

## REVIEW

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## Silver niobate perovskites: structure, properties and multifunctional applications

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AgNbO<sub>3</sub> exhibits "peculiar" anti-/ferroelectricity and narrow bandgap semi-conductivity that lead to active responses to different types of external stimuli, including electric fields, light and mechanical forces. Some of these unique properties are also mutually coupled and could be suited for the development of multifunctional devices for a broad range of applications, including dielectric, piezoelectric, high-power energy storage/conversion, photocatalytic and photovoltaic devices. In this review, recent studies of AgNbO<sub>3</sub> and AgNbO<sub>3</sub>-based materials are summarized. The main scope is to establish the correlations between chemical composition, synthesis conditions, structure, and properties, with an improved understanding of the phase transformations taking place in the three so called "M" phases, to ultimately provide guidance on the materials development in two key sectors: high-power energy storage and photocatalysis. Finally, current challenges in multifunctional applications and future research directions are summarized.

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The present review is organized into six main sections. The first section contains a broad introduction to the generalities of AgNbO<sub>3</sub>-based ceramics and their importance within the lead-free perovskite family. Particular attention is given to the peculiar shapes of the polarization–electric field hysteresis loops, which represent one of the most fascinating aspects of these materials. The second section is dedicated to the synthesis, structure and relevant properties of AgNbO<sub>3</sub>. In particular, the initial part of this section is focused on the procedures used to fabricate bulk ceramics, single crystals and thin films and the main issues encountered in obtaining high quality products. This part is followed by an extensive discussion on the ongoing debate concerning the actual nature of the

crystal structure of AgNbO<sub>3</sub> at room temperature, during which ferroelectric, ferrielectric and antiferroelectric structures are compared and contrasted against each other. After that, a detailed survey on the polymorphic nature of AgNbO<sub>3</sub> is presented, by summarizing the current understanding of the temperature-induced structural modifications, in relation to the anomalies found in the temperature dependence of the dielectric properties. The following part is dedicated to the shapes of the current–polarization–electric field hysteresis loops and their relationships with the electric field-induced structural transformations that take place during the application of an external electric field. The second section is concluded with a survey on the photocatalytic and photovoltaic behavior of AgNbO<sub>3</sub>. The third section is focused on the variations of crystal structure, and dielectric, ferroelectric and hysteresis loop characteristics, induced by chemical modifications using iso- and aliovalent elements. This helps outline general compositional design criteria aimed at obtaining desired properties. Sections four and five are respectively dedicated to two main applications, namely high-power energy storage and photocatalysis. The comparison of key properties obtained in different AgNbO<sub>3</sub>-based systems provides additional information on composition and processing optimization to achieve maximized properties. The review is concluded with a broad summary and outlook presented in section six, in which the main findings are concisely described to increase the focus. At the same time, open questions, outstanding issues, and future research directions that could possibly fill the current knowledge gap are discussed.

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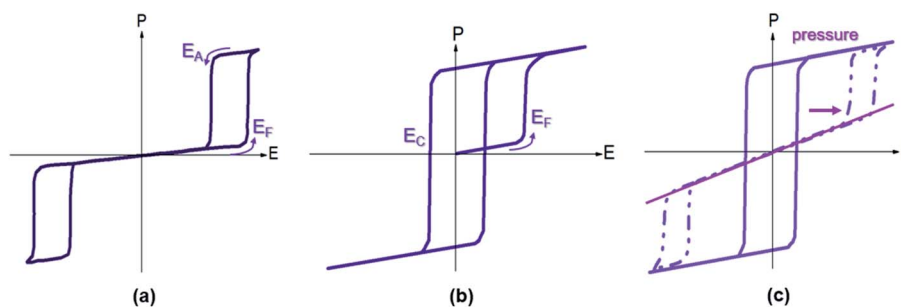


Fig. 1 Polarization vs. electric field loops for an antiferroelectric: (a) double polarization hysteresis loop for a reversible antiferroelectric-ferroelectric (AFE-FE) transition, (b) first-poling of an antiferroelectric displaying an irreversible AFE-FE transition, and (c)  $P$ - $E$  loop evolution under increasing *in situ* pressure (indicated by arrow), showing a pressured-induced transition from metastable FE to AFE.

## 1. Introduction

Pb-containing antiferroelectric compounds such as lead zirconate can transform into ferroelectrics when they are subjected to an electric field beyond a certain threshold. During these electric-field-driven phase transitions, various properties (such as polarization and strain) exhibit significant changes.<sup>1-3</sup> Chemically-modified Pb-based antiferroelectric materials mainly display two types of field-induced transition process.<sup>1,2</sup> The first type is characterized by reversible double polarization-electric field hysteresis loops (Fig. 1a), while the second is distinguished by “ferroelectric-like” hysteresis loops, which are observed after an irreversible transition has taken place during the first electrical loading cycle, as shown in Fig. 1b.

Double hysteresis loops are characterized by high charge storage density, which can be released over very short times, even in nanoseconds.<sup>4</sup> Ferroelectric-like hysteresis loops can become “double” or even “linear” under pressure or shock pressure waves (Fig. 1c),<sup>3</sup> and the charge stored in the electric-field-induced metastable ferroelectric state can be instantly released. These properties have allowed for the widespread application of Pb-containing antiferroelectric functional oxides in energy storage technologies, such as pulse power electronic devices,<sup>5</sup> and in energy conversion systems, such as explosive power exchange.<sup>3,6</sup> Moreover, the electric-field-induced anti-/ferroelectric phase transition determines the on-off properties

and can provide tunable dielectric, piezoelectric, pyroelectric, and electrocaloric properties by varying the magnitude of the applied electric-field. These multifunctional properties make Pb-based antiferroelectrics useful materials for applications in intelligent sensors, actuators, infrared detectors, and solid-state refrigeration systems.<sup>7-9</sup> However, Pb-containing electronic products are being progressively eliminated from these applications, due to increasingly stricter environmental legislation.<sup>10</sup> Thus, in recent years, research has focused on the search for alternative Pb-free materials.

Amongst Pb-free perovskites, silver niobate ( $\text{AgNbO}_3$ ) shows a “peculiar” response to the application of an electric field, that is different to that shown by antiferroelectric lead-based materials, such as  $\text{PbZrO}_3$ .<sup>11-14</sup> It was reported that  $\text{AgNbO}_3$  shows “ferroelectric-like” hysteresis loops under weak electric-fields, as shown in Fig. 2a,<sup>14,15</sup> meanwhile, it displays classic “double” hysteresis (similar to  $\text{PbZrO}_3$ ) at larger electric fields, as shown in Fig. 2b.<sup>13,16</sup>

The discovery of “double” polarization hysteresis under strong electric fields indicated that  $\text{AgNbO}_3$  could potentially replace Pb-based antiferroelectrics in energy storage applications. Based on chemical composition adjustments, the field-induced phase transitions and the associated physical properties can be optimized. Thus, chemically-modified  $\text{AgNbO}_3$ -based materials could not only meet the stringent property requirements for energy storage and energy exchange devices, but also overcome the environmental drawbacks of lead-based materials. In 2016, our research team reported that  $\text{AgNbO}_3$  ceramics show great potential in energy storage applications due to the high recoverable energy density,<sup>15</sup> triggering increased research interest in this material. The energy storage capability in chemically-modified  $\text{AgNbO}_3$  ceramics has been subsequently improved;<sup>17</sup> it is now triple the value first reported for the unmodified composition, and even outperforms many Pb-based antiferroelectrics.

Although a large number of research studies on  $\text{AgNbO}_3$  have been reported since its discovery in 1958, comprehensive reviews on this system are still rare. Early publications dating back about 20 years ago proposed that  $\text{AgNbO}_3$  could be useful for piezoelectric and microwave applications.<sup>18,19</sup> Additionally, following the seminal work on semiconductor-based

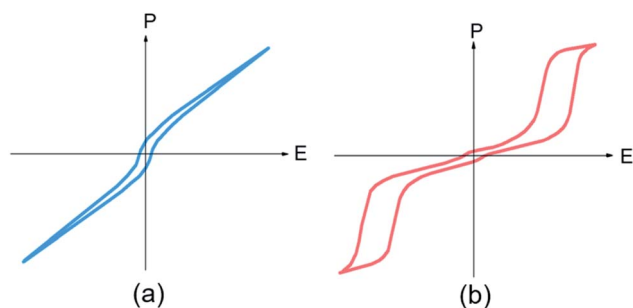


Fig. 2 Schematic polarization-electric field hysteresis loops for  $\text{AgNbO}_3$  ceramic under (a) weak-field ( $\leq 20 \text{ kV cm}^{-1}$ ) and (b) strong-field ( $\geq 170 \text{ kV cm}^{-1}$ ).<sup>15,16</sup>



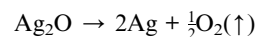
photocatalytic technology in 1972,<sup>20</sup> studies have also reported that AgNbO<sub>3</sub> exhibits attractive visible-light photocatalytic activity related to its narrow band gap ( $\sim 2.8$  eV) that could be suitable for producing hydrogen and for degrading water pollutants.<sup>21</sup> More recently, the photovoltaic properties of AgNbO<sub>3</sub> have also been explored for potential solar cell applications.<sup>22</sup> Therefore, it is timely to comprehensively review the past and present understanding of AgNbO<sub>3</sub> and AgNbO<sub>3</sub>-based materials to guide future development. The scope of this article is to review the synthesis, crystalline structure, microstructure and properties of these systems. In particular, the structure–property relationships of AgNbO<sub>3</sub>-based perovskites are described and recent progress on AgNbO<sub>3</sub>-based materials for high-power energy storage and photocatalysis is extensively discussed. The review is concluded with the proposal of future research, which could further improve the understanding and boost the applications of silver niobate-based materials.

## 2. Silver niobate (AgNbO<sub>3</sub>)

### 2.1 Synthesis

In the binary phase diagram of Ag<sub>2</sub>O–Nb<sub>2</sub>O<sub>5</sub>, there are five confirmed compounds with different Ag<sub>2</sub>O : Nb<sub>2</sub>O<sub>5</sub> stoichiometric ratios. Besides the perovskite-structured AgNbO<sub>3</sub> (1 : 1),<sup>22</sup> the other confirmed phases are the layer-structured natrotantite Ag<sub>2</sub>Nb<sub>4</sub>O<sub>11</sub> (1 : 2),<sup>23</sup> the tungsten bronze-structured AgNb<sub>3</sub>O<sub>8</sub> (1 : 3),<sup>24</sup> and the AgNb<sub>7</sub>O<sub>18</sub> (1 : 7)<sup>25</sup> and AgNb<sub>13</sub>O<sub>33</sub> (1 : 13)<sup>26,27</sup> compounds. It has been reported that these other compounds have similar properties to AgNbO<sub>3</sub>, also exhibiting anti-/ferroelectricity and photocatalytic activity.<sup>23,25</sup> Although these compounds can be synthesized *via* conventional solid-state reaction routes, it has been widely experienced that synthesizing single phase AgNbO<sub>3</sub> in air by solid-state reaction is very difficult, due to the thermodynamic instability of Ag<sup>+</sup> ions at elevated temperatures. To overcome this problem, the synthesis process is often carried out in an atmosphere of pure

oxygen.<sup>13</sup> The successful synthesis of AgNbO<sub>3</sub> in pure O<sub>2</sub> atmosphere using Ag<sub>2</sub>O and Nb<sub>2</sub>O<sub>5</sub> as raw materials has been reported in several publications.<sup>15–20</sup> During heating up to 200 °C, there is a slow reduction of Ag<sub>2</sub>O:<sup>28</sup>



The thermogravimetric curves of metallic Ag, Ag<sub>2</sub>O and Ag<sub>2</sub>CO<sub>3</sub> as different silver sources in the formation of AgNbO<sub>3</sub> indicate that all these Ag precursors can react with Nb<sub>2</sub>O<sub>5</sub>, to finally form AgNbO<sub>3</sub>, as shown in Fig. 3.

By analyzing the three curves, it can be found that Ag<sub>2</sub>O and Ag<sub>2</sub>CO<sub>3</sub> go through a rather fast decomposition, leading to the formation of metallic Ag and O<sub>2</sub> at *ca.* 450 °C. On further heating, the metallic Ag is re-oxidized and reacts with Nb<sub>2</sub>O<sub>5</sub> to form AgNbO<sub>3</sub> at *ca.* 800 °C, which melts and decomposes at *ca.* 1140 °C. The kinetics of the chemical reaction using metallic Ag are much slower than that using Ag<sub>2</sub>O/Ag<sub>2</sub>CO<sub>3</sub> precursors. The temperature for the complete formation of AgNbO<sub>3</sub> using metallic Ag is *ca.* 1060 °C. From Fig. 3, it is evident that only part of the Ag content is reduced in the initial stages of the reactions involving Ag<sub>2</sub>O and Ag<sub>2</sub>CO<sub>3</sub> and that the subsequent re-oxidation stage is completed more rapidly than in the case of the reaction with pure Ag, where the oxidation of the entire Ag content is required. An alternative strategy to synthesize pure AgNbO<sub>3</sub> without O<sub>2</sub> atmosphere is represented by wet chemical methods. It was reported that pure AgNbO<sub>3</sub> powders can be synthesized *via* hydrothermal processing at 160–240 °C using analytical grade NH<sub>4</sub>HF<sub>2</sub>, Nb<sub>2</sub>O<sub>5</sub>, and Ag<sub>2</sub>O as raw materials in an acid solution (pH = 3).<sup>29,30</sup>

Other chemical methods, such as sol-gel and co-precipitation have also been developed, using Ag and Nb alkoxides or citrates as precursors, with a firing step performed at 650 or 800 °C that burns off the organic ligands and leads to the formation of AgNbO<sub>3</sub>.<sup>31</sup> The color of the AgNbO<sub>3</sub> synthesized powders is usually milk-like or faint yellow, but it becomes darker, from red to gray, under visible-light.

AgNbO<sub>3</sub> single crystals can be grown *via* the molten-salt method using NaCl or V<sub>2</sub>O<sub>5</sub> as flux agents.<sup>32,33</sup> Using flux methods, recently, Zhao *et al.* have successfully grown AgNbO<sub>3</sub> single crystals using Ag<sub>2</sub>O (99.9%), Nb<sub>2</sub>O<sub>5</sub> (99.99%), and V<sub>2</sub>O<sub>5</sub> (99.2%) with a molar ratio of 7.4 : 1 : 4 in the starting mixture.<sup>34</sup> Polycrystalline bulk ceramics can be obtained by sintering synthesized AgNbO<sub>3</sub> powders at high temperature (usually from 1060 to 1120 °C) in O<sub>2</sub> atmosphere.<sup>28,35</sup> The prepared AgNbO<sub>3</sub> ceramics are generally yellowish in color. It should be noted that the decomposition temperature of AgNbO<sub>3</sub> is *ca.* 1120 °C, which is very close to the densification temperature of the ceramics, making the sintering of high density AgNbO<sub>3</sub> ceramics rather challenging. Therefore, precise temperature control is essential to prepare high-quality AgNbO<sub>3</sub> bulk ceramics. Generally, the melting or decomposition of AgNbO<sub>3</sub> ceramics begins at the surface when the sintering temperature is too high, and leads to the appearance of impurity phases, such as Ag<sub>2</sub>Nb<sub>4</sub>O<sub>11</sub> and AgNb<sub>3</sub>O<sub>8</sub>, amongst others. These impurities are usually confined within the surface layers and can be removed *via* a fine

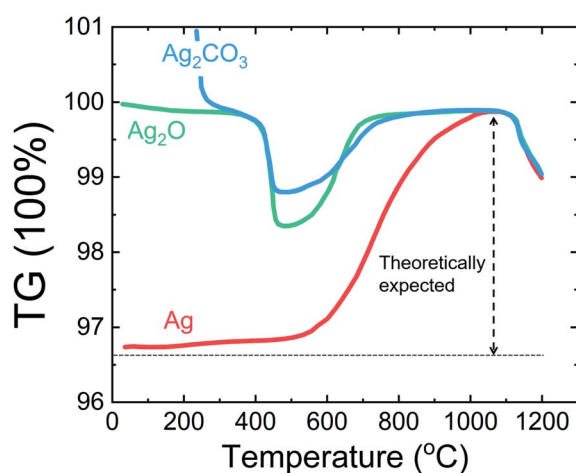


Fig. 3 Thermogravimetric curves of mixtures of Nb<sub>2</sub>O<sub>5</sub> with different silver sources: metallic Ag powder, Ag<sub>2</sub>O, and Ag<sub>2</sub>CO<sub>3</sub>.<sup>28</sup> [Reprinted by permission from Springer Nature: Journal of Materials Research, copyright (2007)].





grinding process. By contrast, rough grinding would lead to the modification of the original lattice and to a number of artificial defects originating from the unstable chemical structure. Recently, it has been reported that Ag quantum-dots/nanoparticles are formed on  $\text{AgNbO}_3$  after irradiation with a Xe lamp.<sup>36</sup> Theoretical modelling and simulations have shown that due to Ag deficiency, Schottky defect clusters consisting of two Ag vacancies and an O vacancy (*i.e.*  $[2V'_{\text{Ag}} + V''_{\text{O}}]$  defect complex using the Kroger-Vink notation) can be formed in  $\text{AgNbO}_3$  perovskites.<sup>37</sup> Kania *et al.* investigated the influence of Ag deficiency on microstructure and found that a few percent excess of  $\text{Ag}_2\text{O}$  in the starting composition can significantly increase the homogeneity of the perovskite structure and can thus improve the quality of  $\text{AgNbO}_3$  ceramics.<sup>38</sup> On the other hand, a large excess of  $\text{Ag}_2\text{O}$  in the starting composition leads to the presence of metallic Ag. To avoid Ag deficiency at high temperature, Kitanaka *et al.* attempted to grow  $\text{AgNbO}_3$  single crystals using slow cooling and Czochralski methods in a high-pressure  $\text{O}_2$  atmosphere. They found that a high oxygen pressure can significantly prevent Ag deficiency, especially in the Czochralski method.<sup>39</sup> The materials used in high-power energy storage systems require high electrical insulation under strong field conditions and large electrical breakdown strength ( $E_b$ ). The  $E_b$  of unmodified  $\text{AgNbO}_3$  ceramics prepared by the solid-state method in pure  $\text{O}_2$  atmosphere is  $200 \pm 20 \text{ kV cm}^{-1}$ .<sup>13,15,16</sup> The effects of Ag deficiency on electrical properties were also reported in our recent publication, where small amounts of Ag deficiency favor increased stability of the AFE state.<sup>40</sup>

Due to the demand for miniaturized electronic devices, the preparation of  $\text{AgNbO}_3$  thin films has also been explored over the past ten years, particularly for microwave applications. The two primary deposition techniques used to prepare  $\text{AgNbO}_3$  films include chemical solution deposition (CSD) and pulsed laser deposition (PLD). Telli *et al.* showed that  $\text{AgNbO}_3$  thin films can be grown on (001)  $\text{SrRuO}_3$ /(001)  $\text{LaAlO}_3$  substrates by CSD.<sup>41</sup> Nevertheless, thin films prepared using this method showed significantly distinct dielectric properties compared to bulk ceramics. Kim *et al.* reported that  $\text{AgNbO}_3$  films can be grown on  $\text{LaAlO}_3$  (001) and sapphire ( $\text{Al}_2\text{O}_3$ -0112, *r*-cut) single crystals substrates by PLD, from stoichiometric  $\text{AgNbO}_3$  ceramic targets.<sup>42</sup> Using PLD technology,<sup>43</sup> thin films grown on  $\text{SrTiO}_3$  (STO) single crystal substrates with different orientation can exhibit ultrahigh  $E_b$  values ( $\sim 400 \text{ kV cm}^{-1}$ ). Furthermore, it is interesting to note that  $\text{AgNbO}_3$  films grown on (001) STO exhibited “double” polarization hysteresis, while  $\text{AgNbO}_3$  films deposited on (110) and (111) STO substrates showed a “ferroelectric-like” polarization hysteresis under strong-field cycling. However, the authors did not perform a structure investigation to clarify the reasons for this difference. More recently, Zhang *et al.* grew  $\text{AgNbO}_3$  epitaxial films on (001) STO substrates using the PLD method.<sup>44</sup> The film showed double polarization hysteresis with a higher AFE-FE phase transition field and a higher  $E_b$  value ( $624 \text{ kV cm}^{-1}$ ), indicating that such methodology may be viable for the fabrication of  $\text{AgNbO}_3$  with high breakdown strength.

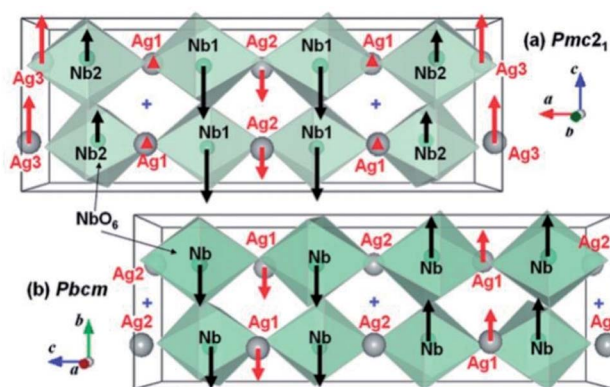


Fig. 4 Refined structure models of non-centrosymmetric  $Pmc2_1$  and centrosymmetric  $Pbcm$  phases. Red and black arrows indicate the displacements of  $\text{Ag}^+$  and  $\text{Nb}^{5+}$  ions, respectively, from the center of the bonded oxygen positions. The “cross” symbol indicates the center of symmetry in the  $Pbcm$  structure.<sup>49</sup> [Reproduced with permission from ref. 49 copyright 2011 American Chemical Society].

## 2.2 Crystal structure

The exact crystallographic structure of  $\text{AgNbO}_3$  is still under debate. Initially, it was suggested that  $\text{AgNbO}_3$  is ferroelectric at room temperature with a tiny “ferroelectric-like” polarization observed under a weak electric field of  $17 \text{ kV cm}^{-1}$  (remnant polarization  $P_r \approx 0.04 \mu\text{C cm}^{-2}$ ) and a low piezoelectric response (piezoelectric coefficient  $d_{33} \approx 0.24 \text{ pC N}^{-1}$ ).<sup>14,45</sup> Subsequent investigations using powder diffraction techniques, including X-ray diffraction (XRD) and neutron diffraction (ND), demonstrated that  $\text{AgNbO}_3$  displays an antipolar ordering with antiparallel displacements of  $\text{Ag}^+$  and  $\text{Nb}^{5+}$  in the centrosymmetric space group,  $Pbcm$ ,<sup>46–48</sup> which is inconsistent with its electrical response. To resolve this ambiguity, in 2011, Yashima *et al.* carried out a comprehensive investigation by analyzing data generated by convergent beam electron diffraction (CBED), ND and synchrotron XRD, and suggested a weak polar or ferroelectric (FIE) structure, modelled with the non-centrosymmetric space group,  $Pmc2_1$ , where both  $\text{Nb}^{5+}$  and  $\text{Ag}^+$  cations show antipolar behavior. Since the unit-cell showed a weak net polarization,  $\text{AgNbO}_3$  was categorized as a “non-compensated AFE”, which is distinct from typical AFEs.<sup>49</sup> In 2012, based on theoretical calculations, Niranjana *et al.*<sup>50</sup> revealed the possibility that  $Pmc2_1$  and  $Pbcm$  phases may coexist, as supported by the small difference between their free energies (only  $0.1 \text{ meV f.u.}^{-1}$ ).

Fig. 4 shows that the refined structure models of  $Pmc2_1$  and  $Pbcm$  space groups are very similar. Both structures show a structural distortion with a  $(a^-b^-c^+)/(a^-b^-c^+)$  tilting using Glazer's notation. Actually, distinguishing the two structures based on average structural analysis methods such as XRD and ND is very difficult, since both structural models can fit the experimental data very accurately and give satisfactory refinement results. The  $Pbcm$  space group is a super-group of  $Pb2_1m$ , with some classes of diffraction peaks theoretically present in the latter but systematically absent in the former.<sup>16</sup> For  $Pbcm$ , the full Hermann-Mauguin symbol is  $P 2/b 2_1/c 2_1/m$  generating



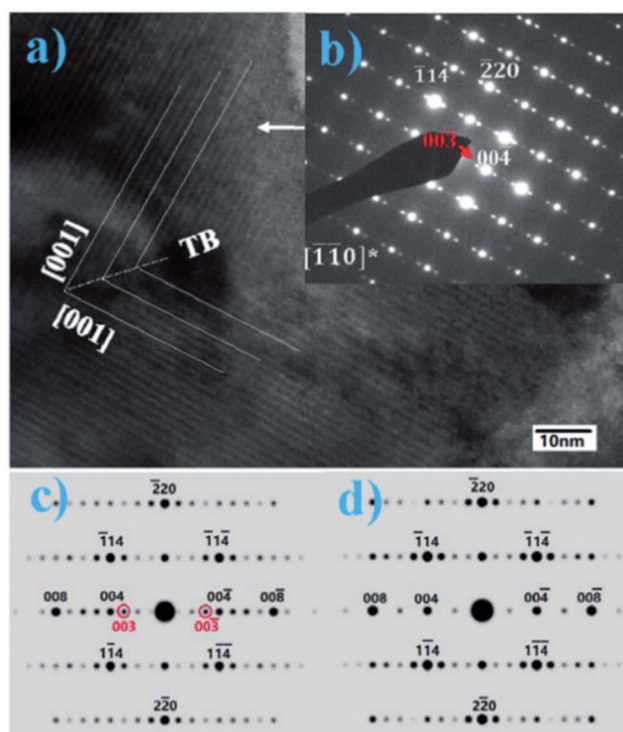


Fig. 5 (a) HREM and (b) observed SAED pattern at the  $[-1-10]$  zone axis; (c) and (d) calculated SAED patterns of  $Pb2_1m$  and  $Pbcm$  space groups, respectively.<sup>16</sup>

systematic absences of the type  $k = \text{odd}$  for  $0kl$ ,  $l = \text{odd}$  for  $h0l$  and hence  $k = \text{odd}$  for  $0k0$ , and  $l = \text{odd}$  for  $00l$  reflections. For  $Pb2_1m$  (transformed from  $Pmc2_1$  to ensure that the axial assignments are consistent with  $Pbcm$ ), the center of symmetry is lost as is much of the translational symmetry leaving only the  $b$ -glide perpendicular to the  $a$ -axis and a  $2_1$  screw axis parallel to the  $b$ -axis, to generate systematic absences of the type  $k = \text{odd}$  for  $0kl$  and  $k = \text{odd}$  for  $0k0$ . Thus, the  $Pb2_1m$  structure can be distinguished by the presence of  $00l$  or  $h0l$  reflections where  $l$  is odd. Indeed in our previous work,<sup>16</sup>  $00l$  diffraction spots with  $l = \text{odd}$  were found in selected area electron diffraction (SAED) patterns, while further tilting experiments confirmed that these were not due to double diffraction (Fig. 5).

The simulated electron diffraction (ED) patterns clearly indicate that the space group is  $Pb2_1m$  rather than  $Pbcm$ , consistent with the early structure investigation reported by Yashima *et al.*<sup>49</sup> Using spherical aberration correction methods, Li *et al.* obtained high-angle annular dark-field (HAADF) images of  $\text{AgNbO}_3$ , where the fast Fourier transform pattern, as well as the ED patterns, indicated that the structure tends to the non-polar  $Pbcm$  space group.<sup>51</sup> However, more recently, using a combination of high-resolution neutron and synchrotron X-ray powder diffraction, Farid *et al.* reported the structure as being polar in space group  $P2_1am$  (related to  $Pb2_1m$  by axial transformation).<sup>52</sup>

Second harmonic generation (SHG) experiments can be used to identify whether or not a structure has inversion symmetry. A very weak SHG signal was found in  $\text{AgNbO}_3$  powder and ceramic

samples, which disappeared on heating at around  $170^\circ\text{C}$ , indicating that at least part of the structure has inversion symmetry below this temperature.<sup>16</sup> Recently, Lu *et al.* systematically demonstrated the difference between the  $Pbcm$  and  $Pmc2_1$  structures in terms of crystallographic symmetry.<sup>53</sup> Each of the two distorted structures can be constructed *via* the coupling/competition between specific “softening” symmetry-modes within the first Brillouin-zone of the non-distorted parent structure. The scheme in Fig. 6 shows the structure distortion induced from the undistorted parent structure ( $Am\bar{m}m$ ) by specific symmetry-modes. The  $T4+$  and  $H2$  modes, located at the zone-boundary, can induce an  $R(\langle 110 \rangle_p)$ -type octahedral rotation (*i.e.*,  $(a^-a^-c^0)$  tilting) and an  $R(\langle 001 \rangle_p)$ -type octahedral rotation (*i.e.*,  $(a^0a^0c^-)/(a^0a^0c^+)$  tilting), respectively. The  $\Lambda 3$  mode is mainly associated with the off-centering displacement of cations, which can result in an anti-parallel alignment of dipoles. The  $\Gamma 4-$  mode, located at the zone-center, is an FE mode, which can induce all ions to move along a specific direction but with different magnitudes. The coupling of  $T4+$  and  $H2$  modes could induce a secondary  $\Lambda 3$  mode and finally construct the  $Pmca$  (axial transformation from  $Pbcm$ ) distorted structure, while an additional “softening” of the primary mode  $\Gamma 4-$  could lower the symmetry from  $Pmca$  to  $Pmc2_1$ . The global amplitude of the  $\Gamma 4-$  mode is far from the amplitude of the  $\Lambda 3$  mode, which is responsible for the observed weak-ferroelectricity in  $\text{AgNbO}_3$ .

