffused into the grain boundary, such as \(x = 1.50\).

### Table 1: Density and average grain size of BS–PTFMn + x% Sb\(_2\)O\(_3\) ceramics (\(d\): average grain size; \(\rho_1\): actual density; \(\rho_2\): theoretical density; \(\rho_r\): relative density)

| Composition | \(d\) (μm) | \(\rho_1\) (g cm\(^{-3}\)) | \(\rho_2\) (g cm\(^{-3}\)) | \(\rho_r\) (%) |
|-------------|------------|----------------------------|----------------------------|--------------|
| \(x = 0.00\) | 10         | 7.23                       | 7.45                       | 97.1         |
| \(x = 0.25\) | 9          | 7.25                       | 7.44                       | 97.4         |
| \(x = 0.50\) | 8          | 7.28                       | 7.41                       | 98.2         |
| \(x = 0.75\) | 5          | 7.32                       | 7.42                       | 98.6         |
| \(x = 1.00\) | 3          | 7.27                       | 7.43                       | 97.8         |
| \(x = 1.50\) | 2          | 7.22                       | 7.48                       | 96.5         |

### 3.2 Dielectric properties

Fig. 3 shows the temperature dependence of dielectric constant (\(\epsilon_r\)) and dielectric loss (\(\tan \delta\)) of BS–PTFMn + x% Sb\(_2\)O\(_3\) ceramics. The values of dielectric constant (\(\epsilon_r\)) at room temperature increased first and then decreased with the increase in Sb\(_2\)O\(_3\), as shown in Table 2. The maximum of \(\epsilon_r\) appeared at BS–PTFMn + 0.75% Sb\(_2\)O\(_3\) ceramics. The increase in
Fig. 3 Temperature dependence of dielectric properties ($\varepsilon_r$, tan $\delta$) of BS–PTFMn + x% Sb2O3 ceramics: (a) 0% Sb2O3 (b) 0.25% Sb2O3, (c) 0.5% Sb2O3 (d) 0.75% Sb2O3, (e) 1% Sb2O3, (f) 1.5% Sb2O3. The insets show the fitting result using semi-empirical Curie–Weiss law.

Table 2 Electrical properties of BS–PTFMn + x% Sb2O3 ceramics

| Composition | $T_c$ (°C) | $\varepsilon_r$ | tan $\delta$ | $P_r$ (µC cm$^{-2}$) | $E_c$ (kV cm$^{-1}$) | $d_{33}$ (pC N$^{-1}$) | $k_p$ | $Q_m$ | $T_d$ (°C) |
|-------------|------------|-----------------|--------------|----------------------|-------------------|-------------------|-------|-------|-------------|
| x = 0.00    | 425        | 920             | 0.008        | 17                   | 28                | 230               | 0.39  | 180   | 380         |
| x = 0.25    | 422        | 960             | 0.009        | 20                   | 26                | 260               | 0.42  | 141   | 365         |
| x = 0.50    | 407        | 1000            | 0.009        | 21                   | 23                | 270               | 0.43  | 125   | 350         |
| x = 0.75    | 404        | 1230            | 0.014        | 24                   | 22                | 305               | 0.45  | 93    | 325         |
| x = 1.00    | 400        | 1060            | 0.015        | 24                   | 24                | 280               | 0.40  | 86    | 310         |
| x = 1.50    | 394        | 1050            | 0.018        | 29                   | 25                | 240               | 0.36  | 72    | 300         |
the dielectric constant is related to the decrease in the grain size (see Table 1). G. Arlt et al. reported that the width of the domain is proportional to the square root of the grain size in ferroelectric ceramics.\textsuperscript{28} It is reasonable to conclude that the domain density increases when the grain size decreases. In addition, the dielectric constant is mainly affected by the extrinsic contribution originated from the movement of the domain wall.\textsuperscript{29,30} Therefore, the extrinsic contribution to dielectric response can be enhanced if the density of domain walls increases, resulting in the enhancement of the dielectric constant. Therefore, the dielectric constant increased first, as shown in Table 2. When the Sb\textsubscript{2}O\textsubscript{3} doping content further increased, the dielectric constant decreased because the extra Sb\textsubscript{2}O\textsubscript{3} doping diffused into the grain boundaries, which hindered the movement of the electric domain.\textsuperscript{26,30} In polycrystalline ferroelectric ceramics, the relationship between dielectric properties and grain size is very common and correspondingly observed.\textsuperscript{31,32} 

