

Strain-induced magnetic domain morphology in Pt/CoFeB/Ir multilayers on BaTiO₃(001)

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Strain exerted by a ferroelectric substrate on a thin ferromagnetic multilayer film with an interfacial Dzyaloshinskii-Moriya interaction is a promising method of manipulating chiral magnetic spin textures. Here, we report on the strain coupling between magnetic domains in a Pt/CoFeB/Ir multilayer and ferroelectric domains in a BaTiO₃(001) substrate. Using wide-field Kerr microscopy to image the demagnetized domain pattern, at room temperature (RT), we observe magnetic stripe domains aligned with the underlying ferroelectric domain pattern. Cooling to 13 K and subsequently warming back up to RT, there are irreversible changes to the ferroelectric domain pattern associated with structural phase transitions, and we find that magnetic stripe domains reorient to match the new ferroelectric domain structure in each phase. These results open a route to electric field manipulation of the domain morphology and chiral spin textures in thin multilayer films.

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I. INTRODUCTION

Some of the most actively researched thin magnetic film systems at present are multilayers with perpendicular magnetic anisotropy (PMA) and an interfacial Dzyaloshinskii-Moriya interaction (iDMI) [1], which contain domain walls (DWs) with a chiral spin texture and, if the iDMI exceeds a critical value, skyrmions [2]. Chiral spin textures are not only of fundamental interest but are also considered potential information carriers in devices for data storage and computing [3]. The advantage of chiral DWs in a device is that their dynamics are more predictable than those of achiral walls [4], while skyrmions, which are nanoscale and particlelike, are proposed for more unconventional computing applications [5,6]. However, to operate any of these devices in an energy-efficient manner, it is increasingly recognized that alternatives to magnetic field or electric current must be used. Magnetoelectric methods that use an electric field to modify exchange coupling or magnetic anisotropy have been demonstrated to manipulate DWs and are considered low energy due to the lack of current flow and therefore minimal Joule heating [7–9]. Strain applied via a voltage-controlled piezoelectric, for example, modifies the magnetic anisotropy and the DW dynamics [10]. In some systems, it has been found that the iDMI itself may be controlled by an electric field, indicating the possibility of electric field tuning of chiral spin textures [11,12].

The strain applied by ferroelectrics also brings about the possibility of local manipulation of the magnetic DWs since ferroelectrics possess local strains associated with polarization at the surface that may be transferred to an adjacent ferromagnetic layer [13]. This has enabled, for example, the manipulation of stripe domains in transcritical Permalloy films, in which magnetic stripe domains with in-plane magnetization emerge above a critical film thickness on the order of 100 nm [14], and the electric field manipulation of DWs in thin films [15–17]. As the strain applied is inherently tied to the crystalline structure, authors of previous work have also used cooling or heating as a way to explore the magnetostrictive coupling in other crystalline phases and at a variety of in-plane lattice strains [18–20].

Authors of the majority of the work on these types of artificial multiferroics have focused on in-plane magnetized thin films. Authors of previous work on films with PMA, such as Ref. [21], used epitaxial layers and focused primarily on the spin-reorientation transition from out-of-plane to in-plane. For modern spintronic applications, such as neuromorphic computing [22], less ordered materials and even amorphous layers are proposed, and a wider range of stimulus and response may be required, so further studies are needed. Annihilation and destruction of skyrmions with a voltage-controlled strain has already been shown [23], which highlights the interest in manipulating spin textures in PMA layers and the need to investigate alternative methods of applying strain in these systems.

Here, we combine magnetic multilayers with PMA and iDMI with BaTiO₃(001) (BTO) single-crystal substrates. We investigate the effect of local strains from the BTO on the domain structure in the multilayer by direct imaging of both the ferroelectric and ferromagnetic domains in a wide-field Kerr microscope. The ferroelectric domain and, hence, the strain pattern are modified by cooling the BTO through its structural

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phase transitions. We find that, similar to the transcritical Permalloy, stripe domains are formed and oriented according to the local strain transfer from the underlying ferroelectric. Thus, we demonstrate the uniform alignment of out-of-plane magnetic stripe domains and chiral DWs by strain over length scales of several micrometers. We also find that temperature cycling introduces irreversible changes in the ferroelectric domain pattern and, thus, in the magnetic domain morphology. Given the significant coupling of the magnetic structure to the substrate strain, which is maintained even across phase transitions, this heterostructure could be a model system for manipulating more complex chiral spin textures by strain.