### 2.3 Dielectric properties and temperature-driven transitions

Besides the controversy regarding its crystallographic structure at room temperature,  $\text{AgNbO}_3$  has also been reported to display a series of temperature-induced phase transitions between

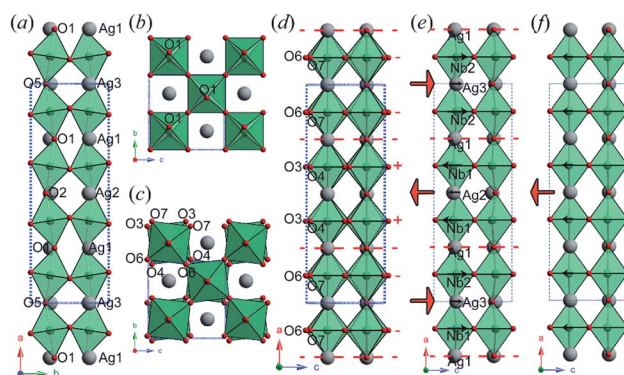


Fig. 6 (a) and (b) Distorted  $\text{AgNbO}_3$  (AN) structure induced by the  $T4+$  mode only, viewed along (a) the  $c$ -axis and (b) the  $a$ -axis. (c) and (d) Distorted AN structure induced by the  $H2$  mode only, viewed along (c) the  $a$ -axis and (d) the  $b$ -axis. The  $+/-$  signs on the right in panel (d) show clockwise/anticlockwise rotations, respectively, of the right-hand column of octahedra around the  $a$ -axis. (e) The distorted structure induced by the  $\Lambda 3$  mode only, and (f) that of the  $\Gamma 4-$  mode only. The black arrows in panel (e) show the off-center displacements of  $\text{Nb}^{5+}$  and  $\text{Ag}^+$  cations, while the red arrows in panels (e) and (f) indicate the local spontaneous polarization. The horizontal dashed red lines represent antiphase boundaries for octahedral rotation around the  $a$ -axis in panel (d), and cation displacements along the  $c$ -axis in panel (e).<sup>53</sup>



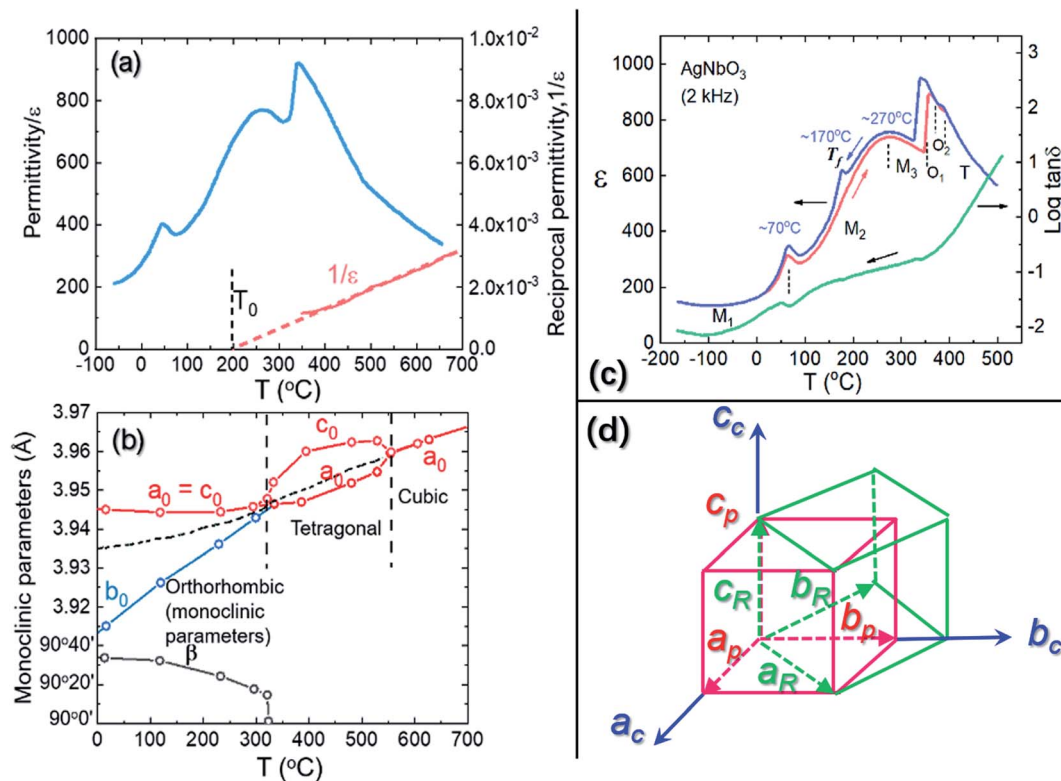
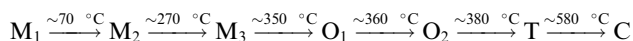


Fig. 7 Thermal dependence of (a) permittivity and (b) lattice parameters of AgNbO<sub>3</sub> reported by Francombe *et al.*;<sup>35</sup> (c) temperature dependence of dielectric permittivity and loss of AgNbO<sub>3</sub> reported by Kania *et al.*;<sup>54</sup> [reproduced from ref. 54 <https://doi.org/10.1088/0953-8984/11/45/316> with permission from copyright IOP Publishing]. (d) Schematic showing the axial relationship between orthorhombic symmetry in rhombic ( $a_R$ ,  $b_R$ ,  $c_R$ ) and parallel orientations ( $a_P$ ,  $b_P$ ,  $c_P$ ), respectively ( $a_C$ ,  $b_C$ ,  $c_C$  represent the axes of the pseudo-cubic unit cell).

different polymorphs, the details of which are still under debate. Generally, the structural transitions for non-linear dielectric materials (such as anti-/ferroelectrics) can be detected by measuring the temperature-dependence of dielectric properties and identifying the presence of dielectric anomalies that could originate from phase transitions. In 1958, Francombe *et al.* synthesized AgNbO<sub>3</sub> and studied its structure and dielectric properties; the relevant results of their study are reported in Fig. 7a and b.<sup>35</sup> In the plots, three dielectric anomalies at *ca.* 60, 260 and 340 °C can be observed; however, the structure analysis indicated that only the dielectric anomaly around *ca.* 340 °C relates to a phase transition (from orthorhombic to tetragonal structures). Furthermore, the structure analysis indicated that when the temperature approaches *ca.* 580 °C, a phase transition from the tetragonal to the cubic phase occurs, although no obvious dielectric anomaly can be found around this temperature (Fig. 7a).

With the improvement of characterization equipment, Kania *et al.* re-investigated the relationship between polymorphic structure and dielectric properties and proposed the following polymorphic phase transition sequence during heating<sup>54</sup> (Fig. 7c):



M<sub>1</sub>, M<sub>2</sub> and M<sub>3</sub> denote phases with orthorhombic (O) symmetry in rhombic orientation. The T and C phases have tetragonal and cubic symmetries, respectively. O<sub>1</sub> and O<sub>2</sub> represent phases with orthorhombic symmetry in parallel orientation. A schematic diagram of the O symmetry with different orientations is shown in Fig. 7d. The transitions between the high-temperature M<sub>3</sub>, O (O<sub>1</sub> and O<sub>2</sub>), T and C phases are related to the thermal evolution of octahedral tilting,<sup>55,56</sup> as observed by XRD<sup>35,55,57</sup> and TEM<sup>56</sup> experiments. However, no evidence of structural change can be observed in XRD or TEM for the O<sub>1</sub> ↔ O<sub>2</sub> transition. It should be noted that the M-type phase transitions were only ascertained *via* analyzing *in situ* diffraction data based on the primitive perovskite pseudo-cubic unit-cell using monoclinic parameters (hence their designation as “M” phases). The changes in structural parameters that distinguish the three M phases are mostly consistent with the dielectric anomalies around *ca.* 70 °C and *ca.* 270 °C. Additionally, dielectric spectroscopy also revealed a peculiar sharp anomaly at *ca.* 170 °C which appeared only in the cooling cycle and was hardly detected during heating, as shown in Fig. 7c. This was attributed to the dipole freezing temperature, T<sub>f</sub>. During the 1980s, it was believed that the M<sub>1</sub> phase was ferroelectric, while the M<sub>2</sub> and M<sub>3</sub> phases were antiferroelectric, and the O<sub>1</sub>, O<sub>2</sub>, T and C phases were paraelectric. With the help of neutron diffraction, the symmetries of the high-temperature phases O, T and C have been

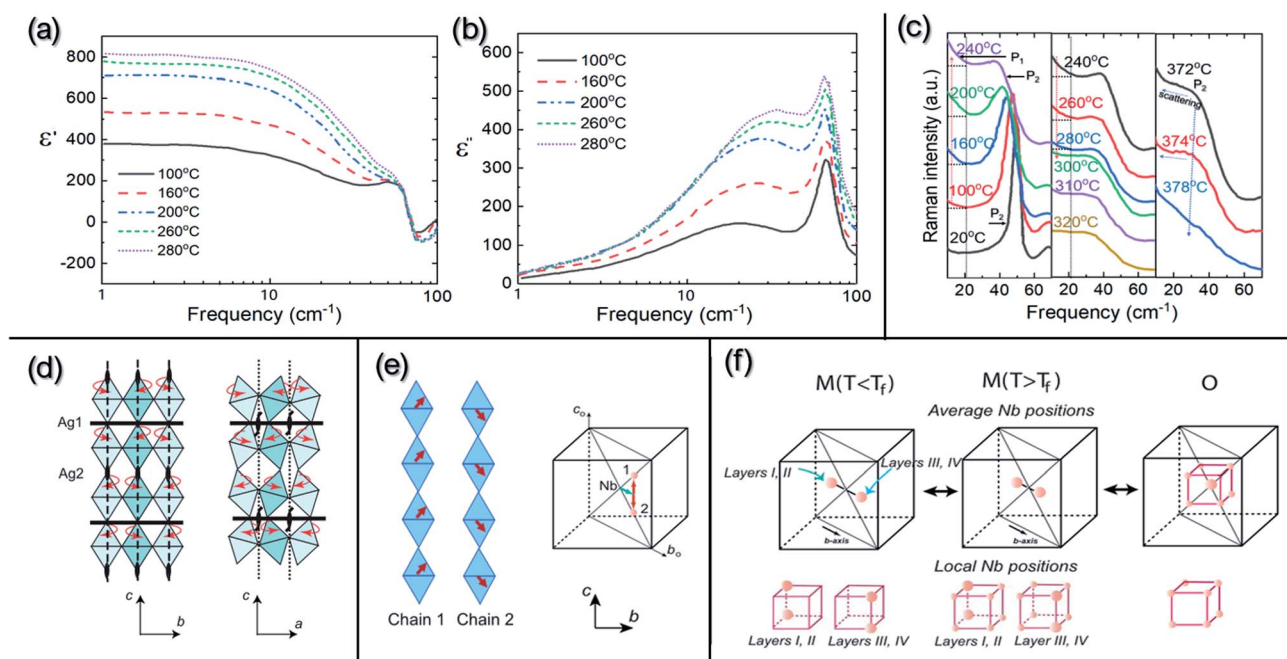




better understood and their space groups have been identified as  $Cmcm$ ,  $P4/mbm$  and  $Pm\bar{3}m$ , respectively.<sup>47</sup> However, the exact symmetry of the M phases, the dielectric anomalies related to M-type phase transitions, as well as the so-called freezing temperature ( $T_f$ ) are still under debate. Based on dielectric studies combined with infrared and Raman spectroscopy results reported in early publications,<sup>58–60</sup> the dielectric anomalies have been related to the different evolutionary stages of displacive  $Nb^{5+}$  ion dynamics, evidenced by the presence of submillimeter relaxation modes in the dielectric spectra shown in Fig. 8a and b, and the center component ( $P_1$  peak) in Raman spectra shown in Fig. 8c.

These connections have been further explored in a systematic structure investigation reported by Levin *et al.* in 2009.<sup>48</sup> Based on various techniques that allow for probing of the local structure, such as neutron pair distribution functions (PDFs) and extended X-ray absorption fine structure (EXAFS), local structural distortions upon heating were detected and correlated to the order–disorder behavior of displacive  $Nb^{5+}$  cations in the octahedral framework of the non-polar  $Pbcm$  matrix. The octahedral framework in the  $Pbcm$  symmetry of  $AgNbO_3$

observed from the  $c$ – $b$  and  $c$ – $a$  planes is shown in Fig. 8d, in which a sequence of two in-phase and two antiphase octahedral rotations about the  $c$ -axis yields a  $4a_c$  periodicity with two crystallographically distinct Ag sites. The chain-like correlation for the local off-centering Nb displacements within two adjacent octahedral columns directed along  $[111]_c$  and  $[11\bar{1}]_c$  can give rise to the average Nb displacements along the  $[110]_c$  direction (*i.e.*, the  $b$ -axis of the orthorhombic  $Pbcm$  structure) as shown in Fig. 8e. At high temperature, the  $Nb^{5+}$  cations are randomly displaced among eight sites along the  $\langle 111 \rangle_c$  directions, which give the ideal positions on average in the O phase. Upon cooling, the off-centered ordered arrays of  $Nb^{5+}$  cations displacement along the  $[111]_c$  and  $[11\bar{1}]_c$  directions in two adjacent octahedral layers gives an average displacement component along the  $[110]_c$  direction (*i.e.*, along the  $b$ -axis of the orthorhombic  $Pbcm$  structure). In the two next adjacent layers, these average displacement components point in opposite directions, leading to AFE structural order. Locally,  $Nb^{5+}$  cations still occupy these eight positions, but two of these locations are preferred. Upon further cooling, the occupancy probabilities for the remaining six sites decrease and vanish below  $T_f$ , leading to



**Fig. 8** (a) Real ( $\epsilon'$ ) and (b) imaginary ( $\epsilon''$ ) parts of the dielectric permittivity as functions of temperature for  $AgNbO_3$  in the 1–100  $cm^{-1}$  frequency range (data as reported by Fortin *et al.*)<sup>60</sup> reproduced from ref. 60 <https://doi.org/10.1063/1.361796> with the permission of AIP Publishing; (c) temperature dependence of the low-frequency Raman spectra of  $AgNbO_3$  (data as reported by Kania *et al.*)<sup>61</sup> reproduced from ref. 61 <https://doi.org/10.1088/0022-3719/19/1/007> with permission from copyright IOP Publishing (note: only Stokes Raman scattering is re-plotted in (c)). The spectra show an intense low-frequency phonon peak ( $P_2$ ) and a strong central component ( $P_1$ ), the intensity of which clearly rises with temperature and reaches a maximum at around 240 °C (corresponding to the  $M_2$ – $M_3$  transition). The phonon peak ( $P_2$ ) shifts to lower frequency with increasing temperature and abruptly disappears around 378 °C (corresponding to the O–T transition). The diffuse peak,  $P_1$ , above 320 °C, is caused by the scattering of the initial phonon peak; (d) schematic of the octahedral framework within the non-polar  $Pbcm$  symmetry of  $AgNbO_3$ , in which octahedral rotations about the  $c$ -axis are indicated using arrows, while Ag1 and Ag2 specify  $c$ -planes occupied by the symmetrically non-equivalent Ag cations; (e) (left) schematic of the chain-like correlations for the local Nb displacements directed along  $[111]_c$  and  $[11\bar{1}]_c$  directions; (e) (right) a random mixture of chains 1 and 2 equivalent to average Nb displacements along the  $[110]_c$  direction (*i.e.*,  $b$ -axis of orthorhombic crystal); (f) schematic for the  $Nb^{5+}$  displacive order–disorder behavior as a function of temperature.<sup>48</sup> [Reprinted with permission from ref. 48 <https://doi.org/10.1103/PhysRevB.79.104113> copyright 2003 by the American Physical Society].

a partial long-range ordered antipolar-like array of  $\text{Nb}^{5+}$  cations and to the observed sharp dielectric response at  $T_f$ . Miga *et al.*<sup>62</sup> suggested that further freezing of the displacive  $\text{Nb}^{5+}$  and  $\text{Ag}^+$  sites on sub-lattice scales could lead to a weakly-polar relaxor ferroelectric or dipolar glass transition corresponding to the  $M_1 \leftrightarrow M_2$  transition. Such disorder–order behavior of displacive cations has been used to successfully explain the polymorphic ferroelectric transitions in  $\text{KNbO}_3$  and  $\text{BaTiO}_3$ , as reported in early studies.<sup>63–65</sup> In these cases, due to the absence of octahedral tilting, the partial ordering of off-center  $\text{Nb}^{5+}/\text{Ti}^{4+}$  displacements, on cooling, triggers a sequence of well-defined thermodynamic phase transitions accompanied by changes in symmetry (*i.e.*, tetragonal  $\rightarrow$  orthorhombic  $\rightarrow$  rhombohedral).

It should be noted that the local structural distortions that induce the dielectric anomalies separating the M-type polymorphs of  $\text{AgNbO}_3$  suggested by Levin *et al.*<sup>48</sup> are based on the lattice matrix with  $Pbcm$  symmetry. However, in 1983, Kania *et al.* reported that the weak ferroelectricity in  $\text{AgNbO}_3$  disappeared during heating at *ca.* 70 °C.<sup>14</sup> After discovering this weak ferroelectricity, researchers preferred to link the  $M_1 \leftrightarrow M_2$  phase transition to a structural change between the  $Pmc2_1$  and  $Pbcm$  symmetries. This phase transition should be accompanied by a soft mode at the first Brillouin zone center, but Raman scattering does not offer any evidence for such an effect near  $\sim 70$  °C (see Fig. 8c). Instead, the central component corresponding to  $P_1$  in Fig. 8c presents more obvious changes as a function of temperature and it progressively disappears approaching *ca.* 300 °C (in the  $M_3$  region). This feature is closely related to the  $\text{Nb}^{5+}$ -ion dynamics. In other words, the  $M_1 \leftrightarrow M_2$  phase transition is possibly related to local structural variations or to an improper ferroelectric transition as defined by Dvorak (*i.e.*, the primary driving force inducing this transition is not the polarization order parameter).<sup>66</sup> Based on *in situ* high-energy X-ray diffraction studies, Yoneda *et al.* provided additional insights into the temperature-driven polymorphic nature of the so-called M-phases.<sup>67</sup> In this study, the authors found that the  $\text{Ag}^+$  ion shows displacive disorder behavior. This discovery indicated the existence of an intricate coupling between octahedral tilting and local displacements of  $\text{Ag}^+$  ions since the octahedral tilting can usually lead to the off-centering of A-site cations within the  $\text{AO}_{12}$  cage of the perovskite structure. The authors suggested that these interactions trigger a series of changes in the cation displacements across the so-called M phases, being similar to the situation in Pb-based solid-solutions near the morphotropic phase boundary (MPB). In

Pb-based perovskites, three kinds of monoclinic phases ( $M_A$ ,  $M_B$ , and  $M_C$ ) can be formed due to the displacive ordering of cations shifting along specific directions (*i.e.*, polarization rotation) from the parent tetragonal phase, as shown in Fig. 9.<sup>68</sup> Considering  $\text{PbZr}_{0.48}\text{Ti}_{0.52}\text{O}_3$  as an example, the monoclinic structure can be derived from the tetragonal structure by shifting the Pb and Zr/Ti cations along the tetragonal [101] axis. The monoclinic phase in Pb-based solid-solutions provides a bridge between the ground states of rhombohedral and tetragonal structures.

However, early ND studies on  $\text{AgNbO}_3$  provided no evidence for a ground state with rhombohedral symmetry, even at low temperatures down to 1.5 K.<sup>47</sup> Nevertheless, a ground state rhombohedral phase can be found in  $(\text{Ag}_{0.94}\text{Li}_{0.06})\text{NbO}_3$ , which suggests that a rhombohedral-like structure may be buried in the local disorder of Ag atoms of the undoped  $\text{AgNbO}_3$ .<sup>69</sup> In 2014, Zhang *et al.*<sup>70</sup> reported the occurrence of an additional dielectric anomaly in the low-frequency range around 250 K in  $\text{AgNbO}_3$  ceramics related to a first-order phase transition (denoted as  $M_0 \leftrightarrow M_1$ ), as also supported by the presence of sharp anomalies in the differential scanning calorimetry (DSC) thermogram (see ref. 70, and Fig. 4 and 5). It should be noted that this ceramic was prepared using an  $\text{AgNbO}_3$  powder synthesized by a hydrothermal method.<sup>71</sup> Kania *et al.* carefully re-measured the DSC and dielectric properties of  $\text{AgNbO}_3$  ceramics and single crystals, and found results to be consistent with previous investigations with no additional events found around 250 K in the dielectric and thermal data.<sup>72</sup> Kania *et al.*<sup>72</sup> argued that the extra dielectric anomaly reported by Zhang *et al.*<sup>70</sup> is more likely to originate from the increasing electrical conductivity around 250 K rather than an intrinsic dielectric process of  $\text{AgNbO}_3$ , indicating that water contamination is not negligible and that its prevention is necessary during experiments to avoid misinterpretation. Recently, our research team compared the variation of the temperature-dependent properties of unpoled samples including dielectric, DSC, dynamic mechanical analysis (DMA), as well as SHG data as shown in Fig. 10.<sup>16</sup> It was found that the dielectric anomaly at the freezing temperature  $T_f$  (see Fig. 7c),<sup>52</sup> also appeared in the dielectric data on heating, suggesting that this dielectric anomaly might be related to a temperature-driven reversible transition.

The comprehensive analysis of the properties shown in Fig. 10 indicated that the dielectric anomaly at “ $T_f$ ” might not be associated with the freezing of Nb-ion dynamics, but could rather be ascribed to a second-order structural transition

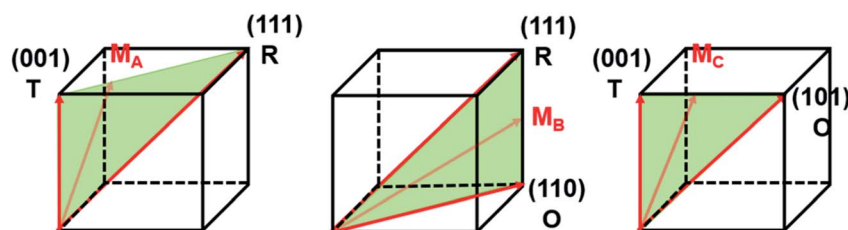


Fig. 9 Direction of the spontaneous polarization in the three types of monoclinic phases with the usual nomenclature  $M_A$ ,  $M_B$  ( $Cm$ ) and  $M_C$  ( $Pm$ ) redrawn, according to ref. 68.





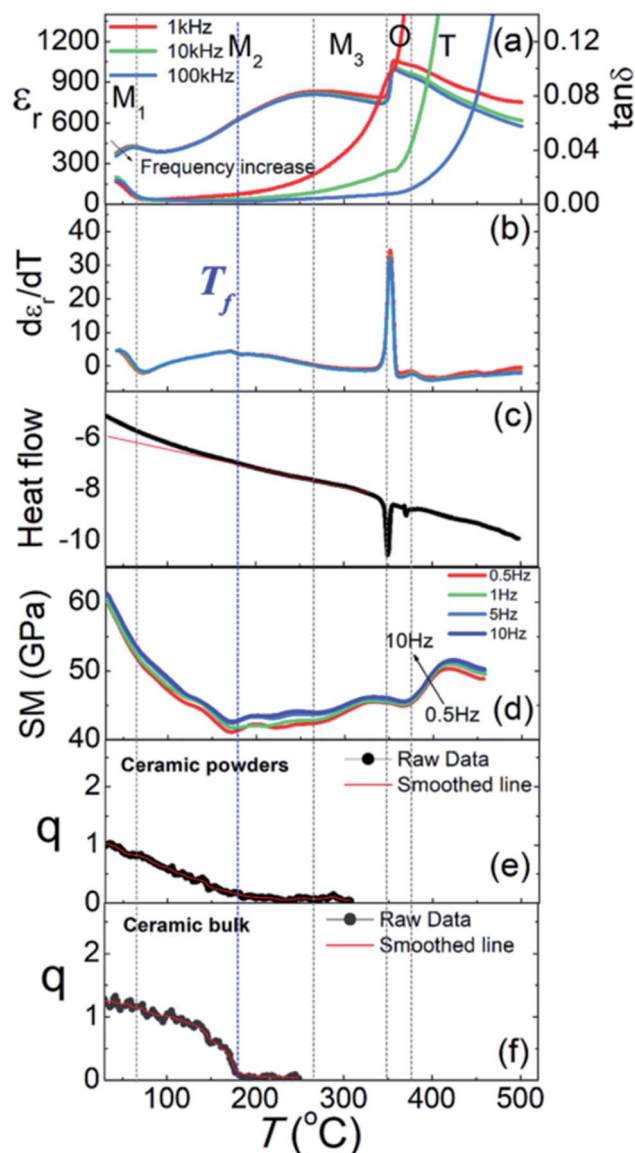


Fig. 10 Thermal dependencies of (a) relative dielectric permittivity ( $\epsilon_r$ ) and dielectric loss tangent ( $\tan \delta$ ) at selected frequencies, (b)  $d\epsilon_r/dT$ , (c) DSC thermogram, (d) storage modulus (SM); (e) SHG for ceramic powders and (f) SHG for bulk ceramics (note:  $q = I_{2w}/I_{2w}(\text{SiO}_2)$ ) of  $\text{AgNbO}_3$ . All plots were generated during heating.<sup>16</sup>

related to the breaking of inversion symmetry (i.e.,  $\text{Pmc}2_1 \leftrightarrow \text{Pbcm}$ ). It is possible that the primary driving force of this transition is the spontaneous strain rather than the spontaneous polarization (i.e., an improper ferroelectric transition),<sup>66,73</sup> since a significant change in the storage modulus can be found at  $T_f$  (see Fig. 10d). Additionally, the dielectric behavior under a DC bias field is very similar to that of typical relaxor-ferroelectrics, such as  $\text{PbMg}_{1/3}\text{Nb}_{2/3}\text{O}_3$  (see ref. 16, Fig. 7). It could be reasonably deduced that the  $\text{M}_1 \leftrightarrow \text{M}_2$  and  $\text{M}_2 \leftrightarrow \text{M}_3$  transitions are related to local structural distortion in the polar ( $\text{Pmc}2_1$ ) and non-polar ( $\text{Pbcm}$ ) phases, respectively. In particular, the diffuse dielectric anomaly associated with the  $\text{M}_2 \leftrightarrow \text{M}_3$  transition can be intimately related to disorder-order

behavior of displacive  $\text{Nb}^{5+}$  dynamics on the sub-lattice scale in the non-polar  $\text{Pbcm}$  structure matrix. Recently, by studying  $\text{Ag}_{1-3x}\text{Bi}_x\text{Nb}_{0.8}\text{Ta}_{0.2}\text{O}_3$  ceramics, Yan *et al.* found that within the  $\text{M}_2$  phase stability region, two structural characteristics can be identified, and were denoted as the  $\text{M}_{2a}$  phase (at  $T < T_f$ ) and the  $\text{M}_{2b}$  phase (at  $T > T_f$ ). The  $\text{M}_{2a}$  phase is polar non-centrosymmetric, while  $\text{M}_{2b}$  is centrosymmetric. Moreover, they also proposed that even in the  $\text{M}_3$  region, a few polar regions still persist, as evidenced by the observed weak satellite spot assigned to the (003) reflection of the  $\text{Pb}2_1\text{m}$  structure.<sup>74</sup> Additionally, Gao *et al.* investigated the local structural features, including interatomic distance distributions and atomic displacements, using neutron small-box pair distribution function (PDF) refinement in conjunction with large-box reverse Monte Carlo modelling, and found that at room temperature (300 K), with increasing the sizes of small-box PDF refinement, the structure tends to be non-polar, while at high temperature ( $T > T_f$ , e.g. 500 K), the non-polar structure provides a better description, regardless of the size of the PDF refinement. The results of this study further support the difference between the structure below and above  $T_f$ , and indicate that the dielectric anomaly assigned as  $T_f$  originates from a phase transition.<sup>75</sup> Most recently, after analyzing the local structure data obtained from spherical aberration-corrected STEM, Li *et al.* suggested that at room temperature, besides the off-centering of  $\text{Nb}^{5+}$  ions along  $\pm[110]_c$  forming AFE  $\text{Pbcm}$  or FIE  $\text{Pmc}2_1$  structures, the deviation of  $\text{Nb}^{5+}$  ions along  $\pm[001]_c$  directions was also found in local regions. They further point out that these local off-centering cations form polar nanoregions (PNRs), which are responsible for the observed relaxor dielectric response assigned to the  $\text{M}_1 \leftrightarrow \text{M}_2$  phase transition. Their results further support the idea that the so-called  $\text{M}_1 \leftrightarrow \text{M}_2$  transition is more likely related to a local structural evolution.<sup>76</sup>

#### 2.4 Ferroelectric properties and field-induced transitions

The earliest proofs of ferroelectricity in  $\text{AgNbO}_3$  were based on the observation of non-linear and hysteretic  $P$ - $E$  loops under a weak-field ( $<20 \text{ kV cm}^{-1}$ )<sup>14,45</sup> that displayed a very small remnant polarization. In 2007, based on high quality  $\text{AgNbO}_3$  ceramic samples, Fu *et al.*<sup>13</sup> observed a “double” polarization hysteresis under high electric fields (see Fig. 11). The different polarization behavior under weak-field and strong-field conditions suggested that the virgin structure undergoes at least two types of field-induced processes, involving FE domain switching and a field-induced phase transition. In other words, the virgin structure should contain either FE or AFE states. To explain the peculiar polarization hysteresis observed in the experiments, Yashima *et al.*<sup>49</sup> suggested an FIE (i.e. “non-compensated AFE”)  $\text{Pmc}2_1$  structure after studying diffraction data including CBED, ND and synchrotron XRD data. Furthermore, Moriwake *et al.* carried out molecular dynamics simulations on monocrystalline  $\text{AgNbO}_3$ , and proposed that polarization switching or a field-induced transition occurs at temperatures around  $-73^\circ\text{C}$  and above. Regardless of whether the simulations commenced from the AFE  $\text{Pbcm}$  or FE  $\text{Pmc}2_1$  structures, above  $-73^\circ\text{C}$  the crystal fluctuates between the two forms.<sup>77</sup>

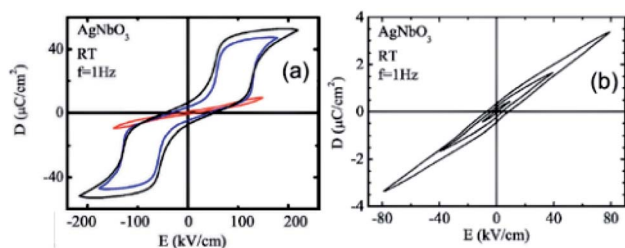


Fig. 11 Electric displacement vs. electric field loops measured under (a) a maximum applied field of  $220 \text{ kV cm}^{-1}$ , and (b) a maximum applied field of  $80 \text{ kV cm}^{-1}$ .<sup>13</sup> [Reproduced from ref. 13 <https://doi.org/10.1063/1.2751136> with the permission of AIP Publishing].

However, detailed studies of polarization–current–electric field hysteresis loops reported by Tian *et al.*<sup>15,16</sup> revealed more complicated electric-field-induced polarization events taking place in  $\text{AgNbO}_3$  ceramics, providing further insights into the virgin structure and electric field-induced transformations, as well as on the temperature dependence of the polymorphic nature of  $\text{AgNbO}_3$  perovskites. At room temperature, a ferroelectric-like polarization hysteresis with weak remnant polarization can be observed upon applying an alternating electric field of  $60 \text{ kV cm}^{-1}$  (Fig. 12a), being similar to that reported in early studies. The  $I$ – $E$  loops revealed two current peaks (denoted as  $P_1$  and  $P_2$ ) located at  $6$  and  $30 \text{ kV cm}^{-1}$ , respectively, which represent two kinds of field-induced polarization events.

