

Electrothermal control of spin-reorientation transition in Co/Fe₃GaTe₂ heterostructures

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Abstract

Electrical control of magnetic anisotropy in van der Waals (vdWs) magnets is a key step toward reconfigurable two-dimensional spintronics, yet how a conventional metallic ferromagnet competes with a van der Waals magnet across a direct interface has remained largely unexplored. Here we demonstrate reversible thermal and electrothermal control of a spin-reorientation transition in Co/Fe₃GaTe₂ (FGaT) heterostructures. As Joule heating weakens the FGaT anisotropy, the heterostructure switches from an out-of-plane- to an in-plane-dominated state at a reorientation temperature of approximately 311 K, well below the Curie temperature, consistent with an exchange-mediated anisotropy competition between the Co overlayer and FGaT. An electrically driven device shows a closely matching loop evolution within an 80-100 mW power window, reversibly over five measurement cycles, consistent with an electrothermal origin. In a Co-free FGaT device, Kerr microscopy traces the switching to a power-tunable domain nucleation barrier and demonstrates power-thresholded, field-assisted magnetization reversal at a threshold near 15 mW. These results demonstrate electrothermal anisotropy competition as a route to heat-assisted and device-level control of vdWs magnetism.

Keywords: Spin-reorientation transition; Fe₃GaTe₂; van der Waals magnet; Co heterostructure; Joule heating; Magnetic anisotropy; Magneto-optical Kerr effect

1. Introduction

Electrical control of magnetism is a central goal in spintronics because it offers a route to reconfigurable magnetic states without relying solely on large write currents. In conventional magnetic heterostructures, electric fields can tune carrier density, interfacial spin-orbit coupling and magnetic anisotropy, while electrically driven heating can transiently reduce magnetic hardness near a phase boundary^[1,2]. The emergence of two-dimensional van der Waals (vdWs) magnets has expanded this idea into atomically thin systems, where clean interfaces, reduced dimensionality and strong anisotropy allow magnetic order to be coupled efficiently to gate fields, strain, optical excitation and interfacial exchange^[3,4].

Among these materials, Fe₃GaTe₂ (FGaT) has recently emerged as a particularly attractive room-temperature ferromagnetic metal. Bulk crystals, exfoliated flakes and wafer-scale films of FGaT exhibit a Curie temperature (T_C) above room temperature, large perpendicular magnetic anisotropy (PMA), and square out-of-plane (OOP) hysteresis loops, making it one of the few vdWs magnets compatible with ambient spintronic operation^[5–7]. Moreover, its magnetic ground state is not rigid. Magnetic force microscopy and Lorentz transmission electron microscopy have revealed thickness- and field-dependent stripe domains, bubble domains and skyrmion-like spin textures in FGaT^[8,9], while non-stoichiometry and local inversion-symmetry breaking

can generate Dzyaloshinskii-Moriya interaction (DMI) and stabilize chiral textures^[10,11]. These findings indicate that FGaT is both a robust ferromagnet and a highly tunable magnetic host.

Interface engineering provides an additional degree of freedom for controlling FGaT. Au capping has been shown to modulate bubble and stripe domains at room temperature^[12], while Pt-, WTe₂- and Ti-based heterostructures enable current-induced, spin-orbit or orbital-torque switching of FGaT^[13–15]. In parallel, conductive atomic force microscopy (cAFM) has demonstrated localized electrothermal writing of bubble domains in FGaT^[16], although such cAFM-based schemes remain intrinsically local and serial rather than addressable at the device level. Ferroelectric and electric-field-coupled FGaT devices further demonstrate that its T_C , coercivity and PMA can be strongly modulated by interfacial polarization, strain or carrier redistribution^[17,18]. Nevertheless, how a conventional metallic ferromagnet such as Co competes with FGaT across a direct magnetic interface remains comparatively unexplored. Such a Co/FGaT bilayer is an useful platform because Co can act as a competing ferromagnetic layer whose relative influence should increase as the FGaT magnetization weakens near its T_C .

This competition is naturally framed in terms of a spin-reorientation transition (SRT). In ultrathin magnetic films, the easy axis is determined by a balance among magnetocrystalline anisotropy, shape

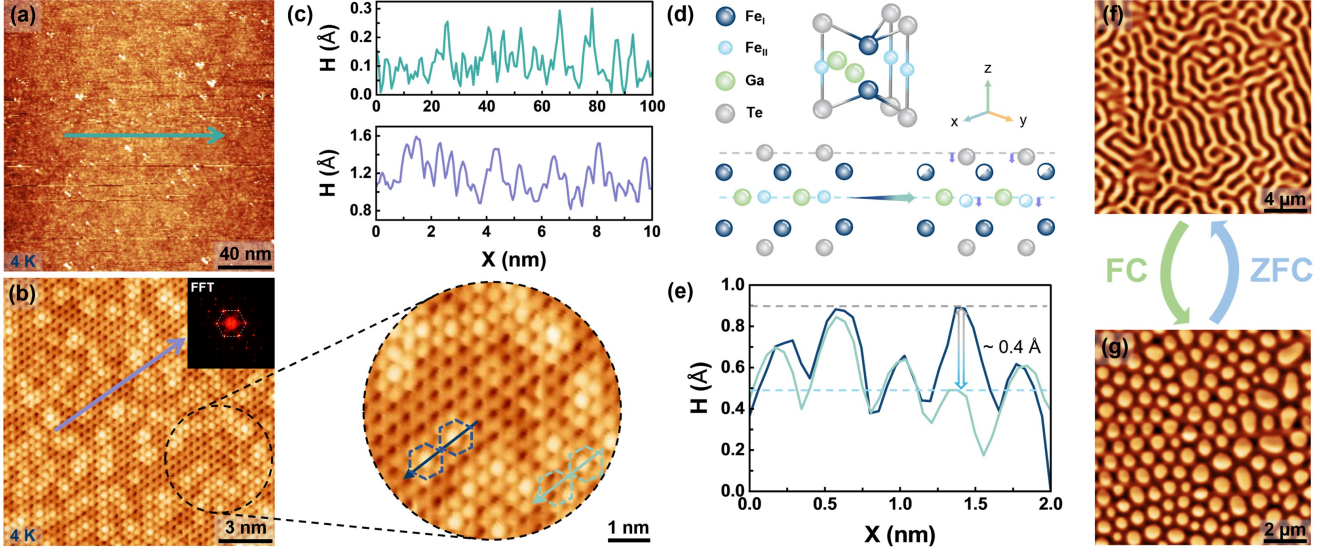


Figure 1. Atomic-scale deficiency signatures characterized by scanning tunneling microscope (STM) on cleaved Fe_3GaTe_2 (FGaT). (a) Large-area STM topograph of cleaved FGaT measured at 4 K, showing a flat terrace-like surface without visible step edges in the scanned region. (b) Atomic-resolution STM image showing a hexagonal Te lattice; the fast Fourier transform (FFT) inset confirms sixfold in-plane (IP) symmetry. The dashed circle marks the region enlarged in the central inset, in which dark- and light-blue hexagons mark two representative local motifs used for line-profile comparison. (c) Height profiles along the arrows marked in (a) (top) and (b) (bottom). The large-area profile varies by less than approximately 0.3 Å over 100 nm, confirming a step-free terrace; the atomic-scale profile resolves the periodic Te corrugation. (d) Schematic illustration of a possible subsurface Fe-deficiency/off-centering configuration that perturbs the apparent height of top-layer Te atoms. (e) Height profiles along the arrows in the central inset, showing an approximately 0.4 Å apparent-height depression at one central Te site. (f) Magnetic force microscope (MFM) image of the FGaT surface measured at room temperature after zero-field cooling (ZFC) from $T > T_C$, showing labyrinthine stripe domains. (g) MFM image after field cooling (FC) under $H = 440$ Oe applied along the c -axis, showing skyrmion-like bubble domains.