II. SAMPLE GROWTH

Magnetic multilayers are fabricated on commercially available BaTiO_3 and on thermally oxidized silicon substrates (SiOx) for comparison, using DC magnetron sputtering with a base pressure below 10^{-7} mbar and an argon growth pressure of 2.4 mTorr. The BTO substrates display ferroelectric domains as received, and no poling procedure was carried out.

As a magnetic multilayer, we use $\text{Ta}(3\text{ nm})/[\text{Pt}(2.3\text{ nm})\text{Co}_{68}\text{Fe}_{22}\text{B}_{10}(0.7\text{ nm})/\text{Ir}(0.5\text{ nm})]_n/\text{Pt}(2.3\text{ nm})$, where the repeat number $n = 2$. This is an analog of the multilayers in previous reports in which skyrmions were the focus [24–26]. The ferromagnet is amorphous CoFeB, originally chosen to try and avoid the problems of pinning at grain boundaries. Pt and Ir are chosen because they give rise to iDMI of opposite signs either side of the ferromagnet and thus a large net DMI [27]. A repeat number $n = 1$ produces a multilayer film with large perpendicular anisotropy, where the domains are several tens of micrometers across, difficult to stabilize and inconvenient for imaging, while $n = 2$ adds to the sample volume, reducing the effective anisotropy enough to produce stable micrometer-scale domains at zero field after demagnetizing the sample. Adding repeats so that $n > 2$ yields samples with increasingly small domains, which can still be imaged by Kerr microscopy, but since strain is exerted from the bottom of the stack, we wish to keep the overall multilayer thickness to a minimum to ensure that strain is present throughout the sample volume. Furthermore, with sufficiently thin magnetic films, it is possible for polarized light to penetrate and image the ferroelectric order in the substrate as well as the ferromagnetic order in the film on top, so $n = 2$ is a good compromise.

III. DOMAIN IMAGING

Imaging is performed using an Evico wide-field Kerr microscope with an optical cryostat attachment. Samples are cooled to a base temperature of 4 K, and images are taken as the sample is warmed to room temperature (RT). For the sample grown on BaTiO_3 , we obtain ferroelectric contrast in the unprocessed images owing to the birefringence of the ferroelectric domains, which originates from the elongation of the lattice associated with the ferroelectric polarization. This has been observed previously in the literature [13,16], and an example is shown for these samples in Fig. S1 in the Supplemental Material [28]. Magnetic contrast is obtained by using an image subtraction technique to remove nonmagnetic contributions, including ferroelectric contrast, and then using

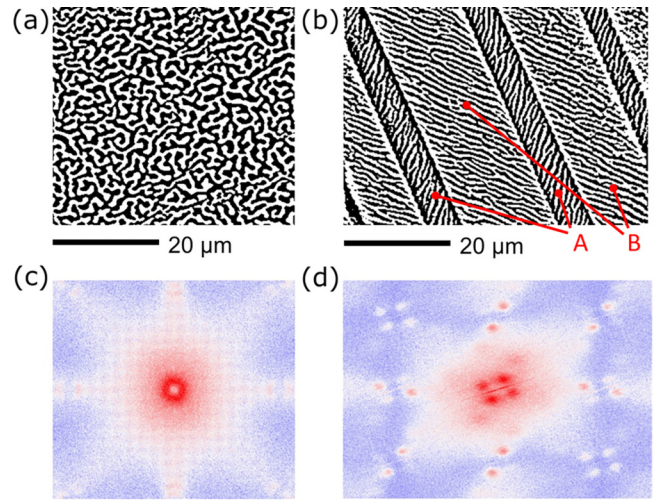


FIG. 1. Demagnetized magnetic domains in $[\text{Pt}/\text{CoFeB}/\text{Ir}]_2$ multilayers grown on (a) SiOx and (b) BTO. A and B correspond to narrow and wide ferroelectric stripe domains in the BTO, respectively. Two-dimensional Fourier transform for the magnetic domain pattern on (c) SiOx, displaying rotational symmetry, and (d) BTO, where each ferroelectric stripe region A and B contributes one preferential axis with complementary poles of periodicity.

a filter convolve and threshold algorithm to enhance the contrast (Figs. S2 and S3 in the Supplemental Material [28]).