The data recorded in the first-poling cycle shown in Fig. 12b clearly reveal a sharp current peak,  $+P_1$ , around  $6 \text{ kV cm}^{-1}$ . The peak  $+P_2$  located at higher electric field is not clearly visible in the first poling cycle, but careful observation reveals a small step around  $30 \text{ kV cm}^{-1}$ . Further insights into the origin of these current peaks were obtained by their temperature-dependence shown in Fig. 12c and d, which confirmed that the current peaks  $P_1$  and  $P_2$  disappeared around the temperatures associated with the  $M_1 \leftrightarrow M_2$  transition and  $T_f$ , respectively. Similar features have been observed in the  $I$ – $E$  curves of  $\text{Ag}(\text{Nb}_{0.8}\text{Ta}_{0.2})\text{O}_3$  ceramics shown in Fig. 13a.<sup>78</sup> In the same study, a sharp

dielectric anomaly, assigned to  $T_f$ , was clearly observed during heating, in contrast to the case of pure  $\text{AgNbO}_3$ . Linear fitting of the reciprocal of permittivity using the Curie–Weiss law gave  $T_0 = 80^\circ\text{C}$ , which is very close to the value of  $T_f$  ( $\sim 75^\circ\text{C}$ ) (see Fig. 13b).

Based on these observations and the data discussed in Section 2.3, the following possible phase scenario can be derived. The phase  $M_1$  would include the presence of polar and weakly-polar domains, both belonging to the  $Pmc2_1$  symmetry. In other words, the virgin microstructure would include two kinds of polar domains with different electric dipole arrangement types in the same  $Pmc2_1$  symmetry, which can be regarded as polar ferroelectric and polar ferrielectric (*i.e.*, a non-compensated antiferroelectric). This suggestion is also supported by the simulations carried out by Moriwake *et al.*,<sup>77</sup> which indicated that the polar ferrielectric phase has the same space group,  $Pmc2_1$ , as the polar phase, but it is characterized by a slightly different ion arrangement that makes its distinction difficult in microscopy and diffraction techniques. Fig. 14(a)–(c) further illustrate the ordering of electric dipole arrangements resulting in AFE  $Pbcm$ , FIE  $Pmc2_1$  and FE  $Pmc2_1$  structures, respectively.<sup>77</sup> Most recently, domain structure observation on  $\text{AgNbO}_3$  single crystals, has also revealed the existence of polar domains, which would further support the deduction.<sup>79</sup>

In this structural context, the current peak  $P_1$  corresponds to the switching of the polar domains in non-centrosymmetric  $Pmc2_1$  symmetry. This switching event is no longer present above the  $M_1 \leftrightarrow M_2$  transition temperature, suggesting that these polar domains are not switchable in the  $M_2$  phase region. It is noted that the size of polar domains should be nanometric and only emerge on local scales as supported by recent results based on STEM investigation.<sup>76</sup> The current peak  $P_2$  could be related to the presence of weakly-polar domains, which are further stabilized by the applied field, as suggested by the sharpening of the peak  $P_2$  after the first cycle. This peak disappears above  $T_f$ , suggesting that these domains do not switch in the  $M_{2b}$  centrosymmetric phase, in agreement with the loss of inversion symmetry and the closeness of  $T_f$  and the

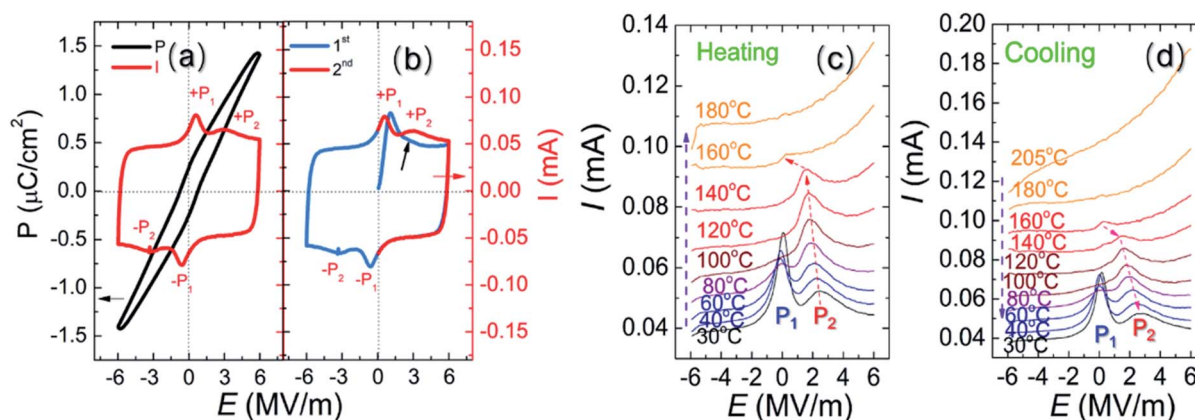


Fig. 12 (a)  $P$ – $E$  and  $I$ – $E$  loops for  $\text{AgNbO}_3$  with successive cycles under a maximum applied field of  $\pm 60 \text{ kV cm}^{-1}$  and a frequency of  $10 \text{ Hz}$ ; (b)  $I$ – $E$  data generated in the initial 1<sup>st</sup> and 2<sup>nd</sup> semi-cycles; thermal dependence of the current peaks on (c) heating and (d) cooling.<sup>15,16</sup>

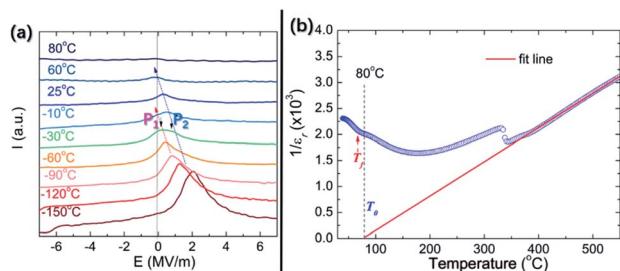


Fig. 13 (a) Detail of the  $I$ - $E$  curves for  $\text{Ag}(\text{Nb}_{0.8}\text{Ta}_{0.2})\text{O}_3$  ceramics at 10 Hz under an applied field of  $8 \text{ MV m}^{-1}$  at different temperatures; (b) temperature-dependence of reciprocal relative permittivity ( $1/\epsilon_r$ ) for unpoled  $\text{Ag}(\text{Nb}_{0.8}\text{Ta}_{0.2})\text{O}_3$  ceramics measured at 1 MHz.<sup>78</sup>

Curie-Weiss temperature. The absence of the peak  $P_2$  in the  $M_{2b}$  phase offers further support to the idea that the  $T_f$  might be identified with the Curie point ( $T_C$ ) of the weakly-polar phase.

Under strong-field conditions, a “double” polarization hysteresis loop is generated (Fig. 15a), which is consistent with the results reported by Fu *et al.*<sup>13</sup> In the  $I$ - $E$  loops, six current peaks can be observed; these are different from those observed in Pb-based AFEs like  $\text{PbZrO}_3$ , which only show four current peaks representing reversible phase transitions.<sup>1</sup> The peaks denoted as  $E_F$  and  $E_B$  correspond to the forward field-induced transition during loading and the backward transition during unloading, respectively. The polarization induced during loading achieves a value of about  $40 \mu\text{C cm}^{-2}$ , while the remnant polarization is about  $4 \mu\text{C cm}^{-2}$ , which is much higher than the  $P_r$  in the low-field loops. This means that the sample is left with a sizable remnant polarization, which is only recovered by reversing the applied field, in correspondence to the extra current peaks  $\pm P$  near  $\pm 50 \text{ kV cm}^{-1}$  (denoted as key fields  $\pm E_U$ ). It is interesting to observe that the peak  $P$  at key field  $E_U$  is absent during the first poling cycle (see Fig. 15b), confirming that these additional current peaks  $\pm P$  should originate from the polarization reversal of the residual strong-field-induced FE state. Interestingly, after strong-field poling of virgin sample, the authors found that the current peak  $P_2$  in the  $I$ - $E$  loop under

weak field conditions becomes stronger.<sup>15,16</sup> This indicates that the weakly-polar phase leading to the sharp dielectric response ( $T_f$ ) is metastable, and can be modified by the strong electric field. At the moment, the mechanisms of the field-induced transitions at  $E_F$  are still unclear and the nature of the high field-induced structure is unknown. Based on the structural model discussed above, the transition at  $E_F$  might involve the weakly-polar ferroelectric phase, as well as the non-polar anti-ferroelectric phase that would both transform to a polar FE phase, the former irreversibly, leading to a larger  $P_r$  and the latter reversibly, recovering at  $E_B$ . However, the first-principles calculations reported by Moriwake *et al.*,<sup>80</sup> indicated that a complete transition from an antiparallel alignment of electric dipoles in the non-polar  $Pbcm$  structure to the parallel alignment in the FE  $Pmc2_1$  phase would require an ultrahigh electric field of  $9 \text{ MV cm}^{-1}$  (see Fig. 14c). Thus, the double polarization hysteresis observed in pure  $\text{AgNbO}_3$  is unlikely to originate from a typical AFE  $\leftrightarrow$  FE transition, but from a metastable weakly-polar FIE  $\leftrightarrow$  FE transition, which is not fully reversible, as suggested by the increased remnant polarization after high field cycling. The metastable FIE structure represents an intermediate state arising from the competition between the AFE  $Pbcm$  and FE  $Pmc2_1$  phases. Additionally, the simulations by Moriwake *et al.* indicated that the field-induced FE phase and the weakly-polar/FIE phase have the same space group,  $Pmc2_1$  (see Fig. 14). However, this deduction is only based on known data reported in the literature. To date, there are no experimental studies on the structure of the strong-field-induced FE phase.

## 2.5 Photocatalytic, photoelectrochemical and photovoltaic properties

Besides showing anti-/ferroelectric behavior,  $\text{AgNbO}_3$  perovskites are semiconductors possessing a narrow band gap, which ranges from 2.08 to 2.93 eV (determined from UV-vis diffuse reflectance spectra).<sup>81</sup> Initially,  $\text{AgNbO}_3$  was reported to be a visible-light-driven photocatalyst possessing the ability to obtain  $\text{H}_2$  or  $\text{O}_2$  from water in the presence of sacrificial reagents. Table 1 lists the photocatalytic activity of  $\text{AgNbO}_3$ .

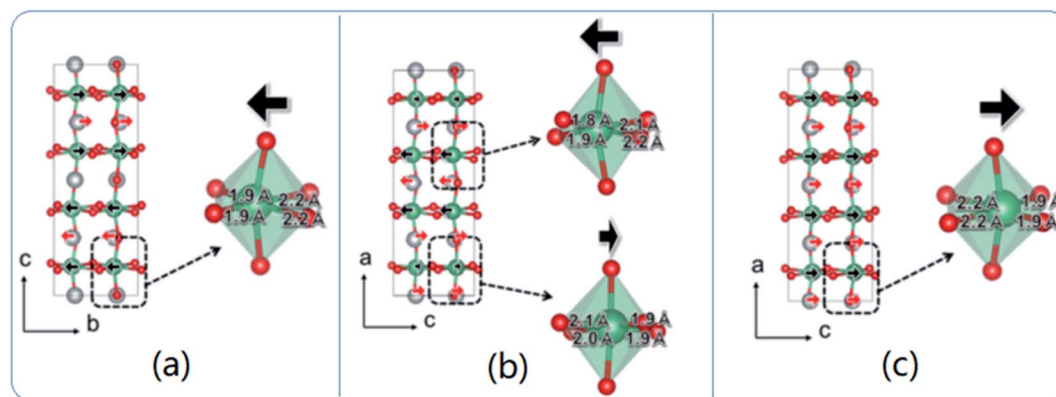


Fig. 14 Comparison of crystal structures of (a)  $Pbcm$  (antiferroelectric), (b)  $Pmc2_1$  (ferrielectric/weakly-polar), and (c)  $Pmc2_1$  (ferroelectric) phases of  $\text{AgNbO}_3$ . The arrows indicate the relative magnitude and direction of atom displacement contributing to the ferroelectricity (or lack thereof) in each structure.<sup>77</sup> [Reprinted with permission from ref. 77 <https://doi.org/10.1103/PhysRevB.67.224104> copyright 2003 by the American Physical Society].





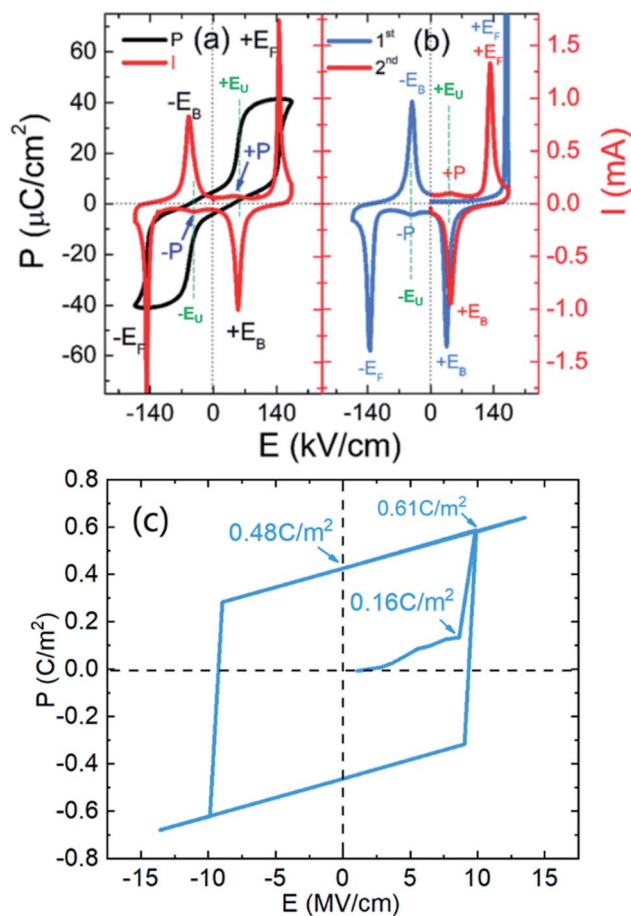


Fig. 15 (a) Polarization–current–field hysteresis loops for AgNbO<sub>3</sub> with successive cycles under a maximum applied field of  $\pm 180$  kV cm<sup>-1</sup> at a frequency of 10 Hz; (b) current–field loops generated in the initial 1st and 2nd cycles;<sup>16</sup> (c) calculated  $P$ – $E$  hysteresis loop for AgNbO<sub>3</sub> induced by an irreversible phase transition from AFE  $Pbcm$  to FE  $Pmc2_1$  phases.<sup>80</sup> [Reproduced from ref. 80 <https://doi.org/10.1063/1.4941319> with the permission of AIP Publishing].

reported by Kato *et al.* under various conditions.<sup>21</sup> In particular, the photocatalytic activity of AgNbO<sub>3</sub> was found to increase on the addition of an excess amount of Ag during preparation (*i.e.*, Ag/Nb > 1) as shown in Table 1. Recently, a series of AgNbO<sub>3</sub>-based photocatalysts, including solid-solutions, composites

and heterojunctions, have been developed for water purification and will be discussed in Section 5.3.

Electronic structure studies based on the plane-wave-based density functional method have revealed that the narrow band gap is due to the fact that the top of the valence band mainly consists of a hybrid orbital of Ag 4d and O 2p. To further understand the photocatalytic activity of AgNbO<sub>3</sub>, Li *et al.* investigated the surface photoelectric properties of Ag/AgNbO<sub>3</sub> photocatalysts and suggested that the two peaks at  $\sim 375$  nm and  $\sim 420$  nm in the surface photovoltage spectrum (SPS) (see Fig. 16) are associated with two electronic transitions.<sup>82</sup> One could correspond to the transition from O 2p to Nb 4d, being related to the UV light absorption property, and the other ( $\sim 420$  nm, located at a shorter wavelength than the light absorption edge) related to visible-light absorption property was proposed to correspond to an electronic transition from Ag 4d to Nb 4d. In particular, the electric field-induced surface photovoltage spectrum (EFISPS) suggested that the second SPS peak (at  $\sim 420$  nm related to the transition from Ag 4d to Nb 4d), determining the visible light absorption, could be enhanced or completely restrained under positive and negative external voltages, respectively, indicating tunable visible-light photocatalytic activity.

In 2014, Zhou *et al.*<sup>83</sup> revealed that AgNbO<sub>3</sub> exhibits surface plasmon resonance (SPR) under visible light. Surface plasmons (SPs) can exist on the surface of noble metals or narrow band gap semiconductors with abundant active current carriers.<sup>84</sup> This indicates that AgNbO<sub>3</sub> can absorb extra visible light with wavelengths ranging from 450 to 700 nm giving it huge potential in light-harvesting technologies. The SPR was suggested to originate from weakly bound Ag atoms in the perovskite structure, favoring a metal-like state of silver ions and self-assembled microstructures of AgNbO<sub>3</sub>. The different absorptions were divided into two parts as shown in the inset of Fig. 17.<sup>83</sup> The electron-density map and the partial density of states of AgNbO<sub>3</sub> reported in earlier studies by Yashima *et al.*<sup>81</sup> reveal a large valence band width ( $\sim 6$  eV). The latter is due to the existence of Ag atoms and Ag–O and Nb–O covalent bonds, which lead to a narrower band gap ( $\sim 1.83$  eV) compared to the experimental values reported in early studies, and a visible-light response. The metal-like characteristics and these transmission paths play a key role in promoting hot-electron generation and transmission.

Table 1 Photocatalytic activity of AgNbO<sub>3</sub> (ref. 21)

Catalyst	Ag/Nb ratio <sup>a</sup>	Incident light/nm	Reaction conditions	Activity/ $\mu\text{mol h}^{-1}$	
				H <sub>2</sub>	O <sub>2</sub>
AgNbO <sub>3</sub>	1.00	>420	0.05 M AgNO <sub>3</sub> (aq)	—	14
AgNbO <sub>3</sub>	1.00	>300	0.05 M AgNO <sub>3</sub> (aq)	—	119
AgNbO <sub>3</sub>	1.05	>420	0.05 M AgNO <sub>3</sub> (aq)	—	37.0
AgNbO <sub>3</sub>	1.05	>300	0.05 M AgNO <sub>3</sub> (aq)	—	240
AgNbO <sub>3</sub>	1.05	>420	H <sub>2</sub> O, CH <sub>3</sub> OH vapor <sup>b</sup>	1.7	—

<sup>a</sup> In the starting materials. <sup>b</sup> Pressures of H<sub>2</sub>O and CH<sub>3</sub>OH were 20 and 70 torr, respectively. Catalyst: 0.3 g for liquid-phase reactions and 1.0 g for gas-phase reactions; reactant solution: 150 mL, 300 W Xe lamp, top window cell made of Pyrex.

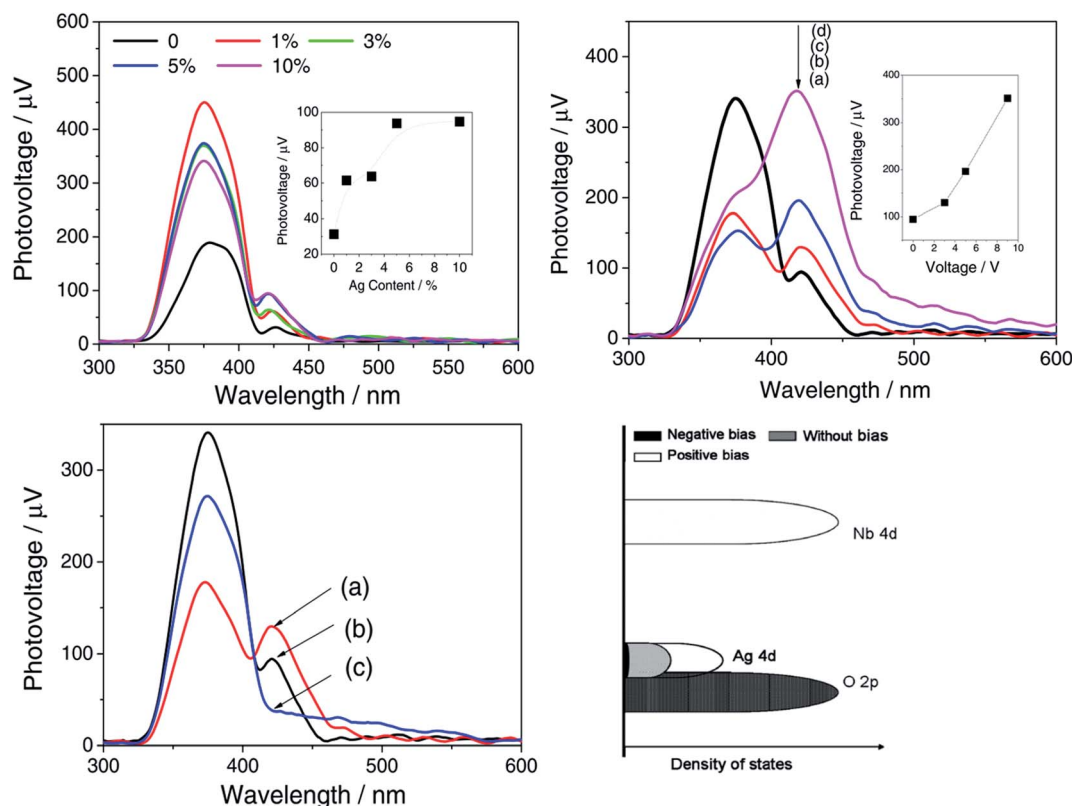


Fig. 16 (a) SPS of Ag/AgNbO<sub>3</sub> with different Ag contents without an external bias. The inset shows the variations in photovoltage and wavelength of the peak in the visible light region with Ag content. (b) EFISPS of Ag/AgNbO<sub>3</sub> (10%) at different external voltages. (c) SPS of Ag/AgNbO<sub>3</sub> (10%) at different external voltages. (d) Schematic band structure under different biases.<sup>82</sup> [Reproduced from ref. 82 <https://doi.org/10.1088/0022-3727/42/23/235503> with permission from copyright IOP Publishing].

Recently, Lu *et al.*<sup>36</sup> discovered that Ag nanoparticles with a size of 10 nm can be grown *in situ* on AgNbO<sub>3</sub> using illumination by Xe lamps, yielding a color change from yellow to brown, as shown in Fig. 18a. A change in color under light illumination was also found in our AgNbO<sub>3</sub> fresh powders. Fig. 18c shows SEM images taken after different irradiation times. Increasing numbers of nanoparticles can be observed on the surface of grains with increasing irradiation time. These Ag nanoparticles appearing on the surface of AgNbO<sub>3</sub> grains bring an extra absorption in the visible-light range as shown in Fig. 18b, which can be attributed to the SPR effect of Ag metal, as proposed by Zhou *et al.*<sup>83</sup> Thus, the SPR effect is responsible for the enhanced photocatalytic activity at increased ratios of Ag to Nb, as shown in Table 1. The appearance of such Ag nanoparticles would lead to the creation of Ag<sup>+</sup> vacancies in the AgNbO<sub>3</sub> surface layers and be consistent with the theoretical prediction of Schottky defects in AgNbO<sub>3</sub>.<sup>37</sup>

Bulk samples of AgNbO<sub>3</sub> have recently been reported to exhibit linear and symmetric current–voltage (*I*–*V*) curves under dark and illuminated conditions.<sup>22</sup> In particular, the extremely weak dark conductivity (in absence of illumination) can be significantly improved by illumination. Further analysis shows that the photocurrent under a certain bias voltage increases linearly with increasing light intensity (see Fig. 19a). The UV-vis absorption spectrum shown in Fig. 19b reveals obvious

absorption in the visible-light wavelength range up to 800 nm, where a weak photoelectric response (compared to the UV-light region) can be noticed up to 550 nm. The authors of the study

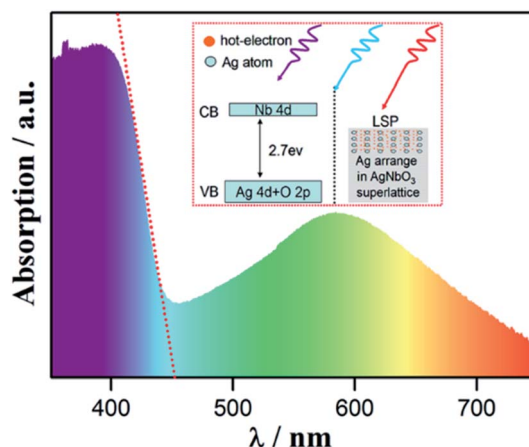
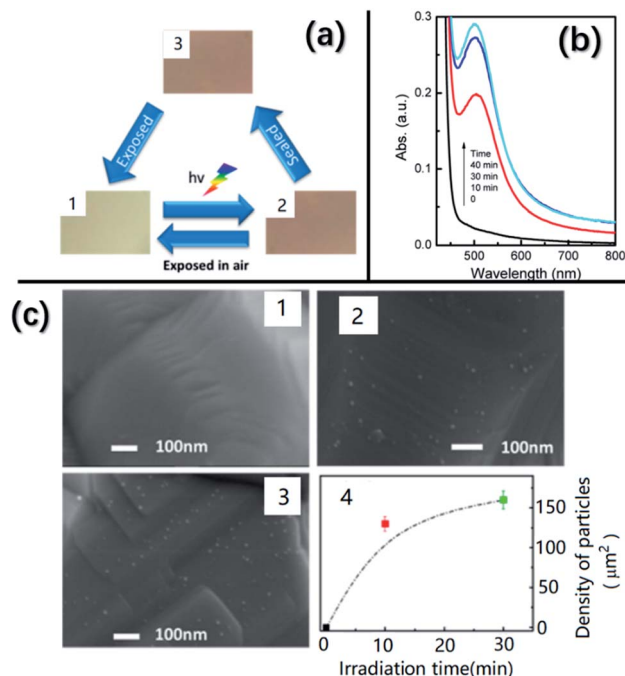


Fig. 17 Ultraviolet-visible diffuse reflectance spectrum of AgNbO<sub>3</sub> samples. The red dashed line is the tangential line of the eigen-absorption band side, showing the location of the band gap. Inset: schematic showing two kinds of absorption mechanism.<sup>83</sup> [Reproduced from ref. 83 <https://doi.org/10.1063/1.4903912> with the permission of AIP Publishing].





**Fig. 18** (a) Color of AgNbO<sub>3</sub> samples in different states: (1) initial, (2) after irradiation, and (3) isolated from air and kept in the dark for 48 h. (b) UV-vis spectra of AgNbO<sub>3</sub> irradiated with a Xe lamp for different times. (c) SEM images obtained from the samples in different states: (1) initial, (2) after 10 min of irradiation, and (3) after 30 min of irradiation. (4) Plot of density of nanoparticles versus irradiation time.<sup>36</sup> Reprinted with permission from ref. 36 <https://doi.org/10.1021/acs.jpcc.6b10961> copyright 2016 by the American Chemical Society.

linked the photoelectric response in the visible-light region to the presence of defects rather than to the SPR behavior mentioned above.<sup>22</sup> The time evolution of the photocurrent response under intermittent on/off illumination shown in Fig. 19c reveals very stable photocurrent values over 1.7 nA under 10 V bias, and a large on/off ratio of 1250. The transient photocurrent curves obtained under the illumination of a nanosecond pulse laser with an emission wavelength of 355 nm (see Fig. 19d) show an instantaneous photoresponse speed of 4.7 ns and highlight the great potential for quick-response optical sensing and detection applications.

AgNbO<sub>3</sub> ceramics have been also reported to exhibit a sizeable photovoltaic (PV) effect after electrical poling or mechanical polishing,<sup>22</sup> evidenced by a non-zero short-circuit current,  $I_{sc}$ , and open-circuit voltage,  $V_{oc}$ , as shown in Fig. 20. However, the poling-induced PV response disappeared when the annealing temperature approached 200 °C, where no polarization switching phenomenon is observed in the  $I$ - $E$  loop, as discussed in Section 2.4. This suggests that the PV response upon poling should be intimately related to the ferroelectric polarization. In contrast to the poling-induced PV effect, the mechanical polishing-induced PV response still persists after annealing the sample at 400 °C, which suggests that ferroelastic domains still exist at 400 °C, as reported in early studies on the domain structure.<sup>56,57</sup> The authors suggested that the flexoelectricity induced by strain gradients could be responsible for the

polishing-induced PV effects, which lead to complex spatial strain gradients distributed along the scratches.

### 3. AgNbO<sub>3</sub>-based solid-solutions

#### 3.1 Anti-/ferroelectricity in perovskite oxides

Chemical modification represents a widely applied strategy to obtain specific physical properties for different applications. Based on the physical properties discovered in AgNbO<sub>3</sub> perovskites, a number of AgNbO<sub>3</sub>-based materials have been explored; in particular, some AgNbO<sub>3</sub>-based solid-solutions were found to hold great potential for microwave applications, as well as for dielectric capacitors and piezoelectric, pyroelectric and high-power energy storage devices.

The structure of the ideal ABO<sub>3</sub> cubic perovskite can be described as being based on a cubic close packed array of oxide ions with  $\frac{1}{4}$  of these replaced by A cations. The B cations are located in  $\frac{1}{4}$  of the octahedral interstices. Generally, the occurrence of anti-/ferroelectricity in perovskite structures is caused by the slight distortion from the high-temperature cubic phase. There is an important question to be clarified: why does a distorted structure form in some perovskite-structured compounds during cooling? Especially for AgNbO<sub>3</sub>, the structural distortion is highly complex. It is well known that the stability of the perovskite structure (see Fig. 21) can be linked to its geometry by the tolerance factor ( $t$ ) as discussed by Bartel *et al.*,<sup>85</sup>

$$t = (r_A + r_O) / \sqrt{2}(r_B + r_O) \quad (1)$$

where  $r_O$ ,  $r_A$ , and  $r_B$  are the ionic radii of the O, A, and B ions, respectively. In some publications, an alternative expression has been proposed:<sup>86,87</sup>

$$t = r_{A-O} / \sqrt{2}r_{B-O} \quad (2)$$

where  $r_{A-O}$  and  $r_{B-O}$  are the bond lengths of A-O and B-O bonds, respectively. In some crystals, the interaction between A/B-site cations and the O anion is not only based on ionic bonds, but also on covalent ones, which play an essential role in the origin of ferroelectricity. The evolution of the perovskite structure and its dependence on the tolerance factor,  $t$ , in various metal oxides is summarized in Fig. 22.

