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XMCD study of magnetism and valence state in iron-substituted strontium titanate

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Room-temperature ferromagnetism was characterized for thin films of SrTi0.6Fe0.4O3 grown by pulsed laser deposition on SrTiO3 and Si substrates under different oxygen pressures and after annealing under oxygen and vacuum conditions. X-ray magnetic circular dichroism demonstrated that the magnetization originated from Fe2+ cations, whereas Fe3+ and Ti4+ did not contribute. Films with the highest magnetic moment (0.8 μB per Fe) had the highest measured Fe2+:Fe3+ ratio of 0.1 corresponding to the largest concentration of oxygen vacancies (δ = 0.19). Postgrowth annealing treatments under oxidizing and reducing conditions demonstrated quenching and partial recovery of magnetism respectively, and a change in Fe valence states. The study elucidates the microscopic origin of magnetism in highly Fe-substituted SrTi1-δFeOδ perovskite oxides and demonstrates that the magnetic moment, which correlates with the relative content of Fe2+ and Fe3+, can be controlled via the oxygen content, either during growth or by postgrowth annealing.

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I. INTRODUCTION

Transition metal oxides are a versatile class of materials with collective charge and spin phenomena including ferro/antiferromagnetism, superconductivity, ferroelectricity, colossal magnetoresistance, and multiferroicity [1–10]. The perovskite structure offers great compositional flexibility owing to the ability of the cation sites to accommodate a range of ionic sizes and cation valencies, permitting fine control of electronic and magnetic properties [3,4,8]. A key aspect of the behavior is the role played by defects and vacancies, both of which can be intrinsic sources of conductivity, magnetism, and other properties [11–14].

Fe-substituted SrTiO3 (STF) is a material where room-temperature ferromagnetism is introduced into a nonferromagnetic SrTiO3 (STO) host [15]. Structurally, Fe replaces Ti in the B site of the perovskite structure as schematically shown in Fig. 1(a). Two regimes may be distinguished: a dilute level of substitution in which nearest-neighbor (Fe–O–Fe) cation configurations are rare, and a highly substituted regime in which nearest-neighbor configurations are common and exchange interactions between Fe cations become important. In bulk STF, the Fe ions typically exist with an Fe3+ or Fe4+ oxidation state [16,17]. However, thin-film growth such as by pulsed laser deposition (PLD) under low oxygen pressures can lead to the presence of Fe2+ [18–22] as observed by surface sensitive techniques, due to the kinetically limited growth process which results in a high concentration of oxygen vacancies that are compensated by changes to the cation valence state [18]. Although x-ray photoelectron spectroscopy (XPS) studies of Fe-substituted STO have been conducted [15,20,21], there is no study of STF using both x-ray absorption (XAS) and x-ray magnetic circular dichroism (XMCD) which can provide a detailed microscopic mechanism for the emergence of magnetism with bulk sensitivity. XPS and XMCD studies on binary oxides such as magnetite [23] exist; however, these materials have qualitatively different structures compared to STF. A systematic investigation of STF with spectroscopic methods will play a key role in unveiling the mechanism responsible for magnetism.

