

弛豫铁电陶瓷电致应变特性研究进展

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摘 要: 压电致动器具有体积小、驱动力大、位移精度高以及响应速度快等优点, 被认为是一种非常有前途的微位移驱动器。弛豫铁电陶瓷凭借出色的电致应变特性, 已成为制备压电致动器的理想材料。作为压电致动器的核心元件, 弛豫铁电陶瓷电场诱导的应变特性会对致动器的行程、精度和效率产生显著影响, 直接决定了器件整体性能。针对弛豫铁电陶瓷的电致应变特性, 文章首先重点介绍了弛豫铁电陶瓷材料电致应变的起源, 包括电致伸缩效应、逆压电效应、畴壁移动、场致相变效应和缺陷工程; 其次, 综述了铅基弛豫铁电陶瓷和无铅弛豫铁电陶瓷电致应变特性的最新研究进展; 最后, 分析了目前铅基弛豫铁电陶瓷和无铅弛豫铁电陶瓷电致应变特性各自存在的问题和未来所面临的挑战。

关键词: 压电致动器; 弛豫铁电陶瓷; 电致应变起源; 应变特性

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压电材料是一种能够实现机械能与电能相互转换的重要功能材料, 这种机械能-电能相互转换的现象被定义为压电效应。压电效应按机电转换顺序的不同分为正、逆压电效应。所谓的正压电效应是指外界机械应力作用于压电材料使其表面产生电荷的效应, 并且外界施加的机械应力越大其表面产生的电荷会越多, 两者基本成正比关系, 该现象由皮埃尔·居里(Pierre Curie)和雅克·居里(Jacques Curie)在 1880 年发现。而逆压电效应是指对压电材料反过来施加电场时, 其会发生形变的效应, 并且施加的电场越大其产生的形变会越显著, 该现象由加布里埃尔·李普曼(Gabriel Lippmann)在 1881 年发现^[1,2]。利用压电效应, 压电材料目前在传感器、致动器、能量收集器、成像设备等方面有着广泛的应用。

作为压电材料最重要的应用领域之一, 压电致动器被认为是一种非常有前景的微位移致动器, 其利用压电材料的逆压电效应, 通过电场控制压电材料的形变, 进而可以实现对外动作。与其他致动器相比, 压电致动器具有许多优点, 如易于小型化、

结构简单灵活、分辨率高、响应快、功耗小等, 目前已经在微电子、精密加工、精密测量、光导纤维的校准、航天航空及生物工程等领域具有广泛的应用^[3]。近年来, 压电器件发展非常迅猛, 预计在 2026 年压电器件的销售额将达到 354 亿美元, 在压电器件市场中压电致动器占有主要份额并且占比在逐年提高^[4], 因此, 其受到越来越多科研工作者的重视。

压电材料是制备压电致动器的核心部件之一, 其电场诱导的应变特性是一个关键参数, 它在很大程度上决定了压电致动器的行程、精度和效率。作为一种特殊的压电材料, 弛豫铁电陶瓷内部存在纳米尺度的极性微区(PNRs), 这些 PNRs 与传统正常铁电体的长程有序铁电畴结构不同, 其取向会随温度、电场等外场条件发生可逆变化。这种独特的结构使其介电、铁电、应变等性能呈现出弛豫特征, 如弥散相变、偏离居里-外斯(Curie-Weiss)定律、介电性能频率依赖性、苗条的电滞回线、相变性质等^[5-7]。与正常铁电陶瓷相比, 弛豫铁电陶瓷由于存在 PNRs 结构会呈现出更加优异的电致应变

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行为, 其非常适合在压电致动器上使用, 是目前制备压电致动器选用最多的一种压电材料^[8]。弛豫铁电陶瓷的电致应变特性, 如应变响应、应变滞后以及应变温度稳定性, 与压电致动器的输出行程和位移精度紧密相关, 对压电致动器件的整体设计与性能优化具有至关重要的作用。因此, 开发具有优异电致应变特性的弛豫铁电陶瓷材料, 已成为突破高精尖压电致动器制备技术瓶颈的关键方向, 持续吸引着科研工作者的深入探索。本文将重点介绍弛豫铁电陶瓷应变响应的起源, 以及近年来在应变特性方面的最新研究成果和面临的主要问题。

1 弛豫铁电陶瓷电致应变起源

铁电陶瓷的电致应变是指当外电场作用于铁电陶瓷时, 由于内部结构和极化状态的变化导致铁电陶瓷会发生宏观变形, 并且变形程度与外电场强度有关。弛豫铁电陶瓷电致应变的贡献包括本征贡献和非本征贡献, 本征贡献源于电致伸缩效应和逆压电效应, 而非本征贡献则是由畴壁移动、场致相变效应和缺陷所产生^[9]。

1.1 电致伸缩效应

电致伸缩效应通常在中心对称和非中心对称的电介质材料中发生, 即所有电介质材料均存在的本征效应。与压电效应相比, 铁电陶瓷由电致伸缩效应引起的应变响应通常都比较小, 因此, 早期对铁电陶瓷中该效应的研究文献较少。直到 20 世纪 80

年代, 由于发现弛豫铁电陶瓷和单晶具有非常高的电致伸缩应变, 如铌镁酸铅-钛酸铅(PMN-PT)陶瓷和单晶, 科研人员才开始重点关注该效应, 并对其进行了系统性的研究。电致伸缩效应是铁电陶瓷受到外加电场作用时, 其晶格中的离子偏离其原来的平衡位置, 导致晶格参数的改变所产生^[10], 因此, 其本质是在电场作用下电介质材料内部正负离子重心发生相对位移, 产生感应极化, 导致晶格结构的畸变, 从而引起铁电材料变形, 并且应变的方向与外电场的方向无关。电致伸缩应变(S)与电场强度(E)的二次方成正比, 可以通过下列方程进行描述^[11]:

$$S_{33} = Q_{33}P^2 \quad (1)$$

式中: Q_{33} 为电致伸缩系数; P 为极化强度。

Q_{33} 是衡量电致伸缩效应的重要指标之一, 提升弛豫铁电陶瓷应变响应的有效途径之一是提高 Q_{33} , 由电致伸缩效应所产生的应变通常其滞后都非常小^[12]。

Li 等^[13]通过建立离子晶体刚性模型来解释电致伸缩效应, 如图 1 所示。图 1 给出了正负离子在不同势能条件下晶格参数的变化。起初正负离子都处于平衡位置。当离子对的伸长势能(U_E)小于压缩势能(U_C)时, 位移是由晶格的伸长所引起; 当 U_E 大于 U_C 时, 位移是由晶格的压缩所引起; 当 U_C 等于 U_E 时, 晶格参数不变, 没有位移产生。通常 U_C 要远高于 U_E , 因此正负离子之间的相对位移更容易通过晶格的伸长所产生。

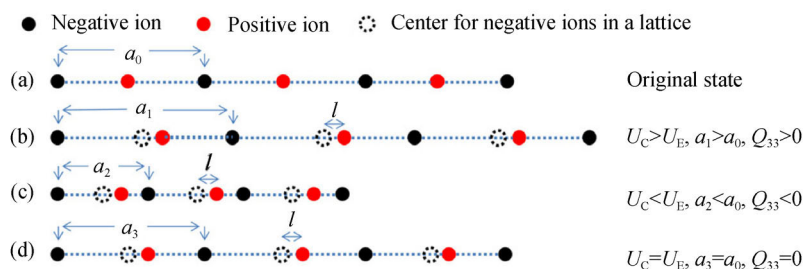


图 1 用于解释电致伸缩效应的离子晶体刚性模型^[13]

Fig. 1 Rigid model of ionic crystal to illustrate the electrostrictive effect^[13]

为了与逆压电效应进行区别, Uchino 等^[14]用一维刚性离子-弹簧模型对晶体中电致伸缩的产生进行了解释, 如图 2a 所示。图 2 使用弹簧来表示晶体中由静电库伦能和量子力学排斥能共同引起的等价结合力, 对于中心对称晶体, 连接离子的是同种弹簧, 称之为软弹簧。理想情况下, 此类弹簧的伸长与压缩幅度可以相互抵消, 不产生应变, 但在实际情况下, 该类弹簧包含非谐和项进而导致其伸长易

于压缩, 因此, 产生了一个较小的应变。

1.2 逆压电效应

逆压电效应是铁电陶瓷在外电场作用下内部铁电畴沿电场方向取向而引发晶格结构变化所产生, 其本质是在电场作用下铁电陶瓷内部的分子发生偶极矩的取向, 从而引起铁电陶瓷变形, 其应变的方向与外电场的方向有关, 具有明显的方向性^[15]。由逆压电效应所产生的应变(压电应变)较电致伸缩效