From the structural geometry point of view, the following three situations can be identified (see Fig. 23): (1) in general, when  $t > 1$ , the B-site cation can spontaneously move off-center within the oxygen octahedron and generate ferroelectric distortion as in the cases of KNbO<sub>3</sub> and BaTiO<sub>3</sub> (see Fig. 22). Since the -A-O-A- atomic links are close packed (*i.e.*, the close packed atomic link is along the  $\langle 011 \rangle$  crystallographic directions of the cubic perovskite), this leads to greater space between the B-site cation and its surrounding O anions (see Fig. 23); (2) when  $t = 1$ , no A/B-site cations can move off-center from their ideal positions in the parent cubic phase, as occurs in the cases of BaZrO<sub>3</sub> and SrTiO<sub>3</sub> (see Fig. 23). In this case both the atomic stacking along the directions of -A-O-A- (*i.e.*,  $\langle 011 \rangle$ ) and -B-O-B- (*i.e.*,  $\langle 001 \rangle$ ) are ideally packed. (3) When  $t < 1$ , the A-site cation is able to spontaneously move off-center in the





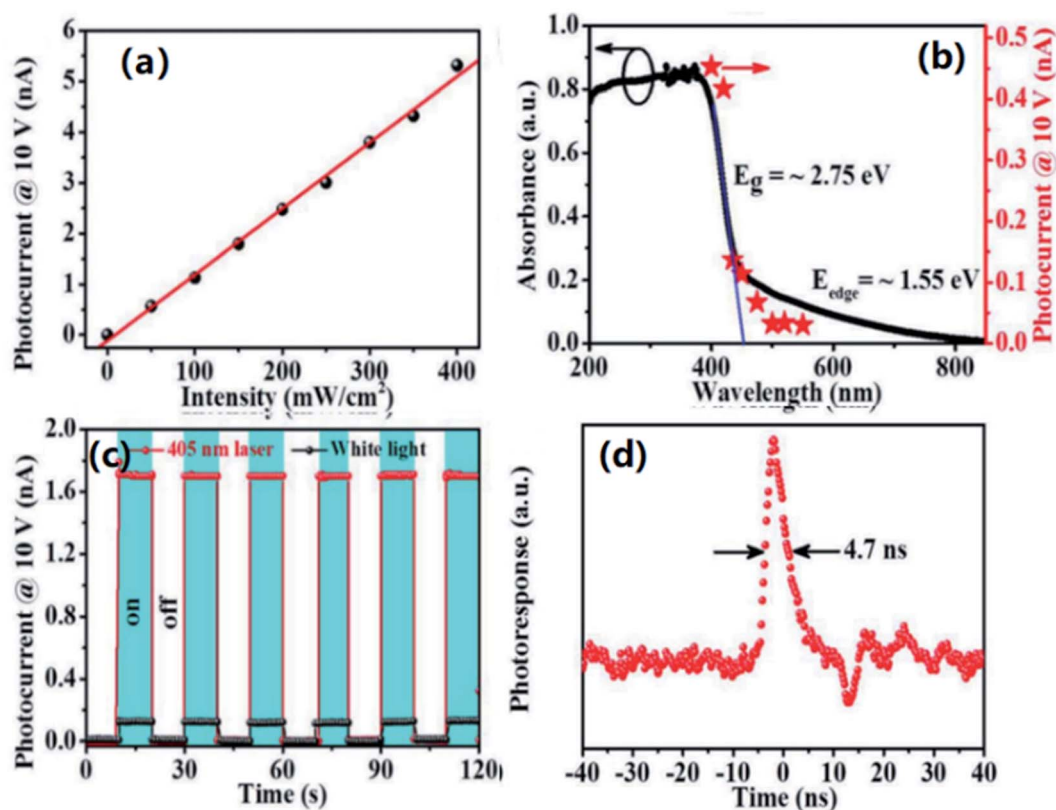


Fig. 19 (a) Light intensity dependence of photocurrent under a 10 V bias on illumination with a 405 nm laser. (b) UV-vis absorption spectrum and photocurrent as a function of wavelength with a 10 V bias and light intensity of  $150 \text{ mW cm}^{-2}$ . (c) Variation of photocurrent with time on intermittent on/off illumination (405 nm laser and white light) under a 10 V bias and a constant light intensity of  $150 \text{ mW cm}^{-2}$ . (d) Transient photocurrent curves under the illumination of a nanosecond pulse laser with an emission wavelength of 355 nm, on tripling the frequency and a pulse duration of 3–4 ns.<sup>22</sup> [Reproduced with permission from ref. 22 copyright 2019 by John Wiley and Sons].

AO<sub>12</sub> cage to generate ferroelectric distortion since the –B–O–B– bonding linkages are shorter, hindering the spontaneous off-center movement of the B-site cation in its oxygen octahedron, leading to a large space between the A-site cation and its surrounding O anions. However, in practical situations, the large stereochemical space results in the A-site cation only partially bonding with the adjacent O anions. Consequently, tilting of two adjacent BO<sub>6</sub> octahedra (*i.e.*, antiferrodistortion, AFD) occurs with the formation of a superlattice. For example, in the case of BiFeO<sub>3</sub>, two adjacent BO<sub>6</sub> octahedra undergo an antiphase rotation around the [111]<sub>p</sub> direction, resulting in an A-site driven FE phase with *R*3*c* symmetry (see Fig. 23). In some cases, the AFD behavior is also accompanied by the antiparallel displacement of A-site cations in two adjacent perovskite cubic units from their ideal positions, without breaking the inversion symmetry. Thus, many perovskite compounds with *t* < 1 cannot display anti-/ferroelectricity, and show only antipolar behavior within supercells (*i.e.*, antipolar phases or commonly, non-polar phases), as in the cases of CaTiO<sub>3</sub>,<sup>89</sup> LnFeO<sub>3</sub> (Ln: rare-earth element)<sup>90</sup> and SrSnO<sub>3</sub> (see Fig. 24a).<sup>91</sup> The difference between antiferroelectric (AFE) and antipolar (AP) lattices is shown in Fig. 24b.<sup>92</sup>

The partial covalent character of compounds such as PbZrO<sub>3</sub> and BiFeO<sub>3</sub> is due to the hybridization between the A-site cation

and O anion caused by the unique electron configuration (*i.e.*, the 6s<sup>2</sup> lone-pair of electrons) and is responsible for the observed anti-/ferroelectricity in compounds with *t* < 1.<sup>93</sup> The *t*-factor of AgNbO<sub>3</sub> calculated using eqn (2) is 0.98. In theory, it should not display anti-/ferroelectricity; however, more detailed calculations reveal a hybridization between Ag and O,<sup>81</sup> although unlike Pb and Bi, the Ag<sup>+</sup> ion does not have a lone-pair of electrons. Additionally, experimental studies based on X-ray photoelectron spectroscopy,<sup>94</sup> have suggested the presence of partial covalent character in the chemical bonds between Ag and O, as well as in those between Nb and O, is responsible for the observed anti-/ferroelectricity in AgNbO<sub>3</sub>.

Analysis based on structural geometry can be used to guide the development of novel AgNbO<sub>3</sub>-based materials with specific polar states. Increasing the *t*-factor *via* suitable elemental substitution could favor the off-centering of B-site cations and induce B-site driven ferroelectricity, while reducing the *t*-factor could promote the stability of A-site driven anti-/ferroelectricity. Both anti- and ferroelectric structures of AgNbO<sub>3</sub> perovskites display off-centering of Ag<sup>+</sup> and Nb<sup>5+</sup> cations. This feature is very rare in anti-/ferroelectric perovskites and should provide wider possibilities to design new solid-solution materials with excellent performance. In Sections 3.2 and 3.3, we will give more



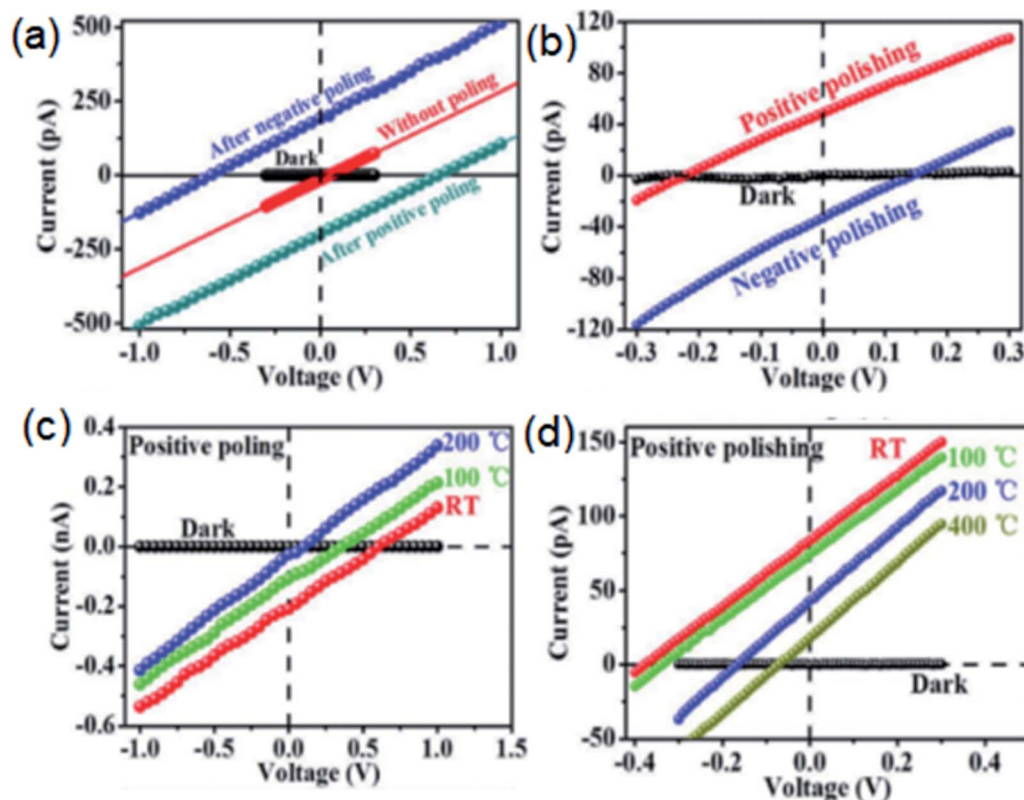


Fig. 20  $I$ - $V$  curves of  $\text{AgNbO}_3$  ceramic samples prepared using different (a) poling and (b) polishing conditions.  $I$ - $V$  curves collected under dark and illuminated ( $405\text{ nm}$  laser,  $200\text{ mW cm}^{-2}$ ) conditions at room temperature after annealing the (c) positively poled and (d) positively polished  $\text{AgNbO}_3$  ceramic at different temperatures for 1 h.<sup>22</sup> [Reproduced with permission from ref. 22 copyright 2019 by John Wiley and Sons].

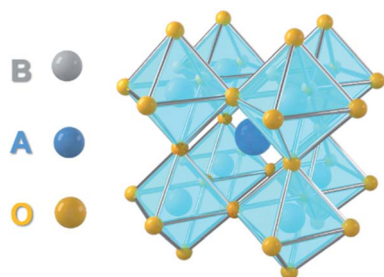


Fig. 21 The parent cubic perovskite structure of  $\text{ABO}_3$ . The A-site cation is located at the interstice within the  $\text{BO}_6$  octahedral framework.

detailed comments on these aspects in relation to recently developed  $\text{AgNbO}_3$ -based materials.

### 3.2 $\text{AgNbO}_3$ -based ferroelectric solid-solutions

Two main types of  $\text{AgNbO}_3$ -based ferroelectric solid-solution with remarkable piezoelectric properties have been developed by A-site isovalent substitution. One type is represented by Li-doped  $\text{AgNbO}_3$  compounds, such as  $(\text{Ag,Li})\text{NbO}_3$ , (ALN), and  $(\text{Ag,Li})(\text{Nb,Ta})\text{O}_3$ , (ANLT) systems, and the other includes K-doped  $\text{AgNbO}_3$  systems, such as  $(\text{Ag}_{1-x}\text{K}_x)\text{NbO}_3$  (AKN).

**3.2.1 Li-modified  $\text{AgNbO}_3$ .** The ionic radius of the  $\text{Li}^+$  ion is much smaller than that of the  $\text{Ag}^+$  ion ( $r = 0.92\text{ \AA}$  and  $1.28\text{ \AA}$ ,

respectively for the ions in 8 coordinate geometry).<sup>95</sup> Thus, the  $t$ -factor is gradually reduced in the ALN system with increasing level of Li substitution. This should enhance the stability of an antiferroelectric structure rather than a ferroelectric structure, as demonstrated in Section 3.1. However, it was found that Li-doping destabilizes the antiferroelectricity of  $\text{AgNbO}_3$  at low Li content, and induces ferroelectric order at high Li doping levels.<sup>52,69,96,97</sup> A theoretical study based on density functional calculations focused on  $(\text{K}_{0.5}\text{Li}_{0.5})\text{NbO}_3$  ( $t < 1$ ) revealed a frustration of tilt instabilities by A-site disorder, leading to a large off-centering of  $\text{Li}^+$  ions and a strong A-site driven ferroelectricity.<sup>98</sup> These results might explain the occurrence of strong ferroelectricity in ALN solid-solutions. Apart from the remarkable piezoelectricity, a composition-driven AFE/FE phase boundary can be found in ALN solid-solutions,<sup>52,69,96,97</sup> which could be suited to various applications, such as energy conversion, infrared detection and electrocaloric refrigeration, owing to irreversible electric field-induced phase transitions occurring near this phase boundary. However, the composition-structure-property relationships reported in previous studies are inconsistent, giving us the motivation to systematically survey and analyze the main factors responsible for these differences.

From the analysis of the strain-electric field simulated curves, it was reported that the piezoelectric property of Li-modified  $\text{AgNbO}_3$  single crystals can reach  $390\text{ pm V}^{-1}$ ,<sup>69</sup> indicating that these compounds are promising lead-free materials



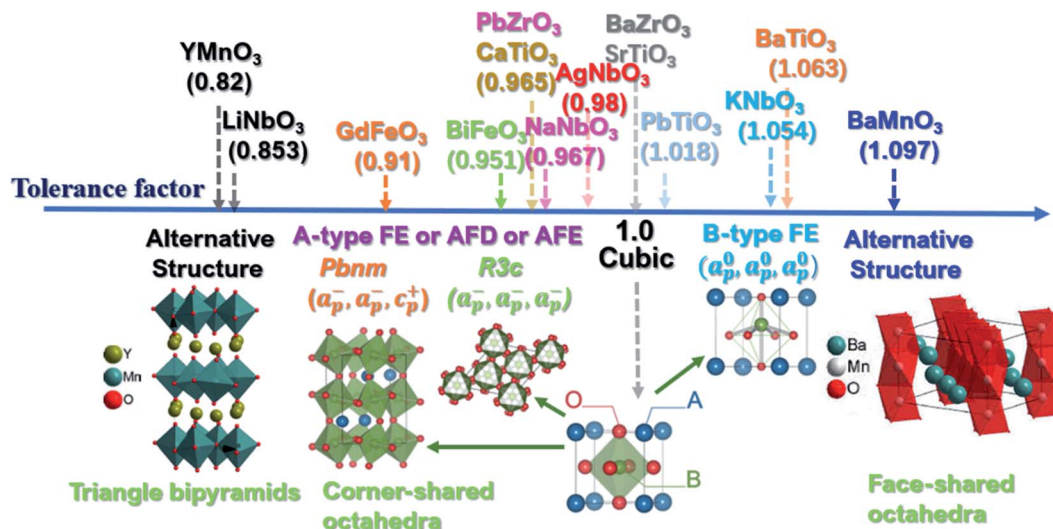


Fig. 22 Perovskite structure evolution with reference to the tolerance factor. The data are from ref. 87 and 88.

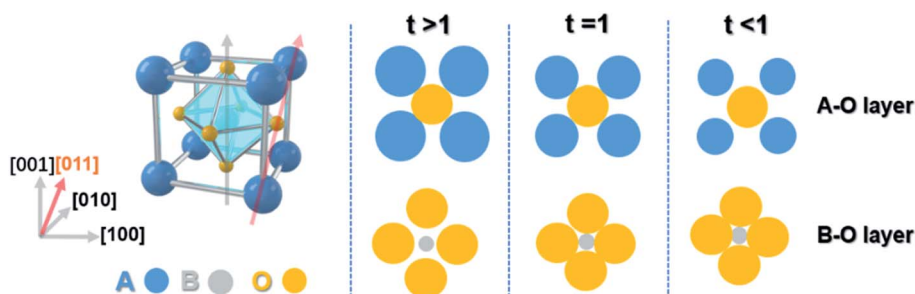


Fig. 23 Schematic showing packing directions of atoms in a perovskite: the  $-A-O-A-$  close packed atom linkage is along the equivalent  $[011]$  direction, while the  $-B-O-B-$  atoms link along the equivalent  $[001]$  direction. The atomic arrangements in each A-O and B-O layers of (pseudo) cubic perovskite when  $t > 1$ ,  $t = 1$  and  $t < 1$  are also drawn.

for piezoelectric applications. Table 2 lists the piezoelectric properties of Li-modified  $\text{AgNbO}_3$  single crystals and ceramics.

**3.2.1.1 Solid-solution synthesis.** It was reported that ALN ceramics can be synthesized via the conventional solid-state method in  $\text{O}_2$  atmosphere using  $\text{Ag}_2\text{O}$ ,  $\text{Li}_2\text{CO}_3$  and  $\text{Nb}_2\text{O}_5$  as raw materials. However, the preparation conditions significantly affect the structure and properties as reported in various studies. In the papers by Niewiadomski *et al.*,<sup>97</sup> and Kania *et al.*,<sup>101</sup> the authors synthesized  $(\text{Ag}_{1-x}\text{Li}_x)\text{NbO}_3$  ceramics using a two-step synthesis strategy as summarized in Fig. 25a. Using this synthesis strategy, a bi-phasic ceramic consisting of ALN solid-solution and  $\text{LiNbO}_3$  was obtained for  $x > 0.03$ . At higher Li concentrations, higher amounts of the  $\text{LiNbO}_3$  phase were found in the X-ray diffraction patterns, as evidenced by the increasing intensity of the peaks belonging to this phase.<sup>97</sup> Based on the compositional dependence of dielectric properties, it could be deduced that at higher Li concentrations, while the amount of  $\text{LiNbO}_3$  increased, more  $\text{Li}^+$  ions also diffused into the  $\text{AgNbO}_3$  lattice.<sup>97</sup> In the papers reported by Fu *et al.* and Khan *et al.*,<sup>102,103</sup> stoichiometric mixtures of  $\text{Ag}_2\text{O}$ ,  $\text{Nb}_2\text{O}_5$  and  $\text{Li}_2\text{CO}_3$  were calcined and subsequently sintered, as summarized in Fig. 25b, and appeared to eliminate the problem with

the secondary  $\text{LiNbO}_3$  phase. The differences between these syntheses are mainly represented by the processing parameters, such as calcination and sintering temperatures. In particular, the ALN solid-solution ceramics reported by Fu *et al.*, exhibited a very high solid solubility of  $\text{Li}^+$  ions (*ca.* 10 mol%), which should reflect the real solid-solution limit.

The synthesis of ALN single crystals using the procedures shown in Fig. 25c and d also results in different structure and properties. It was reported that ALN thin films can be fabricated on (001), (110) and (111)  $\text{SrTiO}_3$  (STO) substrates by pulsed laser deposition (PLD),<sup>104</sup> while the ALN ceramic targets for fabricating thin films were prepared by a solid-state reaction method, being similar to the processing reported by Fu *et al.*<sup>102</sup>

**3.2.1.2 Composition–structure–property relationships.** The composition–structure–property relationships in ALN solid-solutions reported in previous publications<sup>50,67,90,91,99–101</sup> differ with respect to the different processing conditions used. As demonstrated above, using the synthesis approach described in Fig. 24a, bi-phasic ceramics (*i.e.*, ALN solid-solution and  $\text{LiNbO}_3$ ) are obtained when  $x > 0.03$  mol. Diffraction analysis of samples prepared using the synthesis method reported by Fu *et al.*<sup>102</sup> (see Fig. 26b), shows that the solubility limit of Li is near





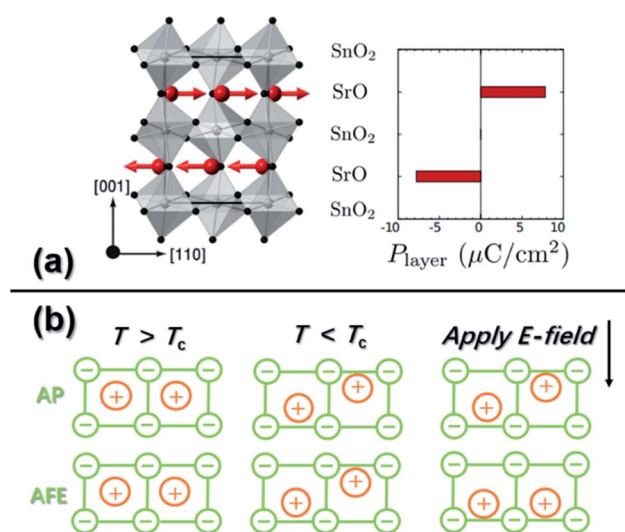


Fig. 24 (a) Schematic layer-resolved polarization of non-polar  $Pnma$   $SrSnO_3$ . The polarizations induced in the Sr–O layers in  $SrSnO_3$  are exactly equal and opposite and hence cancel each other out so that the macroscopic polarization is zero.<sup>91</sup> [Reproduced with permission from ref. 91 copyright 2013 by John Wiley and Sons]. (b) A schematic representation of the differences between antipolar (AP) and anti-ferroelectric (AFE) structures.<sup>92</sup>

$x = 0.1$  and a composition-driven transition from orthorhombic to rhombohedral (O–R) phases was found in the range  $x = 0.05$  to  $0.06$ , above which, the ALN solid-solution exhibits a single rhombohedral ferroelectric phase, as shown in Fig. 26a. Significant property variation was also revealed in ALN ceramics, giving additional evidence to support a composition-driven phase transition. With increasing  $x$ -value, a significant change in the dielectric behavior is observed (Fig. 26b), with a sharp maximum peak found at  $x \geq 0.06$ .<sup>102</sup> This peak shifts to higher temperature with further increase of the  $x$ -value. The temperature corresponding to the dielectric peak is believed to be the Curie point ( $T_C$ ) of the FE R phase. Fu *et al.*<sup>102</sup> proposed that as the  $x$ -value increases, the FE R phase develops from the FE structure originally found in the  $M_1$  region. However, further examination of the dielectric pattern reveals that the dielectric anomaly  $T_1$  (assigned to the  $M_1 \leftrightarrow M_2$  transition) gradually shifts to lower temperature and disappears for  $x > 0.06$ . Thus,

the  $M_1 \leftrightarrow M_2$  transition could correspond to a local structural evolution rather than to a symmetry change, as proposed in our previous study on  $AgNbO_3$ .<sup>16</sup> The double  $P$ – $E$  hysteresis loops evolve into a ferroelectric-type loop when the composition approaches the  $AFE_O/FE_R$  phase boundary, consistent with the structure characterization. The ALN single crystals prepared using the procedure described in Fig. 25d reported by Fu *et al.*<sup>69</sup> also show similar results to the ceramic samples. Interestingly, the investigation on ALN single crystals, prepared using the procedure shown in Fig. 25c reported by Wada *et al.*,<sup>96</sup> highlights a structural change from the AFE orthorhombic ( $Pbcm$ ) phase to an FE orthorhombic ( $Pc2_1b$ ) phase at  $x = 0.07$ . On further increasing the  $x$ -value, an extra composition-driven FE orthorhombic ( $Pc2_1b$ ) to FE rhombohedral ( $R3c$ ) phase transition occurred at  $x = 0.1$  (see Fig. 26d). For  $x > 0.1$ , this phase transition shifts to higher temperature, indicating the possible presence of an intermediate structure connecting the  $AFE_O$  and  $FE_R$  phases. It is possible that the range of this intermediate structure is dependent on the quality of samples.

It should be noted that the  $FE_O Pc2_1b$  structure exhibits the same unit-cell dimensions ( $Z = 8$ ,  $\sqrt{2}a_p \times \sqrt{2}a_p \times 4a_p$ ) as the  $AFE_O Pbcm$  structure (see Fig. 26d). After the FIE structure of  $AgNbO_3$  was identified in 2011,<sup>49</sup> Farid *et al.*<sup>52</sup> re-investigated the structural evolution of ALN powders synthesized by the method described in Fig. 25b, using high-resolution ND and synchrotron XRD. They suggested a new phase diagram for the ALN system, where the coexistence of FIE  $P2_1am$  (axial transformation of  $Pmc2_1$ ) and FE  $R3c$  phases was found for  $x > 0.05$ . In addition, for  $x > 0.05$ , a distinct polymorphic phase behavior compared to that of pure  $AgNbO_3$  on heating was proposed: both  $P2_1am$  and  $R3c$  polar structures evolved into a non-polar  $Pbnm$  phase and caused the maximum dielectric peak ( $T_C$ ) shown in Fig. 26b. By means of electron diffraction (ED), additional structural details for the  $x > 0.05$  compositions were revealed. The ED study by Khan *et al.*<sup>103</sup> reported a new complex modulated oxygen octahedral tilt system with a  $\sqrt{2}a_p \times \sqrt{2}a_p \times 6a_p$  unit-cell, which could be considered as an intermediate structure bridging the  $AFE_O$  phase and the  $FE_R$  phase. Such intermediate modulated structures are not unusual in anti-/ferroelectric perovskites and have been found near the MPB in many solid solutions, such as BNT-based and BFO-based perovskites, due to the complex interaction between off-

Table 2 Partial piezoelectric properties of Li-modified  $AgNbO_3$  ferroelectric materials

Materials	$d_{33}/d_{31}$ (pC N <sup>−1</sup> )	$k_{33}/k_{31}$ (%)	$g^{33}$ (10 <sup>−3</sup> V m N <sup>−1</sup> )	$T_C$ (°C)
<sup>a</sup> Ag <sub>0.949</sub> Li <sub>0.051</sub> NbO <sub>3</sub> (ref. 69)	65/—	—	11.3	173
<sup>a</sup> Ag <sub>0.938</sub> Li <sub>0.062</sub> NbO <sub>3</sub> (ref. 69)	180/—	—	38.4	199
<sup>a</sup> Ag <sub>0.935</sub> Li <sub>0.065</sub> NbO <sub>3</sub> (ref. 69)	190/—	—	53.8	204
<sup>a</sup> Ag <sub>0.914</sub> Li <sub>0.086</sub> NbO <sub>3</sub> (ref. 69)	210/—	—	53.9	275
<sup>a</sup> Ag <sub>0.9</sub> Li <sub>0.1</sub> NbO <sub>3</sub> [010] <sub>o</sub> (ref. 99)	—/−39.2	—/15.5	—	470
<sup>a</sup> Ag <sub>0.9</sub> Li <sub>0.1</sub> NbO <sub>3</sub> [110] <sub>o</sub> (ref. 99)	—/−80.0	—/30.4	—	470
<sup>a</sup> Ag <sub>0.9</sub> Li <sub>0.1</sub> NbO <sub>3</sub> (ref. 99)	—/−130.0	—/70.5	—	470
<sup>b</sup> Ag <sub>0.9</sub> Li <sub>0.1</sub> NbO <sub>3</sub> (ref. 100)	52/—	44/—	—	—

<sup>a</sup> Single crystal. <sup>b</sup> Ceramic.



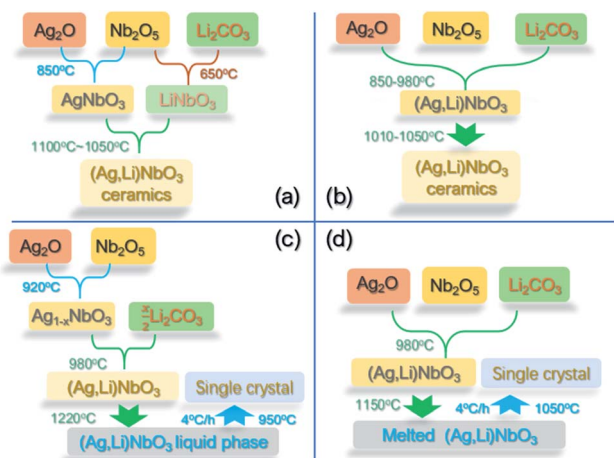


Fig. 25 Schematic synthesis procedures for  $(\text{Ag,Li})\text{NbO}_3$  ceramics reported by (a) Niewiadomski *et al.*,<sup>97</sup> and A. Kania *et al.*,<sup>101</sup> and (b) Fu *et al.* and Khan *et al.*,<sup>102,103</sup> schematic synthesis procedures for  $(\text{Ag,Li})\text{NbO}_3$  single crystals reported by (c) Wada *et al.*<sup>99</sup> and (d) Fu *et al.*<sup>69</sup>

centering cation displacements and oxygen octahedral tilting.<sup>105–109</sup> Thus, it is hard to describe the exact structural nature of the so-called MPB region based on the coexistence of two or more well-defined crystallographic phases. Therefore, more

structural investigations are needed to deepen the understanding of the MPB region in ALN solid solutions.