STF and related materials such as Co-substituted STO have demonstrated room-temperature ferromagnetism when grown under low oxygen partial pressures [12,15,20,21,24,25]. PLD-grown STF thin films on both Si and STO substrates exhibited room-temperature ferromagnetism when deposited at pressures below 5 μTorr, in contrast to films grown at higher oxygen pressures [15,26]. The former study showed higher magnetization for films grown on Si as compared to those on SrTiO3 for nominally identical deposition conditions, and found the presence of Fe3+ but did not detect Fe2+ through Mössbauer spectroscopy and XPS. Laser-irradiated undoped STO single crystals and thermally annealed Co-doped (La, Sr)TiO3 also exhibited ferromagnetism, which was suggested to be related to oxygen vacancies [12,27]. In these materials, oxidizing treatments lowered the magnetic moment while reducing treatments increased it. Density functional theory calculations predict magnetism in SrTiO3, Fe-substituted SrTiO3, and Co-substituted SrTiO3 that can arise from Ti vacancies, oxygen vacancies, or substituted transition metal cations [15,28,29].
Here, we report on the evolution of Fe valence state and magnetic moment in highly Fe-substituted STO films which are grown at different oxygen pressures and on different substrates. By a combination of x-ray diffraction (XRD), transmission electron microscopy (TEM), vibrating sample magnetometry (VSM), XAS [30,31], and XMCD [31–33], we uncover the link between room-temperature magnetization and the electronic configuration of Fe and Ti ions as a function of oxygen vacancy concentration. The magnetization increases with the Fe$^{2+}$:Fe$^{3+}$ ratio, and therefore with the oxygen vacancy concentration. We demonstrate that the magnetism can be quenched on annealing in an oxidizing environment and partly restored upon annealing in a reducing environment. The experimental evidence identifies the origin of magnetism in this class of materials as arising from the presence of Fe$^{2+}$ in an otherwise Fe$^{3+}$ system, which modifies the dominant antiferromagnetic Fe$^{3+}$ interactions. These results extend our understanding of the source of magnetism in Fe-substituted SrTiO$_3$, and are helpful in facilitating the design of oxide materials whose magnetic properties can be manipulated, e.g., by annealing or electrochemical means.

II. EXPERIMENTAL DETAILS

STF films were deposited from a SrTi$_0.6$Fe$_{0.4}$O$_3$ target on single-crystal (100) Si substrates with a native oxide layer and on single-crystal (100) SrTiO$_3$ substrates using a Neo-cera pulsed laser deposition (PLD) system with a KrF laser (248 nm) and a fluence of 1.4 J/cm$^2$. A previous study [24] using the same target and nominally the same deposition conditions indicated that films have less Fe than the target composition, i.e., a 40% Fe target yielded films of composition SrTi$_{0.65}$Fe$_{0.35}$O$_3$. Measurement of a sample of this study grown on Si at 3 μTorr gave a ratio of Ti : Fe = 67 : 33. The substrate temperature was 650 °C for all depositions, and the base and growth pressures varied between 1 and 4 μTorr. Base pressure refers to the pressure in the chamber before the substrate heater was turned on, while growth pressure refers to the pressure in the chamber at the start of deposition after the sample reached the deposition temperature. All depositions were made on 10-mm × 10-mm × 0.5-mm-thick substrates. Annealing treatments carried out in the PLD chamber consisted of (1) oxidation at a substrate temperature of 650 °C

FIG. 1. (a) Model of Fe-substituted SrTiO$_3$ demonstrating cubic perovskite structure, oxygen octahedra surrounding the B sites, and cation substitution of Ti with Fe. (b)–(e) XRD ω-2θ scans for SrTi$_{60}$Fe$_{40}$O$_{3-δ}$ samples on (b) Si and (c)–(e) SrTiO$_3$ substrates, subjected to different base and growth pressures (b)–(d), or annealing treatments (e). XRD ω-2θ scans at 40° < 2θ < 50° (d) are left of their corresponding wide-range scans (c). (f) TEM image of 1–1.2 μTorr (STO) with Fe elemental map showing homogeneous distribution of Fe.
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TABLE I. Base pressures, growth pressures, thicknesses, annealing conditions, and unit-cell volume from XRD and magnetic moment, fraction of Fe\(^{2+}\), and \(\delta\) for the Sr\(_{1-x}\)Fe\(_{x}\)O\(_3\) thin films on Si and on STO.