应的更大^[16]。由逆压电效应所产生的应变可以通过以下方程来描述^[17]:

$$S_{33} = d_{33}^* \cdot E \quad (2)$$

式中: d_{33}^* 为有效压电系数。

由式(2)可知, d_{33}^* 是衡量逆压电效应的重要指标之一, 其可以通过 S 与 E 的线性拟合来获得。

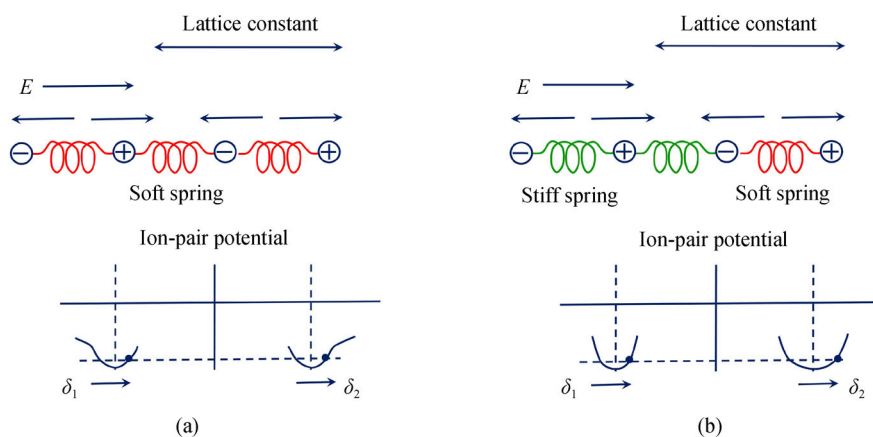


图 2 由(a)电致伸缩效应和(b)逆压电效应产生电致应变的微观模型^[14]

Fig. 2 Microscopic model of electrostrain induced by (a) electrostrictive effect and (b) converse piezoelectric effect^[14]

Uchino 等^[14]也用一维刚性离子-弹簧模型对晶体中逆压电效应的产生进行了解释, 如图 2b 所示。在非中心对称晶体中, 连接离子的是软、硬 2 种弹簧, 并且这 2 种弹簧产生的位移大小是有差异的, 两者对逆压电效应的产生起着关键作用。当施加电场时, 正负离子朝着不同的方向移动, 导致离子的间距发生改变, 此时与硬弹簧相比软弹簧会产生一个更大的形变, 从而产生一个较大的净应变。

1.3 畴壁移动

畴壁移动是指铁电畴边界在电场作用下发生位移或翻转, 从而改变铁电材料整体极化方向的过程。当外加电场很小时, 铁电畴不会翻转, 此时铁电陶瓷的应变主要由电致伸缩效应所贡献。当外加电场超过一定值时, 这些随机取向的铁电畴会克服阻力发生翻转取向与电场方向变得一致, 畴壁的移动会引发晶格畸变, 因此, 畴壁移动贡献了非本征应变。当外加电场移除时, 一部分重新定向的铁电畴会返回到原来的状态, 而另一部分铁电畴状态保持不变, 所以铁电陶瓷的应变曲线通常表现出非线性和具有明显滞后的特征^[18,19]。上述这种应变主要是由非-180°畴在电场作用下的翻转所产生。对于弛豫铁电陶瓷, 其内部存在大量 PNRs, 其畴壁移动所受能量势垒更低, 畴壁移动会变得更加容易, 从而提高铁电陶瓷在低电场下的应变响应^[20]。一般情况下, 弛豫铁电陶瓷的电致应变主要是电致伸缩效应、逆压电效应和畴壁移动三者的综合贡献, 因此应变曲线会显示出复杂的蝶形特征。

1.4 场致相变效应

有些弛豫铁电陶瓷在电场的作用下还会呈现出巨大的应变响应, 该应变产生的物理机制一直受到研究者的重视。目前, 接受度比较高的一种解释是, 大应变响应的产生归因于在电场作用下弛豫铁电陶瓷中会发生弛豫相到铁电相的转变^[21-22], 该理论的模型解释如图 3 所示。对于普通铁电陶瓷, 其内部主要由随机取向的极性长程有序铁电畴组成, 因此, 一般不会发生上述相变。对于弛豫铁电陶瓷, 其内部拥有许多 PNRs, 当外加电场足够高时, 这些 PNRs 能够转变为极性长程铁电畴, 从而产生巨大的应变响应。当电场去除后, 这些极性长程铁电畴变得非常不稳定, 又返回到初始无序的 PNRs 态^[23]。通过上述分析可知, 这种大应变响应产生的机制是外电场对 PNRs 的调控, 使弛豫铁电陶瓷从无序的遍历弛豫态转变为有序的铁电态, 引发剧烈晶格畸变, 因此, 其应变响应主要来自相变过程中的非本征极化变化, 而非畴壁运动。

1.5 缺陷工程

除了上述几种常见的应变响应产生的机理外, 最近几年, 利用缺陷工程使铁电陶瓷产生大应变响应的报道也非常多, 尤其在无铅铁电陶瓷研究中, 构建缺陷偶极子增强应变响应是当前研究的热点^[24-26]。缺陷偶极子能极大地增强应变响应的主要原因如下: 对铁电陶瓷极化时, 缺陷偶极子会产生内偏电场, 当施加电场与内偏电场方向一致时, 铁电陶瓷的应变响应能被极大地提高, 如图 4 所示^[27]。

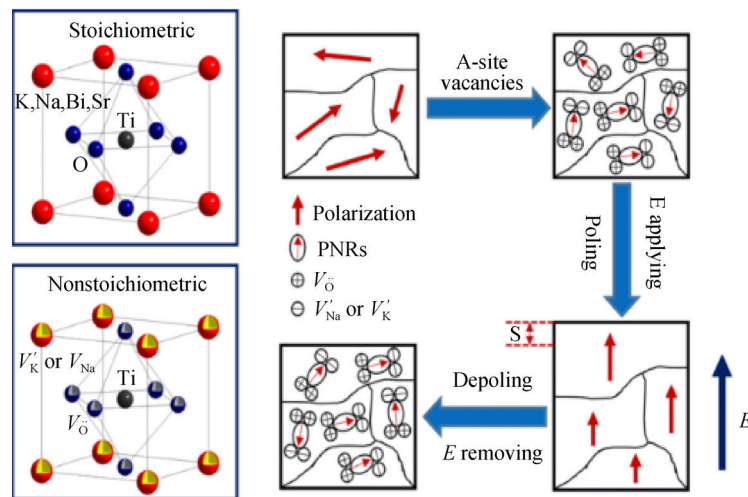
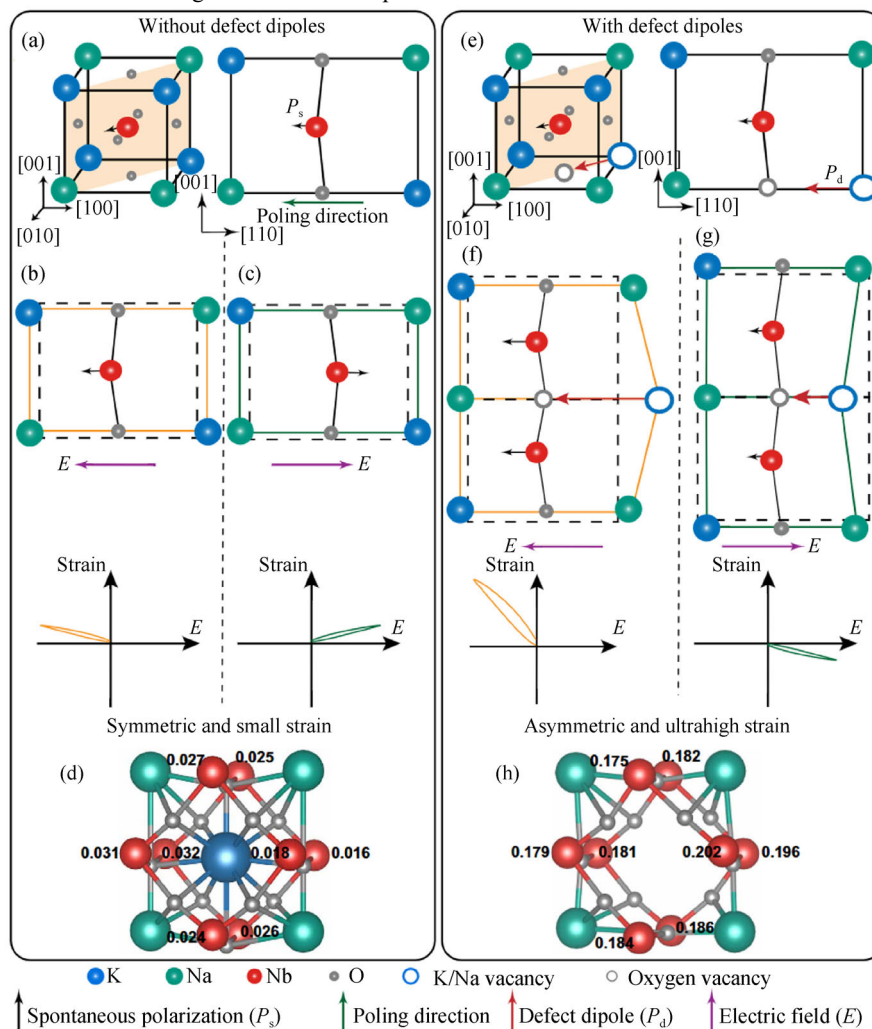
图 3 由相变增强应变响应的模型^[23]Fig. 3 Model of the phase transitions enhanced strain^[23]

