In recent years, studies on magnetic topological materials with kagome lattices have become one of the hottest frontiers of condensed matter research, due to their exotic physical properties in both real space and momentum space [1,2]. In momentum space, angle-resolved photoemission spectroscopy (ARPES) experiments on Mn$_3$Sn, Fe$_3$Sn$_2$, FeSn, and CoSn [3-6] show that kagome lattices give rise to Dirac cones and flat bands which are topologically protected and are of particular interest. In addition, scanning tunneling spectroscopy finds evidence for topological flat bands as a sharp peak in the local density of states [7]. These topologically nontrivial features result in signatures of anomalous transport (e.g. chiral anomaly) in magnetotransport experiments [3-8]. Furthermore, it is theoretically predicted that the band structures of the kagome topological magnets can be controlled by tuning of their magnetic structures [1,3]. In real space, the kagome topological magnets have layered structures with spins occupying corner-sharing triangular lattices, which leads to geometrical spin frustration [9,10]. A surprisingly large anomalous Hall effect (AHE) and magneto-optic Kerr effect (MOKE) have been reported in noncollinear antiferromagnet Mn$_3$Sn, even with vanishingly small net magnetization [10,11]. Skyrmion spin textures have been observed in ferromagnetic Fe$_3$Sn$_2$ resulting from the competition of exchange, dipolar, and Zeeman energies [12,13]. However, most of the studies on the kagome magnets have been done on bulk materials [1,3,10,16] with a few papers reporting the growth and characterization of epitaxial films [17-21]. Looking forward, epitaxial films will provide new opportunities for exploring fundamental physics and potential applications by tuning the dimensionality in thin films and interfacing Fe$_3$Sn$_2$ with different materials. In addition, preparing and visualizing a well-defined domain structure is crucial for studying the spatially resolved physics of kagome magnets, such as spin transfer torque and spin-orbit torque induced domain wall and skyrmion motion [22,23]. While the local magnetic structure of bulk Fe$_3$Sn$_2$ has been recently studied and room temperature magnetic skyrmions were revealed [12,13], there have been no studies probing the domain structure of Fe$_3$Sn$_2$ in thin films.

In this paper, we report the epitaxial growth and magnetic domain imaging of high-quality Fe$_3$Sn$_2$ thin films on Pt(111)/Al$_2$O$_3$(0001) substrates. We utilize atomic layer molecular beam epitaxy (MBE) to synthesize Fe$_3$Sn$_2$ films by sequentially depositing Fe$_3$Sn kagome layers and Sn$_2$ layers (see Figure 1). Structural characterization by in situ reflection high energy electron diffraction (RHEED) and X-ray diffraction (XRD) confirm the crystalline structure of Fe$_3$Sn$_2$. The magnetic properties of Fe$_3$Sn$_2$ are investigated using MOKE and the anomalous Nernst effect (ANE) and consistently observe easy-plane magnetic anisotropy with square hysteresis loops. Using a microscopy technique based on ANE, we success-
fully image the in-plane oriented domain structure of the epitaxial Fe$_3$Sn$_2$ films and investigate the magnetization reversal as a function of applied field.

Fe$_3$Sn$_2$ is a ferromagnet with a high Curie temperature, $T_C = 670$ K [24], and saturation magnetization of 1.9 $\mu B$ per Fe at low temperature [4]. Fig. 1 shows the crystal structure of Fe$_2$Sn$_2$ (space group R3m, with lattice constants $a = 5.338$ Å and $c = 19.789$ Å [24]) which consists of Fe$_3$Sn kagome layers and Sn spacer layers. In each Fe$_3$Sn monolayer, the Fe atoms form corner-sharing equilateral triangles surrounding hexagons, with Sn atoms sitting in the center of the hexagons. The alternating sequence of one Sn$_2$ monolayer with honeycomb lattice and two Fe$_3$Sn kagome layers produces the layered crystal structure of Fe$_3$Sn$_2$.