<table>
<thead>
<tr>
<th>Sample designation</th>
<th>Substrate</th>
<th>Base pressure, (\mu)Torr</th>
<th>Growth pressure, (\mu)Torr</th>
<th>Thickness, nm</th>
<th>Annealing conditions</th>
<th>Unit-cell volume, (\AA^3)</th>
<th>Magnetic moment (VSM), (\mu_B/Fe)</th>
<th>Fraction Fe(^{2+}) (XAS)</th>
<th>(\delta) (XAS)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1-1.2 (\mu)Torr (Si)</td>
<td>Si</td>
<td>1</td>
<td>1.2</td>
<td>129</td>
<td>n/a</td>
<td>60.85 (\pm) 0.87</td>
<td>0.81</td>
<td>8 (\pm) 3%</td>
<td>0.189 (\pm) 0.003</td>
</tr>
<tr>
<td>5-1.2 (\mu)Torr (Si)</td>
<td>Si</td>
<td>5</td>
<td>1.2</td>
<td>118</td>
<td>n/a</td>
<td>60.03 (\pm) 0.86</td>
<td>0.65</td>
<td>8 (\pm) 3%</td>
<td>0.189 (\pm) 0.003</td>
</tr>
<tr>
<td>2-2 (\mu)Torr (Si)</td>
<td>Si</td>
<td>2</td>
<td>2</td>
<td>124</td>
<td>n/a</td>
<td>60.15 (\pm) 0.86</td>
<td>0.43</td>
<td>8 (\pm) 3%</td>
<td>0.189 (\pm) 0.003</td>
</tr>
<tr>
<td>3-3 (\mu)Torr (Si)</td>
<td>Si</td>
<td>3</td>
<td>3</td>
<td>120</td>
<td>n/a</td>
<td>60.79 (\pm) 0.87</td>
<td>0.51</td>
<td>7 (\pm) 3%</td>
<td>0.187 (\pm) 0.003</td>
</tr>
<tr>
<td>4-4 (\mu)Torr (Si)</td>
<td>Si</td>
<td>4</td>
<td>4</td>
<td>173</td>
<td>n/a</td>
<td>60.10 (\pm) 0.86</td>
<td>0.28</td>
<td>5 (\pm) 3%</td>
<td>0.184 (\pm) 0.003</td>
</tr>
<tr>
<td>1-1.2 (\mu)Torr (STO)</td>
<td>SrTiO(_3)</td>
<td>1</td>
<td>1.2</td>
<td>129</td>
<td>n/a</td>
<td>61.34 (\pm) 0.35</td>
<td>0.79</td>
<td>10 (\pm) 3%</td>
<td>0.192 (\pm) 0.003</td>
</tr>
<tr>
<td>5-1.2 (\mu)Torr (STO)</td>
<td>SrTiO(_3)</td>
<td>5</td>
<td>1.2</td>
<td>118</td>
<td>n/a</td>
<td>60.45 (\pm) 0.19</td>
<td>0.074</td>
<td>0 (\pm) 3%</td>
<td>0.175 (\pm) 0.003</td>
</tr>
<tr>
<td>2-2 (\mu)Torr (STO)</td>
<td>SrTiO(_3)</td>
<td>2</td>
<td>2</td>
<td>124</td>
<td>n/a</td>
<td>60.89 (\pm) 0.14</td>
<td>0.11</td>
<td>0 (\pm) 3%</td>
<td>0.175 (\pm) 0.003</td>
</tr>
</tbody>
</table>

1-1.5 \(\mu\)Torr (STO):

As-grown SrTiO\(_3\) | 1 | 1.5 | 102 | n/a | 61.45 \(\pm\) 0.15 | 0.84 | 10 \(\pm\) 3\% | 0.192 \(\pm\) 0.003 |

Oxidized SrTiO\(_3\) | 1 | 1.5 | 102 | 160 Torr/2 h | 61.59 \(\pm\) 0.67 | 0.00 | 0 \(\pm\) 3\% | 0.175 \(\pm\) 0.003 |

Reduced SrTiO\(_3\) | 1 | 1.5 | 102 | 160 Torr/2 h + 1.5 \(\mu\)Torr | 61.24 \(\pm\) 0.18 | 0.66 | 6 \(\pm\) 3\% | 0.185 \(\pm\) 0.003 |

and pressure of 160 Torr of 100% O\(_2\) for 2 h, followed by (2) reduction under vacuum at a substrate temperature of 650 °C and starting pressure of 1.5 \(\mu\)Torr for 45 min.