anisotropy, magnetoelastic contributions and interfacial exchange; changing thickness, temperature or interface chemistry can therefore rotate the preferred magnetization direction^[19–21]. Related Co-containing multilayers show that temperature-driven anisotropy changes in an underlying layer can be transferred through exchange coupling to a Co overlayer, producing correlated changes in hysteresis loops and magnetic easy axes^[22]. In Co/FGaT, the analogous control parameter is the temperature-dependent magnetic strength of FGaT. As Joule heating drives FGaT toward its T_C , its coercivity, remanence and nucleation behavior are expected to soften, allowing the Co layer and the Co/FGaT interface to reshape the reversal process.

Here we investigate electrothermally controlled SRT behavior in Co/FGaT magnetic heterostructures mediated by Joule heating. We combine temperature-dependent and electrically driven magneto-optical Kerr effect (MOKE) measurements to track hysteresis-loop evolution, coercivity, remanent magnetization and nucleation fields, together with STM characterization of surface deficiency and Co growth morphology, magnetic force microscopy (MFM) imaging of the FGaT domain textures, and polar Kerr microscopy of the power-dependent domain reversal. Based on the finite leakage through the top oxide layer and the close correspondence between the power-driven and temperature-driven loop evolutions, we identify the electrical response as predominantly electrothermal rather than

purely electrostatic. This positions the observed reversible magnetic softening as a device-level extension of cAFM-based electrothermal writing in FGaT^[16], and connects it to broader energy-assisted recording concepts such as heat-assisted magnetic recording, where local heating near a magnetic transition reduces coercivity^[23,24]. The loop evolution was almost symmetric with respect to bias polarity (Supplementary Fig. S1), as expected for Joule heating.

2. Results

Fig. 1 shows the surface morphology and atomic-scale defect-related signatures of cleaved FGaT measured by low-temperature scanning tunneling microscope (STM). After cleaving the FGaT crystal in the preparation chamber to minimize surface oxidation and contamination, a large-area STM topograph was first acquired at 4 K. The surface showed an extended, terrace-like morphology without visible step edges within the scanned region, indicating that the exposed vdWs surface was sufficiently flat for atomic-resolution imaging. This step-free morphology is further supported by the height profile in Fig. 1(c) (top panel), which varies by less than approximately 0.3 Å over a 100 nm span without step-like discontinuities. Additionally, sparse bright protrusions and weak streak-like features were observed, which are commonly assigned to local adsorbates, native defects or

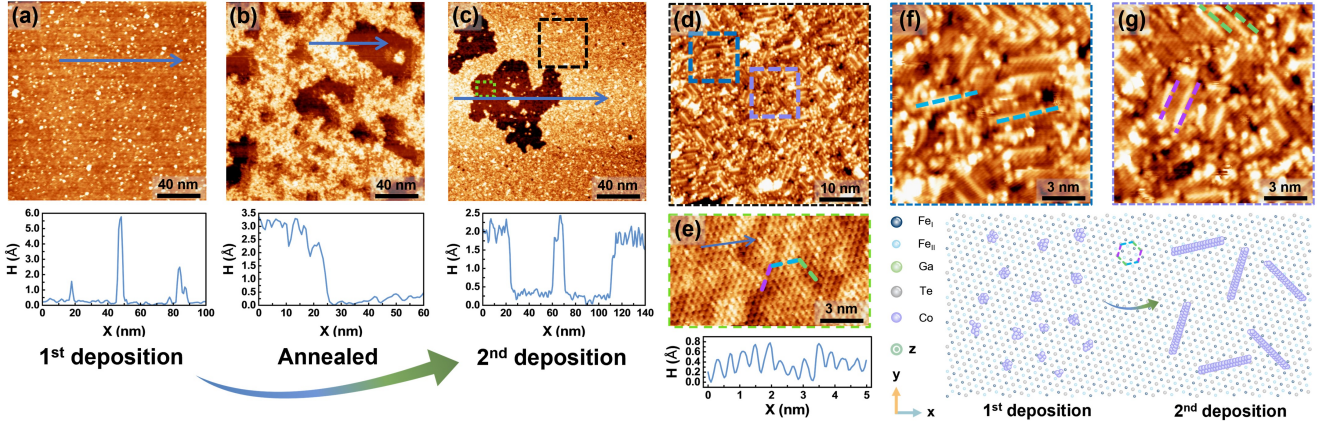


Figure 2. In situ STM characterization of Co growth on cleaved FGT: initially isolated Co clusters evolve into quasi-continuous Co coverage with lattice-aligned stripe-like features upon annealing and higher-dose deposition. Apparent Co-covered areas were estimated from levelled STM topographs by absolute-height-threshold mask analysis in Gwyddion [28]. (a) Large-area STM topograph after low-dose Co deposition on FGaT, showing dense nanoscale Co clusters with an apparent Co-covered area of 13.6%. The height profile along the blue arrow (bottom panel) shows isolated clusters up to approximately 5.5 Å in height. (b) STM topograph after annealing at 180 °C for 1 h. The apparent Co-covered area increases to 77.3% although no additional Co was supplied; the height profile shows extended plateaus approximately 3 Å high, indicating lateral spreading and flattening of the pre-deposited clusters. (c) STM topograph after a second cleave and higher-dose Co deposition without post-annealing, showing a more continuous but still partially discontinuous Co layer with an apparent Co-covered area of 73.3%; the height profile shows connected regions approximately 2 Å high separated by uncovered holes. (d) Zoomed-in STM image of the region marked by the black dashed box in (c), showing coexisting nanoscale clusters and stripe-like features. (e) (left) Atomic-resolution STM image of the exposed FGaT region marked by the green dotted box in (c), with the Te lattice direction indicated; the height profile below resolves the Te periodicity. (right) Schematic of the proposed growth evolution, from isolated Co clusters after the first deposition to lattice-aligned stripe-like Co structures after annealing and the second deposition. (f,g) Enlarged STM images of the regions marked by the blue and purple dashed boxes in (d). The blue, green and purple lines in (e), (f) and (g) mark the same Te lattice direction; the stripe orientations in (f) and (g) are nearly parallel to this direction.

tip-scan artifacts in STM topographs rather than crystallographic step edges [25].