Based on this layered structure, we synthesized Fe$_3$Sn$_2$ thin films on top of epitaxial Pt(111) buffer layers on Al$_2$O$_3$(0001) substrates by atomic layer MBE. The epitaxial growth was performed in an MBE chamber with a base pressure of $4 \times 10^{-10}$ Torr. Films were deposited on Al$_2$O$_3$(0001) substrates (MTI Corporation) prepared by annealing in air at 1000 °C for 3 hours followed by annealing in ultrahigh vacuum (UHV) at 500 °C for 30 minutes. A 5 nm Pt(111) buffer layer was deposited from an e-beam evaporator (Pt: 99.99%, Kurt J. Lesker) onto the Al$_2$O$_3$ (0001) substrate by growing the first 0.6 nm at 440 °C and the rest 4.4 nm while cooling down from 140 °C to 80 °C. The Pt buffer layer was post-annealed at 300 °C to improve the crystallinity and surface roughness. The Fe$_3$Sn$_2$ layer was grown on Pt(111) at room temperature using the following atomic layer MBE sequence: deposit two atomic layers of Fe$_3$Sn with a Fe:Sn flux ratio of 3:1, deposit one atomic layer of Sn$_2$ with the growth time same as two Fe$_3$Sn layers, then repeat. The Fe and Sn fluxes were generated from Knudsen cells (Fe: 99.99%, Alfa Aesar; Sn: 99.998%, Alfa Aesar) and the growth rates were determined using a quartz deposition monitor that was calibrated by x-ray reflectometry. Typical growth rates were $\sim 0.85$ Å/min, $\sim 0.67$ Å/min, and $\sim 0.45$ Å/min for Fe, Sn, and Pt, respectively. To protect the sample from oxidation, a 3 nm Pt or 5 nm CaF$_2$ capping layer was deposited on top of the Fe$_3$Sn$_2$.

RHEED patterns were measured during growth to characterize the epitaxial growth and determine the in-plane lattice constants. Figure 2a shows the RHEED patterns for the Al$_2$O$_3$(0001) substrate (top row), 5 nm Pt buffer layer (middle row), and the Fe$_3$Sn$_2$ layer after 20 nm of growth (bottom row). The left and right columns show patterns taken for the beam along the [1100] and [1120] directions of the substrate. The RHEED intensity is measured within the red box and normalized by the background in the blue box.

![RHEED patterns](image)

**FIG. 2.** (a) RHEED patterns for the Al$_2$O$_3$(0001) substrate, 5 nm Pt film, and 20 nm Fe$_3$Sn$_2$ film measured with the beam along [1100] (left column) and [1120] (right column) directions of the substrate. (b) Oscillations in the normalized RHEED intensity as a function of time. The RHEED intensity is measured within the red box and normalized by the background in the blue box.
ferences in the RHEED pattern between Sn$_2$ and Fe$_3$Sn terminations. Nevertheless, the presence of RHEED oscillations in atomic layer MBE confirms the modulation of the surface termination during growth.

Films grown by this method were studied with XRD to analyze their crystal structure. A representative $\omega$-2$\theta$ scan of a 20 nm film is shown in Figure 3a and includes the Fe$_3$Sn$_2$ (0009) peak with several Laue-oscillations, indicating a high degree of film smoothness. The out-of-plane lattice parameter extracted from analysis of this scan is 19.84 Å which agrees well with previous reports of 19.789 Å [24]. A peak from the 5 nm Pt(111) buffer layer produces a shoulder on the Fe$_3$Sn$_2$ peak. Larger range scans do not show additional peaks from the Fe$_3$Sn$_2$ or any other materials. To further analyze the crystallinity of our films, rocking curve scans about the Fe$_3$Sn$_2$ (0009) peak were taken and analyzed. By scanning the sample angle $\omega$ while keeping the detector fixed, this characterizes the angular distribution of the (0001) orientation relative to the film normal within the x-ray spot. Figure 3b contains a rocking curve or $\omega$-relative scan of the Fe$_3$Sn$_2$ using a triple-axis analyzer to achieve high resolution of the desired peak. Fitting the peak with a standard Gaussian yields a full width half maximum (FWHM) of 0.0027°. Such a sharp peak indicates that films grown by atomic layer MBE have excellent mosaicity.

To investigate the in-plane and out-of-plane magnetic properties of the Fe$_3$Sn$_2$ films, we measured longitudinal and polar MOKE hysteresis loops. The samples were probed using a linearly-polarized He-Ne laser (633 nm wavelength, $\sim$ 100 $\mu$W power, $\sim$ 100 $\mu$m spot size) and a polarizing beamsplitter, photodiode bridge, and lock-
in amplifier (463 Hz intensity modulation) to detect the Kerr rotation. The laser beam had a $\sim45^\circ$ angle of incidence for longitudinal MOKE and normal incidence for polar MOKE. Figure 4a shows a representative longitudinal hysteresis loop (red curve) measured on a 20 nm thick Fe$_3$Sn$_2$ sample (Sample I). The square hysteresis loop with a coercivity of 2.4 mT indicates ferromagnetic order with in-plane magnetization. In contrast, the polar hysteresis loop (blue curve) shows a small Kerr rotation with slight variation with out-of-plane magnetic field. Together, the longitudinal and polar MOKE loops show that the Fe$_3$Sn$_2$ samples have an easy-plane magnetic anisotropy. This agrees with a previous study of Fe$_3$Sn$_2$ films grown by sputter deposition [20]. To check the consistency of the synthesis and magnetic properties, we investigated additional samples grown by the same method with thickness varying from 5 to 20 nm. The results are shown in Fig. 4b. All the samples show very similar magnetic properties, with square in-plane hysteresis loops and similar coercive fields.