Results from nine selected samples are discussed in this paper. Samples are referenced by their base and growth pressures as “x-y \(\mu\)Torr” where x is the base pressure in \(\mu\)Torr and y is the growth pressure in \(\mu\)Torr. The films subjected to the annealing cycle are referred to as “as-grown,” “oxidized,” “reduced,” representing the sample after annealing in oxygen, and “reduced,” representing the sample after oxidation followed by annealing in vacuum. The base pressures, growth pressures, thicknesses, and annealing treatment of the films are shown in Table I. For the first six samples, each pair of films on Si and SrTiO\(_3\) was grown during the same deposition process. All thicknesses were measured by profilometry.

XRD \(\omega\)-29 scans were performed using a Rigaku Smartlab Multipurpose Diffractometer with the Smartlab Guidance data collection program and an incident-beam Ge (022) double bounce monochromator. Reciprocal space mapping (RSM) scans were collected on a Bruker D8 High Resolution Diffractometer. For the films on STO, the unit-cell volume was calculated on the basis that the in-plane lattice parameter matched that of the STO as shown by the RSM data (Supporting Information [34]) and in Refs. [15,21], and taking the out-of-plane lattice parameter from the highest intensity peak in the XRD data. For films on Si we assume a cubic unit cell, based on earlier work showing a c/a ratio of 0.990–0.998 for STF/Si [15]. TEM samples were prepared with a Helios Nanolab 600 Dual Beam Focused Ion Beam Milling System. TEM images were collected with a JEOL 2010 Advanced High Performance TEM at 200 kV and elemental mapping was performed with a JEOL 2010 FEG Analytical Electron Microscope at 250 kV. Magnetic hysteresis loops were measured using a Digital Measurement System 7035B vibrating sample magnetometer (VSM). Composition of 35% Fe and 65% Ti was confirmed by electron microprobe analysis.

III. RESULTS AND DISCUSSION

A. Experimental results

X-ray diffraction \(\omega\)-29 scans between 20° and 80° are shown in Figs. 1(b)–1(d). All films have a perovskite structure with no visible secondary phases. Polycrystalline perovskite films formed on the Si substrates in Fig. 1(b) and single-crystal films on the STO substrates in Figs. 1(c) and 1(d). The relative intensities of the peaks for STF on Si correspond to those of the polycrystalline reference for bulk cubic SrTiO\(_3\),
indicating no preferred texture. No metallic phases, iron oxides, titanium oxides, or strontium oxides were identified for the as-grown films or for the annealed films shown in Fig. 1(e). Multiple or asymmetrical (200) film peaks were seen in the ω-2θ scans which is indicative of a distribution of out-of-plane lattice parameters. Consistent with this result, RSM of films on STO showed that the film peak was broadened along q_x, but the film peak remained matched to the substrate peak along q_y, indicating a coherent interface with in-plane lattice matching to the substrate. TEM imaging of 1–1.2 μTorr (STO) showed the presence of vertical planar defects in the single-crystalline film; however, elemental mapping proved that the component elements including Fe were homogeneously distributed. Additional TEM images can be found in the Supplemental Material [34].

For oxygen-deficient STF, the presence of oxygen vacancies is compensated for by lowering the oxidation state of the cations, leading to higher ionic radii and to chemical expansion of the unit cell compared to that of Sr(Ti,Fe)O_3 [15,21,35]. Unit-cell volumes are shown in Table I. A sample with the highest magnetic moment reported in Ref. [15]. Interestingly, the 5–1.2-μTorr (STO) sample had a lower unit-cell volume than the 1–1.2 μTorr (STO) suggesting a role of base pressure in determining the film structure.

VSM hysteresis curves, XAS spectra, and XMCD hysteresis curves and spectra of the 1–1.2-μTorr (STO) and 1–1.2-μTorr (Si) films of Figs. 1(a) and 1(b) are compared in Fig. 2. It is apparent from VSM that both samples exhibit room-temperature ferromagnetic behavior with an out-of-plane easy axis. The anisotropy in STF/STO has been attributed to magnetoelastic effects [21,24,25,36].