To produce the demagnetized domain images, samples are demagnetized by an AC procedure in which a large out-of-plane magnetic field of 30–50 mT is applied to saturate the sample and the magnitude of the field is decreased in an oscillatory manner over 120 s with a frequency of 0.5 Hz.

A. Room temperature

Figure 1(a) shows the demagnetized domain pattern at RT for $[\text{Pt}/\text{CoFeB}/\text{Ir}]_2$ deposited on a conventional thermally oxidized silicon wafer (SiOx). A maze pattern of up/down magnetized domains typical for this type of multilayer is formed, where the characteristic domain width is determined by minimization of the stray field. This demagnetized domain pattern has been modeled by previous authors [29] who included the iDMI of the multilayer, and an equation was established for the DW energy density in terms of the magnetization M_s and domain period d (which is the length scale conventionally used in modeling, where $d = 2 \times$ domain width). Measuring the domain period directly from the image yields $d = (2.08 \pm 0.01) \mu\text{m}$.

Comparing with the demagnetized domain pattern at RT for the same multilayer deposited on single-crystal BTO(001) in Fig. 1(b), the maze domain pattern is no longer present, and instead, the magnetic domains are striped. This domain morphology has previously been observed in similar PMA multilayer structures of $\text{Pt}/\text{CoFeB}/\text{MgO}$ [30], with a maze-to-stripe domain structure transition resulting from a change in thickness of the Pt seed layer. The Pt thickness used here is below this transition, evidenced by the lack of any preferred orientation in our sample grown on SiOx. The symmetry is broken only by the strain induced by the BTO substrate. Larger scale (10–20 μm) stripe contrast is also present in

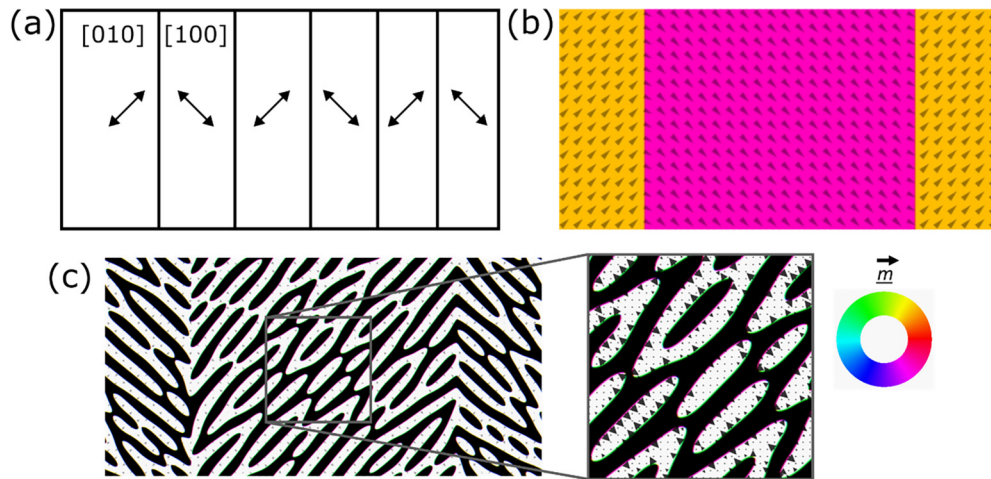


FIG. 2. (a) Schematic representation of lattice elongations within ferroelectric stripe domains. (b) In-plane anisotropy directions used in our micromagnetic simulations for a three-stripe snapshot. (c) Micromagnetic demagnetization pattern obtained with an in-plane uniaxial anisotropy factor, alongside a blown-up region showing the chirality of the DWs aligning with the in-plane positive anisotropy. Color wheel represents the direction of in-plane magnetization.

