Interfacial Dzyaloshinskii-Moriya interaction arising from rare-earth orbital magnetism in insulating magnetic oxides

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The Dzyaloshinskii-Moriya interaction (DMI) is responsible for exotic chiral and topological magnetic states such as spin spirals and skyrmions. DMI manifests at metallic ferromagnet/heavy-metal interfaces, owing to inversion symmetry breaking and spin-orbit coupling by a heavy metal such as Pt. Moreover, in centrosymmetric magnetic oxides interfaced by Pt, DMI-driven topological spin textures and fast current-driven dynamics have been reported, though the origin of this DMI is unclear. While in metallic systems, spin-orbit coupling arises from a proximate heavy metal, we show that in perpendicularly-magnetized iron garnets, rare-earth orbital magnetism gives rise to an intrinsic spin-orbit coupling generating interfacial DMI at mirror symmetry-breaking interfaces. We show that rare-earth ion substitution and strain engineering can significantly alter the DMI. These results provide critical insights into the origins of chiral magnetism in low-damping magnetic oxides and identify paths toward engineering chiral and topological states in centrosymmetric oxides through rare-earth ion substitution.
Inversion symmetry breaking gives rise to rich chiral phenomena that yield new fundamental physics and materials properties\(^2\). For example, in chemistry, single-handedness of biomolecules results in specialized functionality during biological operations\(^5\). In particle physics, a fundamental understanding of electroweak interactions hinges on charge-parity symmetry violations\(^3\). In magnetic materials, a chiral exchange interaction known as the Dzyaloshinskii–Moriya interaction (DMI) can manifest from broken spatial inversion symmetry\(^6,9\). This discovery has led to the observation of chiral and topological spin textures, such as magnetic skyrmions and homochiral spin spirals\(^10,11\). Only a limited number of inversion-asymmetric bulk magnetic materials are known, almost all of which exhibit ordered chiral states only at cryogenic temperatures\(^12,13\). However, mirror symmetry breaking at interfaces has been shown to give rise to an interfacial DMI (iDMI) in thin-film ferromagnets at room temperature, providing a general route to tailor chirality in magnetic materials through interface engineering\(^8,15\). DMI is typically observed in metallic ferromagnet/heavy-metal (HM) bilayers, where strong spin-orbit coupling (SOC) in the HM is responsible for generating the iDMI\(^16,17\). Since iDMI-inducing HMs such as Pt also give rise to a large spin current from the spin Hall effect\(^18,19\), topological spin textures in such materials can be efficiently manipulated by electric current\(^14,15,20\). This has led to large interest in such systems for spintronics applications in which chiral domain walls (DWs) or skyrmions are used to encode bits in racetrack devices\(^14,15,20–22\).

Very recently, iDMI has been discovered in thin films of centrosymmetric insulating iron garnet\(^23–25\), an important class of low-damping magnetic oxides in which its presence had gone undetected for decades despite their ubiquity in magnetics research. Experimental observations of homochiral Néel DWs\(^23,24,26\) and topological Hall-like signals\(^27,28\) in thin-film rare-earth (RE) iron garnet (REIGs) at room temperature suggest that spin–orbit-driven phenomena thought to be restricted to metallic systems might manifest more broadly in insulating magnetic oxides. However, unlike metallic systems in which iDMI stems from SOC in an adjacent HM, experiments in REIGs suggest that a proximate high-SOC layer is not required. The underlying origin of iDMI in insulating oxides hence remains to be understood.

Here we show that in REIGs, iDMI arises intrinsically at mirror symmetry-breaking interfaces due to SOC in the magnetic oxide itself, owing to the orbital magnetism of the RE ion. We show that the iDMI is influenced by the substrate/oxide and oxide/metal interfaces, but the SOC in the symmetry-breaking layers plays little role in the iDMI. Rather, we show that the iDMI scales with the SOC of the RE ion in the oxide and is undetectable if the RE ion is absent. We further show that the iDMI can be tuned by substrate strain, by at least a factor of two, providing a powerful means to tune chiral magnetism in oxide materials. These results suggest that introducing SOC in magnetic oxides through RE ion substitution can provide a general path toward realizing topological spin states in thin films and heterostructures, opening a new door for oxide-based spintronics.