**3.2.2.  $(\text{Ag,Li})(\text{Nb,Ta})\text{O}_3$  ferroelectric solid-solutions.** Recently, our research team prepared  $(1-x)\text{AgNbO}_3-x\text{LiTaO}_3$  (ANLT100x) perovskite solid-solution ceramics,<sup>53</sup> using the same preparation procedure as Fu *et al.*<sup>102</sup> shown in Fig. 25b. Using ND data on the ceramic samples, the simultaneous presence of O and R phases (using FIE  $\text{Pmc}2_1$  and FE  $\text{R}3\text{c}$  models) was detected when  $x \geq 0.05$ . The composition-driven structural evolution of the ANLT system is very similar to that found in the ALN system by Farid *et al.*<sup>52</sup> The ED study also revealed the presence of inhomogeneous local structures including disordered in-/antiphase octahedral tilting and off-centering cation displacements in the O–R phase transition regions.<sup>110</sup> More importantly, in order to understand the origin of the structural evolution from the lattice dynamics view point, our structural refinement used the symmetry-mode-decomposition approach and revealed that the destabilizing AFE mode and the  $a^0a^0c^+/a^0a^0c^-$  AFD mode with increasing  $\text{Li}^+$  ion concentration constitute the driving force to induce the composition-driven  $\text{Pmc}2_1 \leftrightarrow \text{R}3\text{c}$  phase transition, as shown in Fig. 27a.<sup>53</sup> In particular, a sudden change in the lattice parameters of the FIE  $\text{Pmc}2_1$  phase can be found for  $x \geq 0.05$  (see Fig. 27b), indicating that the structure is highly distorted or that an additional unknown intermediate structure is present. The

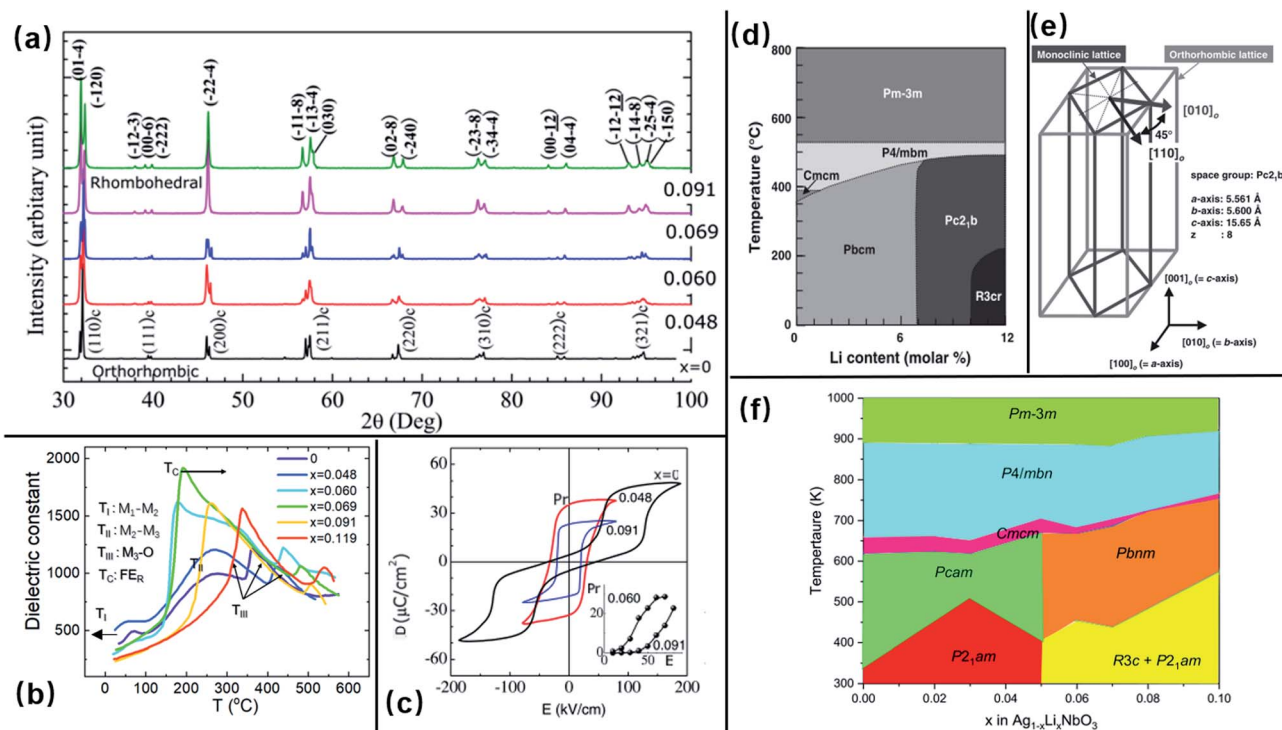


Fig. 26 (a) XRD patterns of synthesized powder, (b) temperature-dependent dielectric permittivity and (c)  $P$ – $E$  loops of  $\text{Ag}_{1-x}\text{Li}_x\text{NbO}_3$  ceramic solid-solutions reported by Fu *et al.* in 2011.<sup>102</sup> [Reproduced from ref. 102 <https://doi.org/10.1088/0953-8984/23/7/075901> with permission from copyright IOP Publishing]. (d) Schematic phase diagram of ALN system (using the preparation procedure shown in Fig. 25c) and (e) schematic configuration of  $[010]_o$  and  $[110]_o$  directions for  $\text{Pcm}$  and  $\text{Pbnm}$  orthorhombic symmetry reported by Wada *et al.* in 2006;<sup>96</sup> [reproduced from ref. 96 <https://doi.org/10.1143/JJAP.45.7389> with permission from copyright IOP Publishing]. (f) Schematic phase diagram of ALN system reported by Farid *et al.* in 2020 (using the preparation procedure shown in Fig. 25b).<sup>52</sup> [Reproduced with permission from ref. 52 copyright 2020 American Chemical Society].

dielectric spectrum of the ANLT system also shows a similar behavior to that of ALN. An additional dielectric peak (denoted as  $T_U$ ) could be found when  $x > 0.05$  (see Fig. 27c). The dielectric peak  $T_U$  is related to a temperature-driven  $FE_R$  phase transition. Moreover, through a detailed analysis of the dielectric loss, it can be found that the dielectric anomaly ( $T_1$ ) assigned to the  $M_1 \leftrightarrow M_2$  transition is still present at  $x = 0.06$  (ANLT6). This feature demonstrated that the composition ANLT6 still possesses the  $AgNbO_3$ -based O symmetry and that the  $M_1 \leftrightarrow M_2$  transition is linked to a local structural evolution, being consistent with our previous studies on pure  $AgNbO_3$ .<sup>16</sup> It is well known that electric field-induced irreversible AFE-FE transitions generally occur near the composition-driven AFE-FE phase boundary once the applied electric-field is larger than the characteristic field indicated by  $E_F$  (see Fig. 1b). In this case, the FE state can persist even after removing the applied electric field. In the ANLT system, our research group has recently discovered an interesting phenomenon considered as a type of ferroelectric “wake-up” effect.<sup>110</sup> In  $x = 0.045$  and 0.06 ceramics near the phase transition regions, ferroelectric loops can be triggered by increasing the number of electric field bipolar cycles, even when the maximum applied field is much smaller than the characteristic field,  $E_F$  (see Fig. 27d and e). The “woken-up” state means that the ferroelectric loops remain almost unchanged on further increase of the number of bipolar cycles. Due to the “wake-up” effect, the dielectric peak  $T_1$  in Fig. 27f is gradually suppressed and the dielectric peak  $T_U$  appears, indicating that the electric field bipolar cycles induce a transition from the FIE  $Pmc2_1$  phase to the FE  $R3c$  phase, as confirmed by the ND data analysis (see Fig. 27g).<sup>110</sup> Moreover, the structure refinements

also reveal that the off-center cation displacements of the  $Ag^+$  and  $Nb^{5+}$  ions gradually increase, indicating enhanced ferroelectricity of the FE  $R3c$  lattice (see Fig. 27h). As discussed in Section 2.4, the field-induced FE structure of pure  $AgNbO_3$  should be iso-symmetric with the initial FIE structure. However, this prediction was not verified in the case of the ALN and ANLT systems. The explanation is that near the phase coexistence region, the FIE ( $P2_1am$  or  $Pmc2_1$ ) lattice is highly distorted, and the disordered octahedral tilting and off-centering cation displacements (originating from destabilized AFE and  $a^0a^0c^+/a^0a^0c^-$  AFD modes<sup>53</sup>) provide a favorable energy path to the  $R3c$  phase. Once the external electric field (smaller than  $E_F$ ) is applied, the phase transition from FIE  $Pmc2_1$  to FE  $R3c$  is induced. The discovered intrinsic FE “wake-up” effect in AN-based perovskite ceramics may represent a potential solution for the long-standing fatigue problem in FE materials. This is evident in Fig. 28b, which shows that after  $10^4$  bipolar cycles, the remnant polarization  $P_r$  of a  $(Bi_{0.5}Na_{0.5})_{0.94}Ba_{0.06}TiO_3$  (BNBT6) capacitor drops by 60%. When connecting the ANLT6 ceramic in parallel with the BNBT6 ceramic to form a BNBT6-ANLT6 dual capacitor, the reduction of  $P_r$  can be controlled within 10% after  $10^4$  cycles.<sup>110</sup>

Li *et al.*<sup>111</sup> recently reported that the ANLT solid-solution also exhibits excellent pyroelectric properties during the depolarization process of the electric field-induced FE state (see Fig. 28c). ANLT compositions with  $x \geq 0.04$  exhibit a high pyroelectric coefficient  $p \approx 3\text{--}3.7 \times 10^{-8} \text{ C cm}^{-2} \text{ K}^{-1}$  at room temperature and an ultrahigh peak value ( $p = 120\text{--}160 \times 10^{-8} \text{ C cm}^{-2} \text{ K}^{-1}$ ) at the depolarization temperature ( $T_d$ ). If  $T_d$  could be adjusted to room temperature, these materials could have

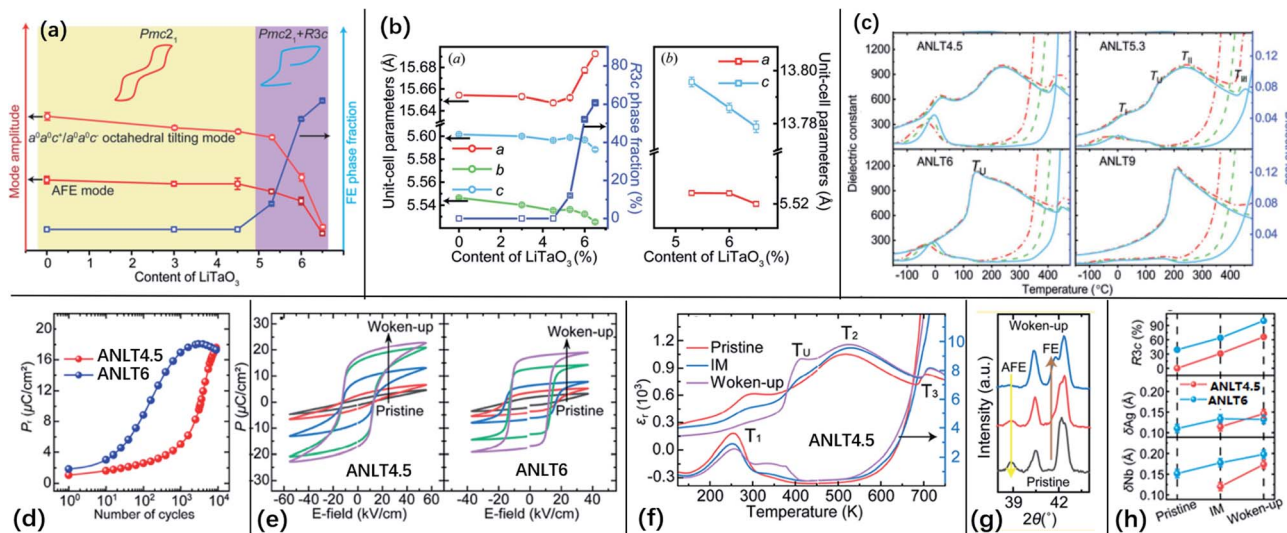


Fig. 27 (a) A schematic drawing reflecting the structure–property relationships present in the ANLT100x system in the form of symmetry modes, phases and electrical properties; (b) composition-driven variation of lattice parameters of orthorhombic  $Pmc2_1$  and rhombohedral  $R3c$  phases, as well as the phase fraction of the  $R3c$  phase; (c) temperature-dependent dielectric spectrum of ANLT100x reported by Lu *et al.*<sup>53</sup> Evolution of  $P_r$  (d) and  $P$ – $E$  hysteresis loops during bipolar cycling measured on ANLT4.5 (50  $\text{kV cm}^{-1}$  and 10 Hz) and ANLT6 (30  $\text{kV cm}^{-1}$  and 10 Hz). (f) Temperature-dependent dielectric permittivity and loss tangent (at 10 kHz) of pristine, intermediate (IM), and “woken-up” (three conditions) ANLT4.5 ceramics; (g) neutron diffraction pattern of selected peaks collected under the three conditions. (h)  $R3c$  phase fraction, with cation displacements along the  $c$ -axis for the pristine, IM, and “woken-up” conditions of ANLT4.5 and ANLT6 ceramic reported by Lu *et al.*<sup>110</sup> [(d–h) are reproduced with permission from ref. 110 copyright 2020 American Chemical Society].



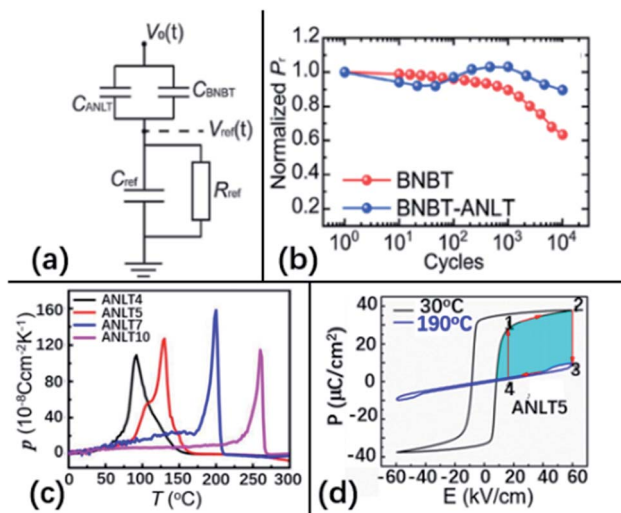


Fig. 28 (a) Schematic diagram of the measurement circuit for BNBT6-ANLT6 dual capacitors and (b) evolution of normalized  $P_r$  during bipolar cycling measured on a BNBT6 capacitor and a BNBT6-ANLT6 dual capacitor reported by Lu *et al.*<sup>110</sup> [Reproduced with permission from ref. 110 copyright 2020 American Chemical Society]. (c) Temperature-dependent pyroelectric coefficient ( $p$ ) for poled ANLT100x ceramics; (d) Olsen cycle diagram of pyroelectric energy harvesting for the ANLT5 ceramic reported by Li *et al.*<sup>111</sup>

important applications as pyroelectric detectors. It is well-known that field-induced transitions can cause an enhancement of dielectric response *via* the creation of an MPB, providing an extra contribution to the total pyroelectric coefficient. Li *et al.*<sup>111</sup> also studied the pyroelectric energy harvesting performance of ANLT5 ceramics. The energy harvesting process can be achieved using an Olsen cycle,<sup>112</sup> as schematically shown in Fig. 28d, while the harvested pyroelectric energy density ( $W$ ) can be estimated from the area of the blue region shown in the figure. The harvested energy between  $30^{\circ}\text{C}$  and  $190^{\circ}\text{C}$  was estimated to be about  $1.4 \text{ J cm}^{-3}$ , which holds promise for energy harvesting.

**3.2.3. (Ag,K)NbO<sub>3</sub> ferroelectric solid-solutions.** KNbO<sub>3</sub> ( $t = 1.06$ ) is a classic ferroelectric without AFD instability, where three well-defined ferroelectric transitions separating tetragonal, orthorhombic and rhombohedral phases are found during cooling.<sup>113</sup> These polymorphic transitions are related to the sudden change of the spontaneous polarization direction due to the B-site Nb<sup>5+</sup> ions. Hence, in the Ag<sub>1-x</sub>K<sub>x</sub>NbO<sub>3</sub> (AKN) solid-solution, by increasing the concentration of K<sup>+</sup> ions, the  $t$ -factor should gradually increase, favoring B-site driven ferroelectricity and reducing the AFD instability. Studies on the AKN system are still rare to date. Fu *et al.* successfully synthesized AKN compositions by solid-state reaction and sintering in an O<sub>2</sub> atmosphere.<sup>114</sup> The structural analysis suggested a perovskite structure with a solubility limit of K<sup>+</sup> ions approaching  $x = 0.2$ . At higher K<sup>+</sup> ion concentrations ( $0.2 < x < 0.8$ ), the perovskite solid-solution does not readily form under these conditions. However, it is possible that the perovskite solid-solution could be formed using high-pressure synthesis.<sup>115</sup> Structural analysis based on the perovskite primitive unit-cell reveals a change in

the structural parameters at  $x = 0.07$ , indicating a phase transition (see Fig. 29a), with the new phase stable up to  $x \approx 0.2$ . An orthorhombic structure is formed when  $x > 0.8$ . In 2016, Xu *et al.* suggested that an AFE<sub>O</sub>-FE<sub>O</sub> transition occurs in the range  $0.04 < x < 0.06$ , evidenced by the vanishing of AFE superlattice satellite reflections in the XRD patterns and by the presence of a ferroelectric-type hysteresis loop.<sup>116</sup> Recently, Liu *et al.*<sup>117</sup> investigated the new phase using ND, revealing a FE structure in space group  $Pmc2_1$ , consisting of four perovskite primitive pseudo-cubic unit-cells.

The dielectric behavior of the AKN system also exhibits significant changes upon composition modification, as shown in Fig. 29b. Observing the variation of the temperature-dependent dielectric behavior with composition, it can be noticed that the dielectric anomalies assigned to the  $M_1 \leftrightarrow M_2$  and  $M_2 \leftrightarrow M_3$  phase transitions are gradually suppressed, and a sharp peak denoted as  $T_{c1}$  can be found at  $x = 0.07$ . Furthermore, another anomaly denoted as  $T_{c2}$  can be noticed near  $150^{\circ}\text{C}$ , the origin of which is still unclear. Below  $T_{c2}$ , the structure could be FE  $Pmc2_1$  as reported in a recent paper.<sup>117</sup> Therefore, the dielectric anomaly denoted as  $T_{c2}$  could correspond to the  $T_f$  transition or to the  $T_C$  of the FIE phase, as reported for the case of AgNbO<sub>3</sub>.<sup>16,74</sup> This would indicate that the FIE phase was gradually replaced by an FE phase. There are no studies clarifying the structural nature in the range between  $T_{c2}$  and  $T_{c1}$ ; hence, the question whether this phase is AFE is still unsettled. Most recently, Liu *et al.*<sup>117</sup> investigated the pressure-induced variations of structure and properties in the AKN system. A composition-pressure phase diagram is outlined in Fig. 30a. The FE-AFE (note: the authors refined the structure using the AFE  $Pbcm$  model and not the FIE  $Pmc2_1$  model) transition can occur when the loading pressure reaches  $350 \text{ MPa}$  for  $x = 0.065$  (Fig. 30a and b). Meanwhile, this material shows a very high charge release density ( $\sim 37 \mu\text{C cm}^{-2}$ ) under a shock pressure of  $6.9 \text{ GPa}$  (Fig. 30c). The simulated energy density of this material is about  $5.401 \text{ J g}^{-1}$ , which is significantly higher than in other FE ceramics (Fig. 30d). The high discharge energy density after shock pressure gives a potential application for explosive energy conversion. It is noticed that after shock-pressure loading, a “double” hysteresis with non-zero remnant polarization can be observed in Fig. 30b. This feature may indicate that the pressure-induced structure is not AFE, but FIE.

### 3.3 AgNbO<sub>3</sub>-based antiferroelectric solid-solutions

In the early studies of AgNbO<sub>3</sub>, various solid-solutions, such as AgNb<sub>1-x</sub>Ta<sub>x</sub>O<sub>3</sub>, were intensively studied due to their excellent dielectric properties.<sup>19</sup> Measurements performed in the  $1 \text{ GHz}$  region showed  $\epsilon_r \approx 400$ ,  $Q \times f \geq 700 \text{ GHz}$ , and  $\Delta f/f = 0-0.22$  in the temperature range  $-40$  to  $60^{\circ}\text{C}$  for a composition close to AgNb<sub>0.5</sub>Ta<sub>0.5</sub>O<sub>3</sub>,<sup>19</sup> which represents an attractive material for high-frequency applications. In recent years, due to the development of new types of energy storage technologies, various research teams have started devoting their attention to AgNbO<sub>3</sub>-based antiferroelectric materials, due to their double  $P$ - $E$  loops. The field-induced phase transition can be significantly



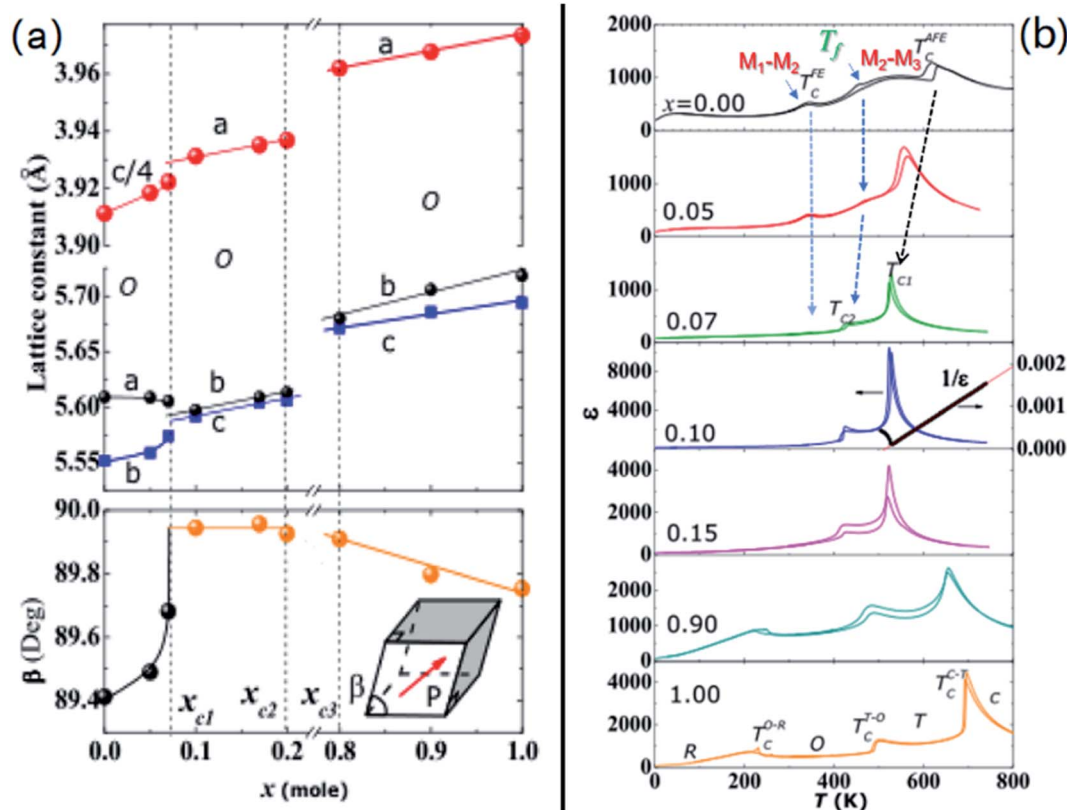


Fig. 29 (a) Composition-driven variation of pseudo-cubic parameters and (b) temperature-dependent dielectric pattern of  $(\text{Ag}_{1-x}\text{K}_x)\text{NbO}_3$  solid-solution reported by Fu *et al.*<sup>114</sup> [Reproduced from ref. 114 <https://doi.org/10.1063/1.3259410> with the permission of AIP Publishing]. The temperature  $T_C^{\text{FE}}$  was assigned by Fu *et al.*<sup>13</sup> to the Curie point, due to the discovered weak-FE/FIE  $\text{Pmc}2_1$  structure reported by Yashima *et al.* in 2011.<sup>49</sup> The temperature  $T_C^{\text{AFE}}$  indicates the Curie temperature of the AFE phase; the other dielectric anomalies were assigned to the  $M_1 \leftrightarrow M_2$  transition,  $T_f$  and the  $M_2 \leftrightarrow M_3$  transition by Kania *et al.* in 1983.<sup>14</sup>

influenced by the concentration of doping elements, giving rise to enhanced energy storage performance compared to non-doped  $\text{AgNbO}_3$ . In this section, a detailed survey of the previous fundamental studies on  $\text{AgNbO}_3$ -based antiferroelectric materials is provided with the aim of identifying a general principle to optimize the related performance, as well as providing guidelines for future studies.

**3.3.1  $\text{Ag}_{1-x}\text{Na}_x\text{NbO}_3$  solid-solution.** The  $t$ -factor of  $\text{NaNbO}_3$  ( $t = 0.967$ ) is slightly smaller than that of  $\text{AgNbO}_3$ , but like  $\text{AgNbO}_3$  ( $t = 0.98$ ),  $\text{NaNbO}_3$  also displays complex polymorphism on heating (see Fig. 31a), some of which are accompanied by the presence of dielectric anomalies.<sup>54,118</sup> However, the exact crystalline structures of these polymorphs, as well as their temperature stability ranges are still controversial.<sup>119</sup> A concise structural evolution of  $\text{NaNbO}_3$  upon cooling is shown in Fig. 31b. The intricate structural evolution indicates that  $\text{NaNbO}_3$  should exhibit more complicated interactions between AFD instabilities and off-centering of cations compared to  $\text{AgNbO}_3$ .

The high temperature phases (cubic, tetragonal and orthorhombic) have been carefully studied and have been found to be isostructural with the three paraelectric phases of  $\text{AgNbO}_3$ .<sup>47,120</sup> At room temperature, the structure of ceramic samples prepared by conventional solid-state reaction is orthorhombic

AFE, which is isostructural with  $\text{AgNbO}_3$ , although some publications have also reported the existence of a metastable  $\text{FE}_0$  phase in local regions.<sup>121,122</sup> At low temperature, the structure of ceramic samples displays a FE  $R3c$  phase, which is isostructural with the FE phase reported in ALN or ANLT systems, as discussed in Section 3.2.1.2. Previous studies on the  $\text{Ag}_{1-x}\text{Na}_x\text{NbO}_3$  (ANN) solid-solution are rare. It has been reported that ANN compositions can be obtained *via* conventional solid-state reaction and sintering in  $\text{O}_2$  atmosphere.<sup>54</sup>  $\text{AgNbO}_3$  and  $\text{NaNbO}_3$  should form a continuous solid-solution due to their similar crystalline structures.<sup>46</sup> However, the temperature-composition diagram proposed by Fu *et al.*<sup>123</sup> indicates that a phase boundary exists when the concentration of  $\text{Na}^+$  ions exceeds 0.8, as shown in Fig. 32a. It is not clear whether in the composition range  $0.8 < x < 0.9$ , a solid-solution can be formed; therefore, further studies on these compositions are needed. Kania *et al.*<sup>54</sup> investigated the dielectric spectroscopy of the ANN system (see Fig. 32b and c) to understand the microscopic mechanisms that contribute to the appearance of a diffuse dielectric peak at around 270 °C (denoted as the  $M_2 \leftrightarrow M_3$  transition in  $\text{AgNbO}_3$ ).

The diffuse dielectric response in the  $\text{Ag}_{1-x}\text{Na}_x\text{NbO}_3$  system has attracted great research interest for high-frequency applications. The temperature dependence of the dielectric

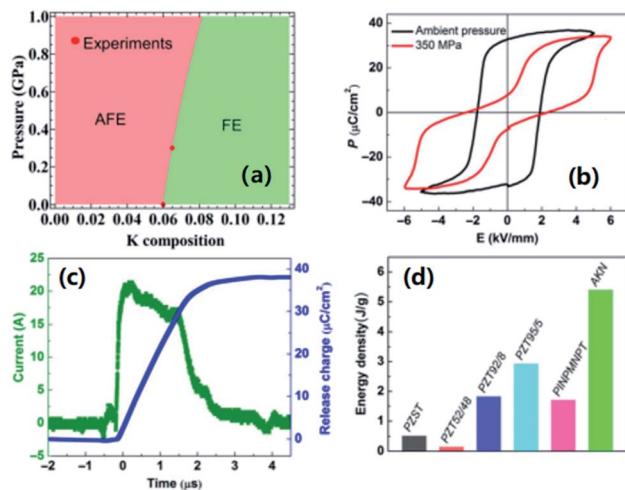


Fig. 30 (a) Pressure–composition phase diagram of the  $(\text{Ag}_{0.935}\text{K}_{0.065})\text{NbO}_3$  (AKN065) system; (b) pressure-dependent polarization–electric field ( $P$ – $E$ ) loops of AKN065 ceramics measured at room temperature; (c) dynamic discharging response of assembled AKN065 ceramic devices collected under a shock pressure of 6.9 GPa; and (d) comparison of the energy storage densities per unit weight of the AKN065 ceramics and other FE materials, reported by Liu *et al.*<sup>117</sup>

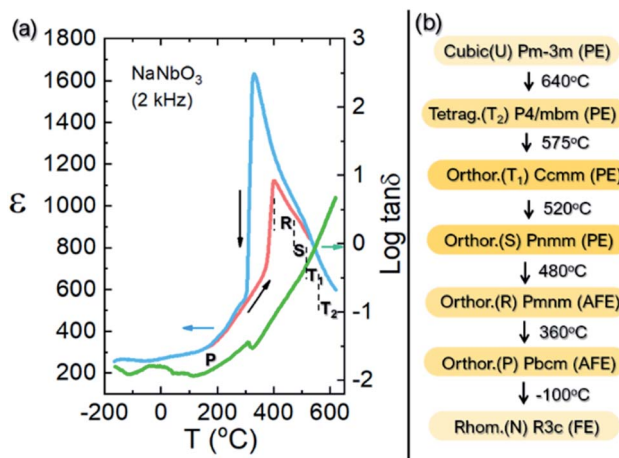


Fig. 31 (a) Temperature-dependence of dielectric permittivity and loss tangent for  $\text{NaNbO}_3$  as reported by Kania *et al.*<sup>54</sup> [reproduced from ref. 54 <https://doi.org/10.1088/0953-8984/11/45/316> with permission from copyright IOP Publishing]. (b) The reported phase transitions of  $\text{NaNbO}_3$  studied using X-ray diffraction. Note: the structural nature of the P phase, as well as its temperature stability range are still controversial (see ref. 119, Fig. 1).

permittivity (see Fig. 32b) shows that the high-temperature anomaly gradually shifts to higher temperature with increasing Na concentration, up to  $x = 0.6$ , and starts shifting to lower temperature at higher Na content, with an increase in the value of the dielectric permittivity near the transition point. The low temperature anomaly monotonously shifts to higher temperature with increasing Na concentration, while the dielectric permittivity decreases near the transition point. The gradual suppression of the diffuse dielectric anomaly with

increasing  $\text{Na}^+$  ion concentration can be more clearly observed in the dielectric pattern obtained during cooling (see Fig. 32c). In particular, upon cooling, the dielectric peak corresponding to the AFE–PE transition presents a negligible shift with composition. Kania *et al.*<sup>54</sup> proposed that Na substitution causes a decrease of the oxygen octahedron size, diminishing the freedom of  $\text{Nb}^{5+}$  ions, resulting in a gradual decrease of the diffuse dielectric anomaly. This viewpoint is consistent with the gradual decrease of the  $t$ -factor due to the occupation of the A-site by  $\text{Na}^+$ , which makes it harder for the  $\text{Nb}^{5+}$  ions to move off-center in their oxygen octahedra. Furthermore, it was reported that this solid-solution exhibits a large electric field-induced strain response of about 0.2% over the wide compositional range  $0.4 < x < 0.8$  (see Fig. 33), which might be interesting for the development of lead-free actuators.