The two films exhibit a similar saturation magnetization (M_s) and remanence (M_r) despite their microstructural differences, with M_s = 43 emu/cm^3 (0.81 μB/Fe) for STF/Si, M_s = 42 emu/cm^3 (0.79 μB/Fe) for STF/STO, and M_r = 33 emu/cm^3 (0.62 μB/Fe) for both. The coercivity H_c for the film on Si, 2.3 kOe, is much higher than for the film on STO, 1.5 kOe. This may be a result of higher pinning at the grain boundaries of the polycrystalline film on Si compared to the single-crystalline film on STO.

Figures 3(a)–3(c) compares the out-of-plane VSM hysteresis loops for the series of films on both STO [Fig. 3(a)] and Si [Fig. 3(b)] as well as the XAS/XMCD spectra for the films on Si [Fig. 3(c)]. All the films grown on Si demonstrated room-temperature ferromagnetism with out-of-plane easy axis, as exemplified by 1–1.2 μTorr (Si) in Fig. 2(a), but the saturation magnetization and remanence decreased with increasing pressure during deposition. In comparison, the STF/STO films of Ref. [15] exhibited a maximum in magnetic moment and
FIG. 3. (a), (b) Out-of-plane VSM hysteresis curves of SrTi$_{60}$Fe$_{40}$O$_{3-\delta}$ thin films grown at different base and growth pressures on Si (a) and SrTiO$_3$ (b). (c) TFY Fe-$L$ edge XAS and XMCD of SrTi$_{0.60}$Fe$_{0.40}$O$_{3-\delta}$ thin films grown on Si at different base and growth pressures. (d) The relation between the Fe valence states determined from XAS and the net magnetic moment measured by VSM. A representative error bar is shown.

A unit-cell volume at a higher base pressure (3–4 μTorr) with a decrease in magnetic moment at higher pressures. Reference [15] also showed evidence of metallic Fe nanorods in films grown at low growth pressures, but metallic Fe was not observed in the present study. The differences may reflect the influence of beam focus, beam intensity, target condition, or other deposition parameters on the film growth.

Figure 2(c) displays XMCD hysteresis curves collected at the energy corresponding to the maximum Fe signal [marked with an arrow in Fig. 2(d)] for 1–1.2 μTorr (STO) and 1–1.2 μTorr (Si). The XMCD hysteresis curves for the two films are similar, unlike the VSM measurements. The XMCD measures a smaller area (1×1 mm$^2$) than the VSM (1×1 cm$^2$), so the difference in coercivity may be due to inhomogeneity in the film. VSM measurements on a smaller piece (5×5 mm$^2$) of the sample indicate some variation in coercivity within the sample (e.g., coercivity of 1700 Oe for the 5×5-mm$^2$ sample compared with 1400 Oe for the 1×1-cm$^2$ sample). The XMCD peak area of Fig. 3(c) scales with the VSM saturation magnetization, decreasing with increasing growth pressure (see the Supplemental Material [34]).

The line shape of the Fe XMCD signal resembles the line shape of Fe$^{2+}$ in FeTiO$_3$, however in our case the dilution and disorder of the Fe ions as well as the weak XMCD signal...
might hinder the detection of the multiplet structures detected in bulk FeTiO$_3$ [37]. While similar XMCD signals are also observed in metallic Fe, an extensive search for Fe metal precipitates using x-ray diffraction, cross-sectional TEM, and x-ray photoelectron spectroscopy strongly hints at the absence of metallic Fe (see the Supplemental Material [34]). The lack of XMCD signal at the Ti-L edge in Fig. 2(e) is consistent with the magnetism residing at the Fe, specifically Fe$^{2+}$, rather than the Ti site as found in oxygen-deficient STO [27,28].

Having established the trend in the magnetic moment of STF vs deposition pressure, we turn to XAS to examine the Fe valence states that give rise to the magnetism. The XAS spectrum, Fig. 2(d), reveals that the film exhibits a mixture of Fe$^{2+}$ and Fe$^{3+}$ (for an extensive comparison of the valence state of Fe see the Supplemental Material [34]). No Fe$^{4+}$ was detected, unlike previous studies [15,21,24,26] on ferromagnetic STF, and similarly there was no signal from metallic Fe. The Ti is present in a 4$^+$ oxidation state [Fig. 2(e)]. The observation of a mix of Fe$^{2+}$ and Fe$^{3+}$ is consistent with earlier studies of mixed cation valencies in ferromagnetic STF films [15,20,21,24] and with the presence of oxygen vacancies.