\(T_{\text{m}}\) defined as the temperature of the maximum of \(\varepsilon_r\), is considered as the Curie temperature \(T_C\). It is clearly seen that Curie temperature is sensitive to Sb doping (listed in Table 2), and decreased linearly from 425 °C to 394 °C with the increase in Sb doping. Chen et al. reported that 0.36BiScO\textsubscript{3}–0.64Pb(Ti\textsubscript{0.97}Fe\textsubscript{0.03})O\textsubscript{3} + 1 mol% MnO\textsubscript{2} ceramics exhibit a high Curie temperature of 492 °C.\textsuperscript{21} There are two main reasons for the gap of Curie temperature between our results and those of previous reported. The first reason is that the deviation of components was induced by Pb and Bi volatilizing during the synthesis process. The second reason is the doping of complex ions. It is well known that Fe and Mn elements have multiple valence states. The Curie temperature can be affected by the change in the valence of Fe and Mn ions. The effects of complex ions on the variation of Curie temperature require further studies.

The peak of maximum \(\varepsilon_r\) became wider and wider as Sb\textsubscript{2}O\textsubscript{3} increased, exhibiting a diffuse phase transition. In addition, with the increase in the Sb\textsubscript{2}O\textsubscript{3} doping, \(T_m\) shifted to a high temperature as frequencies increased, showing frequency dispersion. It is indicated that BS–PTFMn + x% Sb\textsubscript{2}O\textsubscript{3} ceramics exhibit a typical relaxor behavior as Sb\textsubscript{2}O\textsubscript{3} is doped. The relaxor behavior of Sb-doped BS–PTFMn ceramics at a phase transition can be described by the modified Curie–Weiss law:\textsuperscript{33}

\[
\ln\left(1/\varepsilon - 1/\varepsilon_m\right) + \ln C = \gamma \ln(T - T_m)
\]

where \(\varepsilon_m\) is the value of \(\varepsilon_r\) at \(T_m\), \(C\) is the Curie constant, and \(\gamma\) is the degree of phase transition diffuses. When \(\gamma\) is equal to 1, it is usually considered a normal ferroelectric, which satisfies Curie–Weiss law. For \(\gamma = 2\), the phase transition is a canonical relaxor ferroelectric with a completely diffused phase transition.

![Fig. 4](image-url)  
(a) Polarization–electric field (\(P–E\)) hysteresis and (b) current–electric field (\(I–E\)) loops, (c) \(P_r\) and \(E_c\), (d) \(d_{33}\), \(k_p\) and \(Q_m\) as a function of Sb\textsubscript{2}O\textsubscript{3} content of BS–PTFMn + x% Sb\textsubscript{2}O\textsubscript{3} ceramics.
Therefore, a larger value of $\gamma$ represents a larger degree of the diffused phase transition. The insets in Fig. 3 shows the fitting curves of $\ln(1/e - 1/e_m)$ vs. $\ln(T - T_m)$ at 1 kHz and the value of $\gamma$, showing that the value of $\gamma$ for all the ceramic samples was between 1 and 2, indicating typical relaxor ferroelectrics. It can be clearly seen that the value of $\gamma$ gradually increased from 1.49 to 1.95 as Sb$_2$O$_3$ increased. The relaxor behavior was related to the heterogeneity of the solid solution. Small amounts of Sb occupy the B-site in the perovskite structure, leading to a disordered distribution of ions at the B-site. Therefore, the degree of heterogeneity increased with the Sb$_2$O$_3$ doping, which resulted in an increase in the degree of relaxation. Similar phenomena have been observed in other BS–PT based systems.\textsuperscript{20,33}