图 4 KNN100 陶瓷利用缺陷偶极子增强应变响应的机理示意图 (a)无缺陷偶极子陶瓷的自发极化; (b)负电场和(c)正电场下的电应变行为; (d)Nb 离子在电场下的位移; (e)具有缺陷偶极子陶瓷的自发极化; (f)负电场和(g)正电场下的电应变行为; (h)Nb 离子在电场下的位移^[27]

Fig. 4 Mechanism of defect dipole-induced ultrahigh electrostrain of KNN100 ceramics (a) spontaneous polarization of ceramics without defect dipoles; electrostrain behaviors under (b) negative and (c) positive electric fields; (d) Ion displacement of Nb after applying an electric field; (e) Spontaneous polarization of ceramics with defect dipoles. electrostrain behaviors under (f) negative and (g) positive electric fields; (h) Ion displacement of Nb after applying an electric field^[27]

目前, 报道引入缺陷偶极子的方法有多种, 而元素掺杂是最主要的方法。由于掺杂元素在钙钛矿结构(ABO_3)铁电陶瓷晶格中占据位置的不同, 缺陷偶极子分为 A 位缺陷偶极子和 B 位缺陷偶极子, 两者对应变响应的增强机理具有明显差异。A 位缺陷偶极子对铁电畴具有钉扎效应, 其会阻碍畴的连续性, 导致纳米畴的产生和非-180°畴的比例增加, 通过畴壁运动来增强应变响应^[28], 这种缺陷偶极子会导致铁电陶瓷的应变滞后增大和应变温度稳定性变差; 而 B 位缺陷偶极子会导致晶格畸变, 通过增加晶格畸变的程度来增强应变响应^[27], 其应变滞后较小并且应变温度稳定性也比较优异。利用缺陷偶极子增强铁电陶瓷的应变响应, 其双极应变曲线会显示出非对称的、畸形的特征。

2 弛豫铁电陶瓷电致应变滞后

应变滞后(Hys)是指对弛豫铁电陶瓷施加电场和撤去电场的过程中, 其应变曲线不重合, 在应变曲线图中具体表现为撤去电场时的应变曲线高于施加电场时的应变曲线, 即应变曲线呈现出非线性特征。铁电陶瓷应变滞后产生的主要原因是, 在电场的施加下铁电陶瓷的铁电畴中电偶极子随电场进行翻转排列, 当电场撤除后, 电偶极子不能立即恢复到初始状态, 因此, 产生了应变滞后^[29]。在实际应用中, 铁电陶瓷的应变滞后大小对压电致动器的位移精度具有重要的影响, 应变滞后越大, 其位移精度越差。铁电陶瓷的应变滞后的大小与其应变响应产生的类型有关, 由电致伸缩效应产生的应变响应通常其应变滞后都较小, 因为电致伸缩效应与铁电畴翻转无关。由逆压电效应和铁电畴翻转产生的应变响应会表现出明显的应变滞后, 因为铁电畴的不可逆翻转会导致应变随电场非线性变化。由场致相变产生的应变响应会表现出非常大的应变滞后, 例如, 由遍历弛豫相到铁电相的转变或者由反铁电相到铁电相的转变所产生的应变。通常, 应变滞后可以利用以下公式进行计算^[30]:

$$\text{Hys} = \frac{\Delta S_{\max}}{S_{\max}} \times 100\% \quad (3)$$

式中: $\Delta S_{\max}$ 为施加电压与移去电压时应变的最大差值; $S_{\max}$ 为最大应变值。

3 弛豫铁电陶瓷应变特性研究现状

按材料的种类, 弛豫铁电陶瓷分为铅基弛豫铁电陶瓷和无铅弛豫铁电陶瓷 2 大类, 目前, 对这 2

类铁电陶瓷应变响应的研究报道也比较多, 下面分别对这 2 类铁电陶瓷应变特性的研究进展进行介绍。

3.1 铅基弛豫铁电陶瓷

铅基弛豫铁电陶瓷应变曲线通常显示出更小的应变滞后, 大多数低于 20%, 其更有利于在高精度压电致动器上的应用, 因此铅基弛豫铁电陶瓷的研究一直受到研究者的高度重视, 目前铅基弛豫铁电陶瓷的研究主要集中在铌镁酸铅-钛酸铅($\text{Pb}(\text{Mg}_{1/3}\text{Nb}_{2/3})\text{O}_3\text{-PbTiO}_3$, PMN-PT)、锆酸铅-钛酸铅($\text{PbZrO}_3\text{-PbTiO}_3$, PZ-PT)和铈酸铋-钛酸铅($\text{BiScO}_3\text{-PbTiO}_3$, BS-PT)3 种体系上。

3.1.1 铌镁酸铅-钛酸铅铁电陶瓷 PMN 自从 1961 年被第 1 次报道以来, 引起了压电研究者极大的兴趣。PMN 内部具有无序结构并存在极性纳米区域, 这些特征使其呈现出相当大的介电常数, 大的介电常数非常有利于该铁电陶瓷通过电致伸缩效应来提高应变响应, 而且应变滞后也较小, 因此非常适合在高精度位移致动器上应用^[31]。再者, PMN 具有非常低的居里温度, 其室温下无铁电性, 与传统 PT 铁电体复合所形成的 PMN-PT 二元弛豫铁电陶瓷会展示出优异的机电性能, 并且居里温度能提高到大约 140 °C, 因此, 该二元弛豫铁电陶瓷在过去几十年中得到了广泛的研究^[32]。Wongmaneerun 等^[33]在研究 PMN-PT 陶瓷的电学性能时发现, 最佳应变响应在 30%PT 含量的准同型相界(MPB)组分处获得, 在 40 kV·cm⁻¹ 条件下其应变约为 0.12%。Leite 等^[34]报道 PMN-35PT 陶瓷的应变响应在 85 kV·cm⁻¹ 下为 0.34%, 应变滞后约为 10%。根据上述文献报道可知, 纯的 PMN-PT 陶瓷优点是应变滞后较小, 但是其应变响应也非常小。

仅依靠构建 MPB 对 PMN-PT 铁电陶瓷的电学性能进行提高已变得非常艰难。目前, 常用提高应变响应的方法是向 PMN-PT 二元体系中掺杂少量的离子或与其他钙钛矿结构的化合物进行复合形成多元铁电陶瓷。所报道的掺杂离子主要包括: Ba^{2+} ^[35]、 Y^{3+} ^[36]、 Ho^{3+} ^[37]、 Pr^{3+} ^[38]、 Ce^{3+} ^[39, 40]、 Sm^{3+} ^[41]、 Eu^{3+} ^[42-44] 等离子。研究表明, 这些掺杂离子是对 PMN-PT 铁电陶瓷晶格中的 Pb^{2+} 进行部分取代来破坏长程有序铁电畴以提高应变响应。Qin 等^[38]在 PMN-PT 陶瓷中掺杂 Pr^{3+} 时发现, 其应变响应在 20 kV·cm⁻¹ 时达到了 0.12%。Li 等^[41]用稀土 Sm^{3+} 取代 PMN-PT 中的 Pb^{2+} 时发现其压电系数和应变响应能被极大地提高, Sm^{3+} 掺杂后该铁电陶瓷在非常小的电场下其应变

响应很大($2\text{ kV}\cdot\text{cm}^{-1}$, 0.03%), 如图 5 所示。随后几年稀土 Eu^{3+} 也被证明与 Sm^{3+} 具有同样的效果, 其中