A higher-resolution STM image revealed a periodic hexagonal lattice, consistent with the Te-terminated surface expected for layered FGaT. The corresponding fast Fourier transform (FFT) further confirmed the sixfold in-plane (IP) symmetry. Within the hexagonal network, however, the central Te sites displayed two distinct apparent-height contrasts. Line-profile analysis across two neighboring hexagonal motifs showed that the central atoms in one motif had nearly identical apparent heights, whereas another motif contained one bright and one darker central atom. The darker site was depressed by approximately 0.4 Å relative to the bright central sites.

Although STM topography reflects both geometric corrugation and local electronic density of states [26], this sub-Å depression is consistent with a local subsurface deficiency or off-centering scenario. In non-stoichiometric $\text{Fe}_{3-x}\text{GaTe}_2$, Li et al. reported that Fe deficiency can displace Fe_{II} atoms from the centrosymmetric position, break local inversion symmetry and generate DMI that stabilizes Néel-type skyrmions [10]. A related Fe-site/vacancy sensitivity has also been visualized in Fe_5GaTe_2 by STM/STS, where Fe_{I} vacancies and split-site configurations produce local electronic and topographic signatures in the Te-terminated lattice [27]. We therefore associate the observed apparent-height contrast in Fig. 1(b,e) with local perturba-

tions of the top-layer Te atoms that are consistent with subsurface Fe deficiency-induced Fe_{II} off-centering [10], as schematically illustrated in Fig. 1(d). Magnetic force microscopy further shows that zero-field-cooled (ZFC) FGaT exhibits labyrinthine stripe domains (Fig. 1(f)), whereas field cooling (FC) transforms the domain pattern into skyrmion-like bubble domains (Fig. 1(g)). This field-cooling-induced stripe-to-bubble transformation is consistent with previous reports that Fe-deficiency-induced inversion-symmetry breaking can enhance DMI and stabilize bubble or skyrmion-like spin textures in $\text{Fe}_{3-x}\text{GaTe}_2$ [9–11].

Fig. 2 examines the early-stage morphology and thermal redistribution of Co deposited on cleaved FGaT. To avoid oxidation or ambient contamination during metal deposition, Co was evaporated in situ in the preparation chamber of the STM system, and the sample was subsequently transferred back to 4 K for STM imaging. The apparent Co-covered area was estimated from levelled STM topographs using a Gwyddion grain-analysis workflow, in which Co-rich protrusions were marked by an absolute-height threshold, converted into binary masks and normalized by the total scanned area [28]. A small amount of Co was first deposited on FGaT, giving an apparent Co-covered area of 13.6% in Fig. 2(a) (the threshold analysis and the full extraction parameters are given in Supplementary Fig. S2 and Table S1). The resulting surface was decorated by dispersed nanoscale protrusions rather than

a continuous film, indicating that the initial Co deposition proceeded through island nucleation. Such morphology is consistent with Volmer-Weber growth, in which three-dimensional islands form when adatom-adatom cohesion dominates over adatom-substrate wetting^[29]. Similar three-dimensional Co clusters have been reported on pristine graphite surfaces, where weak Co-substrate interaction led to Volmer-Weber-type growth^[30].

The island-like morphology can be rationalized by the weakly interacting van der Waals surface and the geometric mismatch between close-packed Co and the Te-terminated FGaT lattice. Here, bulk hcp Co(0001) was used as a close-packed reference for estimating the geometric mismatch, because hcp Co is the stable low-temperature bulk phase and has a well-defined in-plane lattice constant^[31,32]. Using the IP lattice constant of hcp Co(0001), approximately 2.51 Å^[31,32], and the Te lattice periodicity obtained from the FFT analysis of cleaved FGaT, approximately 4.26 Å, a simple one-to-one comparison gives a large apparent mismatch of approximately 40%. Rotated or higher-order coincidence registries (e.g., $\sqrt{3} \times a_{\text{Co}} \approx 4.35$ Å, within $\sim 2\%$ of the Te periodicity) cannot be excluded^[33], and may underlie the lattice-aligned stripes in Fig. 2(f,g); the large one-to-one mismatch nevertheless disfavors coherent layer-by-layer wetting, consistent with the observed island morphology. The measured Te periodicity is slightly larger than the reported bulk in-plane lattice constant of approximately 3.986 Å^[5], which we attribute to absolute-length calibration uncertainty and piezo nonlinearity in low-temperature STM. Using either the STM-derived value or the reported bulk value gives a large mismatch of approximately 37–40%, so the growth-mode interpretation is unaffected. This large geometric mismatch, together with the weak van der Waals interface, disfavors coherent wetting and is consistent with the observed cluster morphology in the initial Co deposition.

To test whether the deposited Co could undergo longer-range redistribution, the sample was annealed at 180 °C for 1 h. After annealing, the initially isolated clusters were reorganized into broader connected regions, and the height profile showed a more gradual variation across covered areas. The apparent Co-covered area increased from 13.6% to 77.3% after annealing, although no additional Co was supplied. This increase quantifies the lateral spreading and coalescence of the pre-deposited Co clusters and supports thermally activated surface diffusion and long-range redistribution on the FGaT surface. We note two boundaries of this estimate. The apparent covered area is a projected quantity; tip convolution and the absolute-height threshold make it an upper bound for the true covered fraction. In addition, the height change from isolated clusters of up to approximately 5.5 Å to extended plateaus of approximately 3 Å (Fig. 2(a,b)) indicates that annealing flattens as well as spreads the deposit. However, a contribution from interfacial Co–Te intermixing at 180 °C cannot be excluded from topography alone. Annealing-induced diffusion and morphology evolution have been

observed in metal clusters on graphitic surfaces, where STM images after cooling capture the redistribution that occurred at elevated temperature^[34].

After a second cleave, a larger amount of Co was deposited on FGaT, producing an apparent Co-covered area of 73.3% in Fig. 2(c). The comparable high coverage obtained after annealing (77.3%) and after higher-dose redeposition (73.3%) indicates that both thermal activation and sufficient Co supply can promote extended Co redistribution on FGaT. Zoomed-in imaging of the Co-covered region (Fig. 2(d)) showed that the layer still contained nanoscale clusters, together with emerging stripe-like features (Fig. 2(f,g)). Atomic-resolution imaging of an exposed hole (Fig. 2(e)) revealed the underlying Te lattice direction, and the stripe orientation in the Co-covered region was nearly parallel to this crystallographic direction. This directional alignment suggests that once sufficient Co atoms are supplied, Co can redistribute beyond isolated clusters and become partially guided by the Te-terminated FGaT lattice. The coexistence of residual holes, connected Co-rich regions and lattice-aligned stripe textures therefore supports the feasibility of forming a more continuous Co layer on FGaT by optimizing Co dose, annealing and deposition conditions.