Fig. 1(b), corresponding to ferroelectric domains in the BTO substrate. Where there is a ferroelectric domain boundary, the orientation of the magnetic stripe domains changes, representing a pattern transfer between the two materials.

A similar change in the orientation of stripe domains has been reported previously in Permalloy films deposited on BTO(001) [14]. However, in that work, the stripe contrast represented a magnetization that was in-plane, while in Pt/CoFeB/Ir, the magnetization within the domains was out-of-plane (and indeed, our imaging geometry is only sensitive to out-of-plane magnetization). In addition, our films are much thinner, with a total thickness of 7 nm including nonmagnetic material, whereas for the Permalloy films, a thickness of 100 nm was required to see the same effects.

At RT, the BTO substrate is in the tetragonal phase [31,32], which means that the lattice is elongated along a $\langle 001 \rangle$ crystal direction, and polarization occurs due to the relative shift of the Ti ion and the negatively charged oxygen octahedron. The lattice can be elongated along three different axes, leading to six possible orientations of ferroelectric polarization. Here, 90° ferroelectric DWs form in our (001)-oriented substrate, separating $a1$ domains with polarization along [100] from $a2$ domains with polarization along [010] [Fig. 2(a)]. Since the BTO crystal is elongated along the polarization axis, we can expect uniaxial tensile stress in this direction alternating by 90° at each ferroelectric domain boundary oriented at 45° to the ferroelectric stripe length [33].

Uniaxial tensile stress in the plane of a magnetic multilayer with perpendicular anisotropy generates an in-plane easy axis [34]. As the magnetic DWs have an in-plane component of magnetization and are homochiral [35], they will orient preferentially with respect to this easy axis, and we expect that, for multiple DWs with in-plane components oriented in the same direction, the minimum energy state occurs when the DWs align in parallel with their internal magnetization orthogonal to the stress (due to the negative magnetostriction [36]), and thus, the stripe domains themselves orient with their length parallel to the stress.

By taking a Fourier transform of the magnetic domains [Fig. 1(d)], we find that the resulting image representing the periodicity in the domain structure consists of peaks corresponding to the magnetic stripe orientation within individual ferroelectric domains. Focusing on just the central part of the Fourier transform with the highest intensity, we fit the center of the circle and find that the angle between stripe periodicities is 97° , agreeing reasonably well with the 90° rotation of lattice elongation within the BTO. The disagreement may be explained by the competition between in-plane and out-of-plane anisotropies and the relaxation of stress through the multilayer stack. This is in contrast with the same multilayer grown on SiOx [Fig. 1(c)], where there is no in-plane anisotropy, and so the DWs have no preferential orientation, leading to a Fourier transform that is rotationally symmetric.

The domain period d may also be obtained from analysis of the Fourier transform. The average d for magnetic stripes confined within the narrow ferroelectric domains [labeled A in Fig. 1(b)] is $(1.1 \pm 0.2) \mu\text{m}$ and for magnetic stripes confined to the broader ferroelectric domains (labeled B) is $(1.2 \pm 0.3) \mu\text{m}$. These values are smaller than the $d = (2.08 \pm 0.01) \mu\text{m}$ measured for the multilayer on SiOx. Since a smaller d means a lower DW energy density, which is generally linked to a smaller effective magnetic anisotropy K_{eff} , we can expect that the anisotropy for the multilayer on BTO is less than for the same material deposited on SiOx. Within experimental error, there is no difference in magnetic domain width—and thus, we would infer no difference in magnetic anisotropy—between the A and B regions.

To show that magnetic stripe domains can occur due to strain-induced in-plane anisotropy, we perform micromagnetic simulations using MUMAX3 [37] to evaluate the effect of an in-plane uniaxial anisotropy on the Pt/CoFeB/Ir multilayer with parameters obtained from experiments. The simulation is divided into three regions [Fig. 2(b)], corresponding to three ferroelectric stripe domains where the vector of the strain-induced component of anisotropy rotates by 90° from region to region, corresponding to the change in direction of the

lattice elongation. The total simulation size is $32 \times 8 \mu\text{m}$ with a film thickness of 2 nm and a central stripe width of $20 \mu\text{m}$. The in-plane and out-of-plane anisotropies are set up as two separate energy terms, as has been done in previous work in which competing in-plane and out-of-plane anisotropies were investigated [38].