Li 等^[44]制备的 Eu^{3+} 掺杂 PMN-PT 陶瓷的应变响应在 $14\text{ kV}\cdot\text{cm}^{-1}$ 电场下达到了 0.14%。

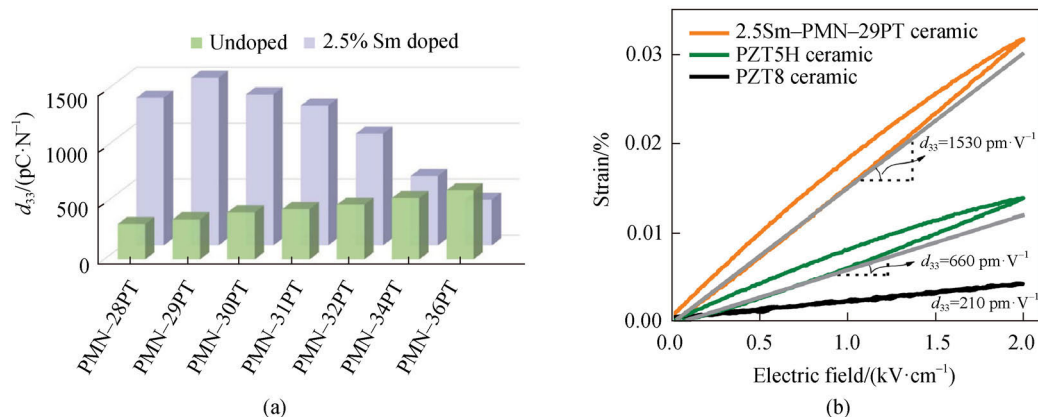


图 5 Sm-PMN-PT 铁电陶瓷的(a)压电系数和(b)单极应变曲线^[41]

Fig. 5 (a) Piezoelectric coefficient and (b) unipolar strains curves of Sm-PMN-PT ferroelectric ceramics^[41]

此外, 向 PMN-PT 二元陶瓷中引入钙钛矿结构的化合物使其形成三元弛豫铁电陶瓷来提高应变响应的报道也非常多, 这些钙钛矿结构的化合物包括: $\text{Pb}(\text{Sc}_{1/2}\text{Nb}_{1/2})\text{O}_3$ (PSN)^[45]、 $\text{Pb}(\text{In}_{1/2}\text{Nb}_{1/2})\text{O}_3$ (PIN)^[46]、 $\text{Pb}(\text{Zn}_{1/3}\text{Nb}_{2/3})\text{O}_3$ (PZN)^[47]、 $\text{Ba}(\text{Zn}_{1/3}\text{Nb}_{2/3})\text{O}_3$ (BZN)^[48]、 BiFeO_3 (BF)^[49]、 $\text{Bi}(\text{Mg}_{1/2}\text{Ti}_{1/2})\text{O}_3$ (BMT)^[50]、 $\text{Bi}(\text{Zn}_{1/2}\text{Ti}_{1/2})\text{O}_3$ (BZT)^[51]、 $(\text{Bi}_{1/2}\text{Na}_{1/2})\text{ZrO}_3$ (BNZ)^[52] 等。Guo 等^[45]向 PMN-PT 中引入 PSN 制备出 PSN-PMN-PT 三元弛豫铁电陶瓷, 该陶瓷在 $20\text{ kV}\cdot\text{cm}^{-1}$ 电场下的应变响应为 0.12%, 在 $30\sim 150\text{ }^\circ\text{C}$ 内应变的最大变化约为 38%。Xia 等^[51]在 PMN-PT 中引入少量的 BZT 组元, 其应变响应在 $50\text{ kV}\cdot\text{cm}^{-1}$ 时达到了 0.24%, 应变滞后约为 15%, 如图 6 所示。

3.1.2 锆酸铅-钛酸铅铁电陶瓷 PMN-PT 铁电陶瓷由于具有较低的居里温度, 其在高温下的应用受到极大的限制。而 PZ-PT 铁电陶瓷具有相对较高的居里温度(约 $380\text{ }^\circ\text{C}$)和优异的综合压电性能, 自从 1954 年被 Jaffe 等发现以来一直是压电陶瓷研究和应用的热点^[53]。直到现在, 对于 PZ-PT 基陶瓷的研究依然比较重视, 并且目前还是制备压电器件的首选材料。Jaffe 等建立了 PZ-PT 体系的二元相图并发现该体系在 MPB 附近显示出异常高的压电和机电性能, 并且最高压电系数和最大应变响应在 Ti/Zr 比值为 47:53 处实现^[53,54]。但是纯 PZ-PT 陶瓷的压电性能和应变响应都相对较低, 因此对其性能进一步改善一直是研究的热点。

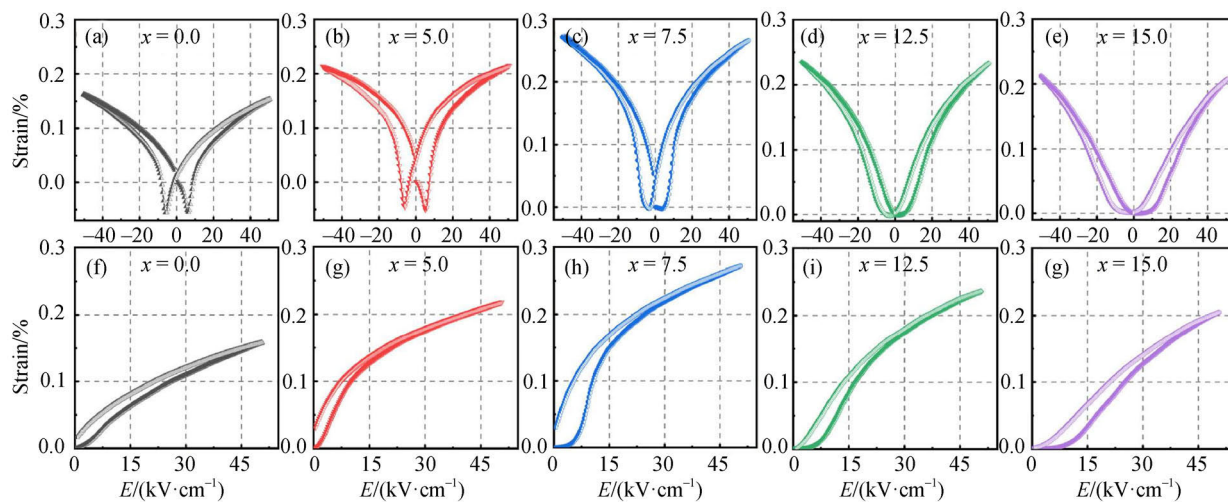


图 6 不同 BZT 含量 PMN-PT 铁电陶瓷的(a)–(e)双极应变曲线和(f)–(g)单极应变曲线^[51]

Fig. 6 (a)–(e) bipolar and (f)–(g) unipolar strain curves for PMN-PT-BZT ceramics with different BZT content^[51]

科研人员同样采用掺杂改性和多元复合的手段对 PZ-PT 陶瓷的压电性能和应变响应的提高进行了大量研究。对 PZ-PT 的掺杂改性又有软、硬之分, 软性掺杂是指用高价阳离子取代其晶格中的 Pb^{2+} 、 Zr^{4+} 、 Ti^{4+} , 而硬性掺杂正好与软性掺杂相反, 则是利用低价离子对其进行取代^[55]。目前, 提高 PZ-PT 陶瓷的压电性能和应变响应主要采用的是软性掺杂。Gao 等^[56]对 PZ-PT 陶瓷掺杂 Sm^{3+} 时发现, 在 $70 \text{ kV}\cdot\text{cm}^{-1}$ 电场下其最大应变响应能达到 0.31%, 而应变滞后仅为 10%。Pu 等^[57]对 PZ-PT 陶瓷进行 Nb^{3+} 掺杂时发现, 在 $50 \text{ kV}\cdot\text{cm}^{-1}$ 条件下其最大应变响应能达到 0.32%, 其应变响应的改善归因于 A 位铅空位和条纹畴结构的出现, 铅空位可以降低内应力并提高 90° 畴壁的移动性, 而条纹畴结构更容易对外部电场刺激做出响应。