To quantify the Fe valence states, we performed a principal component analysis of the Fe$^{2+}$ and Fe$^{3+}$ spectral fingerprints. We extract a phenomenological parameter, the ratio between the XAS features labeled peak A and peak B in Fig. S3, which is a proxy for the Fe$^{2+}$:Fe$^{3+}$ ratio, as detailed in the Supplemental Material, Fig. S4 [34]. From the peak ratio of the measured XAS spectra, we estimate the Fe$^{2+}$:Fe$^{3+}$ ratio of each film. The results are summarized in Table I.

The XAS data reveal a striking correlation between the Fe valence states and the net magnetic moment as measured by VSM, Fig. 3(d). Films grown under different base and deposition pressures showed a general trend of decreasing Fe$^{2+}$ fraction with decreasing magnetic moment. The most magnetic samples, 1–1.2 $\mu$Torr (STO), 1–1.5 $\mu$Torr (STO), and 1–1.2 $\mu$Torr (Si), consisted of $\sim$90% Fe$^{3+}$ and 8–10% (±3%) Fe$^{2+}$. In contrast, the oxidized 1–1.5 $\mu$Torr (STO) and the 2–2.5 $\mu$Torr (STO) films had little or no magnetic moment and no measurable Fe$^{2+}$.

Furthermore, the Fe valence state analysis can be used to infer the fraction of oxygen vacancies $\delta$ based on the composition SrTi$_{0.65}$Fe$_{0.35}$O$_3$.4. We assume that the ions are present as Fe$^{2+}$, Fe$^{3+}$, Ti$^{4+}$, and O$^{2-}$ (not considering fractional oxidation states); all the vacancies are doubly ionized; and that there is an insignificant concentration of free electrons and other defects. This yields $\delta = 0.192$ for the 1–1.2 $\mu$Torr (STO) and 1–1.5 $\mu$Torr (STO) samples, $\delta = 0.189$ for the 1–1.2 $\mu$Torr (Si) sample, and $\delta = 0.175$ in films with no net magnetization (Table I). The higher fraction of Fe$^{3+}$ at higher deposition pressures is consistent with a reduction in the concentration of oxygen vacancies [16,17].

In Fig. 4 we focus on the effect of annealing treatments, in particular the reversibility of the process. Figure 4(a) reports the $\omega$-2$\theta$ scans between 40 and 50$^\circ$, Figs. 4(b) and 4(c) highlight the out-of-plane and in-plane VSM loops, and Figs. 4(d) and 4(e) depict XAS and XMCD data. As with the samples grown at differing pressures, multiple or asymmetrical (200) film peaks are indicative of a distribution of out-of-plane lattice parameters. RSM scans likewise show the film peak spread along $q_z$, while it is aligned with the substrate peak along $q_x$, indicating a coherent interface. In prior work on perovskite films, the presence of multiple peaks was attributed to partial strain relaxation [36]. The oxidized film has a broader peak and a higher out-of-plane lattice parameter than the reduced film.

The sample was initially magnetic with a moment of 0.84 $\mu_B$/Fe, but annealing in 160 mTorr oxygen lowered the magnetization to zero, while annealing in vacuum (1.5 $\mu$Torr) partially restored the magnetic moment. These magnetometry observations are corroborated by the XMCD data. The loop shapes also changed irreversibly after the oxidation/reduction cycle with the out-of-plane loop having lower squareness and the in-plane loop showing a small hysteresis compared to the as-grown sample. The data demonstrate incomplete reversibility of the changes resulting from oxidation, similar to the irreversibility observed in thermally annealed Co-doped (La, Sr)TiO$_3$ [12]. Consistent with the trends from as-grown films, the Fe$^{2+}$ content was lowered on oxidation and increased when the film was vacuum annealed. The annealing treatments conducted in this study therefore demonstrate the efficacy of postgrowth processing as a technique to control ferromagnetism in highly substituted STF.