**3.3.2  $\text{Ag}(\text{Nb,Ta})\text{O}_3$  solid-solutions.** Among  $\text{AgNbO}_3$ -based materials, the system  $\text{Ag}(\text{Nb,Ta})\text{O}_3$  (ANT) has been extensively investigated for its excellent dielectric properties in the microwave range, as mentioned above.<sup>19,124–126</sup> Recent studies have demonstrated that this solid-solution also exhibits excellent energy storage performance.<sup>17</sup> At 0.64 Å, the radius of the  $\text{Ta}^{5+}$  ion is identical to that of the  $\text{Nb}^{5+}$  ion. However, various Ta-containing perovskites, such as  $\text{KTaO}_3$  and  $\text{AgTaO}_3$ , do not exhibit anti-/ferroelectricity.<sup>127,128</sup> The reason can be attributed to weak/non-existent off-centering of the  $\text{Ta}^{5+}$  cation in the oxygen octahedron due to the nature of the Ta–O bond. Valant *et al.* reported a survey on these systems in 2007, with discussions focused on the synthesis methods, crystal chemistry and dielectric properties.<sup>126</sup> Similarly to  $\text{AgNbO}_3$ , the ANT solid-solution can be synthesized by solid-state reaction and conventional sintering in  $\text{O}_2$  atmosphere. In order to obtain a homogenous distribution of  $\text{Nb}^{5+}$  and  $\text{Ta}^{5+}$  and avoid the formation of undesired phases, it has been frequently proposed that  $\text{Nb}_2\text{O}_5$  and  $\text{Ta}_2\text{O}_5$  should be initially mixed and heated above 1200 °C to form a chemically homogenous  $(\text{Nb,Ta})_2\text{O}_5$  powder, which could be used as a precursor and subsequently mixed with  $\text{Ag}_2\text{O}$ . Wet chemical methods have also been used to synthesize ANT solid-solutions, to reduce the sintering temperature or to prepare thin films using spin-coating techniques.<sup>31</sup> In early studies, the temperature–composition phase diagram of ANT were proposed by Pawełczyk *et al.*<sup>55</sup> as shown in Fig. 34a. It can be observed that with increasing  $\text{Ta}^{5+}$  ion concentration, the dielectric peak related to the  $\text{M}_3 \leftrightarrow \text{O}_1$  or AFE–PE transition shows a slight shift toward high temperature, while the dielectric anomalies assigned to the M-type phase transitions, significantly shift toward lower temperature. In particular, the broad dielectric response ( $\epsilon \approx 400$ ), corresponding to the  $\text{M}_2 \leftrightarrow \text{M}_3$  transition, shifts to around room temperature in compositions with  $0.4 \leq x \leq 0.6$  (see Fig. 34b).<sup>126,129</sup> Furthermore, the high-frequency dielectric spectrum reveals that the dielectric permittivity is very stable up to the submillimeter wave region (see Fig. 34c), and a dielectric dispersion is present when the temperature approaches the broad dielectric anomaly related to the  $\text{M}_2 \leftrightarrow \text{M}_3$  transition. Further Raman spectroscopic studies have suggested that the presence of a dielectric dispersion in the submillimeter range is intimately related to the disordered off-centering motion of



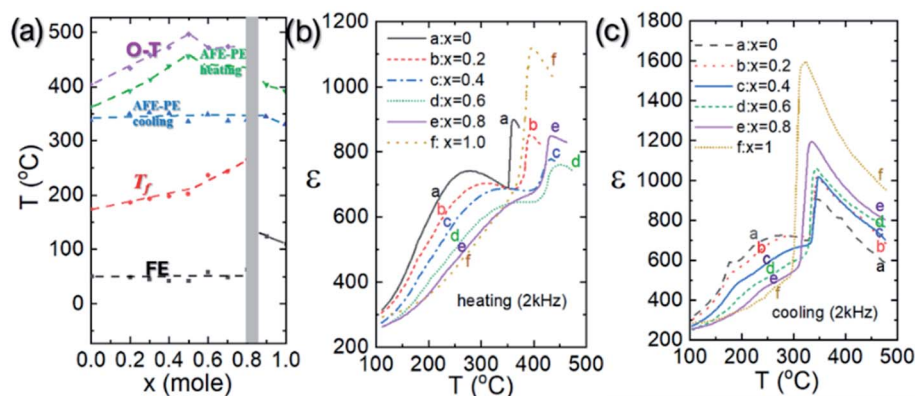


Fig. 32 (a) Variation of transition temperatures in the  $\text{Ag}_{1-x}\text{Na}_x\text{NbO}_3$  system derived from the dielectric measurements reported by Fu *et al.*<sup>123</sup> [Reproduced from ref. 123 <https://doi.org/10.1063/1.3609234> with the permission of AIP Publishing]. Temperature-dependence of dielectric permittivity in  $\text{Ag}_{1-x}\text{Na}_x\text{NbO}_3$  (b) on heating and (c) on cooling, reported by Kania *et al.*<sup>54</sup> [Reproduced from ref. 54 <https://doi.org/10.1088/0953-8984/11/45/316> with permission from copyright IOP Publishing].

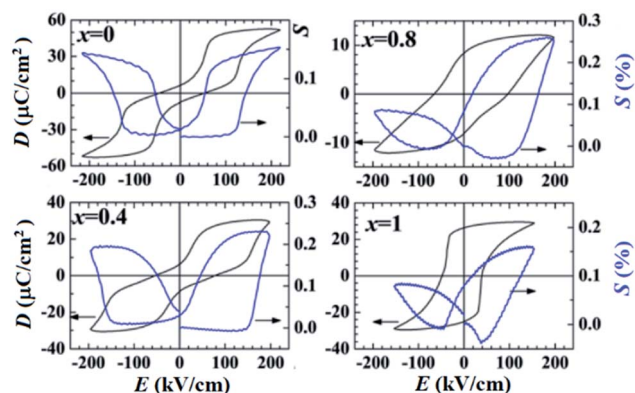


Fig. 33  $P$ - $E$  and  $S$ - $E$  hysteresis loops of ANN solid-solution reported by Fu *et al.*<sup>123</sup> [Reproduced from ref. 123 <https://doi.org/10.1063/1.3609234> with the permission of AIP Publishing].

$\text{Nb}^{5+}$  ions, which could be the origin of the broad dielectric response. This deduction was further confirmed by a structure investigation carried out on ANT50 samples by Levin *et al.*,<sup>130</sup> who proposed the presence of partial ordering of local B-cation displacements, when the temperature approaches the broad maximum of dielectric permittivity corresponding to the  $M_2 \leftrightarrow M_3$  transition. Tantalum substitution suppresses this ordering because of the dissimilar off-centering trends for  $\text{Ta}^{5+}$  and  $\text{Nb}^{5+}$ , as revealed by EXAFS results.<sup>130</sup> Therefore, the dielectric anomalies assigned to the M-type transitions shift to lower temperature with increasing  $\text{Ta}^{5+}$  ion concentration. Additionally, it was suggested that  $\text{Ag}^+$  cations also exhibit displacive disorder, and on cooling, undergo ordering coupled to that of the B-cations, which could explain the enhanced dielectric response assigned to the  $M_1 \leftrightarrow M_2$  transition. Although in an early study it was suggested that all of the three M phases show the same structure with the non-polar  $Pbcm$  space group,<sup>47</sup> the dielectric response under DC electric field reported by Li *et al.*<sup>131</sup> (shown in Fig. 34d) reveals that a composition-driven FIE-AFE transition occurs with increasing  $\text{Ta}^{5+}$  ion concentration. For

pure  $\text{AgNbO}_3$ , the dielectric permittivity gradually decreases until the DC bias field reaches  $50 \text{ kV cm}^{-1}$ , above which it gradually increases with increasing DC bias. The  $P$ - $E$  loop exhibits a large hysteresis, consistent with the recently-discovered ferroelectric structure in space group  $Pmc2_1$ .<sup>49</sup> The unique dielectric tunability might be of interest for dielectric tunable devices. For ANT50, the dielectric permittivity hardly changes under weak DC electric fields, but significantly increases under higher DC fields, indicating that the virgin structure does not present ferroelectricity, but antiferroelectricity. The ferroelectric nature has also been evidenced in our recent work.<sup>16,74</sup> Based on high-quality ANT samples, Zhao *et al.* characterized the  $P$ - $E$  loops of the ANT100x system with  $x \leq 0.20$ ,<sup>132</sup> extrapolating the key parameters listed in Table 3. One can notice that with increasing Ta concentration, the characteristic fields  $E_B$  and  $E_F$  are enhanced, the remnant polarization gradually decreases and the field-induced FE polarization presents negligible changes, indicating a more stable antiferroelectric state. Most recently, Luo *et al.* synthesized ANT100x ceramic samples with ultrahigh quality, and systematically characterized their ferroelectric properties under high electric fields.<sup>17</sup> As shown in Fig. 35a, the field-induced double  $P$ - $E$  loops of ANT100x gradually evolve into linear-like loops accompanied by reduced polarization with increasing Ta concentration, indicating the gradual decrease or even the suppression of the field-induced ferroelectric state under strong-field conditions. For ANT55 ceramics, the  $P$ - $E$  loop exhibits an almost linear trend, while the unipolar  $I$ - $E$  loop of ANT55 ceramic reveals a couple of weak current peaks, indicating that there are still a few AFE regions in the  $Pbcm$  non-polar lattice.

**3.3.3 A-site deficient  $\text{AgNbO}_3$ -based solid-solutions.** As discussed above, reducing the  $t$ -factor can favor AFD instabilities, promoting antiferroelectricity. Hetero-valent ( $3+$  or  $2+$ ) elements have been considered for A-site dopants as an alternative strategy to develop  $\text{AgNbO}_3$ -based antiferroelectric solid-solutions. In these systems, A-site vacancies are created to maintain charge neutrality, such as in  $\text{BiNb}_3\text{O}_9$  (*i.e.*,  $\text{Bi}_{1/3}\text{NbO}_3$ )



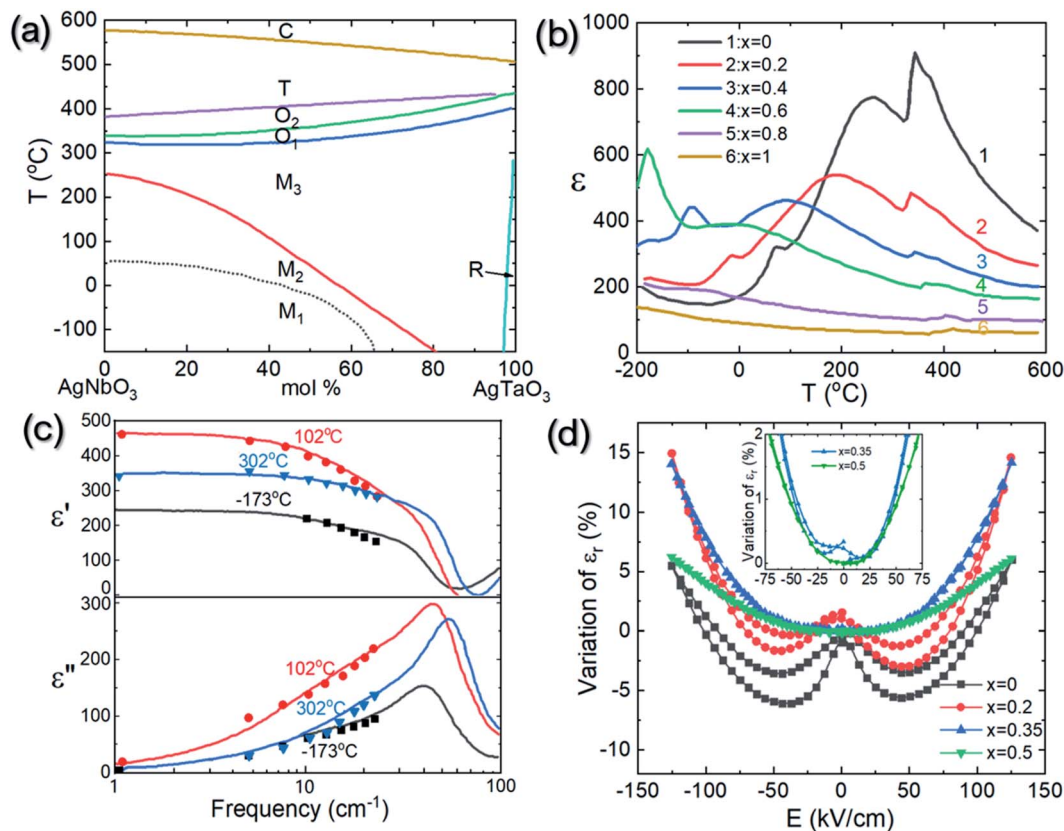


Fig. 34 (a) Temperature–composition phase diagram of the AgNb<sub>1-x</sub>Ta<sub>x</sub>O<sub>3</sub> reported by Pawełczyk *et al.*<sup>55</sup>; [reproduced from ref. 55 <https://doi.org/10.1080/01411598708220073> with permission from Taylor & Francis Informa Ltd]. (b) Temperature-dependent dielectric pattern of the AgNb<sub>1-x</sub>Ta<sub>x</sub>O<sub>3</sub> system original reported by Kania *et al.*<sup>129</sup> [reproduced with permission from ref. 129 <https://doi.org/10.1080/01411598308244116> with permission from Taylor & Francis Informa Ltd]. (c) The dielectric dispersion in the millimeter and submillimeter wave region of AgNb<sub>0.6</sub>Ta<sub>0.4</sub>O<sub>3</sub> as calculated from the oscillator and relaxor additive model (solid lines) at different temperatures; points are experimental values reported by Volkov *et al.*<sup>58</sup> [reproduced from ref. 58 <https://doi.org/10.1088/0953-8984/7/4/009> with permission from copyright IOP Publishing]. (d) Variation of dielectric permittivity of AgNb<sub>1-x</sub>Ta<sub>x</sub>O<sub>3</sub> ceramics at 100 kHz with a maximum electric field bias of 125 kV cm<sup>-1</sup> as reported by Li *et al.*<sup>131</sup> [Reproduced from Ref. 131 <https://doi.org/10.1063/1.4875581> with the permission of AIP Publishing].

Table 3 List of the key parameters of field-induced transitions in ANT100x systems reported by Zhao *et al.*<sup>132</sup>

Samples	$P_{\max}$ [ $\mu\text{C cm}^{-2}$ ]	$P_r$ [ $\mu\text{C cm}^{-2}$ ]	$E_F$ [ $\text{kV cm}^{-1}$ ]	$E_B$ [ $\text{kV cm}^{-1}$ ]	$\Delta E$ [ $\text{kV cm}^{-1}$ ]
AN	35.9	3.6	110	52	58
ANT5	37.2	2.0	153	95	58
ANT10	35.5	1.9	166	110	56
ANT15	36.8	1.7	189	130	59
ANT20	31.5	1.2	222	163	59

and LaNb<sub>3</sub>O<sub>9</sub> (*i.e.*, La<sub>1/3</sub>NbO<sub>3</sub>).<sup>133,134</sup> Structural investigations revealed that these materials exhibit ordering of A-site cations and vacancies. Table 4 lists various M<sub>1/3</sub>NbO<sub>3</sub> compounds and their crystalline structures.<sup>135</sup>

In 2006, Hu *et al.* successfully prepared a bismuth-modified AgNbO<sub>3</sub> solid-solution (Ag<sub>1-x</sub>Bi<sub>x/3</sub>NbO<sub>3</sub>).<sup>136</sup> The structure analysis suggested that about 20 mol% of Bi<sup>3+</sup> ions can be incorporated into the lattice of AgNbO<sub>3</sub> to form a perovskite structured solid-solution without impurity phases, implying that 40% of the A-sites are vacant. A dielectric study on this

solid-solution indicated that all dielectric anomalies shift to lower temperature and the diffuse dielectric peak corresponding to the M<sub>2</sub> ↔ M<sub>3</sub> transition becomes broader with increasing Bi concentration (see Fig. 36). In particular, the dielectric permittivity is almost independent on temperature in the range from -50 °C to 50 °C for some compositions, making them of interest for capacitor applications. When  $x \geq 0.4$ , the sharp maximum corresponding to the AFE–PE phase transition disappeared and the response of a typical relaxor was observed. Our research team has also successfully prepared

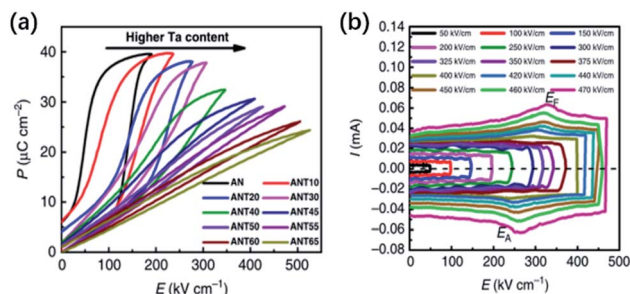


Fig. 35 (a)  $P$ - $E$  loops of ANT100 $x$  ceramics and (b)  $I$ - $E$  loops of ANT55 ceramics reported by Luo *et al.*<sup>17</sup>

Table 4 List of various A-site deficient niobates<sup>135</sup>

Chemical formula	Mineral	Crystal system	Space group
$\text{Bi}_{1/3}\text{NbO}_3$ ( $\text{Bi}_3\text{NbO}_9$ )	Perovskite	Tetragonal	$P4/mmm$
$\text{La}_{1/3}\text{NbO}_3$	Perovskite	Orthorhombic	$Pmmm$ or $Cmmm$
$\text{Ce}_{1/3}\text{NbO}_3$	Perovskite	Monoclinic	$P2_1/m$
$\text{Pr}_{1/3}\text{NbO}_3$	Perovskite	Orthorhombic	$Cmmm$
$\text{Nd}_{1/3}\text{NbO}_3$	Perovskite	Orthorhombic	$Cmmm$

$\text{Ag}_{1-3x}\text{Bi}_x\text{NbO}_3$  solid-solutions with high-quality.<sup>137</sup> The dielectric response under DC field reveals that the ferroelectric/weak-ferroelectric behavior progressively disappears when  $x$  reaches 0.04 (corresponding to  $x = 0.12$  according to the chemical formula reported by Hu *et al.*<sup>136</sup>), as shown in Fig. 37a. Above  $x = 0.04$ , the ceramic samples exhibit a single AFE phase. The enhanced antiferroelectric stability is evidenced by the  $D$ - $E$  loops under strong-field cycling as shown in Fig. 37b. By increasing the  $\text{Bi}^{3+}$  ion concentration, the characteristic fields of the AFE-FE field-induced transition increase, indicating a higher energy barrier between AFE and FE phases.

Based on the design and the successful preparation of the A-site deficient  $\text{Ag}_{1-3x}\text{Bi}_x\text{NbO}_3$  solid-solution, a series of other A-site deficient  $\text{Ag}_{1-3x}\text{Ln}_x\text{NbO}_3$  (Ln: rare-earth elements,  $x < 0.05$ ) ceramics have been subsequently reported,<sup>138-141</sup> in the search for new materials with improved energy storage performance.

The temperature dependence of the dielectric permittivity of various A-site deficient compounds is reported in Fig. 38. It can be noticed that  $\text{La}^{3+}$  doping induces similar trends to those produced by  $\text{Bi}^{3+}$  doping, while  $\text{Sm}^{3+}$  and  $\text{Gd}^{3+}$  doping shifts the maximum dielectric peak related to the AFE-PE transition with a decrease of the dielectric permittivity. The transition almost disappears for 4 mol%  $\text{Sm}^{3+}$  and  $\text{Gd}^{3+}$  doping. This might suggest that the dopant elements are also present on the B-site of the  $\text{AgNbO}_3$  lattice, resulting in composition fluctuation and breaking of the AFE distortion. These features also suggest that the solubility limit of  $\text{Sm}^{3+}/\text{Gd}^{3+}$  in  $\text{AgNbO}_3$  is around 4 mol%. As with Bi-doped  $\text{AgNbO}_3$ , the characteristic field  $E_F$  increases with increasing  $\text{Ln}^{3+}$  concentration, while the field-induced polarization visibly reduces as shown in Fig. 39a-c. The composition-driven property variation is very similar to the changes observed in the ferroelectric loops of AKN6 ceramics

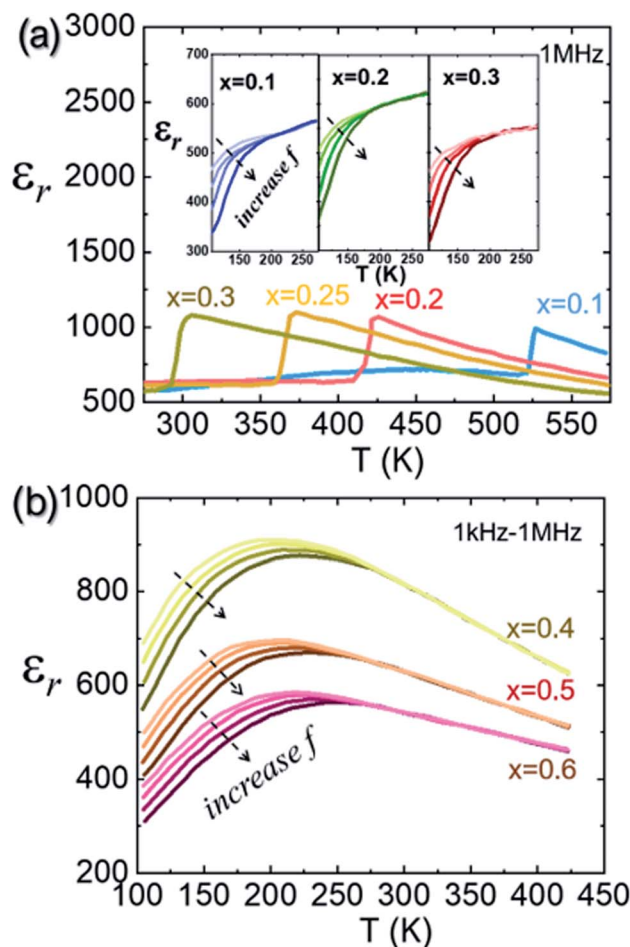


Fig. 36 Temperature-dependent dielectric spectra of  $\text{Ag}_{1-x}\text{Bi}_{x/3}\text{NbO}_3$  ceramics reported by Hu *et al.*<sup>136</sup> [Reproduced from ref. 136 <https://doi.org/10.1063/1.2209552> with the permission of AIP Publishing].

with increasing pressure shown in Fig. 39d.<sup>117</sup> Similar phenomena have also been observed in rare-earth substituted  $\text{BiFeO}_3$ , in which the field-induced FE polarization is significantly suppressed at high concentrations of rare-earth dopants.<sup>108,109</sup> It can be inferred that the reduced  $t$ -factor weakens the AFE distortion and reduces the freedom of cation displacement. In addition to  $\text{Ln}^{3+}$ -doped  $\text{AgNbO}_3$ , alkaline earth elements such as  $\text{Ca}^{2+}$  have been also used as dopants for  $\text{AgNbO}_3$  and showed similar effects to  $\text{La}^{3+}$  doping.<sup>143</sup>

### 3.3.4 Other $\text{AgNbO}_3$ -based antiferroelectric solid-solutions.

Besides the chemical modification of  $\text{AgNbO}_3$  via A- or B-site doping, co-doping of A- and B- sites through formation of solid-solutions between  $\text{AgNbO}_3$  and  $\text{A}^{2+}\text{B}^{4+}\text{O}_3$  or  $\text{A}^{3+}\text{B}^{3+}\text{O}_3$  perovskites with smaller  $t$ -factors (such as  $\text{CaTiO}_3$ ,  $\text{Bi}_{0.5}\text{Na}_{0.5}\text{TiO}_3$ , and  $\text{BiMnO}_3$ ) represents an alternative strategy to achieve enhanced stability of the AFE state and develop new materials with improved energy storage performance.<sup>144-146</sup> These materials also show enhanced antiferroelectric stability with increasing concentration of the second perovskite compound added, as shown in Fig. 40a and b. These perovskite dopants can severely interrupt the ordered arrangement of antipolar A-





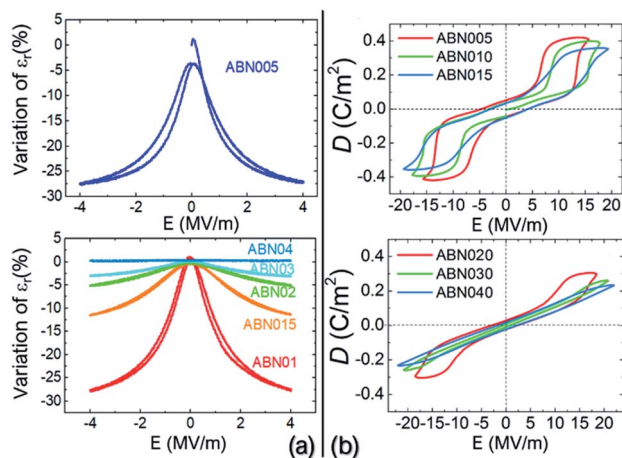


Fig. 37 (a) Percentage variation in relative permittivity ( $\epsilon_r$ ) at low DC field in  $(\text{Ag}_{1-3x}\text{Bi}_x)\text{NbO}_3$  (ABNx) ceramics, and (b) high-field ferroelectric  $D$ - $E$  loops of ABNx ceramics reported by Tian *et al.*<sup>137</sup>

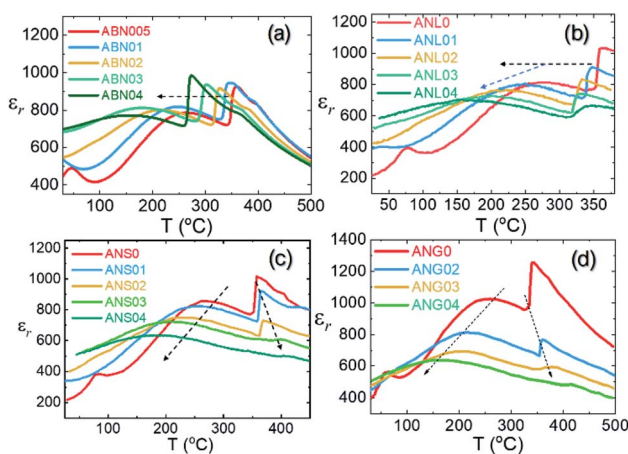


Fig. 38 Temperature-dependent dielectric spectra of (a)  $\text{Ag}_{1-3x}\text{Bi}_x\text{NbO}_3$  (reported by Tian *et al.*<sup>137</sup>); (b)  $\text{Ag}_{1-3x}\text{La}_x\text{NbO}_3$  (reported by Gao *et al.*<sup>142</sup>); (c)  $\text{Ag}_{1-3x}\text{Sm}_x\text{NbO}_3$  (reported by Gao *et al.*<sup>138</sup>) [reproduced with permission from ref. 138 copyright 2020 American Chemical Society] and (d)  $\text{Ag}_{1-3x}\text{Gd}_x\text{NbO}_3$  (reported by Li *et al.*<sup>140</sup>).

site and B-site cations and reduce the domain size, giving rise to relaxor-like antiferroelectric behavior accompanied by a gradual decrease of the field-induced FE polarization. Recently, Yan *et al.* reported on the  $(1-x)\text{AgNbO}_3-x\text{Bi}(\text{Zn}_{2/3}\text{Nb}_{1/3})\text{O}_3$  system, highlighting that the changes in relative permittivity before and after DC poling can be also used to study the polarization behavior in  $\text{AgNbO}_3$ -based ceramics, as shown in Fig. 40c.<sup>147</sup> For  $x = 0.000$  and  $x = 0.005$  compositions, the relative permittivity shows a small decrease after DC poling, which can be attributed to decreased domain wall density. In contrast, for  $x = 0.030$ , the relative permittivity increased on successive polarization under low and high electric fields due to the contribution of the field-induced polar structure. In the case of the  $x = 0.010$  composition, intermediate behavior is observed, with a decrease in relative permittivity on the initial polarization at low field, followed by an increase at high field. The initial decrease in

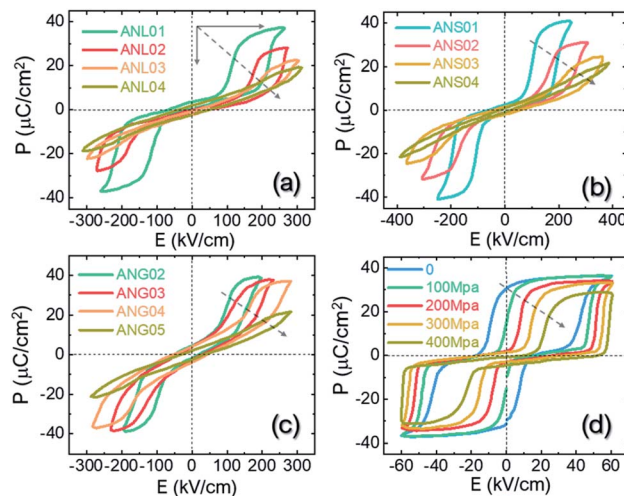


Fig. 39  $P$ - $E$  hysteresis loops of (a)  $\text{Ag}_{1-3x}\text{La}_x\text{NbO}_3$ ,<sup>139</sup> (b)  $\text{Ag}_{1-3x}\text{Sm}_x\text{NbO}_3$ ,<sup>138</sup> [reproduced with permission from ref. 138 copyright 2020 American Chemical Society] (c)  $\text{Ag}_{1-3x}\text{Gd}_x\text{NbO}_3$  (ref. 140) and (d)  $\text{Ag}_{0.94}\text{K}_{0.06}\text{NbO}_3$  (ref. 117) under different pressures.

permittivity is attributed to the decrease in FE domain wall density, while the increase in permittivity at higher field arises from the metastable FIE structure.

#### 4. $\text{AgNbO}_3$ -based materials for high-power energy storage

The direct way to calculate the total stored energy density ( $W$ ) of a dielectric material is through integration of the  $D$ - $E$  curve:

$$W = \int_0^D E dD \quad (3)$$

where  $E$  is the electric field and  $D$  is the electric displacement. In dielectrics,  $D = \epsilon_0 E + P$ ; however, since the polarization of ferroelectrics is very high,  $D \approx P$ . Thus, the recoverable energy storage density ( $W_{\text{rec}}$ ) for anti-/ferroelectrics can be obtained by a modified expression as:

$$W_{\text{rec}} = \int_{P_r}^{P_{\text{max}}} E dP \quad (4)$$

where  $P_{\text{max}}$  is the polarization under the maximum external applied electric field, while  $P_r$  is the remnant polarization under zero electric field. Due to the difference between  $W$  and  $W_{\text{rec}}$ , a parameter to simulate the energy efficiency ( $\eta$ ) can be expressed as:

$$\eta = \frac{W_{\text{rec}}}{W} \quad (5)$$

where  $W$  is the total stored energy density. A schematic showing the electrical energy storage of dielectrics is shown in Fig. 41, in which the colored area represents the recoverable energy density  $W_{\text{rec}}$  of the dielectric material.