另一种提高 PZ-PT 应变响应的重要手段是采用复合组元的设计方法, 该方法主要通过可逆相转变、弛豫特性增强等作用来提高应变响应。1965 年, 日本松下公司率先成功研制出性能优异的 PMN-PZ-PT 三元铁电陶瓷, 该三元陶瓷的成功研制在当时引起了人们广泛的关注。受此启发, 对其他 PZ-PT 基三元陶瓷的开发也受到高度重视, 此后一系列具有优异压电性能和应变响应的 PZ-PT 基三元铁电陶瓷被成功制备, 包括: $\text{Pb}(\text{Zn}_{1/3}\text{Nb}_{2/3})\text{O}_3\text{-PbZrO}_3\text{-PbTiO}_3$ (PZN-PZ-PT)^[58]、 $\text{Pb}(\text{Ni}_{1/3}\text{Nb}_{2/3})\text{O}_3\text{-PbZrO}_3\text{-PbTiO}_3$ (PNN-PZ-PT)^[59]、 $\text{Pb}(\text{In}_{1/2}\text{Nb}_{1/2})\text{O}_3\text{-PbZrO}_3\text{-PbTiO}_3$ (PIN-PZ-PT)^[60]、 $(\text{Bi}_{1/2}\text{Na}_{1/2})\text{TiO}_3\text{-PbZrO}_3\text{-PbTiO}_3$ (BNT-PZ-PT)^[61-63]、 $\text{Pb}(\text{Ni}_{1/3}\text{Ta}_{2/3})\text{O}_3\text{-PbZrO}_3\text{-PbTiO}_3$ (PNT-PZ-PT)^[64]、 $(\text{Bi}_{1/2}\text{K}_{1/2})\text{TiO}_3\text{-PbZrO}_3\text{-PbTiO}_3$ (BKT-PZ-PT)^[65] 等。Zhang 等^[58]报道 PZN-PZ-PT 三元铁电陶瓷的应变响应在 $30 \text{ kV}\cdot\text{cm}^{-1}$ 时为 0.38%, 大应变响应主要由电场诱导单斜相和四方相的可逆转变引起。Liu 等^[61]报道把 BNT 组元引入到 PZ-PT 陶瓷中应变响应能被提高, 在 $30 \text{ kV}\cdot\text{cm}^{-1}$ 电场下的应变响应约为 0.21%, 并且在 $25\sim 250^\circ\text{C}$ 温度范围内应变的变化小于 8%, 如图 7 所示, 应变响应的改善归因于将 BNT 引入到 PZ-PT 陶瓷中增强了弛豫行为, 并且电畴变得短而宽, 在电场作用下响应会更快。

3.1.3 钽酸铋-钛酸铅铁电陶瓷 随着能源与航天等领域的快速发展, 压电器件在极端环境下使用的频率变得日益频繁, 因此开发具有更高居里温度的压电陶瓷变得至关重要。在已成功研制的高温压电陶瓷中, BS-PT 陶瓷被发现性能是最杰出的, 其

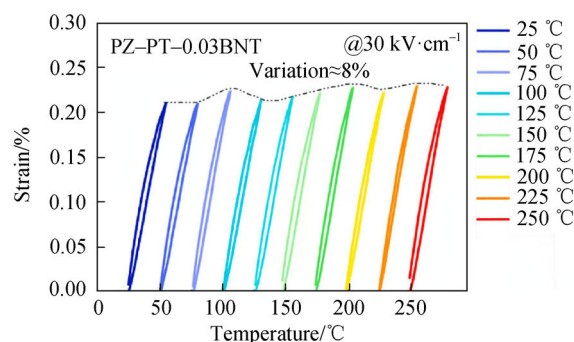


图 7 PZ-PT-BNT 铁电陶瓷在不同温度下应变曲线的变化^[61]
Fig. 7 Temperature-dependent strain of PZ-PT-BNT ceramics^[61]

不仅展示出较高的居里温度(450°C), 而且压电系数也非常大(约为 460 pC/N), 因此该陶瓷一直受到研究者的极大关注和广泛研究。2001 年, Eitel 等^[66]首先报道了 BS-PT 二元铁电陶瓷的 MPB 区域位于 64PT 的组分处, 并且证明了 MPB 区域处的 BS-PT 陶瓷会呈现出最高的压电性能和应变响应。随后, 在 2005 年, Zhang 等^[67]报道在 PT 含量为 64% 时, BS-PT 陶瓷的压电系数为 460 pC/N , 在 $40 \text{ kV}\cdot\text{cm}^{-1}$ 下应变响应大约为 0.17%。

为了改善 BS-PT 陶瓷的压电性能和应变响应, 第 3 组元的引入是近年来文献报道最多的手段, 这些第 3 组元主要包括: $\text{Bi}(\text{Ni}_{1/2}\text{Zr}_{1/2})\text{O}_3$ (BNZ)^[68]、 $(\text{Sr}_{0.7}\text{Bi}_{0.2})\text{TiO}_3$ (SBT)^[69]、 $\text{Bi}(\text{Zn}_{1/2}\text{Zr}_{1/2})\text{O}_3$ (BZZ)^[70]、 $\text{Pb}(\text{Sn}_{1/3}\text{Nb}_{2/3})\text{O}_3$ (PSN)^[71]、 $\text{Bi}(\text{Zn}_{1/2}\text{Ti}_{1/2})\text{O}_3$ (BZT)^[72]、 $\text{Pb}(\text{Sc}_{1/2}\text{Nb}_{1/2})\text{O}_3$ (PSN)^[73]、 $\text{Bi}(\text{Sn}_{1/3}\text{Nb}_{2/3})\text{O}_3$ (BSN)^[74] 等。Wang 等^[71]在 BS-PT 中引入 PSN, 其应变响应在 $40 \text{ kV}\cdot\text{cm}^{-1}$ 能达到 0.22%, 如图 8 所示, 其应变提高归因于构建三方相和四方相共存的 MPB 相界。Dong 等^[74]在 BS-PT 陶瓷中引入 BSN 组元, 在 MPB 组分中应变达到最大, $50 \text{ kV}\cdot\text{cm}^{-1}$ 电场下应变响应达到 0.3%, 大应变响应主要由于在 MPB 成分中存

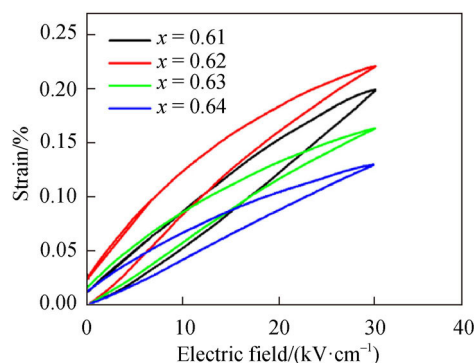


图 8 不同 PT 含量下 BS-PT-PSN 的单极应变曲线^[71]
Fig. 8 Unipolar strain curves of BS-PT-PSN ceramics with different PT^[71]

在的纳米畴结构, 并且优异的高温热稳定性也和铁电畴及相结构的稳定性相关。

3.2 无铅弛豫铁电陶瓷

众所周知, 铅元素对人体和环境有极大的危害, 随着全球对环保意识的增强, 考虑到反对使用铅元素的环境法规, 科研人员已经做出了大量努力寻找无铅铁电陶瓷体系来取代主流的铅基铁电陶瓷。在无铅弛豫铁电陶瓷中, 目前, 研究应变特性最多的体系是钛酸铋钠(BNT)基、铌酸钾钠(KNN)基和铁酸铋-钛酸钡(BF-BT)3种铁电陶瓷。

3.2.1 钛酸铋钠基铁电陶瓷 钛酸铋钠(BNT)基铁电陶瓷因其能实现高应变响应方面的潜力而受到重视, 纯的 BNT 铁电陶瓷由于在室温下矫顽场(E_c)较大且烧成温度较高, 因此, 压电性能和应变响应较低, 在 $70 \text{ kV} \cdot \text{cm}^{-1}$ 下其应变响应仅为 0.07% ^[75,76]。为了改善 BNT 陶瓷的应变响应, 在 BNT 陶瓷中复合第二组元已证明是一个有效的方法, 例如 $\text{BaTiO}_3(\text{BT})$ ^[77]、 $(\text{K}_{1/2}\text{Bi}_{1/2})\text{TiO}_3(\text{BKT})$ ^[78]、 $\text{SrTiO}_3(\text{ST})$ ^[79,80]、 $\text{BaAlO}_{2.5}(\text{BA})$ ^[81]等。Guo 等^[77]在 BNT 中复合 BT 组元, 当 BT 含量为 6%时, 在 $60 \text{ kV} \cdot \text{cm}^{-1}$ 下其应变响应约为 0.14% , 并且在 $30 \sim 150^\circ\text{C}$ 温度范围内应变响应的变化非常大。Sowmya 等^[78]在 BNT 中复合 BKT 组元, 在 $60 \text{ kV} \cdot \text{cm}^{-1}$ 电场下的应变响应约为 0.3% 。