To obtain a Co/FGaT heterostructure suitable for magnetic measurements, a nominal 10 nm Co layer was deposited on a ~ 198 nm FGaT nanoflake (determined by AFM, Supplementary Fig. S3(a)) supported on an Al₂O₃(0001) substrate, and the stack was capped with 3 nm Pd, giving a Pd(3 nm)/Co(10 nm)/FGaT(198 nm)/Al₂O₃(0001) structure. Temperature-dependent MOKE hysteresis loops were then measured in OOP and IP geometries. In this heterostructure, the magnetic response is governed by competing anisotropy contributions from the Co overlayer and the FGaT flake. For Co films, the demagnetizing field generally favors IP magnetization unless interfacial anisotropy is sufficiently strong to stabilize PMA, and Co-based ultrathin films are known to exhibit reorientation when the balance between interface and volume/shape anisotropies changes^[19,35]. In contrast, FGaT possesses strong PMA and an above-room-temperature T_C of approximately 350–380 K^[5]. Because Co remains ferromagnetic far above the temperature window used here, heating mainly weakens the FGaT contribution and drives the heterostructure through a temperature-dependent anisotropy competition, consistent with established models of temperature-induced SRT in thin films^[36].

Fig. 3(a) shows representative temperature-dependent OOP MOKE hysteresis loops of Pd(3 nm)/Co(10 nm)/FGaT(198 nm) heterostructure (the complete temperature series in both geometries is shown in Supplementary Fig. S4). At 297.5 K, the loop exhibited a large remanent Kerr signal and an abrupt switching feature, indicating an FGaT-dominated PMA state. Upon heating, the loop became progressively less square, and the remanent Kerr signal was reduced. At 377.0 K, the OOP response approached a nearly linear, hard-axis-like loop, showing that the perpendicular compo-

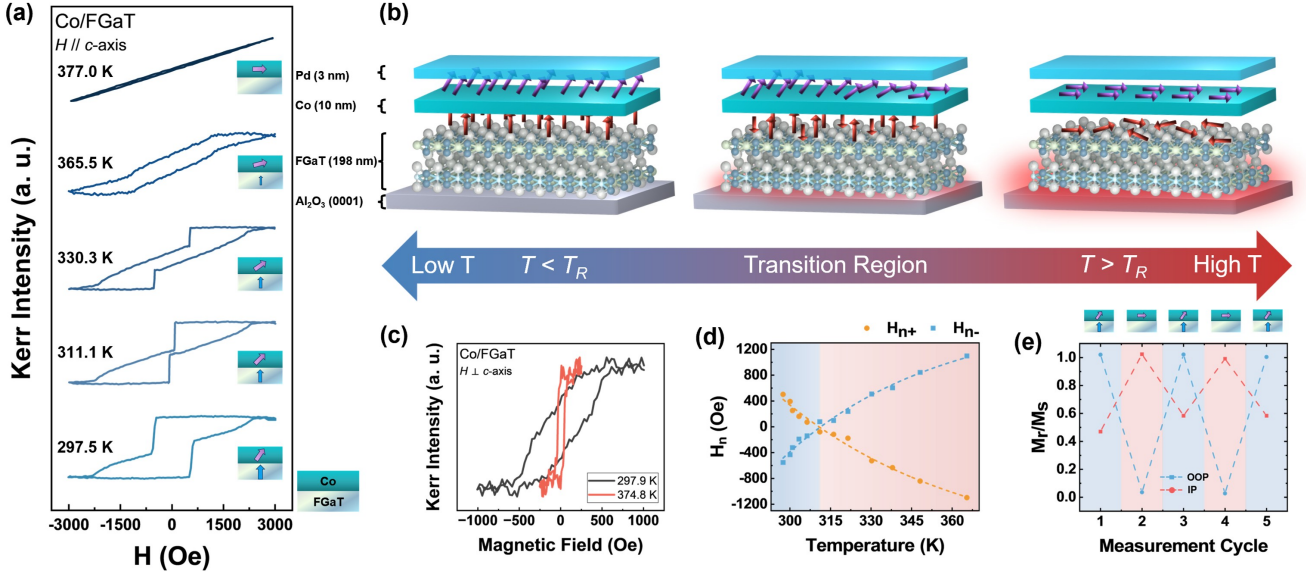


Figure 3. Temperature-driven spin-reorientation transition in Co/FGaT. (a) Temperature-dependent out-of-plane (OOP) MOKE hysteresis loops of a Pd(3 nm)/Co(10 nm)/FGaT(198 nm) heterostructure on an Al_2O_3 (0001) substrate. The magnetic field was applied along the crystallographic c-axis. The OOP loop evolves from a square, remanent state at low temperature to a nearly linear hard-axis-like response at high temperature, indicating a progressive reduction of the perpendicular magnetic contribution. (b) Schematic illustration of the temperature-driven spin-reorientation transition (SRT) process. Below the reorientation temperature T_R , the heterostructure is dominated by the perpendicular magnetic anisotropy of FGaT. Across the transition region, OOP and IP components coexist, whereas above T_R , the IP contribution associated with the Co overlayer becomes dominant. (c) IP MOKE hysteresis loops of the same heterostructure at 297.9 K and 374.8 K. The IP response evolves from a partially open, low-remanence loop at 297.9 K to a more square loop at 374.8 K, complementary to the OOP evolution in (a). (d) Temperature dependence of the positive and negative nucleation fields, H_{n+} and H_{n-} , extracted from the abrupt switching features in the MOKE loops. Above approximately 311 K, H_{n+} and H_{n-} change sign, indicating that reversed-domain nucleation occurs before field reversal. The systematic shift of H_n with temperature indicates thermally tunable magnetic-domain nucleation during the reorientation process. (e) M_r/M_s extracted from OOP and IP MOKE loops over repeated measurement cycles. The alternating recovery of the OOP- and IP-dominated states demonstrates the reversible nature of the temperature-driven SRT in the Co/FGaT heterostructure.

ment was no longer the dominant easy-axis contribution. This evolution is schematically summarized in Fig. 3(b): below the reorientation temperature T_R , the heterostructure is dominated by the PMA of FGaT; across the transition region, the OOP and IP components coexist; above T_R , the IP contribution associated with Co becomes dominant. The IP loops showed the complementary evolution (Fig. 3(c)). At 297.9 K, the IP response was a partially open loop with reduced remanence, whereas at 374.8 K it developed a more square shape, indicating that the in-plane direction became the easier magnetization direction at elevated temperature.