Results and structural characterization. We first examine the thickness scaling of the iDMI in perpendicularly magnetized Tm\(_3\)Fe\(_5\)O\(_{12}\) (TmIG) to verify its interfacial origin and quantify its magnitude. A series of TmIG films were epitaxially grown on (111)-oriented Gd\(_3\)Ga\(_5\)O\(_{12}\) (GGG) substrates via pulsed laser deposition (PLD) and covered by a 4.0-nm-thick Pt overlayer (Fig. 1a, see Methods). The nominal TmIG thickness \(t_{\text{tmiG}}\) ranged from 2.4 to 24 nm (approximately 2 unit cells to 20 unit cells). TmIG is ferrimagnetic, with Fe\(^{3+}\) ions occupying three tetrahedral and two octahedral lattice sites per formula unit with oppositely oriented magnetic moments. The moment of \(3\mu_\text{B}\) in dodecahedral sites is oriented in the same direction as that of the octahedral Fe\(^{3+}\) (Fig. 1a). Figure 1b shows X-ray diffraction performed on exemplary 20 and 40-nm-thick GGG/TmIG samples, revealing (hhh)-type reflections. Laue fringes around the TmIG peak of the thicker film indicate the high crystalline quality and thickness uniformity of the magnetic garnet film. Fringes are broader and spaced further apart in the thinner TmIG film. The epitaxial growth of the fully strained garnet film on the GGG substrate is confirmed by reciprocal space mapping\(^29\) and by high-resolution scanning transmission electron microscopy (Fig. 1c). Complementary electron energy loss spectroscopy (EELS) is shown in Fig. 1d, e, revealing an interface region between the substrate and TmIG of ~1 nm width where Gd and Ga have diffused into the TmIG layer, similar to other studies\(^30–33\). Perpendicular magnetization in all films is confirmed by vibrating sample magnetometry, shown in Fig. 1f for a 12-nm GGG/TmIG sample. Figure 1g shows the measured magnetic moment per unit area versus \(t_{\text{Pt}}\), where the fitted slope of the data (solid curve) yields the saturation magnetization \(M_s\) of the films. We find that TmIG films as thin as 2.4 nm retain near-bulk magnetization with \(M_s \approx 115\ \text{emu cm}^{-3}\) beyond an interfacial nonmagnetic layer. The non-zero horizontal intercept indicates a thickness for this layer of \(t_{\text{mix}} = 1.4\ \text{nm}\), approximately one unit cell of TmIG, consistent with the STEM and EELS results showing interdiffusion of Fe, Ga, Tm, and Gd at the substrate/film interface. We presume that the non-magnetic dead layer corresponds to the ~1 nm intermixed region at the substrate/garnet interface, where the substitution of Ga for Fe renders the garnet paramagnetic.

Interfacial DMI in REIGs. iDMI was measured using the spin Hall torque magnetometry technique introduced in ref. \(^34\), wherein current-induced torque from a spin Hall current injected from the Pt layer is used to probe DW orientation as a function of in-plane applied magnetic field. Samples were lithographically patterned into DW racetracks (Fig. 2a), with Au contacts at either end for current injection. An orthogonal Au strip line was used to nucleate DWs via an Oersted field from a short current pulse (see Methods). DW propagation driven by an out-of-plane field \(H_z\) in the presence of a longitudinal current density \(j\) was probed using a scanning polar magneto-optical Kerr effect (MOKE) polarimeter (see Methods).