The simulation is initialized in a random configuration, and the energy is minimized to mimic the effects of demagnetization on the domain structure. A micromagnetic cell size of $8 \times 8 \times 2 \text{ nm}$ was used with periodic boundary conditions in the x and y directions. Micromagnetic parameters of saturation magnetization $M_{\text{sat}} = 1 \times 10^6 \text{ A/m}$, exchange stiffness $A_{\text{ex}} = 4 \times 10^{-11} \text{ J/m}$, interfacial DMI $D_{\text{ind}} = 0.9 \text{ mJ/m}^2$, and anisotropy $K_{\text{me}} = 30 \text{ kJ/m}^3$ in-plane and $K = 680 \text{ kJ/m}^3$ out-of-plane were used. The magnetoelastic anisotropy K_{me} induced by the strain takes a value in line with previous reports [39,40].

Figure 2(c) shows the result, with a micromagnetic domain structure that matches well with the domain structure observed in Fig. 1(b), where the magnetic stripes are aligned at approximately 90° between ferroelectric domains. The orientation of the stripe length is orthogonal to the direction of the easy-axis anisotropy, with the in-plane chiral component of the DW preferring to align along the direction of in-plane strain. In the case of negative magnetostriction, we would therefore expect a hard axis to be imprinted parallel to the ferroelectric polarization, creating an easy axis orthogonal to the polarization, and so the chirality of the DW would be oriented orthogonal to the strain. The stripe domains observed in the simulations are topologically similar to stripe skyrmions observed in simulations elsewhere [41,42]. The DWs on either side of the stripe have opposing in-plane components, and there is a topological charge at each end of the stripe where the curvature is largest [Fig. 2(c)].

B. Cryogenic temperatures

When BTO is cooled from RT to $<180 \text{ K}$, it undergoes phase transitions from tetragonal (T) to orthorhombic (O) to rhombohedral (R) structure, with consequent changes in polarization within ferroelectric domains and strain at the interface with the magnetic multilayer. The lattice constant decreases in a stepwise fashion at the transition temperatures [43,44]. The phase transitions exhibit a small temperature hysteresis, and here, we give the transition temperatures on warming, to match the magnetic characterization we have performed as a function of temperature from 13 to 320 K. In the R phase, below 180 K, the a and b lattice constants are $4.004 \text{ \AA}$, and they increase abruptly in the O phase, just above 180 K to $4.018 \text{ \AA}$. On entering the T phase at approximately 280 K, the a lattice constant increases abruptly to $4.035 \text{ \AA}$, while b decreases to $3.991 \text{ \AA}$. For a film deposited onto the a - b plane at RT, due to the different a and b , we thus expect a tensile stress in the a direction, as discussed above. Cooling to the O phase, there is a compression in a and tension in b , i.e., 90° change in direction of tensile stress; and cooling further to the R phase, there is further compression in a and relaxation of the tension in b . Thus, cooling from T to R phase, we would expect a 90° change in tensile stress and therefore in

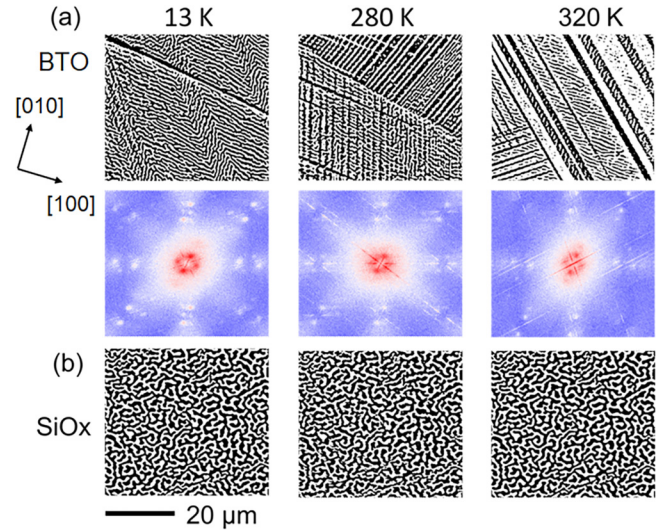


FIG. 3. (a) Temperature dependence of demagnetized domain patterns in $[\text{Pt}/\text{CoFeB}/\text{Ir}]_2$ multilayers grown on BTO and the corresponding fast Fourier transform (FFT) images. (b) Maze domains in the control sample on SiOx.