对于 BNT 基体系, 目前研究最多的复合二元体系是 BNT-BT(BNBT)和 BNT-BKT(BNKT), 为了改善 BNBT 和 BNKT 铁电陶瓷的应变响应, 目前有 2 种方法被采取。一种是向 MPB 附近的 BNBT 或 BNKT 陶瓷中掺杂少量离子, 已报道的掺杂离子主要包括: La^{3+} ^[82]、 Hf^{4+} ^[83]、 Zr^{4+} ^[84]、 Sb^{3+} ^[85-86]、 Sn^{4+} ^[87]等。Dong 等^[86]在 BNBT 铁电陶瓷中掺杂 Sb^{3+} , 其应变响应在 $80 \text{ kV} \cdot \text{cm}^{-1}$ 下从 0.22% 提高到 0.47% , 在 $20 \sim 140^\circ\text{C}$ 温度范围内应变响应的波动仅为 10.1% 。Li 等^[88]在 BNBT 中同时掺杂 K^+ 、 Na^+ 、 Li^+ 、 Nb^{5+} 、 Sb^{3+} 等多种离子后, 该铁电陶瓷的应变特性被显著改善, 在 $55 \text{ kV} \cdot \text{cm}^{-1}$ 电场下最大应变响应为 0.51% , 应变滞后约为 60% , 在 $25 \sim 125^\circ\text{C}$ 应变响应的最大变化约为 37% , 如图 9 所示。

另一种有效的方法是向 BNBT 或 BNKT 陶瓷中引入少量具有钙钛矿结构的氧化物来提高其应变响应, 最近已报道的这些氧化物主要包括: $\text{SrTiO}_3(\text{ST})$ ^[89]、 $\text{Bi}(\text{Mg}_{1/2}\text{Ti}_{1/2})\text{O}_3(\text{BMT})$ ^[90]、 $\text{NaNbO}_3(\text{NN})$ ^[91-93]、 $\text{BiScO}_3(\text{BS})$ ^[94]、 $\text{LiNbO}_3(\text{LN})$ ^[95]、 $\text{Nb}_{1/2}\text{Ga}_{1/2}\text{O}_3(\text{NG})$ ^[96]、 $\text{NaTaO}_3(\text{NT})$ ^[97]、 $\text{Sr}_{0.7}\text{La}_{0.2}\text{TiO}_3(\text{SLT})$ ^[98]、 $\text{Ba}(\text{Sn}_{0.70}\text{Nb}_{0.24})$

$\text{O}_3(\text{BSN})$ ^[99]、 $\text{AgNbO}_3(\text{AN})$ ^[100]、 $\text{BiCoO}_3(\text{BCO})$ ^[101]、 $\text{Fe}_{1/2}\text{Nb}_{1/2}\text{O}_3(\text{FN})$ ^[102]、 $\text{Bi}(\text{Zn}_{2/3}\text{Ta}_{1/3})\text{O}_3(\text{BZT})$ ^[103]、 $\text{Ba}(\text{Fe}_{1/2}\text{Sb}_{1/2})\text{O}_3(\text{BFS})$ ^[104]、 $\text{Sr}(\text{Hf}_{0.5}\text{Sb}_{0.4})\text{O}_3$ ^[105]、 $(\text{K}_{1/2}\text{Na}_{1/2})\text{NbO}_3(\text{KNN})$ ^[106]等。Wang 等^[92]在 BNBT 陶瓷中引入 NN 时, 发现在 $80 \text{ kV} \cdot \text{cm}^{-1}$ 电场下其应变响应从 0.2% 提高到 0.47% , 但是应变滞后极大地增加。Yang 等^[93]在 BNKT 中引入 NN, 在 $60 \text{ kV} \cdot \text{cm}^{-1}$ 电场下其应变响应为 0.41% 。Wang 等^[100]在 BNBT 中引入 AN, 在 $65 \text{ kV} \cdot \text{cm}^{-1}$ 电场下应变响应从 0.2% 提高到 0.48% , 其应变滞后也极大地增加, 如图 10 所示。Hao 等^[106]报道在 BNKT 中引入 KNN, 在 $80 \text{ kV} \cdot \text{cm}^{-1}$ 电场下其应变响应能从 0.27% 提高到 0.46% , 应变滞后达到了 62% , 在 $20 \sim 140^\circ\text{C}$ 温度范围内应变响应的变化也非常大。

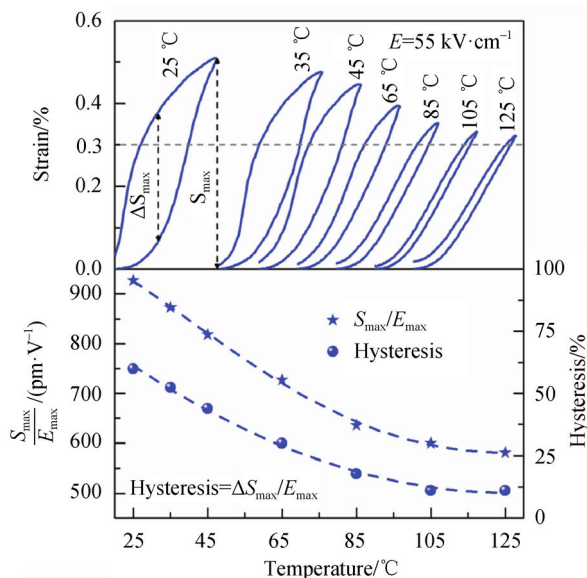


图 9 不同温度下 BNBT-KNLNS 陶瓷应变特性的变化^[88]
Fig. 9 Temperature-dependent strain properties of BNBT-KNLNS ceramics^[88]

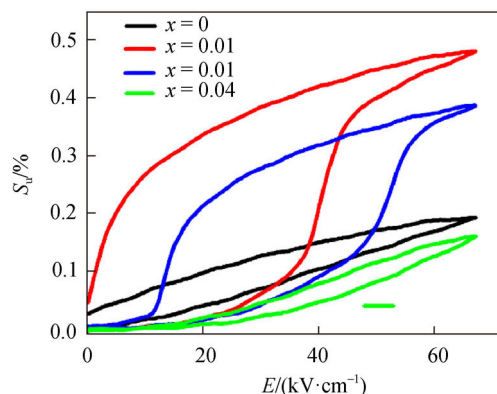


图 10 不同 AN 含量的 BNBT-100xAN 铁电陶瓷的单极应变曲线^[100]
Fig. 10 Unipolar strain curves for BNBT-100xAN ceramics with different AN content^[100]

3.2.2 铌酸钾钠基铁电陶瓷 铌酸钾钠(KNN)铁电陶瓷由于具有较好的温度稳定性和优异的压电性能, 被认为是最具有潜力代替商用铅基陶瓷的压电材料之一^[107]。纯 KNN 铁电陶瓷应变响应较小并且去除电场后的剩余应变较高, 这严重限制了其实际应用^[108]。在过去几年里, 各国研究者专注于 KNN 基铁电陶瓷应变特性的改善, 并取得了一系列进展。研究发现构筑缺陷偶极子是提升 KNN 陶瓷应变响应的一种有效策略, 其主要是通过向 KNN 陶瓷中掺杂离子来引入, 最近报道的掺杂离子包括 Li^+ ^[109]、 $\text{Li}^+/\text{Sb}^{3+}$ ^[110]、 Sr^{2+} ^[111]、 $\text{Ca}^{2+}/\text{Ba}^{2+}$ ^[112]、 Na^+ ^[113]、 Fe^{2+} ^[114]、 Al^{3+} ^[115]等。Guo 等^[111]通过 Science 期刊报道在 KNN 铁电陶瓷中掺杂 Sr^{2+} 获得了 0.8% ($40 \text{ kV} \cdot \text{cm}^{-1}$) 超高的应变响应, 如图 11 所示。研究发现, Sr^{2+} 的掺入不仅可以调控 KNN 陶瓷中 K^+/Na^+ 空位和氧空位含量, 而且还可以调控 KNN 陶瓷的相结构和铁电畴结构, 在电场的作用下该铁电陶瓷内部可以形成缺陷偶极子, 超高应变响应正是由缺陷偶极子与铁电畴的强耦合作用所致。