From Fig. 41, it is clear that under the same applied field, antiferroelectrics should display higher energy density compared to the other types of dielectrics. A summary of the



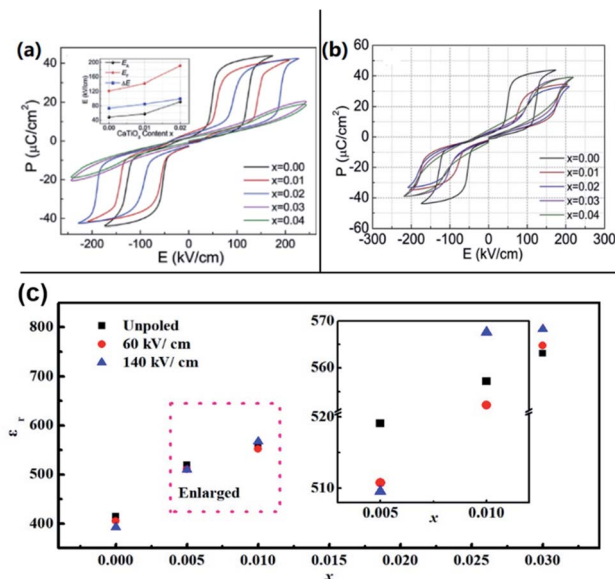


Fig. 40  $P$ - $E$  loops of (a)  $(1-x)\text{AgNbO}_3-x\text{CaTiO}_3$  (ref. 144) [reproduced with permission from ref. 144 copyright 2020 by Elsevier]; (b)  $(1-x)\text{AgNbO}_3-x(\text{Bi}_{0.5}\text{Na}_{0.5})\text{TiO}_3$ ,<sup>145</sup> [reproduced with permission from ref. 145 copyright 2020 by Elsevier] and (c)  $\text{Bi}(\text{Zn}_{2/3}\text{Nb}_{1/3})\text{O}_3$  (BZN) concentration dependence of relative permittivity of  $(1-x)\text{AN}-x\text{BZN}$  ceramics before and after poling.<sup>147</sup>

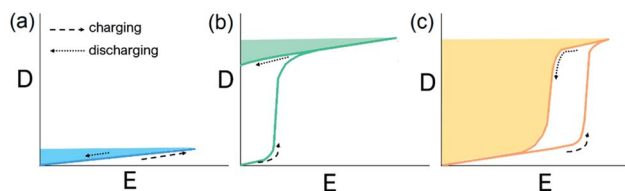


Fig. 41 Schematic energy storage density (colored area) of (a) linear dielectrics, (b) ferroelectrics, and (c) antiferroelectrics under the same applied electric field.

energy storage performances of  $\text{AgNbO}_3$ -based antiferroelectric materials is reported in Table 5. It can be noticed that the value of  $W_{\text{rec}}$  in chemically-modified  $\text{AgNbO}_3$  ceramics can reach  $6 \text{ J cm}^{-3}$ , which is three times that of unmodified  $\text{AgNbO}_3$  ceramics. Besides the composition modification, the  $W_{\text{rec}}$  of  $\text{AgNbO}_3$ -based ceramic capacitors can be greatly improved through multilayer chip ceramic technology, as reported by Zhu *et al.* in a recent study.<sup>148</sup> In particular, an  $\text{AgNbO}_3$  epitaxial film grown on a (001) STO substrate reported by Zhang *et al.* showed an ultrahigh characteristic field  $E_F$  ( $\sim 250 \text{ kV cm}^{-1}$ ) while the  $E_B$  remained unchanged compared to an  $\text{AgNbO}_3$  ceramic sample. Nevertheless, the film showed a higher  $W_{\text{rec}}$  than the ceramic sample of  $\text{AgNbO}_3$ .<sup>44</sup>

The energy storage efficiency in these materials is usually less than 70%. The main advantage of antiferroelectric capacitors compared to linear dielectrics is the possibility of a large field-induced polarization under certain applied electric fields. An increase of the characteristic field  $E_F$  can give rise to an enhancement of  $W_{\text{rec}}$  for an antiferroelectric material. Another

key parameter to impact the energy storage performance is the degree of hysteresis. Nano-structured domains or polar nano-regions (PNRs) generally found in relaxor-ferroelectrics lead to slim  $D$ - $E$  loops accompanied by a large FE polarization, as observed in  $\text{PbMg}_{1/3}\text{Nb}_{2/3}\text{O}_3$  and  $\text{BaTiO}_3$ -based compounds. Therefore, further efforts should be directed to develop  $\text{AgNbO}_3$ -based relaxors/anti-/ferroelectrics as proposed in Fig. 42 to improve the energy efficiency  $\eta$ .

Considering practical applications of dielectric capacitors, the energy storage capability per unit volume of a linear dielectric material can generally be estimated using the following expression:<sup>160</sup>

$$\overline{W} = \frac{1}{2} \frac{CV^2}{\text{volume}} = \frac{1}{2} \varepsilon_0 \eta^2 E_b^2 \quad (6)$$

where  $C$  is the capacitance,  $V$  is the voltage,  $\varepsilon$  is the dielectric permittivity of the material,  $\varepsilon_0$  is the dielectric permittivity of a vacuum, and  $E_b$  is the breakdown field. The dielectric is susceptible to breakdown under the application of high external fields. To avoid any premature breakdown of the capacitor, it is advised that the maximum applied external field should be about half of the value of  $E_b$  (*i.e.*,  $V = \eta V_b$  ( $0 < \eta < 1$ )).<sup>161</sup> This means that the value of  $E_b$  should be at least two times higher than the characteristic field  $E_F$  of the antiferroelectric capacitor. For dielectrics,  $E_b$  is affected by several factors including intrinsic factors such as the band gap, and extrinsic factors such as grain size, porosity and sample geometry. Here, we briefly summarize these factors to provide guidance for obtaining larger  $E_b$  values. Widening the band gap will result in a higher  $E_b$ , because the narrower the band gap, the easier is the jump of the electrons from the valence band to the conduction band, increasing the possibility of intrinsic breakdown.<sup>162</sup> Research has shown that increasing the sample density, *i.e.* decreasing the cavities existing in the ceramic body, is beneficial to obtaining high  $E_b$  values.<sup>163,164</sup> In addition, lowering grain size is also particularly useful in increasing  $E_b$ . Tunkasiri and Rujjanagul proposed an empirical relationship between breakdown strength and grain size, *i.e.*,  $E_b \propto (\text{grain size})^{-0.5}$ , in the micron range for  $\text{BaTiO}_3$  ceramics.<sup>165</sup> Similar results were also reported for  $\text{MgO}$  ceramics with grains in the sub-micron size range.<sup>166</sup> Recently, Yang *et al.* systematically fitted the dependence of  $E_b$  on the grain size in various ceramic systems, including  $\text{MgO}$ ,  $\text{Al}_2\text{O}_3$  and  $\text{BaTiO}_3$ -based systems, and proposed a relationship of the type  $E_b \propto (\text{grain size})^{-a}$ , where  $a$  is in the range of 0.2–0.4.<sup>167</sup> From these two relationships, it can be concluded that reducing grain size leads to an improved  $E_b$ . This is mainly due to the fact that smaller grains can lead to higher grain boundary density and provide more depletion regions, which could hinder charge carrier transport, increasing the resistivity of the material.<sup>165–169</sup> The trend described above has also been observed in  $\text{AgNbO}_3$ -based ceramics, as shown in Fig. 43. The  $E_b$  values can be evaluated using the Weibull distribution, which has been widely used in ceramics and polymer-based systems.<sup>132,139–141,144,147,149,153,154</sup>

Additionally, it is generally believed that the sample thickness has a significant effect on the  $E_b$ , because fewer defects

Table 5 Key parameters and energy storage performance of currently reported AgNbO<sub>3</sub>-based ceramics<sup>a</sup>

Materials compositions	$D_{\max}$ ( $\mu\text{C cm}^{-2}$ )	$E_F$ ( $\text{kV cm}^{-1}$ )	$\Delta E$ ( $\text{kV cm}^{-1}$ )	$E_b$ ( $\text{kV cm}^{-1}$ )	$W_{\text{rec}}$ ( $\text{J cm}^{-3}$ )	$\eta$ (%)	Ref.
AgNbO <sub>3</sub> (AN)	41	125	75	175	2.1	39	15
AN epitaxial film	~40	~250	~200	624	5.8	55.8	44
AN-0.1 wt% WO <sub>3</sub>	~42	~165	75	200	3.3	50	149
AN-0.1 wt% MnO <sub>2</sub>	37	117	54	150	2.5	57	150
Ag <sub>0.97</sub> Bi <sub>0.03</sub> NbO <sub>3</sub>	26	—	—	205	2.6	86	137
Ag <sub>0.94</sub> La <sub>0.02</sub> NbO <sub>3</sub>	~28	~250	~75	273	4.4	73	139
Ag <sub>0.94</sub> La <sub>0.02</sub> NbO <sub>3</sub>	~30	~175	~75	230	3.12	63	151
Ag <sub>0.94</sub> Sm <sub>0.02</sub> NbO <sub>3</sub>	~30	~270	~80	~300	4.5	63	138
Ag <sub>0.91</sub> Sm <sub>0.03</sub> NbO <sub>3</sub>	~38	~225	~75	~290	5.2	68	141
Ag <sub>0.97</sub> Sm <sub>0.01</sub> NbO <sub>3</sub>	~40	~125	~40	~175	3.8	73	152
Ag <sub>0.88</sub> Gd <sub>0.04</sub> NbO <sub>3</sub>	~34	~250	80	290	4.5	63	140
Ag <sub>0.97</sub> Nd <sub>0.01</sub> NbO <sub>3</sub>	~40	~150	~70	210	3.2	52	153
Ag <sub>0.92</sub> Ca <sub>0.04</sub> NbO <sub>3</sub>	39	179	80	220	3.5	56	143
Ag(Nb <sub>0.85</sub> Ta <sub>0.15</sub> )O <sub>3</sub>	~37	189	59	240	4.2	69	132
Ag(Nb <sub>0.8</sub> Ta <sub>0.2</sub> )O <sub>3</sub>	37	175	55	270	3.7	58	78
Ag(Nb <sub>0.45</sub> Ta <sub>0.55</sub> )O <sub>3</sub>	~28	~300	~60	470	6.3	90	17
ANT15 multilayer ceramic	~44	~190	~90	1020	7.9	71	148
Sm <sub>0.02</sub> Ag <sub>0.94</sub> Nb <sub>0.9</sub> Ta <sub>0.1</sub> O <sub>3</sub>	38.5	~220	~80	280	4.87	63.5	154
Nd <sub>0.01</sub> Ag <sub>0.97</sub> Ta <sub>0.2</sub> Nb <sub>0.8</sub> O <sub>3</sub>	~40	~275	~75	370	6.5	71	155
Ag <sub>0.94</sub> La <sub>0.02</sub> Nb <sub>0.8</sub> Ta <sub>0.2</sub> O <sub>3</sub>	~50	~300	~100	~400	6.9	71	156
Ag <sub>0.9</sub> Ca <sub>0.05</sub> Nb <sub>0.95</sub> Ta <sub>0.05</sub> O <sub>3</sub>	~30	~190	~120	210	3.36	58.3	157
0.99AN-0.01BZN	54	198	80	220	4.6	57	147
0.97AN-0.03BNT	35	180	80	220	3.4	62	145
0.98AN-0.02CT	40	190	90	220	3.7	52	144
0.94AN-0.06CTH	40	266	98	300	5.4	66	158

<sup>a</sup> BZN: Bi(Zn<sub>1/3</sub>Nb<sub>2/3</sub>)O<sub>3</sub>; BNT: Bi<sub>0.5</sub>Na<sub>0.5</sub>TiO<sub>3</sub>; CT: CaTiO<sub>3</sub>; CTH: Ca(Ti,Hf)O<sub>3</sub>; ANT15: Ag(Nb<sub>0.85</sub>Ta<sub>0.15</sub>)O<sub>3</sub>.

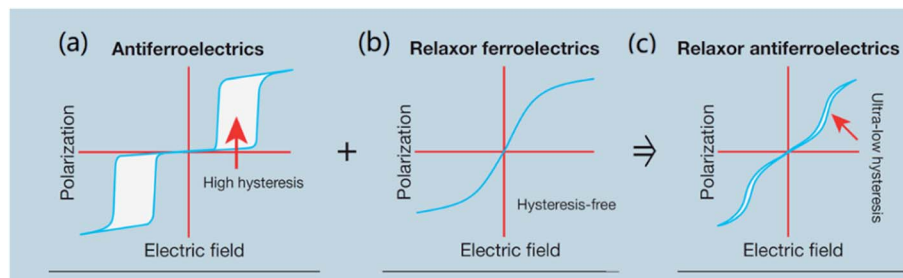


Fig. 42 Schematic strategy to develop AgNbO<sub>3</sub>-based antiferroelectrics for practical applications<sup>159</sup> [reproduced with permission from ref. 159 copyright 2019 by Elsevier].

exist in thinner layers.<sup>170–172</sup> More interestingly,  $E_b$  can be significantly improved through microstructural design, such as introducing core-shell features, with a ferroelectric core and a non-ferroelectric (*e.g.* paraelectric) shell.<sup>173,174</sup>

## 5. AgNbO<sub>3</sub>-based materials for photocatalysis

Semiconductor-based photocatalytic technologies have been extensively investigated since 1972,<sup>20</sup> for the production of renewable hydrogen from water and for the degradation of organic pollutants by photo-generation of electricity and photoelectrochemical reactions.<sup>175–177</sup> AgNbO<sub>3</sub>-based perovskites were reported to exhibit visible-light-driven photocatalysis

activity due to their narrow band gap (2.08–2.93 eV),<sup>81</sup> which has attracted increased research interest over the past decade. The following sections will give a detailed review of photocatalysis associated with AgNbO<sub>3</sub>-based materials.

### 5.1 Fundamental principles of semiconductor photocatalysis

The underlying mechanisms of the photocatalysis process have been extensively discussed in textbooks and in primary literature.<sup>178–180</sup> Here, a brief introduction to the fundamental principles of semiconductor photocatalysis is provided. The principle of photocatalytic degradation of water pollutants is schematically shown in Fig. 44a. Upon light-illumination, with photons of larger energy ( $E = h\nu$ ) than the band gap ( $E_g$ ) of the





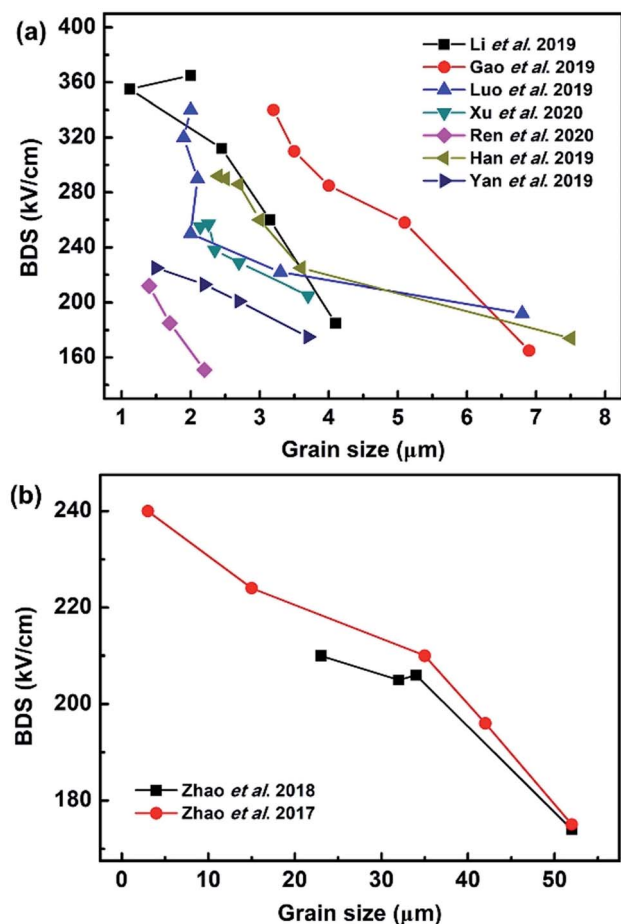


Fig. 43 Electrical breakdown strength ( $E_b$ , BDS) of AgNbO<sub>3</sub>-based ceramics as a function of grain size: (a) small grain sizes and (b) large grain sizes.

semiconductor, electrons will be excited from the valence band to the conduction band, generating an equal number of free electrons and holes. The excited electrons and holes will be separated and will migrate to the surface of the semiconductor. The photo-generated electrons could reduce a dye or react with electron acceptors, such as the O<sub>2</sub> molecules adsorbed on the surfaces of the semiconductor, generating superoxide radical anions ( $\cdot\text{O}_2^-$ ). The photo-generated holes can oxidize organic substances or react with electron donors, such as OH<sup>-</sup>, resulting in hydroxyl radicals ( $\cdot\text{OH}$ ). The hydroxyl radicals can oxidize most dyes into mineral end-products due to their strong tendency to oxidation (standard redox potential +2.8 V).

The processes of generating hydrogen from water under light irradiation can be mainly classified into two reaction categories: (1) photocatalytic; and (2) photoelectrochemical.<sup>180</sup> The photocatalytic reaction process for hydrogen production is very similar to the process of water pollutant degradation, but usually involves a heterogeneous co-catalyst (such as TiO<sub>2</sub>-Pt powder).<sup>181</sup> The photocatalytic reaction of water splitting is schematically shown in Fig. 44b. During light-illumination, the semiconductor photocatalyst is first excited by absorbing photons, generating an equal number of electrons and holes; then, water molecules are reduced

by the photo-generated electrons to form H<sub>2</sub> and are oxidized by the photo-generated holes to form O<sub>2</sub>. To ensure the occurrence of these photocatalytic reactions and achieve water-splitting, the energy band of a semiconductor-based photocatalyst should straddle the redox potentials of water. In particular, the bottom of the conduction band has to be more negative than the redox potential of H<sup>+</sup>/H<sub>2</sub> (0 V versus standard hydrogen electrode (SHE)), while the top of the valence band has to be more positive than the redox potential of O<sub>2</sub>/H<sub>2</sub>O (1.23 V), as shown in Fig. 44c.

The generation of hydrogen by photoelectrochemical reaction is usually achieved in a photoelectrochemical cell (PEC), which consists of a photoelectrode (*i.e.*, a semiconductor-based photocatalyst such as TiO<sub>2</sub>), a counter electrode (typically Pt or graphite) and an electrolyte solution.<sup>182,183</sup> The fundamental photoelectrochemical reaction is schematically shown in Fig. 44d for a TiO<sub>2</sub>-Pt PEC as an example. TiO<sub>2</sub> is a well-known n-type semiconductor, and can work as a photoanode in the PEC, while Pt serves as a counter electrode. Upon light-illumination, photo-excited electrons and holes are generated in the conduction band and valence band of TiO<sub>2</sub>, respectively, and then, the photo-generated electrons are transferred and transported onto the surface of the counter electrode, Pt, through an external circuit with ohmic contacts, where H<sup>+</sup> is reduced to H<sub>2</sub>. Concurrently, the photo-generated holes can favor the oxygen evolution reaction at the surface of the TiO<sub>2</sub> photoanode, generating O<sub>2</sub>. It is noted that if a p-type semiconductor is used as the photocatalyst, it would usually act as a photocathode for hydrogen evolution. When comparing the PEC to powdered photocatalytic processes using a co-catalyst, the powdered co-catalyst system can be considered as a short-circuit version of a PEC since the photocatalytic reactions are spatially separated on the powdered catalyst and occur on a much shorter length scale than that for the PEC. The advantage of the PEC is that it can be more effective in separating photo-generated electrons and holes on the macroscale using different electrodes and in decreasing recombination, but the cost for industrial production of long-life and efficient PECs is high. The powdered catalyst, instead, shows an easier process and holds great potential due to its much lower costs, although its catalytic efficiency is lower than that of a PEC.

## 5.2 Fundamentals of ferroelectric photocatalysis

Currently, great efforts have been devoted to improve the photocatalytic activity of powdered catalysts, mainly including: (1) the search for semiconductors with a narrow band gap (for providing a maximum utilization of solar energy especially in the visible light range);<sup>185-187</sup> (2) microstructure optimization by controlling the particle size, and by establishing a balance between large specific surface areas (providing a high density of available reaction sites) and transport pathways for photo-generated charge carriers (affecting recombination rates and back reaction);<sup>176,188</sup> (3) decoration of the surface with a metallic co-catalyst; (4) introduction of heterogeneous interfaces such as hetero-junctions, to create an internal electric field for enhancing the separation rate of photo-generated charge carriers.<sup>180,189-191</sup>

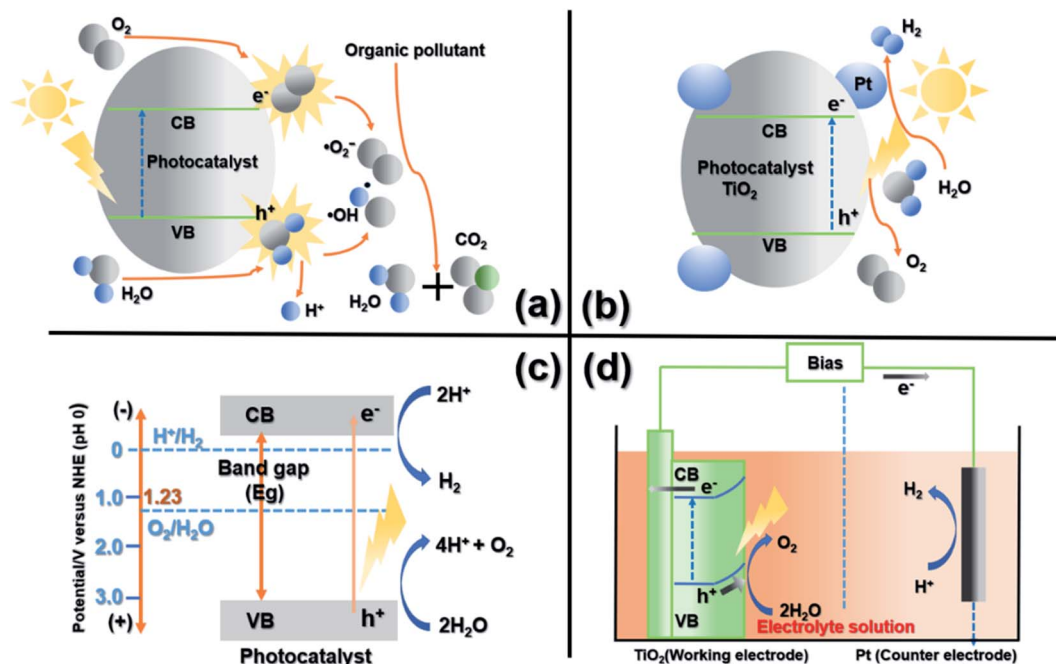


Fig. 44 (a) A schematic of the photocatalytic mechanism in semiconductors for degrading an organic pollutant; (b) a schematic illustration of the fundamental principles of a semiconducting photocatalytic water-splitting process; (c) energy diagrams for photocatalytic water-splitting, and (d) a schematic illustration of the fundamental principles for a semiconducting photoelectrochemical water-splitting process.<sup>184</sup> [Reproduced with permission from ref. 184 copyright 2018 by John Wiley and Sons].

Additionally, the internal electric field induced by the spontaneous polarization ( $P_s$ ) of a ferroelectric has been considered an important factor in photocatalytic performance.<sup>192,193</sup> AgNbO<sub>3</sub> is a weak-ferroelectric/ferrielectric compound possessing a narrow band gap ((2.08–2.93 eV)).<sup>21,81</sup> Ambient ferroelectricity with large  $P_s$  can be obtained in doped compounds, such as ALN and AKN solid-solutions,<sup>69,117</sup> which could represent potential candidates for ferroelectric photocatalysts.

In ferroelectrics, the intrinsic  $P_s$  can generate bound charges on the ferroelectric surfaces/interfaces. These bound charges are not energetically stable and can be compensated in two ways: internal screening and/or external screening,<sup>194</sup> as schematically shown in Fig. 45a. The internal screening can be realized by the flow of free charge carriers within the ferroelectric crystal, whereas external screening is achieved by the adsorption of charged ions and molecules on the surface of the crystal from the surrounding media. Internal screening by the flow of free charge carriers redistributes the charge carriers near the surface, leading to band bending and a space charge region (SCR),<sup>195</sup> as schematically shown in Fig. 45b. If the ferroelectric surface with negative polarity is exposed, electrons would be depleted from the surface, giving rise to a SCR called the depletion layer and to an upward bending of the band. The depletion means that the transfer of photo-generated electrons toward the surface is blocked, but the movement of photo-generated holes toward the surface is promoted. Conversely, if the surface with positive polarity is exposed, electrons would be accumulated for screening the surface-bound positive charges,

resulting in a SCR called the accumulation layer and a downward bending of the band.

The charged surface due to the spontaneous polarization deforms the movements of photo-generated electron–hole pairs, and thus determines the unique photochemical properties of the ferroelectric. The length of the SCR (note:  $L_d$  is the length of the depletion layer and  $L_a$  is that of the accumulation layer) can be written as a function of the surface potential  $V_s$  and the Debye length ( $L_D$ ).<sup>196,197</sup> The voltage  $V_s$  depends not only on  $P_s$ , but also on the conduction band edge and the medium surrounding the surface. The Debye length  $L_D$  depends on the dielectric permittivity of the material and the donor concentration according to the following equation:<sup>184,185</sup>

$$L_D = \left( \frac{\epsilon_0 \epsilon_r k_B T}{e^2 N_D} \right)^{1/2} \quad (7)$$

where  $\epsilon_0$  is the permittivity of free space,  $\epsilon_r$  is the relative permittivity of the material, and  $N_D$  is the donor density. The widths of the space charge layers are then:

$$L_d = \sqrt{\frac{2eV_s}{k_B T}} L_D \quad (8)$$

and

$$L_a = \sqrt{2} \left( 1 - \exp \left[ \frac{eV_s}{2k_B T} \right] \right) L_D \quad (9)$$

for the depletion and accumulation layers, respectively. The term  $e$  is the electronic charge,  $k_B$  is Boltzmann's constant, and  $T$  is the absolute temperature. From the above three equations,

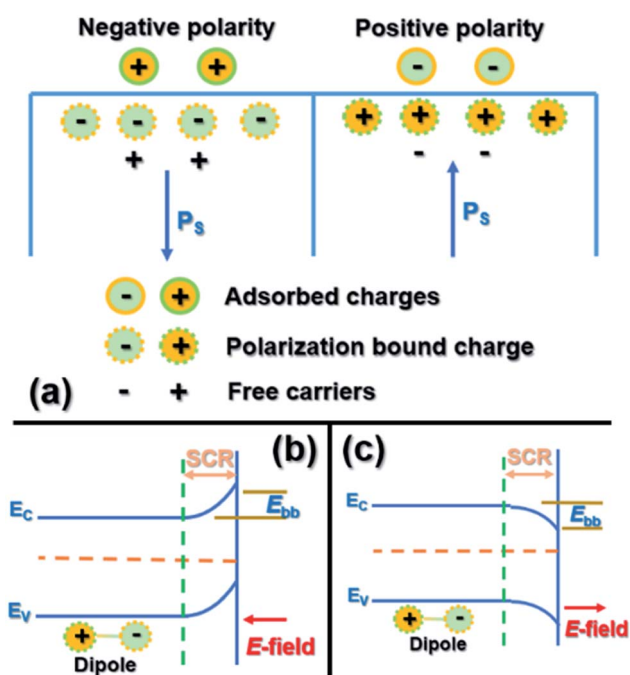


Fig. 45 Schematic diagrams of (a) external/internal screening by adsorbing charges or free carriers/defects in ferroelectric crystals,<sup>194</sup> [reproduced from ref. 194 <https://doi.org/10.1088/0953-8984/17/16/012> with permission from copyright IOP Publishing], and band bending in the exposed surface of ferroelectrics with (b) negative or (c) positive polarity.<sup>195</sup> SCR: space charge region. [Reproduced from ref. 195 <https://doi.org/10.1088/0957-4484/18/18/185702> with permission from copyright IOP Publishing].

it can be deduced that a wider SCR requires a longer  $L_D$ . Thus, a ferroelectric material possessing high dielectric permittivity and low density of free charge carriers could enhance the

separation of photo-generated carriers, benefiting the photochemical properties. If the charged surfaces cannot be fully neutralized by the internal screening mechanism, external screening by adsorption of charged molecules or ions from the environment would occur. The above discussions are based on theoretical aspects. Indeed, to minimize the depolarization field for neutralizing the charges on the surface, the ferroelectric crystal splits into domains separated by domain walls.<sup>198</sup> Thereby, the motion of domain walls could have a significant influence on the separation of charge carriers. Previous studies have suggested that domain walls affect the local surface potential and surface free energy, and may influence the surface chemistry.<sup>199,200</sup> Recent experimental evidence implies that the enhanced photoactivity arises from the inherent  $P_s$  of domains rather than from domain walls.<sup>201</sup> In particular, an experimental investigation conducted on  $\text{PbTiO}_3$  nanoplates exhibiting single domains further supports the contribution of domains as mentioned above.<sup>202</sup> Moreover, the authors found that increasing the thickness along the direction of  $P_s$  can enhance the charge separation ability. The  $P_s$  decreases with decreasing grain size as found in  $\text{PbTiO}_3$  and  $\text{BaTiO}_3$ .<sup>203,204</sup> Thus, for the design of powdered ferroelectric photocatalysts, the particle size is a crucial parameter to be considered. Appropriate control of the particle size would allow for more efficient conduction of the photo-generated charge carriers to the sites where redox reactions take place.

### 5.3 Progress on $\text{AgNbO}_3$ -based photocatalysts

Since Kato *et al.*<sup>21</sup> reported the photocatalytic activity of  $\text{AgNbO}_3$  in 2006, many strategies have been applied to improve its photocatalytic performance. These strategies include surface morphological control, metal coating, design of solid-solutions or composites, and doping, among others. Indeed, these

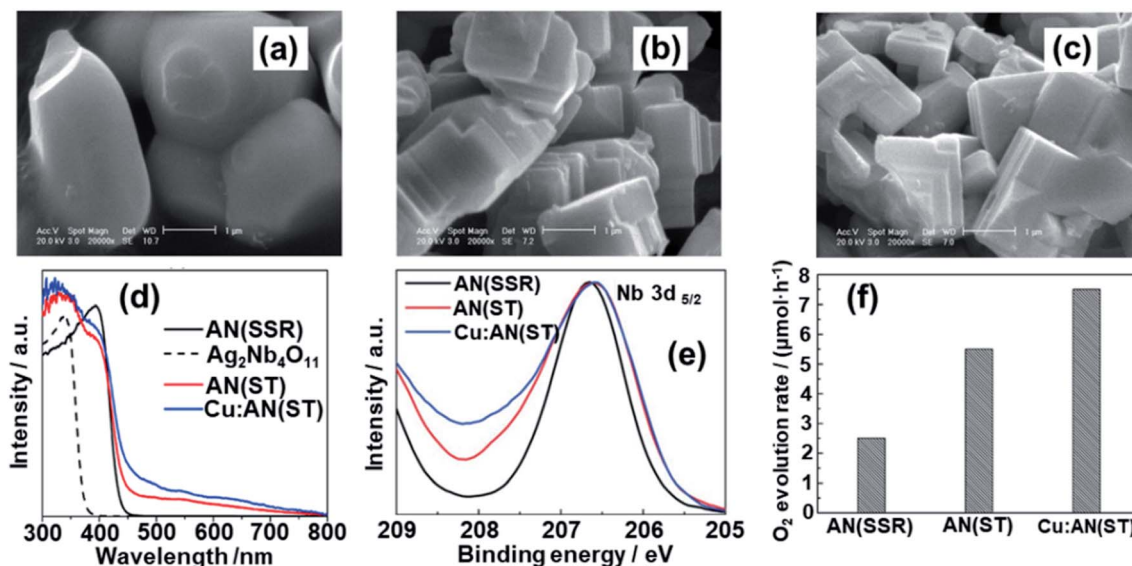


Fig. 46 (a) SEM images of  $\text{AgNbO}_3$  prepared by solid-state reaction, AN (SSR); (b)  $\text{AgNbO}_3$  prepared by a solvothermal method, AN (ST); and (c) Cu-doped  $\text{AgNbO}_3$  prepared by a solvothermal method, Cu : AN (ST); (d) UV-vis diffuse reflection spectra of AN (SSR),  $\text{Ag}_2\text{Nb}_4\text{O}_{11}$ , AN (ST) and Cu : AN (ST); (e) Nb 3d<sub>5/2</sub> XPS line obtained from AN (SSR) (black line), AN (ST) (red line) and Cu : AN (ST) (blue line); (f) photocatalytic  $\text{O}_2$  evolution rates over AN (SSR), AN (ST) and Cu : AN (ST) under visible light irradiation ( $\lambda \geq 420$  nm) reported by Li *et al.*<sup>206</sup>



strategies do not differ from the four strategies for improving the photocatalytic performance of a given semiconductor, as mentioned in Section 5.2.