Current flowing in the Pt strip generates a damping-like torque on DWs that acts like an out-of-plane magnetic field \(H_{\text{eff}} \equiv \chi j = \frac{\chi_0 j}{2} \cos(\psi)\), where \(\chi_0\) is proportional to the spin Hall angle of Pt (see Methods), and \(\psi\) is the angle between the DW moment and the current flow direction. \(H_{\text{eff}}\) was extracted from the variation \(\Delta H_{\text{dp}}\) of the DW depinning field under small d.c. current injection (see Methods). The slope of \(\Delta H_{\text{dp}}\) versus \(j\) (Fig. 2c) yields \(\chi_0\), which is finite at zero in-plane field, indicating an equilibrium Néel character to the DW \((\cos(\psi) > 0)\). \(\chi_0\) changes sign under positive \(H_z\) for up-down DWs, as shown in Fig. 2c, d, indicating reversal of the DW chirality from an initially left-handed spiral (leftward-oriented moment). See Supplementary Figs. 3–5 and Supplementary Note 2 for a full explanation of the measurement methodology.

Figure 3a shows \(\chi / \chi_0 \propto \cos(\psi)\) versus \(H_z\) for up-down and down-up DWs in GGG/TmIG (2.4 nm)/Pt (4.0 nm). The data indicate that both DW types are fully Néel at \(H_z = 0\), with oppositely oriented moments and hence with the same left-handed chirality, in agreement with other work\(^23,24\). The curves
in Fig. 3a are analogous to shifted hard-axis hysteresis loops of the DWs: the field required to rotate the DW moment from Néel ($\chi = \pm 1$) to Bloch ($\chi = 0$) is related to the DW shape anisotropy field $H_a$, and the zero-crossing of the loops corresponds to the DMI effective field $H_D$ that biases the DW in a Néel configuration. The solid lines in Fig. 3a are a fits of $\cos(\psi)$ versus $H_a$ using a simple 1D DW energetics model, yielding $H_a = 80$ Oe and $H_D = 70$ Oe (see Methods).

Figure 3b shows similar results for down-up DWs in films with varying $t_{TmIG}$, from which we find that $H_D$ decreases with increasing $t_{TmIG}$. We computed the DMI energy density $D = \mu_0 H_D M_r D$ using the DW width $\Delta$ obtained from $\Delta = M_r t_{TmIG} \ln(2)/\pi H_a$. Figure 3c shows that $D$ scales as $D/t$, where $t = (t_{TmIG} - t_{dead})$ is the effective magnetic layer thickness and $D$ is the interfacial DMI energy density. The $1/t$ dependence of the data confirms the interfacial origin of the DMI, with $D = 8.99 \times 10^{-12}$ mJ/m. We also find that $\chi$ scales inversely with $t$, implying a constant effective spin Hall angle $\theta_{SH}$ of 0.5% (see Methods). Together with the data in Fig. 1c, this indicates the absence of dimensional crossover effects at small $t_{TmIG}$ in our films, in contrast to the conclusions drawn from the data presented in ref. 30.

As shown previously$^{23,24}$, the DW chirality is opposite to that expected if the Pt interface were the source, since here the DWs are left-handed despite Pt being on top. This suggests that the iDMI in garnets has a different origin than in metallic ferromagnet/HM systems, since the SOC in Pt would be expected to generate right-handed chirality in the present layer geometry. Figure 4a shows spin Hall torque magnetometry data for TbIG (7.1 nm)/Pt (4.0 nm) and TbIG (7.1 nm)/Cu (2.0 nm)/Pt (4.0 nm) films, which show the influence of the RE ion and of the adjacent HM on the iDMI. Both films exhibit sizable iDMI, but when accounting for $M_r$ and $D$, we find $D = 0.0013$ mJ/m$^2$ for TbIG (7.1 nm)/Pt (4.0 nm), approximately four times smaller than for TbIG/Pt of similar thickness. Hence, changing the RE ion results in a substantial change in $D$. By contrast, inserting a thin Cu spacer only slightly changes the iDMI. Hence, the relative difference between having a strong SOC metal in direct contact with the garnet or not is small compared to the effect of changing the RE ion in the oxide itself. Indeed, the iDMI actually increases slightly upon Cu layer insertion, to $D = 0.0018$ mJ/m$^2$, which suggests that symmetry breaking at the oxide/metal interface plays a role in the iDMI, but that the strength of the SOC in the
metal at this interface does not play a significant role. The fact that the magnitude of the iDMI changes only weakly when changing the interfacing metal, whereas it varies strongly with changes in the RE orbital moment, indicates that it is the SOC of the RE ion at mirror symmetry-breaking interfaces that primarily determines the iDMI strength.