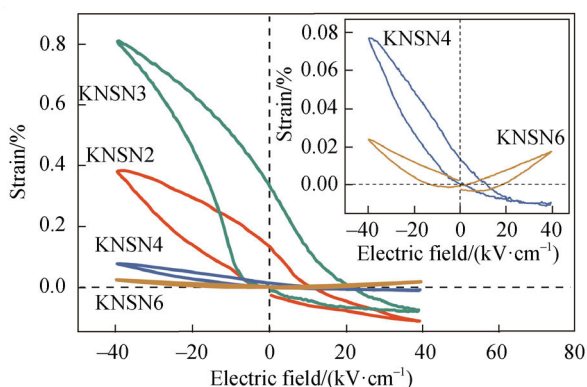


图 11 KNSN 陶瓷的双极应变曲线^[111]

Fig. 11 Bipolar strain curves of KNSN ceramics^[111]

另一种策略是通过构建多晶型相界(PPB)来提高 KNN 陶瓷的压电和应变特性。把钙钛矿型氧化物引入到 KNN 陶瓷中可实现三方-正交-四方(R-O-T)三相共存, 进而可以大幅度提高 KNN 基陶瓷的应变响应。最近报道的这些氧化物主要包括 SrTiO_3 ^[116]、 CaZrO_3 ^[117]、 BaZrO_3 ^[118]、 $(\text{Bi}, \text{Na})\text{ZrO}_3$ ^[119-121]、 $(\text{Bi}, \text{K})\text{ZrO}_3$ ^[122]、 $(\text{Bi}, \text{Na})\text{HfO}_3$ ^[123]、 BiFeO_3 ^[124]、 $(\text{Bi}, \text{In})\text{NaZrO}_3$ ^[125]等。例如, Lv 等^[119]把 $(\text{Bi}, \text{Na})\text{ZrO}_3$ 氧化物引入到 $(\text{K}, \text{Na})(\text{Nb}, \text{Sb})\text{O}_3$ 中, 通过调节 Bi 的含量实现了长程有序与极性纳米区域之间的平衡和 R-O-T 三相共存, 大幅度提高了该铁电陶瓷的压电系数(475 pC/N)和应变响应(约 0.14% , $30 \text{ kV} \cdot \text{cm}^{-1}$),

如图 12 所示, 并且还降低了压电系数对成分的敏感性, 此外由于扩散的多相共存以及畴转换贡献的减弱, 导致该铁电陶瓷应变响应的温度稳定性也被增强。

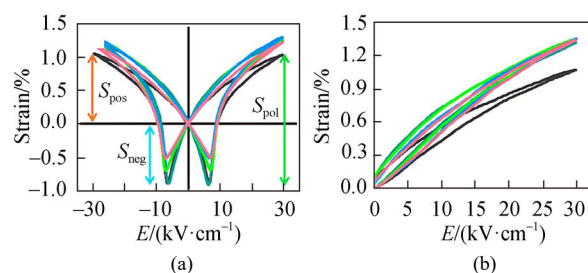


图 12 KNNS-BNZ 铁电陶瓷的(a)双极应变曲线和(b)单极应变曲线^[119]

Fig. 12 (a) Bipolar strain curves and (b) unipolar strain curves of KNNS-BNZ ferroelectric ceramics^[119]

此外, 近 2 年研究发现 KNN 陶瓷样品的厚度对应变特性具有显著的影响, 通常铁电陶瓷样品的厚度小于 $500 \mu\text{m}$ 时, 样品极化后会产生弯曲形变, 并且在交变电场作用下还会表现出交替弯曲现象, 而这一特性使铁电陶瓷的应变响应得到大幅度提升^[126-129]。Chen 等报道^[129]极化的 KNNCZ-Cu 陶瓷在不同样品厚度下的应变响应差别较大。随着样品厚度的减小, 其应变响应在不断增大, 如应变响应从 $600 \mu\text{m}$ 厚度时的 0.15% 增加到 $230 \mu\text{m}$ 时的 1.3% ($50 \text{ kV} \cdot \text{cm}^{-1}$), 如图 13 所示。除了 KNN 基陶瓷, 电致弯曲形变现象导致铁电陶瓷应变响应的提升, 最近在其它无铅陶瓷中也有报道^[130-132]。

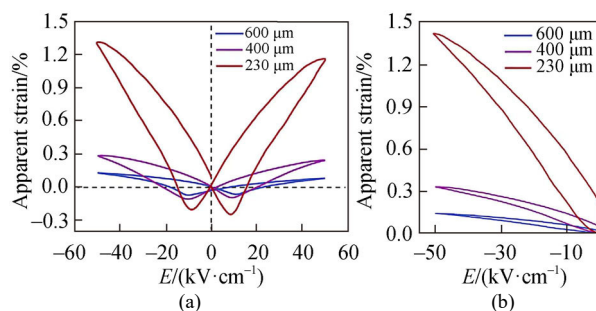


图 13 KNNCZ-Cu 陶瓷在不同厚度下的应变曲线(a)双极应变曲线和(b)单极应变曲线^[129]

Fig. 13 (a) Bipolar strain curves and (b) unipolar strain curves of KNNCZ-Cu ceramics with different thicknesses^[129]

Dai 等^[127]揭示了弯曲变形导致大应变响应的物理本质, 研究发现, 缺陷偶极子易在铁电陶瓷表面产生, 并且由于其低形成能都朝着铁电陶瓷内部取向, 在电场的作用下通过不对称响应产生应力差异, 导致陶瓷弯曲变形, 进而导致应变响应增强, 如图 14 所示。

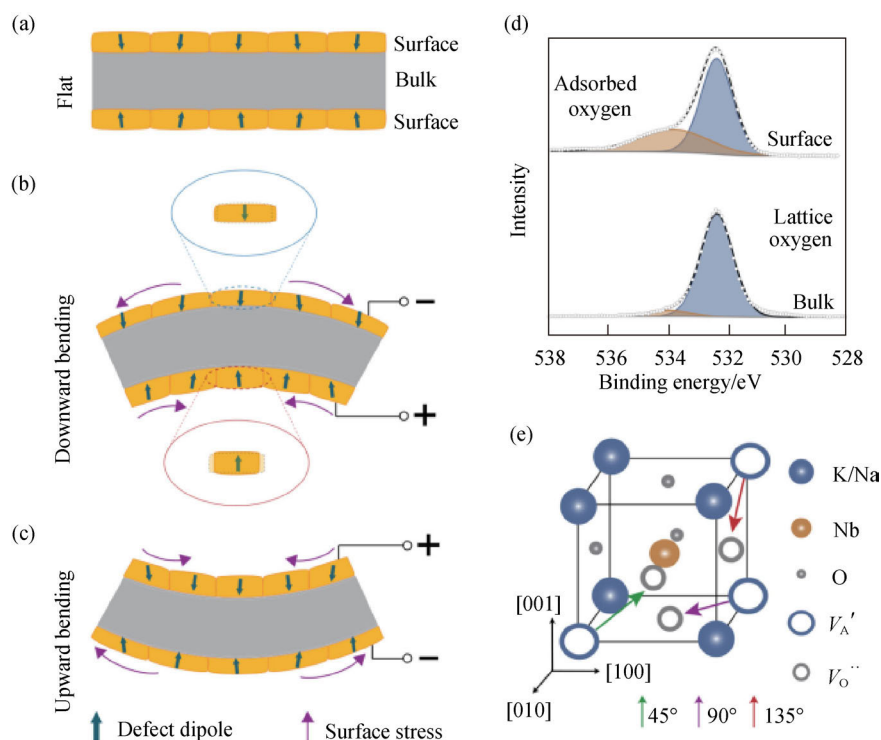


图 14 电致弯曲机理分析示意图 (a)~(c) 具有缺陷偶极子的 KNN 陶瓷在电场作用下的电致弯曲原理图; (d) O1s 轨道的 X-射线光电电子能谱; (e) 露面为(001)晶面的不同取向缺陷偶极子示意图^[127]

Fig. 14 Schematic diagram of working mechanism for electrobending; (a)~(c) Schematic diagram of electrobending in KNN ceramics with defect dipoles under an electric field; (d) X-ray photoelectron spectroscopy of O1s orbitals; (e) Schematic diagram of defect dipoles with different orientations when the exposed surface is the (001) crystal plane^[127]

3.2.3 铁酸铌-钛酸钡铁电陶瓷 铁酸铌-钛酸钡(BF-BT)铁电陶瓷由于具有较高的居里温度($T_C > 500\text{ }^\circ\text{C}$)和良好的铁电与压电性能, 目前成为新型高温压电器件的理想材料之一, 特别是在压电致动器的高温应用中脱颖而出。大量文献报道 BF-BT 铁电陶瓷具有较大的应变响应, 其大应变响应的产生主要归因于 2 个方面^[133]。