the magnetic stripe orientation and potentially other features corresponding to local changes in stress.

Figure 3(a) shows the magnetic stripe domains at three temperatures corresponding to the BTO R phase (13 K), O phase (280 K), and T phase (320 K). Fourier transforms indicate that there are approximately 90° changes in stripe orientation at ferroelectric domain boundaries. This is clear in the R and T phases where the magnetic domains are (mostly) of smaller width than the ferroelectric domains. In the O phase, the situation is less clear because the ferroelectric and magnetic domain widths are similar, and there is an almost one-for-one correspondence in parts of the image, where the ferroelectric domains lead directly to a magnetic stripe pattern that changes angle by approximately 45° between upper and lower parts of the image, leading to a splitting of the peaks in the Fourier transform. We also note that the control sample grown on SiOx shows no significant changes with temperature and that the domain periodicity has no preferential direction.

As the sample is cooled from RT [Fig. 1(b)] to 13 K [Fig. 3(a)], the magnetic stripe domains remain in the same principal directions at 90° to each other, as at RT. This is similar to the result of Lahtinen *et al.* [18] who found that, for $\text{CoFe}/\text{BTO}(001)$, the magnetic stripe pattern set during the growth at RT was preserved upon cooling and heating through all BTO phase transitions. Warming up to the O phase (280 K), the ferroelectric domains are a similar size to the magnetic domains, except in a few regions where the ferroelectric domains are wide enough to permit short magnetic stripes in one of the original, growth-induced directions. In the T phase (320 K) reached after one temperature cycle, the width, spacing, and regularity of ferroelectric domains has changed as compared with the as-grown state [Fig. 1(b)]. There is even a region (lower left in the image at 320 K) where the orientation of the ferroelectric domains has changed by nearly 90° . The irreversible changes in BTO domains after cooling and heating cycles were also seen before [18].

Nevertheless, the magnetic stripe domains remain in similar orientations to those in Fig. 1(b) (confined to narrow A or broad B ferroelectric domains).

C. Domain period

As discussed in previous work [29,30,35], the domain period can be used to calculate the DW energy, which is itself intrinsically related to the exchange stiffness, anisotropy, and iDMI. We refer to Eqs. (2) and (3) in Ref. [35] for a more detailed discussion and derivation of this. Using images obtained of the demagnetized domains as a function of temperature, we perform a Fourier transform to obtain the domain period. For the sample grown on BTO, we determine the center of the pairs of frequency poles, then determine the radial distance and convert this into the domain period. With the sample grown on SiOx, there is no preferred direction, so we simply determine the radial distance to the center maximum of the frequency distribution. The peak position is obtained by fitting to a Gaussian function and converted into the domain period.

Measurements of the effective PMA are performed using a Quantum Design MPMS 3 SQUID-VSM magnetometer. Hysteresis loops are performed at varying temperatures in an in-plane configuration, and the saturation magnetization and anisotropy field are extracted from the resulting loops. The effective perpendicular anisotropy is then calculated using the previously measured value of saturation magnetization as

$$K_{\text{eff}}(T) = \frac{1}{2}\mu_0 H_K(T) M_S(T). \quad (1)$$

The $M_S(T)$'s for both datasets are shown in Fig. S4 in the Supplemental Material [28].