The reversible change in magnetic anisotropy was quantified by the remanence ratio M_r/M_s in Fig. 3(e). In the OOP-dominated state, the OOP M_r/M_s recovered close to 1.0, whereas the IP remanence ratio remained near 0.5 rather than vanishing. This residual IP remanence is consistent with an in-plane-magnetized fraction of the Co layer that is not fully rotated out of plane by interfacial exchange. After heating across the transition region, the IP M_r/M_s approached 1.0 and the OOP remanence was strongly suppressed to near zero. The recovery of these two limiting states over five alternating low- and high-temperature measurements shows that the thermally driven SRT is re-

versible within the measurement window and that the Co/FGaT heterostructure does not undergo substantial irreversible magnetic degradation during repeated heating and cooling.

The nucleation fields H_{n+} and H_{n-} , defined as the fields at which the abrupt Kerr-intensity jumps occur on the ascending and descending field branches, were identified from the two peaks of the field derivative dM/dH of each hysteresis loop (Supplementary Fig. S4(b)). They also shifted systematically with temperature (Fig. 3(d)). With increasing temperature, the magnitudes of H_{n+} and H_{n-} first decreased, crossed zero at approximately 311 K, and then reversed sign. This sign reversal indicates that, at elevated temperatures, reversed domains nucleate before the applied field changes polarity. The remanent OOP state is thus no longer stable at zero field, and reversal is initiated already during field reduction. In PMA films, magnetization reversal commonly proceeds through reversed-domain nucleation followed by domain-wall expansion; thus H_n reflects the field threshold for initiating irreversible domain formation rather than the field required to complete the full reversal^[37]. The continuous displacement of H_n across the transition window indicates that domain nucleation is governed by the same thermal weakening of the FGaT PMA contribution. These features

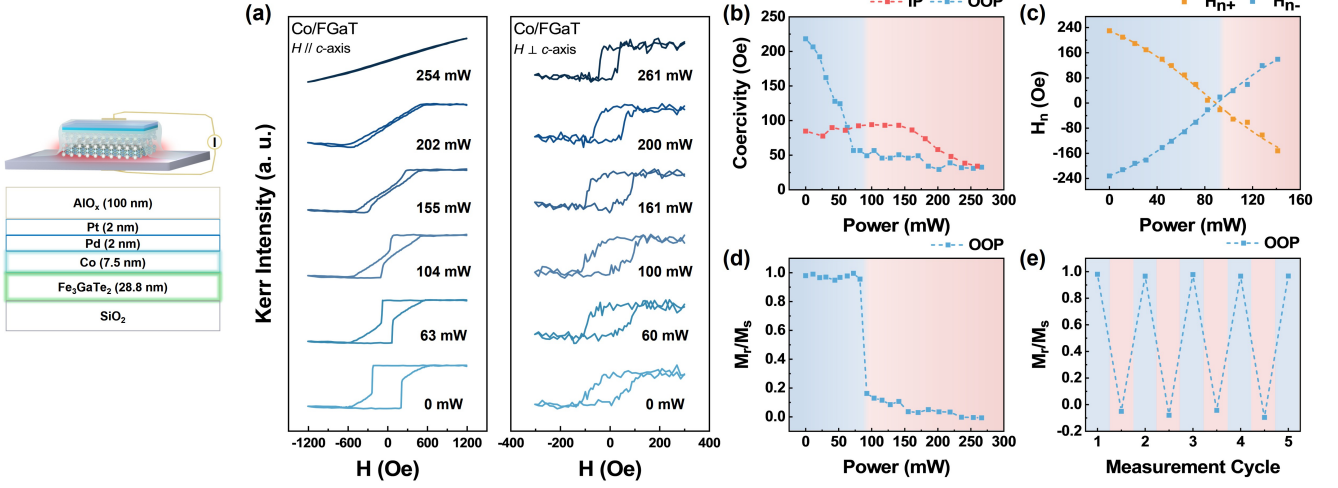


Figure 4. Electrothermal control of spin-reorientation transition in a Co/FGaT device. Left: schematic and layer structure of the $\text{AlO}_x(100 \text{ nm})/\text{Pt}(2 \text{ nm})/\text{Pd}(2 \text{ nm})/\text{Co}(7.5 \text{ nm})/\text{FGaT}(28.8 \text{ nm})/\text{SiO}_2$ multilayer device. The electrical input produces Joule heating in the active heterostructure because of leakage through the AlO_x layer. (a) Power-dependent MOKE hysteresis loops measured in the OOP and IP geometries; OOP and IP loops were recorded in separate runs at nominally matched input powers. The OOP loop evolves from a square PMA-dominated loop at 0 mW to a nearly linear hard-axis-like response at high power, whereas the IP loop becomes more square with increasing power. (b) Coercivity extracted from OOP and IP MOKE loops as a function of input power. (c) Positive and negative nucleation fields, H_{n+} and H_{n-} , extracted from the abrupt switching features in the OOP loops. (d) OOP remanence ratio M_r/M_s as a function of power, showing the collapse of the perpendicular remanent state across the electrothermal transition window. (e) Repeated power-cycling measurement of OOP M_r/M_s , demonstrating reversible switching between high- and low-remanence states. Small negative values of M_r/M_s reflect the residual Kerr-signal offset when the remanent signal is close to zero.

provide an operational definition of the reorientation temperature. We define T_R as the temperature at which H_{n+} and H_{n-} cross zero, giving $T_R \approx 311 \text{ K}$ for this heterostructure; the operational criterion marks the loss of the remanent OOP state, i.e., the onset of the reorientation process, while the full OOP-to-IP evolution is completed over a broader temperature window (Figs. 3(a)). This tunable nucleation field provides a practical handle for later field-cooling or pulsed-heating protocols aimed at accessing bubble- or skyrmion-like nucleation regimes.

To extend the temperature-driven SRT into an electrically addressable device geometry, we fabricated an $\text{AlO}_x(100 \text{ nm})/\text{Pt}(2 \text{ nm})/\text{Pd}(2 \text{ nm})/\text{Co}(7.5 \text{ nm})/\text{FGaT}(28.8 \text{ nm})/\text{SiO}_2$ multilayer device. The electrical input was applied in a two-terminal configuration through two silver-paste contacts, one on the top AlO_x surface and one on the abraded back of the SiO_2 substrate; silver paste was used instead of wire bonding to avoid puncturing the thin oxide layer (Fig. 4, left), and the dissipated power $P = IV$ was used as the control parameter. Compared with the sample in Fig. 3, the device employs a thinner FGaT flake (28.8 nm, Supplementary Fig. S3(b)) and a slightly thinner Co layer (7.5 nm). The same OOP-to-IP crossover phenomenology in both samples indicates that the anisotropy competition is not specific to one thickness combination, although T_R is expected to shift with FGaT thickness. The Pt/Pd bilayer was inserted as a metallic protection layer between Co/FGaT and the oxide deposition environment. The Pd/Co and Pt/Pd interfaces could in

principle add interfacial PMA. At a Co thickness of 7.5 nm, however, such interface terms are much smaller than the shape anisotropy^[19,35] and do not alter the IP preference of the Co layer. The AlO_x layer was designed to electrically isolate the active magnetic stack. However, leakage measurements on a Co-free control device carrying a similarly prepared AlO_x/Pt top layer showed that the oxide did not fully block current flow (Supplementary Fig. S5(a)); the leakage curve exhibits hysteretic behavior with large resistance, confirming that the applied bias dissipates as Joule heating in the stack. We therefore treated the electrical input as an electrothermal drive and used the dissipated power as the control parameter. For this Co-free device the extracted nucleation field and coercivity depend on the bias magnitude but not on its polarity (Supplementary Fig. S5(b-d)), consistent with the polarity-symmetric loop evolution in Supplementary Fig. S1. This assignment is consistent with heat-assisted magnetic writing concepts, in which transient heating reduces magnetic coercivity and enables field-assisted reversal near the magnetic transition regime^[23,24].