**Role of rare-earth orbital magnetism in DMI.** To verify the role of the RE ion in generating iDMI, we perform similar measurements using perpendicularly magnetized Bi-substituted yttrium iron garnet (BiYIG, see Methods) that lacks a RE ion. Here Bi and Y ions occupy the dodecahedral sites, with Bi yielding PMA when the film is under in-plane tensile strain. We examined a GSGG/BiYIG (6.9 nm)/Pt (4.0 nm) film, where Sc-doped GGG (GSGG, lattice parameter 12.554 Å) was used as the substrate to achieve the necessary tensile strain state, similar to the magnetic anisotropy, which originates from magnetoelastic coupling mediated by the RE ion. Figure 4c compares spin Hall torque magnetometry data for TmIG (6.0 nm)/Pt (4.0 nm) films grown on GGG and on substituted-GGG (SGGG), with lattice constants $a_{GGG} = 12.376\text{Å}$ and $a_{SGGG} = 12.480\text{Å}$, respectively. We find $D = 0.0026\text{mJ/m}^2$ for the film grown on SGGG, half the value that it takes in the less-strained film on GGG, which shows the clear influence of strain on the iDMI. A summary of these results is shown as a bar chart in Fig. 4d.

DMI requires both broken inversion symmetry and strong SOC. In thin film heterostructures, an HM layer typically provides both of these ingredients. However, the results above suggest that in the RE garnets, the RE ion in the magnetic film itself provides the critical SOC responsible for DMI, irrespective of the SOC of the interfacing material. Although RE ions exist in...
Fig. 3 Thickness dependence of the chiral exchange energy in GGG/TmIG/Pt. a Normalized spin Hall efficiency ($\chi/\chi_0$) as a function of in-plane field ($H_x$) for up-down (open blue squares) and down-up (open red circles) domain walls. Insets are schematics of up-down and down-up domain walls at zero applied field. b $\chi/\chi_0$ for down-up domain walls in various thicknesses of GGG/TmIG/Pt films. TmIG thickness (t) dependence of the c Dzyaloshinskii–Moriya interaction energy ($D$) and d the spin Hall efficiency. Error bars in c are the propagated error of measured $M_s$, $\Delta$, and $H_D$ at each thickness, while the error bars in d are the standard error of 4–6 measurements at each thickness. Source data for c and d are provided as a Source Data file.

Fig. 4 In-plane field dependence of the spin Hall efficiency in Pt-covered insulating magnetic garnets. Normalized spin Hall efficiency ($\chi/\chi_0$) performed on up-down domain walls in a GGG/TbIG (7.1 nm)/Pt (4.0 nm) and GGG/TbIG (7.1 nm)/Cu (2.0 nm)/Pt (4.0 nm), b GGG/TmIG (6.0 nm)/Pt (4.0 nm) and GSGG/BiYIG (6.9 nm)/Pt (4.0 nm), and c GGG/TmIG (6.0 nm)/Pt (4.0 nm) and SGGG/TmIG (6.0 nm)/Pt (4.0 nm). Insets are layer schematics of each thin film system. d Summary of DMI ($D$) strengths in each system in a–c.
and negative (\( \mu^- \)) helicity and a difference spectrum (\( \mu^+ - \mu^- \)). z-component of the orbital (\( L_z \)) and spin (\( S_z \)) angular momentum and m (XMCD) to directly measure both contributions to the angular momentum from the integrated XMCD signal, which are plotted in Fig. 5b as a function of temperature (\( T \)). From these data, we computed the Tm\(^{3+}\) moment per formula unit, \( m_z = -\mu_B L_z - 2\mu_B S_z \) (Fig. 5c), and the normalized SOC L \cdot S (\( \propto H_{SO} \)) per formula unit (Fig. 5d) as functions of \( T \). As L is quenched in Fe\(^{3+}\), it is not expected to contribute to the SOC in TmIG, whereas the SOC from Tm\(^{3+}\) is significant and strongly \( T \)-dependent. An XMCD-measured Gd hysteresis loop performed on the GGG/TmIG(2.4 nm)/Pt sample is shown in the Supplementary Fig. 6 and described in Supplementary Note 3, illustrating Gd paramagnetism. Thus, we can confirm that substrate Gd and any Gd that diffused into the interfacial dead layer is not ferrimagnetic at room temperature.