B. Mechanism for room-temperature ferromagnetism

The previous analysis has shown that regardless of substrate, room-temperature ferromagnetism in these STF films is correlated with the presence of Fe$^{2+}$, whose concentration can be controlled either during or after growth via manipulation of the oxygen content. Prior studies have suggested the importance of mixed cation valencies in room-temperature ferromagnetism [14,15,20,24,26], and our results support this model in which the magnetic samples contain a mixture of Fe$^{2+}$ and Fe$^{3+}$. While we are unable to exclude the possibility of the oxygen vacancy itself also contributing to the magnetism [12,27,28,38], we note that even the nonmagnetic samples include oxygen vacancies, suggesting that this contribution is minor. This conclusion is consistent with a study on La- and Ce-doped STF in which net magnetization increased with La doping, which lowered the Fe valence state from 3$+$ to 2$+$ without nominally changing oxygen vacancy content [20]. Although the sensitivity of the magnetic moment to the Fe$^{2+}$ content is large, the actual change in $\delta$ required to balance the Fe$^{3+}$ $\rightarrow$ Fe$^{2+}$ valence state change is small because of the low fraction of Fe$^{2+}$ and the presence of only 35% of Fe on the B sites. This could also explain why the unit-cell volume was similar across the set of samples.

Magnetism in oxides has been explained from superexchange interactions as semianipically formulated in the Goodenough-Kanamori-Anderson rules [39,40] or from carrier-mediated mechanisms [9] originally developed to describe dilute magnetic semiconductors, such as double exchange or Ruderman-Kittel-Kasuya-Yosida coupling. Well-known carrier-mediated mechanisms were ruled unlikely in earlier studies on STF where it was noted that ferromagnetism persisted in highly insulating films [14,26]. Consequently, we focus on mechanisms that are more likely to dominate magnetism at the levels of B-site substitution present in this study.
The decrease of the magnetization with increasing growth or annealing pressure and the consequent lowering of oxygen vacancy concentration can be explained if one considers that neighboring Fe$^{3+}$ ions at the B site of perovskites align antiferromagnetically as suggested by the Goodenough-Kanamori-Anderson rules [39,40] and as reported in DFT calculations on STF. [15] Since Fe$^{4+}$ was not detected in the XAS data, we consider mechanisms involving only Fe$^{2+}$ and Fe$^{3+}$. Typically, octahedrally coordinated Fe$^{3+}$ in perovskites is present in a high spin state ($3d^5$, $S = 5/2$), resulting in a magnetic moment of $5\mu_B$/Fe. Without Fe$^{4+}$ or Fe$^{2+}$, compensation of the antiferromagnetically arranged spins occurs, obliterating any XMCD or VSM signatures of remanent magnetization. We assume that the Fe$^{3+}$ is also high spin. The presence of low spin Fe$^{2+}$ ($3d^6$, $0\mu_B$) is inconsistent with the XMCD result that the magnetism resides on the Fe$^{2+}$ ions rather than on the Fe$^{3+}$. Furthermore, it is unusual to find low spin Fe$^{2+}$ in the octahedral site [21,39], and low spin Fe$^{2+}$ (like high spin Fe$^{3+}$) is a nonmagnetoelastic ion [21] which cannot account for the out-of-plane anisotropy observed in the magnetic films.

When Fe$^{2+}$ replaces Fe$^{3+}$, we can calculate the fraction of Fe$^{2+}$ that must be reduced to Fe$^{2+}$ to give the experimentally observed magnetization by considering the ratio of the
magnetic moment measured via VSM to the theoretical moment of high spin Fe\(^{2+}\) and Fe\(^{3+}\) ions. The measured moment for 1–1.2 \(\mu\)Torr (STO) according to VSM is 0.8 \(\mu_B/\text{Fe}\). A high spin Fe\(^{2+}\) ion (3\(d^5\), \(S = 2\)) is magnetoelastic [21] and has a magnetic moment of 4 \(\mu_B/\text{Fe}\). In this case, magnetism could originate from Fe\(^{2+}\) ions embedded in an antiferromagnetic Fe\(^{3+}\) sublattice resulting in a net magnetization of around 0.4 \(\mu_B\) (at 10\% Fe\(^{2+}\)). While the resulting magnetization is not high enough to explain the experimentally observed magnetic moment given the concentration of Fe\(^{2+}\), this discrepancy is resolved if the presence of an adjacent oxygen vacancy enhances the Fe moment [15]. The Fe\(^{2+}\) content may also be higher than the values reported in Table I as a result of the linear calibration being affected by interference effects in the XAS signal. These effects are difficult to account for, and hence are not usually considered in these calculations.