**5.3.1 Surface morphology control.** Arney *et al.*<sup>205</sup> investigated flux synthetic routes to obtain smaller particle sizes of  $\text{AgNbO}_3$ , and compared the resultant surface areas and photocatalytic activities with those obtained from traditional solid-state methods. The results showed that controlling the surface morphology *via* flux synthesis methods can significantly enhance the visible-light photocatalytic rates of  $\text{H}_2$  formation. The enhanced photocatalytic performance *via* morphology control was further evidenced by Li *et al.*,<sup>206</sup> who synthesized polyhedron-shaped  $\text{AgNbO}_3$  and Cu-doped  $\text{AgNbO}_3$  particles with a “nano-ladder” surface microstructure *via* a solvothermal (ST) method (see Fig. 46b and c), achieving enhanced visible-light absorption capability and photocatalytic rates of  $\text{O}_2$  evolution. The structural analysis revealed that the powdered photocatalyst consists of  $\text{AgNbO}_3/\text{Ag}_2\text{Nb}_4\text{O}_{11}$  composites. Furthermore, XPS results implied the existence of  $\text{Nb}^{4+}$ , suggesting the presence of defects. This defect structure might be intimately related to the formation of the “nano-ladder” surface morphology during the solvothermal reaction. Upon  $\text{Cu}^{2+}$ -doping, a highly asymmetric peak in the XPS spectrum was observed (see Fig. 46e), indicating a higher concentration of defect structures/clusters associated with  $\text{Nb}^{4+}$  leading to higher visible-light photocatalytic activity. A similar behavior was also observed in  $\text{La}^{3+}$ -doped  $\text{AgNbO}_3$ .<sup>207</sup> Wang *et al.*<sup>208</sup> synthesized  $\text{AgNbO}_3/\text{Ag}_2\text{Nb}_4\text{O}_{11}$  and  $\text{AgNbO}_3/\text{Ag}_2\text{Nb}_4\text{O}_{11}/\text{Nb}_2\text{O}_5$  composites *via* a solvothermal process and further investigated the temperature impact on the photochemical reaction. These composite particles showed similar surface morphology to those reported by Li *et al.*<sup>206</sup> However, the UV-vis absorption spectra revealed an enhanced absorption tail, consisting of two absorption peaks (one near 500 nm and the other near 600 nm), which cover the visible-light range. This suggests that there are two different mechanisms to generate extra absorption in the visible-light range.

Recently, Cao *et al.*<sup>209</sup> synthesized one-dimensional  $\text{AgNbO}_3$  nanostructures *via* a topochemical transition process using  $\text{K}_2\text{Nb}_2\text{O}_6 \cdot n\text{H}_2\text{O}$  filiform crystals as precursors (see Fig. 47a and

b). These nanostructured samples also showed a broadened absorption peak in the visible-light range (see Fig. 47c). As mentioned in Section 2.5, in the UV-vis diffuse reflection spectrum of pure  $\text{AgNbO}_3$ , Zhou *et al.* found an absorption peak in the same wavelength range, which was proposed to originate from the SPR effect (see Fig. 17).<sup>83</sup> This effect could be responsible for the enhanced photocatalytic activity under visible-light irradiation.

Lu *et al.*<sup>210</sup> synthesized one-dimensional  $\text{AgNbO}_3$  using Ag nanowires as the raw material template. In contrast to the results reported by Cao *et al.*, the final product was a one-dimensional nanostructured composite of  $\text{Ag}_2\text{Nb}_4\text{O}_{11}$  and  $\text{AgNbO}_3$ , which showed an enhanced activity to degradation of 2,4-dichlorophene and rhodamine B (RhB). Most recently, Gao *et al.*<sup>211</sup> synthesized submicron cubes of  $\text{AgNbO}_3$  *via* hydrothermal treatment (HT) of  $\text{Ag}_2\text{O}-\text{Nb}_2\text{O}_5-\text{NH}_4\text{HF}_2$  aqueous suspensions, which exhibited excellent photocatalytic activity under visible-light irradiation ( $\lambda \geq 400$  nm) (see Fig. 48g). The microstructure analysis indicated that a (001) facet with high surface energy was exposed (see Fig. 48c and d). Further photoelectrochemical and photocatalytic performance analysis suggested that the cubes with an exposed (001) facet exhibit higher charge carrier separation efficiency with the greater redox ability of photogenerated electrons and holes being responsible for the superior photocatalytic activity.

**5.3.2 Solid-solution semiconductors with modulated band structure.** It is expected that solid-solutions of  $\text{AgNbO}_3$  with another perovskite semiconductor could exhibit improved photocatalytic properties compared to each of the end members as a result of the electronic structure that would arise. Based on this idea, two solid-solution systems have been explored so far, namely  $\text{AgNbO}_3-\text{SrTiO}_3$  and  $\text{AgNbO}_3-\text{NaNbO}_3$ . In the  $\text{AgNbO}_3-\text{SrTiO}_3$  solid-solution, the hybridization between the Ag 4d and O 2p orbitals and between the Nb 4d and Ti 3d orbitals plays a crucial role in tuning the band structure (see Fig. 49a reported by Wang *et al.*) and in tailoring the photocatalytic properties.<sup>212</sup> As a result of the competition between the reduction/oxidation abilities and the absorption activity, the enhanced visible-light activities for the evolution of  $\text{O}_2$  and the decomposition of organic compounds can be realized in specific compositions. In

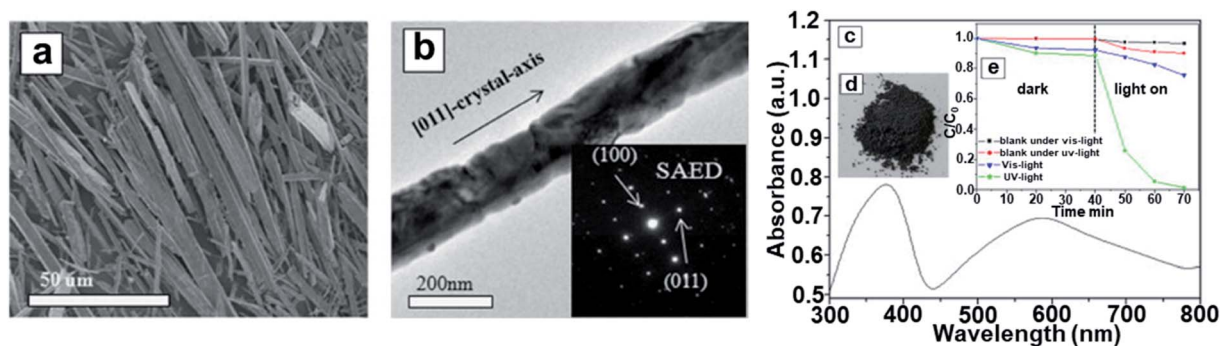


Fig. 47 (a) FE-SEM and (b) TEM images of  $\text{AgNbO}_3$ ; (c) UV-visible diffuse reflectance spectra, (d) photographic image of  $\text{AgNbO}_3$  powder and (e) photocatalytic degradation of methylene blue (MB) by  $\text{AgNbO}_3$  nanostructures reported by Cao *et al.*<sup>209</sup> [Reproduced with permission from ref. 209 copyright 2014 by Elsevier].



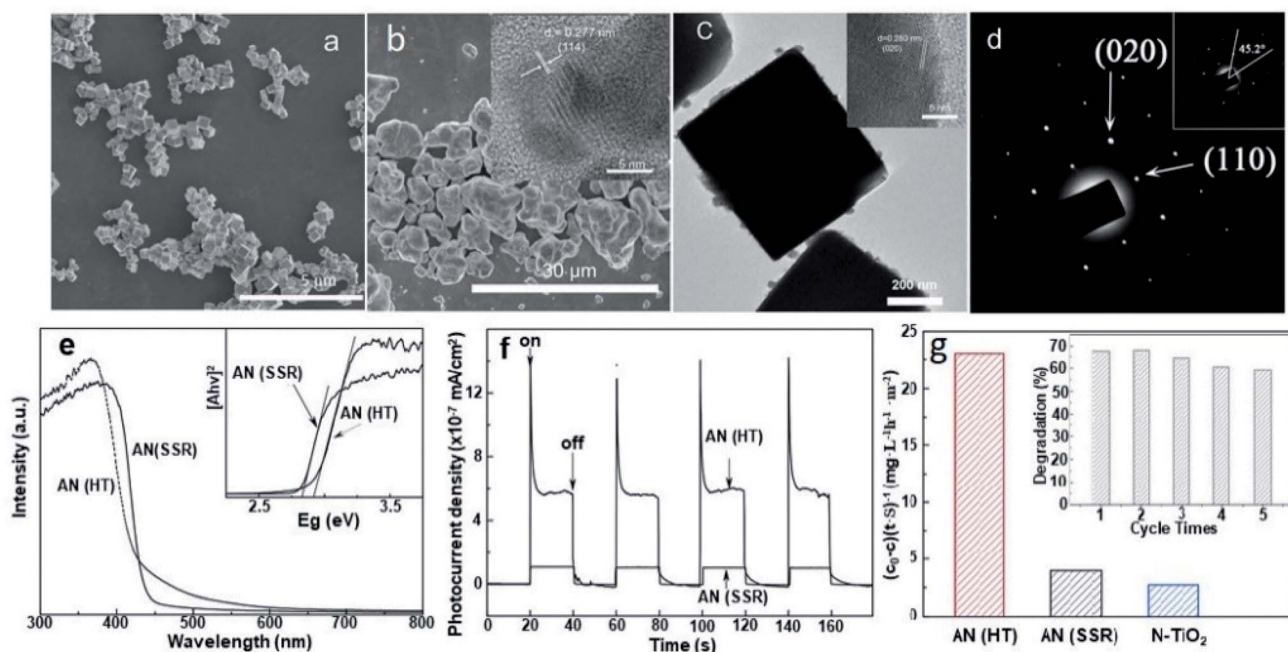


Fig. 48 Field-emission scanning electron microscopy images of  $\text{AgNbO}_3$  (AN) prepared by (a) hydrothermal (HT) and (b) solid state (SSR) methods; (c) transmission electron microscopy image and (d) selected-area electron diffraction pattern of AN (HT); the insets in (b) and (c) show high-resolution transmission electron microscopy images of AN (SSR) and AN (HT), respectively; the inset in (d) shows the angle between the (020) and (110) facets in the AN (HT) cube; (e) UV-vis diffuse reflectance spectra of AN (HT) and AN (SSR); (f) transient photocurrent of AN (HT) and AN (SSR) irradiated with visible light ( $\lambda \geq 400$  nm); (g) specific photocatalytic activity of the catalysts. The inset in (g) shows repeated photocatalytic degradation of tetracycline in the presence of AN (HT) and nitrogen-doped  $\text{TiO}_2$ , reported by Gao *et al.*<sup>211</sup> [Reproduced with permission from ref. 211 copyright 2019 by Elsevier].

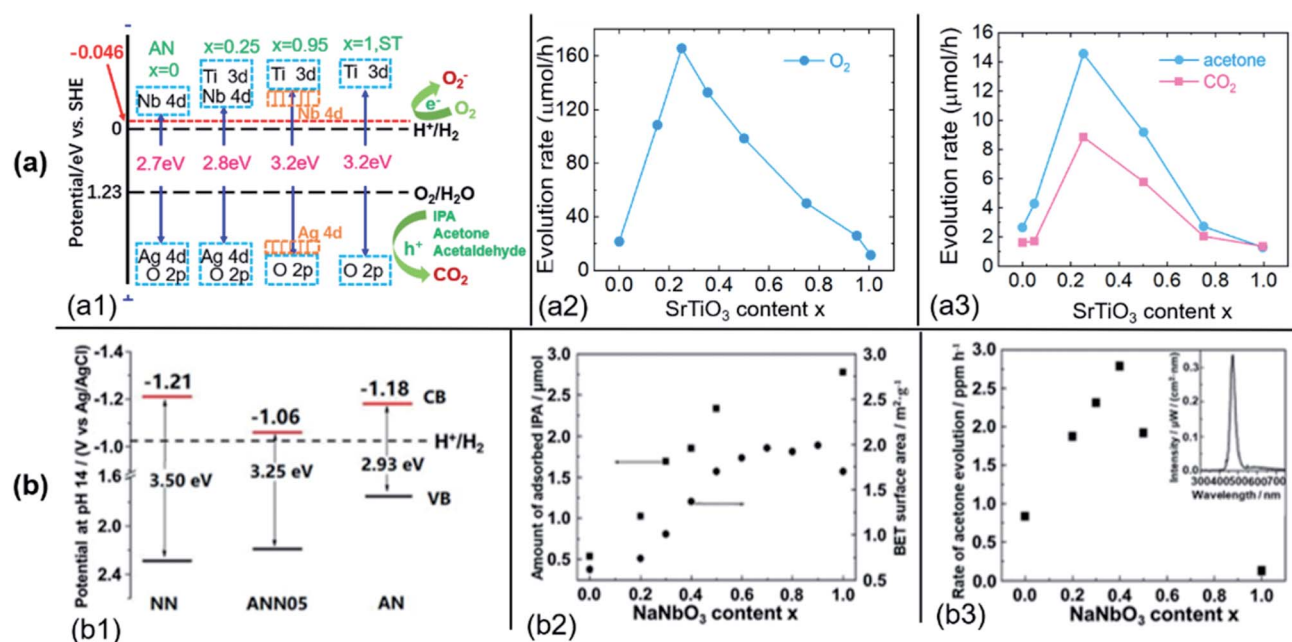


Fig. 49 (a) Schematic band structure (a1) and experimental photocatalytic performance (a2 and a3) of  $(\text{AgNbO}_3)_{1-x}(\text{SrTiO}_3)_x$  (ANST) solid-solution.<sup>212</sup> [Reproduced with permission from ref. 212 copyright 2009 American Chemical Society]. The hybridization of the Ag 4d with O 2p lifts the top of the valence band. In the sample  $x = 0.25$ , both the conduction band and valence band are continuous. In the case of  $x = 0.95$ , discontinuous inter-bands (or doping levels) are formed by the small number of dopants. (b) Schematic band structure (b1)<sup>213</sup> [reproduced with permission from ref. 213 copyright 2010 by Elsevier] and experimental photocatalytic performance (b2 and b3) [reproduced with permission from ref. 215 copyright 2007 by Elsevier] of  $(\text{AgNbO}_3)_{1-x}(\text{NaNbO}_3)_x$  (ANN) solid-solution.<sup>213,215</sup>

the  $\text{AgNbO}_3$ - $\text{NaNbO}_3$  solid-solution, an increased content of  $\text{NaNbO}_3$  decreases the hybridization ratio of Ag 4d and O 2p and the top of the valence band shifts upwards, leading to a blue shift in the optical absorption edge and to an enhancement of oxidizing power (see Fig. 49b).<sup>213</sup> In this solid-solution, Li *et al.*<sup>214</sup> reported an enhancement of visible-light photocatalytic activity for the decomposition of isopropanol (IPA) into acetone at an intermediate composition. This was considered to be related to the competition between oxidizing power, enhanced surface area and visible-light absorption. In particular, the authors fabricated  $\text{AgNbO}_3$ - $\text{NaNbO}_3$  solid-solution films on fluorine-doped tin oxide (FTO) conducting glass substrates and investigated their band structure and photoelectrochemical behavior. The schematic band structure is shown in Fig. 49b, which highlights that the bottom of the conduction band potential in the intermediate composition is significantly more positive than that of the two end compounds.<sup>213</sup>

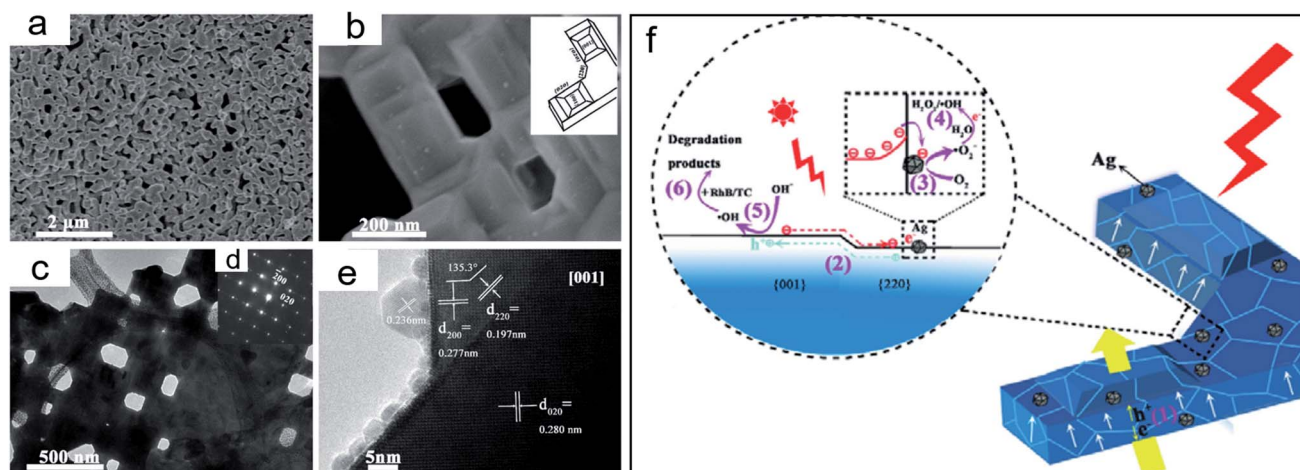
#### 5.4 Composites possessing hetero-junction interfaces

The presence of interfaces in composite materials may bring an enhancement of photocatalytic activity. To date, several  $\text{AgNbO}_3$ -based composites that show an enhancement of photocatalytic performance compared to the pristine bulk have been reported. Table 6 summarizes the photocatalytic properties of various  $\text{AgNbO}_3$ -based composites. In  $\text{AgBr}/\text{AgNbO}_3$  composites, Wang *et al.*<sup>216</sup> suggested that the enhancement of photocatalytic activity can be attributed to the  $\text{AgBr}/\text{AgNbO}_3$  heterostructure (formed at the interface), which creates an internal electric field, modulates the band structure, and favors an efficient separation of electrons and holes. Similar mechanisms have contributed to the enhanced photocatalytic activity of  $\text{Ag}_2\text{O}/\text{AgNbO}_3$  p-n junctions as reported by Yang *et al.*<sup>217</sup> Recently, Yu *et al.*<sup>218</sup> synthesized an  $\text{Ag}/\text{AgNbO}_3$  plasmonic photocatalyst *via* a facile combustion method, obtaining a special porous nano-structure where high energy {001} and {220} facets were exposed (see Fig. 50). The authors believed that the enhanced photocatalytic efficiency is related to the

**Table 6** Photocatalytic performance of  $\text{AgNbO}_3$ -based composites for degrading organic pollutants<sup>a</sup>

Catalysts	Organics	UV/vis	Degrad./time (h)	<i>k</i>	Ref.
$\text{AgNbO}_3/\text{AgSbO}_3$	RhB (2.5 mg L <sup>-1</sup> )	Vis	—	0.02667	219
$\text{Ag@AgCl}/\text{AgNbO}_3$	MB (10 mg L <sup>-1</sup> )	Vis	~65%/2 h	0.007	220
$\text{NiO}/\text{AgNbO}_3$	MB (10 mg L <sup>-1</sup> )	UV	~84%/3 h	—	221
$\text{AgBr}/\text{AgNbO}_3$	MB (10 mg L <sup>-1</sup> )	Vis	74.62%/3 h	0.0083	216
$\text{Ag}_2\text{O}/\text{AgNbO}_3$	RhB (10 mg L <sup>-1</sup> )	Vis	95.4%/1.5 h	—	217
$\text{Ag}/\text{AgNbO}_3$	RhB (5 mg L <sup>-1</sup> )	—	—	0.0447	218
	TC (10 mg L <sup>-1</sup> )	—	70.2%/1 h	0.0293	

<sup>a</sup> RhB: rhodamine B; TC: tetracycline; MB: methylene blue.



**Fig. 50** (a) SEM image of  $\text{Ag}/\text{AgNbO}_3$  porous nano-structure, (b) enlargement of (a) with the structure model (the inset illustration) of the as-prepared particle; (c) typical HRTEM image of  $\text{Ag}/\text{AgNbO}_3$  porous nano-structure and the corresponding SAED pattern in (d); (e) HRTEM image of the prepared  $\text{Ag}/\text{AgNbO}_3$ . (f) Schematic drawing illustrating the possible photocatalytic mechanism of the  $\text{Ag}/\text{AgNbO}_3$  plasmonic photocatalysts: (1) spatial separation of photo-generated carriers along the direction of potentials originating from ferroelectric polarization; (2) orientation transfer of photo-generated carriers onto {001} and {220} facets; (3) photo-generated electrons transfer from the {220} facets to Ag nano-particles to reduce  $\text{O}_2$ ; (4) formation of free  $\cdot\text{OH}$  via the reaction of  $\cdot\text{O}_2^-$  with  $\text{H}_2\text{O}$ ; (5) formation of free  $\cdot\text{OH}$  via the reaction of photo-generated holes with  $\text{OH}^-$  in the solution; (6) rapid oxidation of RhB or (tetracycline)TC by free  $\cdot\text{OH}$ .<sup>218</sup> [Reproduced with permission from ref. 218 copyright 2019 by Elsevier].





synergy between surface plasmonic resonance (SPR), ferroelectric polarization, specific exposed crystal-facets and interfacial crystallinity.

## 6. Summary and outlook

In this review, progress in research on AgNbO<sub>3</sub> and AgNbO<sub>3</sub>-based perovskites since their discovery in 1958 has been summarized. The review covers various aspects, from fundamental science to emerging applications, and allows us to draw some important conclusions in the following areas.

### 6.1 Structures, polymorphic phases, and electric field-induced transitions

AgNbO<sub>3</sub> is polymorphic and shows complex phase behavior, with intricate transformations induced by temperature variations and the application of an electric field. The M<sub>1</sub> phase, stable up to *ca.* 70 °C, consists of an antiferroelectric-like matrix, in which ferroelectric clusters are locally present, which can successfully explain the non-linear polarization under weak and strong electric fields. Both structure and analysis of properties suggest that the dielectric anomaly denoted as (*T<sub>f</sub>*) could originate from a phase transition associated with inversion symmetry. Electrical loading under high electric fields produces double polarization hysteresis loops, but the structure induced under high electric fields is still experimentally unknown. Future studies should be aimed at revealing the nature of the electric field-induced structure, by *in situ* diffraction techniques which have still not been carried out in this important material. Above the temperature *T<sub>f</sub>* (~170 °C in pure AgNbO<sub>3</sub>), the microstructures of both M<sub>2</sub> (stable in the range of ~70–270 °C) and M<sub>3</sub> (stable in the range of ~270–350 °C) phases still show the presence of polar clusters, but field-induced transformations are significantly suppressed and are difficult to drive, even at high electric fields. The frequency-independent diffuse dielectric anomaly near 270 °C is intimately related to the local disorder–order behavior of off-center cation displacements (especially Nb<sup>5+</sup> ions) in the sub-lattice of its average AFE matrix. This emphasizes the importance of local structural heterogeneity at the sub-lattice scale to improve the dielectric response of a given material, especially at high frequency, and will guide the property optimization for specific applications, such as microwave capacitors.

### 6.2 Effects of compositional modification

Chemical modifications by suitable elements can be guided by the perovskite tolerance factor and may result in the suppression of the local displacive behavior, as well as in an increased local structural disorder down to very low temperatures.

Isovalent substitutions on the A-site, using Li, restrains antiferroelectricity and leads to the promotion of a ferroelectric-type order with rhombohedral structure possessing high piezoelectric, pyroelectric and fatigue performances. Gradual substitution of Ag<sup>+</sup> with isovalent Na<sup>+</sup> ions with a smaller radius than Ag<sup>+</sup> allows for the formation of a perovskite solid solution and induces the crossover from double polarization hysteresis

to ferroelectric-like loops. The presence of Na restricts the displacement of Nb<sup>5+</sup> ions and suppresses the diffuse dielectric dispersion of AgNbO<sub>3</sub>. Additionally, the replacement of Ag<sup>+</sup> by isovalent K<sup>+</sup> has the effect of reducing antiferrodistortive instabilities and promotes ferroelectricity, characterized by orthorhombic symmetry and an emerging potential for explosive energy conversion due to the possibility of driving pressure-induced FE → AFE transitions.

A possible strategy to develop AgNbO<sub>3</sub>-based antiferroelectric solid-solutions is to substitute Ag<sup>+</sup> with cations of different radius and valence, including bi- and tri-valent elements, such as Ca<sup>2+</sup>, Bi<sup>3+</sup> and various rare-earth cations, which can provide chemical pressure to suppress ferroelectricity and increase the energy barrier between AFE and FE phases. Besides, the doping also creates defects and enhances local structural heterogeneity that can have negative impacts on the field-induced polar ordering. Similar effects can be obtained by the isovalent substitution of Nb<sup>5+</sup> by Ta<sup>5+</sup> on the B-site producing an increase in the characteristic fields *E<sub>B</sub>* and *E<sub>F</sub>*, and a decrease of the remnant polarization, with *P–E* loops tending to a linear shape with increasing Ta content, evidencing the stabilization of an antiferroelectric state.

### 6.3 Applications in dielectric energy storage

AgNbO<sub>3</sub>-based materials are promising lead-free candidates for energy storage capacitors, due to their large energy density values. These range from 2.1 J cm<sup>−3</sup> in AgNbO<sub>3</sub> to 6.3 J cm<sup>−3</sup> in Ag(Nb<sub>0.45</sub>Ta<sub>0.55</sub>)O<sub>3</sub> ceramics obtained at room temperature, attributed to the enhanced stability of antiferroelectricity and improved breakdown strength. Two methods can be used to further enhance *E<sub>b</sub>* and are based on either intrinsic or extrinsic effects. One method consists of modulating the band structure and increasing the band gap to prevent the excitation of electrons from the valence band to the conduction band (intrinsic). The other method is based on reducing the grain size to increase the resistivity by increasing the amount of grain boundaries (extrinsic).

One of the main issues that should be overcome in future research is to increase the energy storage efficiency, which is often lower than 70%. This can be achieved by reducing the hysteresis area of the polarization–electric field loop. Previous investigations on relaxor-ferroelectrics such as Pb(Mg<sub>1/3</sub>Nb<sub>2/3</sub>)O<sub>3</sub> may provide guidance for designing the AgNbO<sub>3</sub>-based materials displaying field-induced reversible polar ordering with low hysteresis (*i.e.* ergodic relaxor-antiferroelectrics). Systematic studies combining experimental techniques and modelling should be able to provide improved compositional design criteria to obtain increased energy storage efficiency.

A better synergy between research groups in academia and companies engaged in the energy storage capacitor sector should be promoted to overcome additional technical issues that are currently preventing AgNbO<sub>3</sub>-based devices entering the energy storage market. These issues include the difficulties related to the processing of AgNbO<sub>3</sub> in oxygen-rich environments with accurate temperature control and specific heat treatment schedules. Metallic Ag can easily form near the



AgNbO<sub>3</sub> surface when the oxygen content in the processing environment is not sufficient, inducing an n-type semiconductor state. Additionally, the cost of the Ag precursors and the compatibility with electrode materials are also important factors to consider in future development.

#### 6.4 Applications in photocatalysis and photovoltaics

As a functional semiconductor, AgNbO<sub>3</sub> is emerging not only as a potential photocatalyst for splitting water and degrading organic pollutants, but also as a photovoltaic material for solar-cells due to its narrow  $E_g$ , which can extend the absorption edge of the solar spectrum to the visible-light range, providing enhanced photocatalytic and photovoltaic performances. Theoretical calculations also suggest a narrower band gap ( $E_g \approx 1.83$  eV) compared to the reported experimental values ( $E_g \approx 2.08$ – $2.93$  eV). This is attributed to the existence of unbonded Ag atoms and the Ag–O/Nb–O covalent bonds formed in the AgNbO<sub>3</sub> lattice. The presence of metallic Ag is believed to be responsible for the surface plasmon resonance effect and for the observed extra visible-light absorption event and thus for ultimately enhancing the photocatalytic and photovoltaic performances. Furthermore, the visible-light photocatalytic activity can be tuned by the external voltage bias, which can modify the width of surface space charge regions (SCRs) in AgNbO<sub>3</sub>, allowing for the regulation of photocatalytic activity.

To further improve the photocatalytic performance for specific applications, chemically-modified AgNbO<sub>3</sub>-based materials have been explored since 2006 according to the following three materials design strategies: (1) control of the nanostructured surface morphology by wet chemical methods to provide more reaction-sites; (2) change of the surface chemical potential by modulating the band structure to enhance redox reactions; and (3) introduction of heterogeneous interfaces to create an internal electric-field to promote the separation of photo-generated electron-hole pairs.

To date, rare studies have been carried out integrating two or three of these approaches to further improve the photocatalytic performance. In particular, AgNbO<sub>3</sub>-based ferroelectric materials would attract interest for photocatalytic application due to the large  $P_s$  coupled with the narrow  $E_g$  and SPR response. Besides AgNbO<sub>3</sub>-based ferroelectric photocatalysts, AgNbO<sub>3</sub>-based piezo- and pyro-electric semiconducting nanomaterials could also emerge as potential candidates for converting mechanical and thermal energy into chemical energy. Additionally, research focused on ferroelectric photocatalysis of AgNbO<sub>3</sub> is still rare. However, previous studies reported intriguing photovoltaic properties found in electrically poled and mechanically polished AgNbO<sub>3</sub> ceramics, which were proposed to originate from intrinsic FE polarization and strain gradient-induced flexoelectricity, respectively. At the same time, electrical poling and/or mechanical polishing can also cause the precipitation of metallic Ag from the lattice of AgNbO<sub>3</sub> and change the surface electronic structure, such as the width of SCRs, influencing the photovoltaic properties. Therefore, more fundamental studies should be carried out in future to further understand the photovoltaic mechanism.

## Conflicts of interest

There are no conflicts to declare.

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