If the RE-mediated SOC is responsible for the iDMI in these materials, one would expect a similarly strong dependence of the iDMI on \( T \). To confirm this, we measured the DMI effective field \( H_D \) in GGG/TmIG(\( f_{\text{TmIG}} \))/Pt (4.0 nm) as a function of \( T \), as shown in Fig. 6a, b for \( f_{\text{TmIG}} = 2.4 \) and 6.0 nm, respectively.
Similar to the SOC in Tm$^{3+}$, we find that $H_2O$ in both samples scales nearly linearly with $T$ in the measured range, providing further evidence that the IDMI in REIG films originates from SOC coupling in the RE ion.

Discussion

In summary, we examined the chiral Dzyaloshinskii–Moriya exchange interaction in ultrathin centrosymmetric insulating magnetic garnets. We find that the origin of the DMI is interfacial, and that in substrate/REIG/metal trilayers, both the substrate and the metal overlayers contribute to the strength of the DMI. Nevertheless, our experiments cannot separate the IDMI contribution from the top and bottom interfaces, since we only measure a single, net IDMI. However, in contrast to metallic heavy-metal/ferromagnet heterostructures, in which the SOC in the heavy metal plays a critical role, in magnetic garnets, it is the SOC in the oxide itself, arising from RE orbital magnetism, that is responsible for the IDMI. We find that the IDMI can be tuned with substrate strain, and that in TmIG it is strongly temperature dependent, tending to vanish at about 350 K. Intriguingly, we find that the Tm$^{3+}$ SOC increases by about a factor of 40 at low temperature compared to room temperature, which suggests that the IDMI at low $T$ may likewise show a giant enhancement. While these results may give pause to the interpretation of anomalies in the Hall effect signals at elevated temperatures as resulting from chiral magnetic states, they may point to chiral phases such as spontaneous skyrmion lattices that could emerge at low $T$ where IDMI is expected to be substantial. These results also provide critical insights that could allow the IDMI to be substantially enhanced, by engineering the RE content and SOC in garnets, and perhaps in magnetic oxides generally. Our observations hence point to new avenues and opportunities to engineer chiral magnetism in magnetic oxides through RE ion substitution.

Methods

Growth, characterization, and patterning of materials. TmIG and TbIG films were deposited using PLD on single-side-polished, single-crystal GGG and SGGG substrates following a previously reported method. The PLD used a 248 nm wavelength KrF excimer laser with 10 Hz repetition rate and a heated substrate stage. The TbIG was commercially available from a TbIG target with a 99.9% elemental purity. The target–substrate distance was fixed at 8 cm. Epitaxial growth of the films was confirmed using X-ray diffraction 2θ–ω scan of the (444) reflection. Film thickness was determined by X-ray reflectometry. Atomic force microscopy RMS roughness measurements were carried out in a Digital Instruments NanoScope IV with a 1 μm × 1 μm scan size and XRD measurements were carried out in a Bruker D8 Discover HRXRD. Tapping-mode AFM of the 2.4-nm-thick TmIG sample used in this study gave an RMS roughness of 0.64 nm. Cross-sectional samples for scanning transmission electron microscopy (STEM) were prepared using Ga$^+$ ion milling by a focused ion beam (Thermo Fisher Strata 400). The ion beam was reduced to 2 keV for the final thinning step to reduce sample damage. STEM experiments were carried out using a probe aberration corrected microscope (Thermo Fisher Titan Temis) operating at 300 keV and with a probe-forming semi-angle of 21.4 mrad.