Figure 4 shows the domain period as a function of temperature. The domain period of the multilayer grown on SiOx remains stable across the temperature range with an average value of $(2.08 \pm 0.01) \mu\text{m}$. For the sample grown on BTO, we distinguish between magnetic stripe domains that produce frequency poles in the Fourier transform that are oriented approximately on an east-west axis (E) and those that yield frequency poles that are oriented approximately on a north-south axis (N), corresponding to the two magnetic stripe orientations. At RT in the as-grown state [Fig. 1(b)], there is no difference within experimental uncertainty in the domain period between magnetic domains in the A positions (yielding E, where $d = (1.1 \pm 0.2) \mu\text{m}$) and those in the B positions (yielding N, where $d = (1.2 \pm 0.3) \mu\text{m}$). However, once the sample is cooled to 13 K and then warmed back up to RT, there is a difference in the E and N domain periods. This is a mark of the irreversibility of the ferroelectric domain pattern after a temperature cycle and its effect on the magnetic multilayer.

In the *R* phase, the d calculated from the E and N poles is not significantly different. In the *O* phase, the d calculated from the E poles is artificially high because the magnetic domains obtain their periodicity from the ferroelectric domains, while the d calculated from the N poles is similar to that in the *R* phase. Returning to the *T* phase, the N poles yield a smaller d than those in the *R* or *O* phase, indicating a reduction of the magnetic DW energy here. The E poles are less distinct than in the initial image at RT [Fig. 1(d)] because there are few stripes in the corresponding orientation in the image.

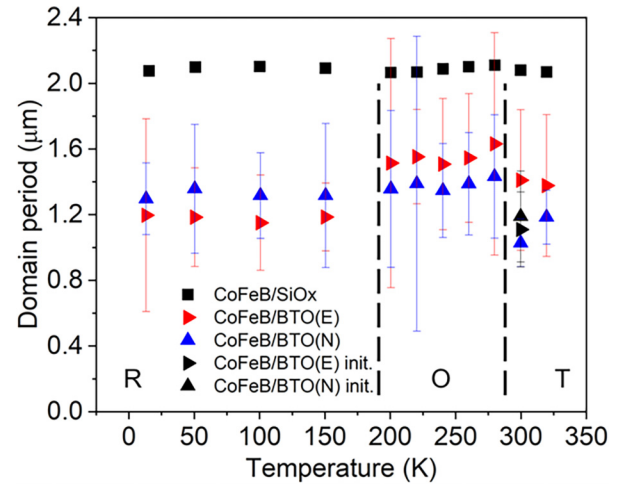


FIG. 4. Temperature dependence of the magnetic domain period for $[\text{Pt}/\text{CoFeB}/\text{Ir}]_2$ multilayers grown on SiOx and BTO. For simplicity, the samples are referred to as CoFeB/SiOx and CoFeB/BTO. The dashed lines are the phase transitions of the BTO, with the rhombohedral (*R*), orthorhombic (*O*), and tetragonal (*T*) phases indicated. E and N correspond to frequency poles in the FFT images oriented along approximately east-west and north-south axes, respectively, while init. corresponds to the initial, as-grown state shown in Fig. 1(b). Error bars for the multilayers grown on BTO originate from the full width at half maximum of the Gaussian peak fitted to the frequency poles in the FFT. Error bars for the multilayers grown on SiOx are smaller than the data points.

We draw some general conclusions about the domain period as a function of temperature. First, since the domain period at all temperatures is larger for the multilayer on SiOx, we expect the DW energy density and consequently the anisotropy K_{eff} to be larger than for the sample grown on BTO at all temperatures. Figure 5 depicting $K_{\text{eff}}(T)$ shows that this is indeed the case. Second, although the size of the error bars in Fig. 4 prevents any strong statement, the RT data supports the observation from images in Fig. 3(a) that a temperature cycle produces an irreversible change not only