The magnetic domain structure of Fe$_3$Sn$_2$ films are of interest due to the observation of skyrmions in bulk Fe$_3$Sn$_2$, but has not yet been studied in thin films. Longitudinal MOKE microscopy with oblique angle incidence can detect the in-plane magnetization and therefore determine in-plane domain structure of our Fe$_3$Sn$_2$ films. However, in this manuscript, we choose to use thermal gradient microscopy (TGM) [25–27] over longitudinal MOKE to image domain structure because we found that it has a better signal-to-noise ratio in our experimental setup.

TGM is based on moving a laser spot over the sample surface, and recording a voltage induced by the local laser heating. The thermal gradient generated in the out-of-plane direction $Z$ and a component of magnetization in the $X$ direction give rise to the anomalous Nernst effect, which is detected as a voltage along the $Y$ direction, $V_{ANE} \sim [\nabla T \times M]$ (see Fig. 5a).

For the ANE imaging, we fabricated 10 $\mu$m wide Hall bar devices by a combination of photolithography and argon ion milling (Fig. 5b). The laser excitation for the thermal gradient was produced by a frequency-doubled (BaB$_2$O$_4$ crystal) mode-locked Ti:Sapphire laser for a wavelength of 400 nm. The laser beam with 0.7 mW power was focused by a 50× objective lens (NA of 0.6) to a spot size of 0.9 $\mu$m, and a fast steering mirror in the 4f alignment scheme was used for scanning the laser spot over the sample surface. The intensity of the beam was modulated at a frequency of 120 kHz and the generated ANE voltage was detected using a lock-in amplifier.
We first utilized the ANE microscope to measure a detailed hysteresis loop at a fixed position. As shown in Fig. 5c for magnetic field along the X direction, the hysteresis loop shows a gradual reversal followed by a sharp switching behavior with coercivity of 1.9 mT. This has a similar coercivity but more gradual initial reversal than the in-plane hysteresis loops obtained by MOKE (Fig. 4b).

The origin of the different hysteresis properties is revealed by imaging the magnetic domain structure of Fe$_3$Sn$_2$ films at a series of magnetic fields. A representative sequence during the magnetization reversal is shown in Fig. 5d. Starting at -5.0 mT, the magnetization is in a saturated state along -X (blue). The reversal initiates with the nucleation of white regions with $M_x \approx 0$, mainly at the edges of the sample. This can be explained by the minimization of domain wall energy as the edge boundary does not contribute a domain wall energy cost. The nucleation at the edges initiates magnetization reversal which results in a more rounded hysteresis loop compared to the uniform films. With increasing magnetic field, domains of opposite polarity grow inward and coalesce across the channel. At about +1.0 mT, the magnetic structure is in a multidomain state with characteristic features (e.g. blue and red regions) ranging from 1 to 10 microns in size. By +2.0 mT, most of the magnetic moments have switched to +X direction, with only a few regions remaining along -X. Finally, at +5.0 mT the magnetization reversal is complete and the films is fully saturated along +X.

In conclusion, we report the growth and characterization of kagome ferromagnet Fe$_3$Sn$_2$ thin films on Pt(111)/Al$_2$O$_3$(0001) by atomic layer molecular beam epitaxy. Structural characterization by in situ RHEED and XRD confirm the high quality of the epitaxial Fe$_3$Sn$_2$ films. The magnetic properties were investigated by magneto-optical Kerr effect and anomalous Nernst effect, confirming the easy-plane magnetic anisotropy of the thin films. Finally, the local magnetic structure was probed by ANE microscopy revealing the presence of in-plane oriented micrometer size domains during magnetization reversal. These results highlight the potential for epitaxial growth to enable new scientific research in kagome magnets at the intersection of topology and magnetism.

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**AUTHOR CONTRIBUTIONS**

S.C., I.L., and R.K.K. conceived the experiments. S.C. conducted the MBE growth and MOKE measurements. I.L. conducted the ANE measurements. A.J.B. conducted the XRD measurements. All authors participated in data analysis and preparation of the manuscript.

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