### 3.3 Ferroelectric and piezoelectric properties

Fig. 4(a) shows the polarization–electric field ($P$–$E$) hysteresis loops of BS–PTFMn + x% Sb$_2$O$_3$ ceramics. The $P$–$E$ hysteresis loops were measured under an electric field of ±70 kV cm$^{-1}$ at room temperature with a frequency of 1 Hz. According to the $P$–$E$ hysteresis loops, the variation of the values of remnant polarization ($P_r$) and coercive field ($E_c$) was obtained, as shown in Fig. 4(c). The value of $P_r$ gradually increased as the Sb content increased. However, the value of $E_c$ decreased first and reached a minimum value (22 kV cm$^{-1}$) at $x = 0.75$. When the Sb$_2$O$_3$ doping content was higher than 0.75%, the value of $E_c$ increased slightly. The first decrease in $E_c$ was mainly due to the “soft” doping of Sb$_2$O$_3$, while the second increase was mainly due to the excessive Sb$_2$O$_3$ at the grain boundary.\textsuperscript{46} Fig. 4(b) shows the $I$–$E$ loops of all the samples at room temperature. The clear positive and negative current peaks, which correspond to the switching of polarization reveals the well-shaped saturated ferroelectric hysteresis loops, confirming the high quality of $P$–$E$ hysteresis loops.

The tendency of piezoelectric coefficient ($d_{33}$), planar electromechanical coupling factor ($k_p$), and mechanical quality factor ($Q_m$) as a function of the Sb$_2$O$_3$ doping content are shown in Fig. 4(d). The variation of $d_{33}$ shows an opposite trend compared to that of $E_c$, showing optimal properties at $x = 0.75$ (305 pC N$^{-1}$), which indicates that Sb$_2$O$_3$ doping can improve the piezoelectric properties of BS–PT-based ceramics. Similarly, the value of $k_p$ showed the same tendency, exhibiting the maximum at $x = 0.75%$. However, the value of $Q_m$ monotonously declined from 180 to 70 as Sb$_2$O$_3$ increased, showing typical characteristics of a “soft” modifier. The detailed ferroelectric and piezoelectric properties are listed in Table 2. The optimal piezoelectric properties appeared for BS–PTFMn + 0.75% Sb$_2$O$_3$ ceramics, exhibiting $e_r = 1230$, tan $\delta = 0.014$, $E_c = 22$ kV cm$^{-1}$, $k_p = 0.45$ and $d_{33} = 305$ pC N$^{-1}$. It is well known that Sb$_2$O$_3$ is a type of soft doping in which Sh$^{3+}$ will replace Ti$^{4+}$ ions, reducing the formation of lead vacancy, which leads to the domain movement becoming easier.\textsuperscript{38} Compared with undoped samples, the enhancement of $d_{33}$ can be explained by a qualitative formula: $d_{33} \sim Q_{33} e_r \varepsilon_0 P_t$ for the piezoelectric materials,\textsuperscript{38} where $e_r$ is the dielectric permittivity of free space, and $Q_{33}$ is the electrostriction coefficient. The $d_{33}$, $e_r$, $P_t$ are interrelated, showing high $e_r$ and high $P_t$ for high piezoelectricity. All values of $e_r$, $P_t$ and $d_{33}$ showed the same change as Sb$_2$O$_3$ increased to 0.75%. Therefore, the variation of $d_{33}$ conforms to the formula. There are two reasons for an appreciable reduction of $d_{33}$ upon further increase of Sb$_2$O$_3$. The one is that extra Sb$_2$O$_3$ doping diffused into the grain boundaries, which hindered the movement of the electric domain, reducing the extrinsic contribution to piezoelectricity. The other is due to the phase transition from the tetragonal phase to the rhombohedral phase, as shown in Fig. 1. The complex perovskite ferroelectrics exhibit optimal piezoelectricity near the MPB region. It has been reported that BS–PT based ceramics exhibit optimal piezoelectricity in the MPB region close to the tetragonal phase.\textsuperscript{44} Therefore, the $d_{33}$ decreased when the structure changed to the rhombohedral phase.

Fig. 5 shows the unipolar strain vs. electric field ($S$–$E$) curves of BS–PTFMn + x% Sb$_2$O$_3$ ceramics measured under an electric field of 70 kV cm$^{-1}$. The unipolar strain first increased with the increase in the Sb$_2$O$_3$ content, the maximum strain of 0.31% was obtained at BS–PTFMn + 0.75% Sb$_2$O$_3$ ceramics, and then decreased. For the undoped samples, a relatively small strain
resulted from the reduction of the external effect, which was mainly due to the domain pinning caused by Fe/Mn doping. While the increase in strain after Sb₂O₃ doping was attributed to releasing the domain wall pinning effect, which lead to an easier switch of the non-180°/C₄₄ domain wall.