Fig. 4(a) shows the power-dependent MOKE hysteresis loops measured in the OOP and IP geometries. At 0 mW, the OOP loop displayed a large remanent Kerr signal and an abrupt switching feature, indicating an FGaT-dominated PMA state. As the input power increased, the OOP loop became progressively less square and the remanent Kerr signal was suppressed. At 254 mW, the OOP response approached a nearly linear hard-axis-like loop, closely resembling

the high-temperature response observed in Fig. 3. This similarity supports the interpretation that the electrical input drives the same anisotropy competition as external heating, namely the weakening of the FGaT PMA contribution relative to the Co-dominated IP contribution^[5,36].

The IP loops showed the complementary evolution. With increasing power, the IP hysteresis became more square, indicating that the IP response gradually evolved from a partially open, low-remanence loop toward an easy-axis-like square loop. The finite IP coercivity already present at 0 mW (approximately 85 Oe, Fig. 4(b)) is consistent with this residual IP component. This OOP-to-IP crossover is further quantified in Fig. 4(b–d). The OOP coercivity decreased rapidly at low power, whereas the IP coercivity remained nearly unchanged until approximately 160 mW before decreasing. This offset indicates that the IP reversal is not simply the inverse of the OOP coercivity collapse. Instead, it is consistent with a residual FGaT PMA contribution that continues to constrain the IP reversal until stronger Joule heating further suppresses the FGaT-dominated perpendicular component.

The nucleation fields H_{n+} and H_{n-} also shifted systematically with power (Fig. 4(c)). As in the temperature-driven case (Fig. 3(d)), H_n provides a sensitive measure of the field threshold for initiating irreversible reversal^[37]. The monotonic displacement of H_{n+} and H_{n-} with increasing power, including a sign reversal similar to that observed in the thermal measurements, shows that electrothermal driving continuously modifies the domain-nucleation condition (H_n could be extracted only below approximately 160 mW, where the OOP loops retain an abrupt switching feature). Using the same criterion, H_{n+} and H_{n-} cross zero at approximately 90–100 mW (Fig. 4(c)), within the same 80–100 mW window in which the OOP M_r/M_s drops sharply (Fig. 4(d)).

Finally, repeated power-cycling measurements confirmed that the electrothermal modulation was reversible within the tested range. As shown in Fig. 4(e), the OOP M_r/M_s repeatedly recovered to a high-remanence state at low power and returned to a suppressed-remanence state at high power over five measurement cycles. These results demonstrate that Joule-heating-driven electrical input can reversibly switch the dominant magnetic anisotropy of Co/FGaT between OOP- and IP-dominated states, providing a device-level route for controllable SRT in the heterostructure.

3. Discussion

Two scenarios could in principle produce the observed loop evolution. In the first, the coupled Co/FGaT system undergoes a genuine spin reorientation. In the second, the FGaT Kerr signal simply vanishes near T_C while the Co layer remains in-plane magnetized at all temperatures, so that the apparent transition reflects only a change in signal weighting. Several observations favor the first scenario. First, a 10 nm Co film can-

not sustain a square OOP loop on its own. At this thickness, the shape anisotropy of Co exceeds typical interfacial anisotropy contributions by roughly an order of magnitude^[19,35]. An uncoupled Co layer would therefore contribute only a linear, zero-remanence OOP background. The measured OOP M_r/M_s at 297.5 K is nearly 1.0. Within the optical penetration depth, a 10 nm top layer contributes substantially to the Kerr signal. Together, these observations indicate that interfacial exchange coupling pulls the Co magnetization out of plane at low temperature. Second, the low-temperature OOP loops switch in a single abrupt step rather than in two separated steps (Fig. 3(a)). This is the behavior expected for an exchange-coupled bilayer reversing collectively, not for two independent layers. Third, the transfer of a temperature-driven reorientation from an underlying layer to a Co overlayer through interfacial exchange has been demonstrated in Co/NiO/Fe trilayers^[22]; although that coupling proceeds through an antiferromagnetic NiO spacer rather than the direct ferromagnet-to-ferromagnet contact studied here, it establishes the general mechanism of exchange-transferred reorientation. In addition, the reorientation is centered near 311 K, well below the T_C of thick FGaT flakes (approximately 350–380 K^[5]). The crossover therefore occurs while FGaT remains ferromagnetic, as required for an anisotropy-competition-driven SRT. Furthermore, Co-free control devices retain a square OOP loop with a coercivity of up to ~ 1 kOe at low power (Fig. S5(d)) across the power range where the Co/FGaT device loses its perpendicular remanence, and their coercivity collapses only at substantially higher power. This is shown both for a device with the same FGaT thickness as the Co/FGaT device (28.8 nm; Supplementary Fig. S5) and for a second Co-free device with an estimated FGaT thickness of approximately 100 nm (Supplementary Fig. S1), confirming that the crossover in the heterostructure is not set by the loss of FGaT ferromagnetism alone. However, because the device temperature was not measured during electrical driving, the dissipated power axis serves as a proxy for the local temperature; the correspondence between the 80–100 mW window and the thermal $T_R \approx 311$ K is therefore qualitative.