Among the different superexchange interactions, it is possible to attain ferromagnetic alignment through exchange interactions with non-180\(^\circ\) bond angles or non-\(\delta\)/semicovalent bond lengths [39,40], which is facilitated by the presence of a neighboring oxygen vacancy that distorts the octahedral symmetry surrounding the Fe. This can change the interaction between Fe\(^{2+}\) and Fe\(^{3+}\) to be ferromagnetic, or the distortion could change the crystal-field splitting of neighboring Fe and result in some low spin ions.

Although the quantity of Fe and oxygen vacancies is high, there is no evidence that they will necessarily form an ordered sublattice. With a random or disordered distribution of Fe ions and oxygen vacancies in the STF lattice, we can expect a range of interactions among Fe\(^{2+}\) and Fe\(^{3+}\) ions through O\(^{2-}\) or possibly V\(_O\)^\(\delta+\) that will result in competing antiferromagnetic and ferromagnetic interactions, as well as differing enhancements to the Fe magnetic moment. The magnetism likely stems from some mix of the aforementioned mechanisms, which all rely on the mixed-valence state identified here and in prior studies [14,15,20,24,26] as crucial to inducing room-temperature ferromagnetism.

IV. CONCLUSION

To summarize, we have described the evolution of room-temperature ferromagnetism in PLD-grown single-crystalline and polycrystalline SrTi\(_{1-x}\)Fe\(_{0.5}\)O\(_{3.5}\) thin films as a function of deposition and annealing pressure. The key result is that the net magnetic moment, measured by magnetometry and by XMCD, showed a strong correlation with the Fe\(^{2+}\) content as measured by XAS, and decreased with increasing growth pressure and with oxygen annealing. A principal component analysis is used to establish an empirical relationship between the ratio of the two XAS peak areas and the Fe\(^{2+}\):Fe\(^{3+}\) ratio.

We were able to quantify the fraction of Fe\(^{2+}\) through XAS and relate it to the observed VSM results, and subsequently derive the concentration of oxygen vacancies required for charge balance. We further observed the quenching and partial recovery of magnetization in a SrTi\(_{0.65}\)Fe\(_{0.35}\)O\(_{3.8}\) thin film upon a cycle of oxidizing and reducing anneals, confirming that annealing can be used to actively change the magnetic moment and paving the way for studies utilizing electrochemical manipulation of the oxygen content.

The major source of magnetism in substituted STF appears to be the mixed Fe valence state consisting of Fe\(^{2+}\) and Fe\(^{3+}\), and putative mechanisms for the emergence of ferromagnetism are discussed. The presence of Fe\(^{2+}\) disrupts the antiferromagnetic interactions between Fe\(^{3+}\), facilitating a number of ferro- and ferrimagnetic interactions, and the presence of oxygen vacancies may enhance the resulting magnetic moment. The sensitivity of the magnetism to the amount of Fe\(^{2+}\) and to small changes in base and growth pressures is surprisingly high. Our study underscores the importance of oxygen vacancies in highly substituted transition metal oxides, and their powerful role in determining magnetization. The present study also motivates other methods to increase the Fe\(^{2+}\):Fe\(^{3+}\) ratio such as cosubstitution of A-site or B-site cations with higher valence states than that of Sr\(^{2+}\).

Postgrowth techniques can provide an additional pathway to engineer materials with desirable properties and also facilitate new applications. XMCD at high magnetic field or magnetic scattering studies could clarify the nature of the exchange interactions and the spin state of the ion species in the class of heavily magnetically substituted perovskites.

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