In addition, the large-signal piezoelectric coefficient $d_{33}^*$ was calculated using the formula, $d_{33}^* = S_{\text{max}}/E_{\text{max}}$, from the $S$–$E$ curves, as shown in Fig. 5(b). It was found that the variation of $d_{33}^*$ with the Sb₂O₃ content was similar to that of the value of $d_{33}$ measured by a quasi-static $d_{33}$ meter. The maximum value of $d_{33}^*$ was 442 pm V⁻¹ for BS–PTFMn + 0.75% Sb₂O₃ ceramics. The value of $d_{33}^*$ was about 1.45-fold the $d_{33}$ value (305 pC N⁻¹). The piezoelectric contributions of the ferroelectric materials contain two aspects: one is the intrinsic contribution due to the lattice distortion; the other is an extrinsic contribution due to the ferroelectric non-180° domain switching. The domain wall motion under a high electric field may provide a more extrinsic contribution to the overall piezoelectric properties. Accordingly, the large-signal $d_{33}^*$ value is larger than that derived from the quasi-static piezoelectric meter.$^{39,40}$

![Fig. 6](image_url) (a) Temperature dependence of $P$–$E$ hysteresis loops, (b) temperature dependence of $I$–$E$ loops, (c) changes in resistivity, and (d) temperature dependence of $k_p$ of BS–PTFMn + 0.75% Sb₂O₃ ceramics.

| Sample                      | $T_c$ (°C) | $e_r$ | $\tan \delta$ | $d_{33}$ (pC N⁻¹) | $k_p$ | $T_d$ (°C) | Ref. |
|-----------------------------|------------|------|---------------|-------------------|------|------------|------|
| 0.36BS–0.64 PT              | 450        | 2010 | 4%            | 460               | 0.56 | 250        | 14   |
| 0.36BS–0.64PTM              | 428        | 865  | 1%            | 254               | 0.47 | 230        | 17   |
| Fe : 0.35BS–0.6PT–0.05PZN   | 440        | 1350 | 1.5%          | 220               | 0.33 | —          | 41   |
| 0.10PT–0.36BS–0.54 PT       | 393        | 1890 | 8%            | 350               | —    | —          | 19   |
| 0.10PMS–0.3BS–0.60 PT       | 320        | 1000 | 4%            | 210               | 0.32 | —          | 18   |
| 0.07PMS–0.33BS–0.60 PT      | 330        | 993  | 1.4%          | 225               | 0.38 | —          | 20   |
| BS–PTFMn + 0.75% Sb₂O₃     | 404        | 1230 | 1.4%          | 305               | 0.45 | 325        | This work |
3.4 Thermal stability

Thermal stability is interesting for sensors and actuators in various industrial applications. The temperature dependence of the P–E hysteresis loops of the BS–PTFMn + 0.75% Sb2O3 ceramics was measured with a frequency of 1 Hz, as shown in Fig. 6(a). BS–PTFMn + 0.75% Sb2O3 ceramics exhibited $P_r = 24 \mu C \cdot cm^{-2}$ and $E_c = 22 \text{kV cm}^{-1}$ at room temperature. It was found that the P–E loops maintained a good square-shaped loop as the temperature increased, and there was no obvious leakage current caused by the reduction of the ceramic volume resistivity under high temperatures. At 300 °C, it is clearly seen that BS–PTFMn + 0.75% Sb2O3 ceramics maintained the values of $P_r$ and $E_c$, as high as 20 μC cm⁻² and 8.5 kV cm⁻¹, respectively. Fig. 6(b) shows the temperature dependence of the I–E loops of BS–PTFMn + 0.75% Sb2O3 ceramics. It is easily seen that standard I–E loops are shown below 300 °C. At 300 °C, the current curves fluctuated with several peaks under a high electric field, indicating that the leakage current starts to increase. Therefore, it is indicated that the ferroelectric properties of BS–PTFMn + 0.75% Sb2O3 ceramics start to degenerate at this temperature.