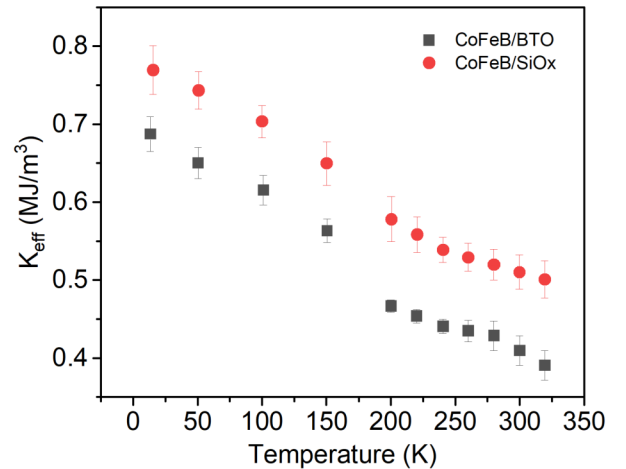


FIG. 5. Effective PMA K_{eff} for equivalent samples of $[\text{Pt}/\text{CoFeB}/\text{Ir}]_2$ multilayers grown on BTO and SiOx.

in the ferroelectric domain pattern but in the magnetic domain morphology as well. This could lead to adjustments in DW energy and K_{eff} depending on the temperature and the nature of the underlying ferroelectric domains. As the ferroelectric domains impart a strain that is at most 1.1% and the strain is unlikely to be fully transferred [39,40], the effect on the iDMI is expected to be on the order of 1% also. Kim *et al.* [45] deduced a relation $D_{\text{at}} \sim 1/z^2$, where D_{at} is the interatomic antisymmetric exchange energy, and z is the distance between magnetic and heavy metal atoms perpendicular to the plane. Approximating a simple cubic lattice where $D = D_{\text{at}}/a t_{\text{eff}}$, with lattice constant a and magnetic layer thickness t_{eff} , and assuming a Poisson ratio of 1, a $\sim 1\%$ in-plane strain gives $\Delta D/D \sim 2\%$. This change in iDMI would be below the sensitivity of most experiments. Using the same method as in Ref. [35], we calculate $D = (1.2 \pm 0.5) \text{ mJ/m}^2$ and $(1.3 \pm 0.5) \text{ mJ/m}^2$ at RT for the multilayers on BTO and SiOx, respectively, the same within the experimental uncertainty. On one hand, while K_{eff} is significantly different between the two substrates, as reflected in the different domain periods, D remains the same. On the other hand, the iDMI is important in that it causes the magnetic DWs to be homochiral, i.e., it gives them Néel character, so that when there is in-plane anisotropy as in the case of the BTO, the energy of DWs in some orientations is lowered, hence the formation of stripes out of the (isotropic) maze domain state.

IV. CONCLUSIONS

In summary, we have found that it is possible to couple the magnetic domain structure of a PMA multilayer to the strain in a BTO substrate. Our domain imaging has shown that the demagnetized magnetic domains we obtain are spatially dependent on the ferroelectric domains in the BTO. The magnitude of the strain is large enough that it breaks the radial symmetry of the domain periodicity, creating long stripes that, during deposition at RT, preferentially orient parallel to the direction of ferroelectric polarization due to the negative magnetostriction of the film. The magnetic domain structure is sensitive to the change in orientation of the strain that occurs

across ferroelectric domain boundaries, leading to magnetic stripes oriented at approximately 45° to the ferroelectric domain boundaries.

We find that, as the BTO is cooled through its structural phase transitions, the ferroelectric and magnetic domain morphology change due to the changes in the lattice parameters. Irreversibility is expected because, whenever a phase boundary is crossed, the domain history is lost, except for any effect that defects may have on the energetics of domain formation in the ferroelectric.

BTO has previously been used as a substrate for CoFe [13] and Fe [15] films with in-plane magnetization, where ferroelectric domain patterns are imprinted into the ferromagnet, and ferroelectric and ferromagnetic DWs are coupled. An electric field has been applied in these cases to write or erase ferromagnetic domain patterns and promote magnetic DW motion. BTO domain switching by electric field has also been demonstrated to drive changes in the magnetic anisotropy of coupled Cu/Ni multilayers [21] and Ni films [46], leading to reversible magnetization reorientation from in-plane to out-of-plane. Based on these prior studies, we can infer that an electric field applied to the BTO will modify the magnetic domain morphology in our magnetic multilayers too, provided that the temperature is not varied too much that a structural phase transition occurs. This is a potential avenue for manipulating chiral spin textures such as DWs and skyrmions.

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DATA AVAILABILITY

The data that support the findings of this article are openly available [47].

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