Beyond the loop-level evidence, direct domain imaging corroborates the nucleation picture inferred from H_n . Polar Kerr microscopy of $\text{AlO}_x(100 \text{ nm})/\text{Pt}(3 \text{ nm})/\text{FGaT}(39.3 \text{ nm})$, a Co-free FGaT device that isolates the electrothermal softening of the FGaT layer itself, was performed at fixed input powers between 0 and 106 mW while sweeping the magnetic field (Fig. 5(a)). At each power, reversal proceeded through reversed-domain nucleation followed by domain expansion, consistent with the H_n analysis in Figs. 3 and 4^[37]. The field required to complete the reversal decreased systematically with increasing power, from approximately 1.2 kOe at 0 mW to below approximately 220 Oe at 106 mW. Joule heating thus lowers the domain-nucleation barrier and facilitates field-driven reversal, providing a direct real-space counterpart of the nucleation-field analysis in Figs. 3(d) and 4(c): the same electrother-

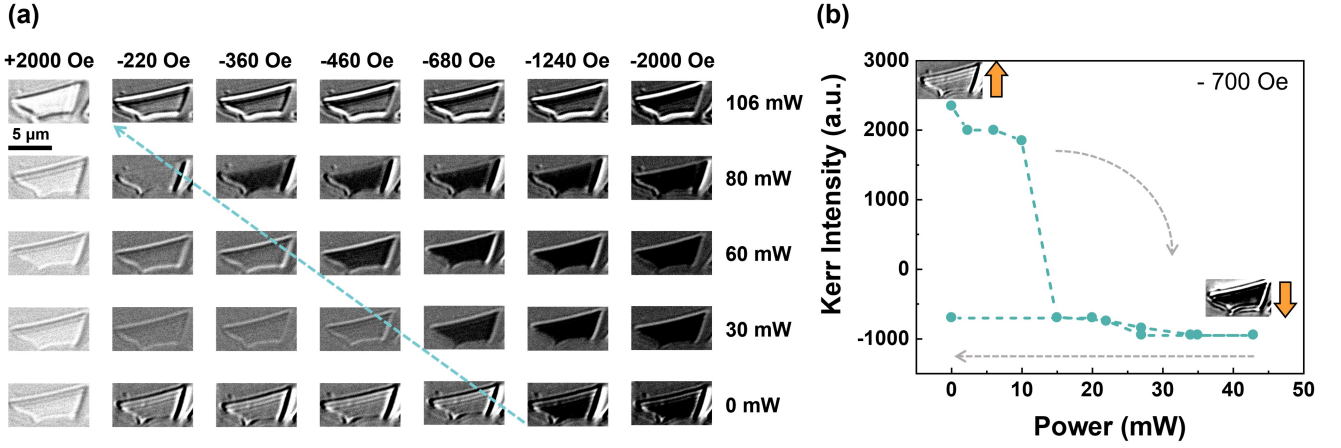


Figure 5. Power-dependent domain reversal in a Co-free $\text{AlO}_x(100 \text{ nm})/\text{Pt}(3 \text{ nm})/\text{FGaT}(39.3 \text{ nm})$ device (thickness determined by AFM, Supplementary Fig. S3(c)), imaged by polar Kerr microscopy. (a) Each row shows the field-driven reversal at a fixed input power; each column corresponds to the labeled applied field, starting from saturation at +2000 Oe. Bright and dark contrast correspond to positive and negative out-of-plane magnetization, respectively. The field required to complete the reversal decreases systematically with increasing power. (b) Power-thresholded switching at a fixed assist field of -700 Oe . Kerr intensity of the flake in (a) as a function of input power. The intensity switches abruptly between 10 and 15 mW and saturates near 30 mW, corresponding to complete reversal from the positive (upper inset) to the negative (lower inset) magnetization state. Gray dashed arrows indicate the measurement sequence; the reversed state persists as the power is returned to zero.

mal softening that displaces H_n in the loop measurements appears here as domain nucleation and expansion at progressively lower fields. This power-tunable stability enables thresholded switching at a fixed assist field. With the field held at -700 Oe , the Kerr intensity switched abruptly between 10 and 15 mW and saturated near 30 mW, corresponding to complete reversal from the positive to the negative magnetization state (Fig. 5(b)). The reversed state was retained as the input power was returned to zero under the -700 Oe assist field, so that the input power sets a threshold for field-assisted reversal. This behavior provides a device-level realization of the heat-assisted writing scheme anticipated above: a transient electrothermal input reduces the switching threshold so that a moderate field writes the magnetization state^[23,24], complementing tip-based local writing in FGaT^[16].

To test whether the heating that drives the SRT additionally modifies the electronic structure or local atomic arrangement of FGaT, we performed X-ray absorption spectroscopy (XAS) on a related PMA heterostructure, $\text{Pd}(3 \text{ nm})/[\text{Co}(0.3 \text{ nm})/\text{Pd}(0.8 \text{ nm})]_{10}/\text{Pd}(3 \text{ nm})$ grown on a bulk FGaT crystal (Supplementary Fig. S6). The sample was heated above the FGaT T_C and measured at the Fe and Ga K edges after FC and ZFC, protocols analogous to those used for the MFM imaging in Fig. 1(f,g). The near-edge spectra, sensitive to the local electronic structure, coincide for the two protocols at both edges, indicating no detectable change in the Fe or Ga valence states. The extended fine structure was Fourier transformed following standard procedures^[38,39]; because the data set did not permit quantitative shell fitting, the comparison is restricted to the positions and amplitudes of the Fourier-transform peaks, which coincide for the FC and ZFC data and are consistent with the Fe- and Ga-centered coordina-

tion shells of the FGaT lattice, together with a weak Fe-O/Ga-O first-shell contribution from surface oxidation of the ex situ crystal (Supplementary Note 1). The slightly higher FC amplitude at the Fe K edge is attributed to small differences in the measurement geometry and to progressive surface oxidation of the ex situ crystal over the extended acquisition time, rather than to an electronic-structure change. This surface-oxidation contribution is unrelated to the oxide-free Co/FGaT interface prepared in situ by STM (Fig. 2). The near-coincidence of the FC and ZFC spectra at both edges further indicates that repeated thermal cycling does not produce progressive chemical changes between the metallic overlayer and FGaT. The contrast between the history-dependent domain patterns (Fig. 1(f,g)) and the history-independent spectra is consistent with reports that stripe and skyrmion-like states in FGaT and Fe_3GeTe_2 coexist through the competition between dipolar interactions and DMI, and can be selected by the magnetic history alone at a fixed temperature^[9,40]. At this level of sensitivity, heating across T_C produces no detectable irreversible change in either the electronic structure or the local lattice environment. The reversible SRT therefore reflects a magnetic anisotropy competition rather than a structural or chemical transformation, consistent with the recovery of the two limiting states over repeated cycles (Figs. 3(e) and 4(e)).

Finally, the STM results frame the microscopic ingredients of this competition. The in situ growth study shows that Co nucleates directly on the freshly cleaved, oxide-free Te-terminated surface, evolving from Volmer-Weber clusters to quasi-continuous, lattice-guided coverage (Fig. 2). Deposition on such an oxide-free surface is a prerequisite for the intimate interfacial contact that would mediate the ex-

change coupling inferred from the magnetic measurements. In parallel, the subsurface Fe-deficiency signatures and the field-cooling-induced bubble domains (Fig. 1) suggest that the same crystals host DMI-active disorder^[10,11]. Combining this defect landscape with the power-tunable nucleation barrier suggests a route to electrothermally assisted nucleation of bubble or skyrmion-like textures, extending local writing schemes^[16] toward device geometries.