The high resistivity is necessary for high-temperature applications, particularly under a high electric field. The resistivity of BS–PTFMn + 0.75% Sb2O3 ceramics under a high electric field is shown in Fig. 6(c). It can be easily seen that the resistivity does not change at room temperature although the electric field was as high as 70 kV cm⁻¹, showing the resistivity of $7.2 \times 10^8 \Omega \text{m}$, which satisfies the poling condition without any breakdown. The resistivity gradually decreased as the temperature increased. At 350 °C, though resistivity declined a lot under 70 kV cm⁻¹, the resistivity under 10 kV cm⁻¹ still maintained the value of $2.0 \times 10^8 \Omega \text{m}$, showing a feature of an insulator. It is known that thermal stability is much affected by the large conductance in applications because it must work repeatedly under high temperatures. Therefore, the degradation of resistivity is responsible for the degradation of the piezoelectric performance of piezoelectric ceramics. It is concluded that the high resistivity of the BS–PTFMn + 0.75% Sb2O3 ceramics is in favor of the enhancement of the thermal stability of BS–PT ceramics.

To further characterize the temperature stability of the samples, the temperature dependence of $k_p$ of the BS–PTFMn + x% Sb2O3 ceramics was measured (Fig. 6(d)). In this study, the value of the thermal depolarization temperature ($T_d$) of each sample was calculated, as listed in Table 2. Mostly, the fluctuation of the piezoelectric coefficient is required to be less than 10% for the industrial applications of piezoelectric ceramics. Therefore, $T_d$ is defined as the temperature at which the value of kp dropped by 10%. It can be easily seen that the $T_d$ value of BS–PTFMn + x% Sb2O3 ceramics decreased slightly as the Sb2O3 content increased, which was consistent with the change in the Curie temperature. However, $T_d$ of all samples was still above 300 °C. For example, BS–PTFMn + 0.75% Sb2O3 ceramics exhibited $T_d$ of 325 °C.

Table 3 shows the comparison of the electrical properties of BS–PTFMn + x% Sb2O3 and other BS–PT based ceramics. Compared with these systems with high $d_{33}$ (≥300 pC N⁻¹), BS–PTFMn + 0.75% Sb2O3 ceramics exhibited a relatively low dielectric loss. In addition, BS–PTFMn + 0.75% Sb2O3 ceramics exhibited the highest $d_{33}$ among these systems with a low dielectric loss (≤1.5%). Therefore, the complex ion doping can improve the piezoelectric properties of BS–PT based ceramics. In addition, BS–PTFMn + 0.75% Sb2O3 ceramics exhibited a relatively high depolarization temperature. These results also confirm that the BS–PTFMn + x% Sb2O3 ceramics are promising piezoelectric materials for high-temperature applications.

4. Conclusions

The dielectric, ferroelectric, piezoelectric, and electromechanical coupling properties of the BS–PTFMn + x% Sb2O3 ceramics are presented in this study. The addition of Sb2O3 leads to BS–PTFMn + x% Sb2O3 ceramics transformed from the tetragonal phase to the rhombohedral phase. The relaxation enhancement was caused by the increase in Sb2O3, which was attributed to the increase in disorder in the perovskite structure. The BS–PTFMn + 0.75% Sb2O3 ceramics possessed optimal piezoelectric properties at room temperature, exhibiting $d_{33} = 442 \text{pm V}^{-1}$. Although the Curie temperature of the system dropped slightly when Sb2O3 content increased, the thermal depolarization temperature ($T_d$) of BS–PTFMn + x% Sb2O3 ceramics remained above 300 °C, such as 325 °C for BS–PTFMn + x% Sb2O3 ceramics. These results also confirmed that doping with Sb2O3 can improve the piezoelectric properties of BS–PT piezoelectric ceramics. The results indicated that the dielectric constant, piezoelectric coefficient, electromechanical coupling factor increased and mechanical quality factor decreased after Sb2O3 doping. Therefore, the role of “soft” doping is the main effect of Sb2O3 on electrical performance. The enhanced piezoelectric properties and good thermal stabilities indicated that the BS–PTFMn + x% Sb2O3 ceramics are promising for high-temperature piezoelectric applications.

Conflicts of interest

There are no conflicts of interest to declare.

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