一方面, 通过组分调控 BF-BT 铁电陶瓷的畴结构, 使其从铁电态转变为弛豫铁电态, 此时许多 PNRs 存在于铁电陶瓷结构中, 与长程铁电畴相比, PNRs 具有更高的动态响应能力, 从而导致应变响应增加^[134]。Li 等^[135]通过两步法在 BF-BT 陶瓷中引入 $\text{Bi}(\text{Zn}_{2/3}\text{Nb}_{1/3})\text{O}_3(\text{BZN})$, 构建出弛豫相与铁电相共存结构, 该铁电陶瓷在 $60\text{ kV}\cdot\text{cm}^{-1}$ 的电场下应变响应达到了 0.28%, 如图 15 所示。Hao 等^[136]在 BF-BT 铁电陶瓷中掺杂 Sm^{3+} 实现了 0.367% 的高应变响应和 13.3% 的低应变滞后, 通过分析可知, 该铁电陶瓷应变响应的提高源于通过微调 Sm^{3+} 掺杂而产生的 PNRs 和 PNRs 畴壁移动的贡献。Li 等^[137]在 BF-BT 陶瓷中引入 Eu^{3+} 使其产生弛豫相与铁电相共存结构, 其应变响应也被极大地提高。

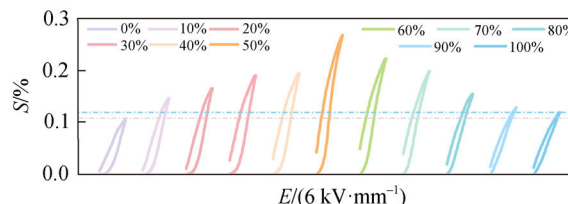


图 15 BF-BT-BZN 铁电陶瓷的单极应变曲线^[135]

Fig. 15 Unipolar strain curves of BF-BT-BZN ceramics^[135]

另一方面, 在 BF-BT 铁电陶瓷中通过构建相界面工程和缺陷偶极子工程来提高该铁电陶瓷的应变响应^[138]。例如, Kang 等^[139]在 BF-BT 陶瓷中引入 SrTiO_3 构建了一个伪立方相与三方相的相界, 在 $65\text{ kV}\cdot\text{cm}^{-1}$ 的电场下获得了 0.25% 的应变响应和 43% 的应变滞后。Lin 等^[140]对 BF-BT 铁电陶瓷进行反复极化和老化处理使其产生一个大的内偏电场, 导致该铁电陶瓷表现出不对称的宏观电学性质, 在 $20\text{ kV}\cdot\text{cm}^{-1}$ 的低电场驱动下, 获得的应变响应达到了 0.21%, 如图 16 所示, 通过研究发现内偏电场的产生源于高度取向的缺陷偶极子。

此外, 具有伪立方结构的 BF-BT 铁电陶瓷在电场作用下会发生从伪立方相转变为三方相的场致相

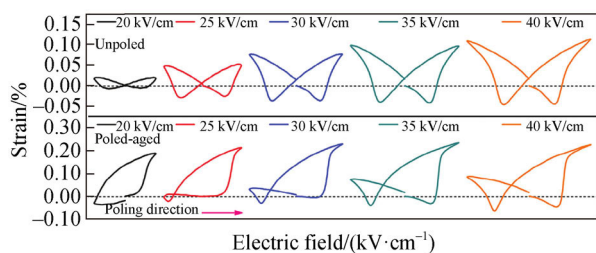


图16 未极化和极化老化BF-BT陶瓷的双极应变曲线^[140]
Fig. 16 Bipolar strain curves of the unpoled and poled-aged BF-BT ceramics^[140]

变,这种相变也能显著提高BF-BT陶瓷的应变响应^[139]。Chen等^[141]在BF-BT陶瓷中掺杂 Mn^{2+} 时研究发现,其最大应变响应出现在伪立方相成分中而非相界成分中,这是由于伪立方相与三方相能发生可逆相变所导致。然而,由场致相变和PNRs引起的大应变响应往往伴随着大的应变滞后(>50%)。因此,BF-BT铁电陶瓷的应变特性相较于铅基压电陶瓷仍然存在较大的差距。

4 面临的挑战

铅基弛豫铁电陶瓷的应变滞后较小、应变温度稳定性也较好,在电致应变特性方面具有显著优势。在高端压电致动器领域,铅基弛豫铁电陶瓷仍占主导地位,但是铅元素对人体和环境具有极大的危害,国内外有许多政策明确限制铅的使用,因此,降低铁电陶瓷的铅含量甚至实现无铅化成为必然趋势。对于压电致动器的应用,目前,无铅弛豫铁电陶瓷的电致应变特性在综合指标上仍然无法与铅基弛豫铁电陶瓷相竞争,但是无铅弛豫铁电陶瓷因规避了铅的环境毒性,成为压电材料领域的研究热点。无铅弛豫铁电陶瓷通过相变可实现较高的电致应变,但其应变滞后非常大,并且应变的温度稳定性也较差,如何实现电致应变特性均衡仍然是无铅弛豫铁电陶瓷的核心难点。尽管挑战重重,未来无铅弛豫铁电陶瓷通过组分设计、工艺优化与可靠性提升,有望在中低端领域实现对铅基弛豫铁电陶瓷的替代,但在高端精密、极端环境场景中,无铅弛豫铁电陶瓷仍难以完全替代铅基陶瓷,还需要长期技术突破。

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Recent Progress on Electric Field-Induced Strain Properties of Relaxor Ferroelectric Ceramics

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Extended Abstract

Piezoelectric actuators, which are recognized as highly promising micro-displacement driving devices due to their merits of small size, large driving force, high-precision displacement, and fast response, are widely applied in fields such as microelectronics, precision machining, precision measurement, optical fiber calibration, aerospace and biomedical engineering. Relaxor ferroelectric ceramics with their outstanding electrostrain properties emerge as ideal materials for fabricating piezoelectric actuators. As a core component of piezoelectric actuators, the electric field-induced strain properties of relaxor ferroelectric ceramics have a significant influence on the stroke, precision and efficiency of the actuators, directly determining their overall performance. This review represents the research progress on the electrostrain characteristics of relaxor ferroelectric ceramics. This review focuses on the origins of electrostrain in relaxor ferroelectric ceramics, encompassing the electrostrictive effect, inverse piezoelectric effect, domain wall motion, field-induced phase transition, and defect engineering. The causes and calculation of strain hysteresis are also represented. Subsequently, this review describes the research progress on the electrostrain properties of both lead-based and lead-free relaxor ferroelectric ceramics. For lead-based ferroelectric ceramics, the focus is on three representative systems, *i.e.*, PMN–PT, PZ–PT, and BS–PT, while lead-free ferroelectric ceramics primarily center on BNT–based, KNN–based, and BF–BT–based ceramics. In particular, this review emphasizes the recently discovered method of electric field-induced bending deformation in KNN–based ferroelectric ceramics, which is proven to significantly enhance the strain response of relaxor ferroelectric ceramics. Finally, the

review analyzes the existing challenges and future prospects for the electrostrain properties of both lead-based and lead-free relaxor ferroelectric ceramics.

Summary and Prospects Lead-based relaxor ferroelectric ceramics exhibit relatively low strain hysteresis (mostly below 20%) and superior strain temperature stability, demonstrating significant advantages in electrostrain properties, which are highly favorable for applications in high-precision piezoelectric actuators. The existing lead-based relaxor ferroelectric ceramics still dominate the high-end piezoelectric actuator market. However, lead is extremely harmful to human health and the environment, and numerous domestic and international policies have explicitly restricted the usage of lead. Consequently, reducing the lead content or even achieving lead-free compositions in ferroelectric ceramics becomes an inevitable trend. Although lead-free relaxor ferroelectric ceramics, which avoid the environmental toxicity of lead, become a research hotspot in the piezoelectric materials field, their comprehensive electrostrain properties still cannot compete with those of lead-based ferroelectric ceramics for piezoelectric actuator applications. Lead-free relaxor ferroelectric ceramics can achieve high electrostrain responses through reversible phase transitions between relaxor and ferroelectric phase, they suffer from excessively large strain hysteresis and poor strain temperature stability. How to realize a balanced electrostrain properties remains a core challenge for lead-free relaxor ferroelectric ceramics. Despite these challenges, lead-free relaxor ferroelectric ceramics are expected to replace lead-based ferroelectric ceramics in mid-to-low-end applications through component design, process optimization, and reliability improvements. Nevertheless, in high precision or extreme environment scenarios, lead-free relaxor ferroelectric ceramics still struggle to fully substitute lead-based ceramics, necessitating long-term technological breakthroughs.

Keywords piezoelectric actuator; relaxor ferroelectric ceramics; strain characteristics; origin