4. Conclusion

In summary, we have shown that the dominant magnetic anisotropy of a Co/FGaT heterostructure can be reversibly switched between OOP- and IP-dominated states, both by external heating and by Joule heating in a device geometry. In situ STM showed that Co deposits directly on the freshly cleaved, oxide-free FGaT surface and revealed subsurface Fe-deficiency signatures consistent with DMI-active disorder^[10,11]. Temperature-dependent MOKE identified a spin-reorientation transition at $T_R \approx 311$ K, well below the FGaT T_C , consistent with an exchange-mediated anisotropy competition in the Co/FGaT bilayer. The electrothermal device showed a closely matching loop evolution within an 80–100 mW power window, reversibly over five measurement cycles, consistent with the same electrothermally driven anisotropy competition. In a Co-free FGaT device, Kerr microscopy traced this softening to a power-tunable domain-nucleation barrier and demonstrated power-thresholded, field-assisted magnetization reversal at a threshold near 15 mW. X-ray absorption showed no detectable irreversible electronic or local structural change in FGaT after thermal cycling across T_C .

These results demonstrate electrothermal anisotropy competition as a practical, device-level handle for van der Waals magnets. The immediate boundaries of this work also define its next steps. Pulsed operation would quantify the switching energy and dynamics beyond the quasi-static drive used here. Element-resolved probes such as XMCD could resolve the layer-by-layer reorientation directly. Combining the power-tunable nucleation barrier with field-cooling protocols may enable on-demand writing of bubble- or skyrmion-like textures as a device-level counterpart to cAFM-based local writing^[16], and wafer-scale FGaT growth^[7] offers a route to extending these functions beyond exfoliated flakes.

5. Methods

Crystal growth and structural characterization. High-quality Fe_3GaTe_2 single crystals were grown by the chemical vapor transport (CVT) method. High-purity Fe granules, Ga lumps and Te pieces were mixed in a 3:1:2 stoichiometric ratio and sealed in an evacuated quartz tube with iodine (2 mg/cm^3) as the transport agent. All preparation steps were performed in a glove box to ensure material purity. The sealed tube was held in a furnace under a temperature gradient of 760–

710 °C for 7–10 days and then cooled naturally to room temperature inside the furnace.

Sample and device fabrication. FGaT flakes were mechanically exfoliated from the bulk crystals with adhesive tape and transferred onto sapphire or SiO_2 substrates. The Co and Pd layers were deposited at room temperature by home-built electron-beam evaporation at a base pressure of 5×10^{-9} mbar. The Pt and AlO_x layers were deposited by pulsed laser deposition (LOTIS TII Nd:YAG laser; wavelength 266 nm, repetition rate 10 Hz, pulse width 16–18 ns) at room temperature and a base pressure of 10^{-7} mbar, with a target-to-substrate distance of 5 cm; the Pt layer was ablated from a Pt target at an energy density of 2.8 J/cm^2 and the AlO_x layer from an Al_2O_3 target at an energy density of 1.6 J/cm^2 . The crystal structure and lattice parameters were characterized by X-ray diffraction, and the refined interatomic distances are used for the EXAFS shell assignments in Supplementary Note 1.

Scanning tunneling microscopy. Surface morphology and atomic structure were characterized with a low-temperature ultrahigh-vacuum STM (Unisoku USM-1300S equipped with an RHK R9 controller) operated at 4 K^[41], using electrochemically etched W tips. To minimize oxidation, the crystals were cleaved in a preparation chamber at 10^{-6} mbar, then transferred to the UHV STM stage. Co deposition and post-annealing (180 °C, 1 h) were performed in the same preparation chamber, and the sample was returned to the 4 K stage for imaging. Apparent Co-covered areas were estimated from levelled topographs in Gwyddion^[28] using absolute height thresholds of 0.6, 2, 1.5 Å for the images in Fig. 2(a), (b) and (c), respectively; the threshold masks and the full extraction parameters, including the grain-area filters, are given in Supplementary Fig. S2 and Table S1.

Atomic force and magnetic force microscopy. Topography and flake thicknesses were measured with a Bruker Innova AFM using silicon tips with a resonant frequency of 300 kHz and a spring constant of 26 N/m. MFM was performed on the same system with CoCr-coated tips. For the domain-imaging protocols, the FGaT bulk crystal was heated above T_C and cooled to room temperature either in zero field or in a field of 440 Oe applied along the c-axis. MFM images were then acquired at room temperature under ambient conditions.

Magneto-optical Kerr effect measurements. Hysteresis loops and magnetic domain images were recorded with a magneto-optical Kerr microscope (Evico Magnetics GmbH) in the polar and longitudinal geometries^[42,43] under ambient conditions. For temperature-dependent measurements, the sample was mounted on a home-built heating stage and the temperature was monitored with a K-type thermocouple. For electrothermal measurements, the electrical input was applied in the two-terminal configuration described in the main text, and the dissipated power $P = IV$ was used as the control parameter, with the bias applied and the current recorded using a Keithley 2400 source-measure unit. Domain images were acquired as background-subtracted difference

images referenced to the saturated state.

For the electrothermal and leakage measurements, contacts were made with silver paste; the SiO_2 substrate backside was abraded before applying the paste, and silver paste (rather than wire bonding) was used on the AlO_x surface to avoid puncturing the thin oxide layer.

X-ray absorption spectra for Fe and Ga were obtained at beamline BL12B2 (SPring-8) in fluorescence mode at room temperature. The XAS and EXAFS data were processed with the ATHENA package^[38] following standard XAFS analysis procedures^[39].

6. CRediT author contribution statement

Po-Wei Chen: Conceptualization, Data curation, Investigation, Methodology, Writing – original draft. **Ming-Hsien Hsu:** Conceptualization, Data curation, Investigation, Methodology. **Cheng-Ying Hsiao:** Data curation, Investigation. **Ming-Yang Ho:** Data curation, Investigation. **Masahiro Haze:** Data curation, Investigation. **Yan-Ru Chu:** Data curation, Investigation. **Yu-Chen Shao:** Investigation. **Po-Chun Chang:** Conceptualization, Methodology. **Chen-Yu Ou:** Investigation. **Ruei Chen:** Investigation. **Ke-Fan Chen:** Data curation, Investigation. **Chung-Ting Ke:** Methodology, Project administration, Resources, Supervision. **Chao-Hung Du:** Methodology, Project administration, Resources. **Yukio Hasegawa:** Methodology, Project administration, Resources. **Wen-Chin Lin:** Conceptualization, Data curation, Formal analysis, Funding acquisition, Investigation, Project administration, Resources, Supervision, Validation, Visualization, Writing – original draft, Writing – review & editing.

7. Declaration of competing interest

The authors declare no competing interests.

8. Acknowledgments

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9. AI Disclosure

AI-assisted tools were used to support language editing, formatting, and LaTeX preparation. No AI-assisted tool was used to generate, analyze, or interpret scientific data. The authors reviewed and revised all AI-assisted outputs and take full responsibility for the final content of the manuscript.

10. Appendix. Supplementary materials

Supplementary material associated with this article can be found in the online version.

11. Data availability

The data that support the findings of this study are available from the corresponding author upon reasonable request.

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