The BiYG (6.9 nm) film was deposited by PLD on single-side-polished, single crystal GGG (111) substrates using a 248 nm KrF excimer laser of fluence ~2 J/cm$^2$ and a laser repetition rate of 10 Hz. The stoichiometric Bi$_{1.8}$Fe$_{1.7}$O$_{12}$ target was prepared from Fe$_2$O$_3$ and Bi$_2$O$_3$ powder by a mixed oxide sintering method. The chamber was pumped to 5 × 10$^{-7}$ Torr base pressure prior to oxygen introduction and depositing the targets. The target–substrate distance was fixed at 6 cm. During deposition the substrate temperature was 560 °C and the oxygen pressure was 100 mTorr. The films were cooled to room temperature at 10 °C min$^{-1}$ and 225 Torr oxygen pressure. High-resolution X-ray diffraction (HRXRD) 2θ–ω scans of the (444) reflection reveal the epitaxial growth of the films and film thicknesses were determined by X-ray reflectometry.

The perpendicular easy axis of the REIG films is due to magnetoelectric anisotropy arising from epitaxial growth on the GGG and SGGG substrates. TmIG is under in-plane tensile strain with a negative magnetostriiction coefficient $\lambda_{111}$ (ref. 39); TbIG is under in-plane compression with a positive $\lambda_{111}$ (ref. 37). Replacing Y with Bi in dodecadial sites of BiYG in sufficient quantities expands the lattice parameter and increases the magnitude of the negative magnetostriiction coefficient; growth on substituted garnet substrates leads to tensile strain yielding an out-of-plane magnetic easy axis.

Cu and Pt metallic overlayers were grown using d.c. magnetron sputtering with an Ar sputter gas pressure of 3.5 and 3.5 mTorr, respectively, and a background base pressure of 10$^{-7}$ Torr. Deposition rates were calibrated using X-ray reflectivity measurements. DW motion tracks were patterned using standard photolithography and ion milling. The contact pads (Ta (6 nm)/Au (150 nm)) were patterned using photolithography and lift-off processes. Spin Hall Torque magnetometry measurements were performed on 50 μm × 40 μm tracks.

Magnetooptical Kerr effect (MOKE). Polar MOKE measurements were acquired on a custom-built, three-axis scanning Kerr microscope with independent out-of-plane and in-plane magnetic field control. The light source was a continuous-wave 445 nm diode laser focused with a x10 objective to a spot size of ~8 μm. The laser was attenuated to ~3 mW to prevent heating on the sample. Temperature was controlled via an integrated flow cryostat.

Spin Hall torque magnetometry. To characterize the effective field $H_{ext}$ from the spin current generated by the Pt overlayer, we followed ref. 34 and measured the depinning field $H_{dp}$ of a DW under a d.c. bias current in lithographically defined DW racetracks. The black curve in Fig. 2b shows a polar MOKE hysteresis loop for GGG/TmIG (6.0 nm)/Pt (4.0 nm). During the field sweep, a DW is nucleated by passing a short current pulse through the overlaid Au stripe during the zero-field crossing of the positive field sweep. As the field ramps, the DW propagates towards the focused MOKE laser spot (Fig. 2a). This results in a asymmetric hysteresis loop, where the negative switching field corresponds to the DW nucleation field of the sample $H_{SW}$, and the positive switching field is the magnetic field required to transition an already nucleated DW to the DW $H_{dp}$. To measure the effect of the spin–orbit torque from the Pt overlayer, a current is injected concurrently with the propagating DW to help or hinder its motion. This results in a decrease or increase in $H_{dp}$ (Fig. 2b; red and blue curves, respectively). The change in propagation field $\Delta H_{prop}$ with in-plane field is the spin Hall efficiency $\chi$.

We fit the spin Hall torque magnetometry data to $\chi = \frac{\psi}{2 \pi} \cos(\psi)$, where $\psi$ is the angle between the DW moment and the x-axis, $\psi$ is computed as a function of $H_x$ and $H_y$ using a 1D DW model with DMI (ref. 39), where the DW surface energy density $\sigma$ is written as

$$\sigma = \frac{H_{c}^{2} \cos^{2}(\psi)}{2 \pi} - \frac{\pi}{2} H_{K} \cos(\psi) - \frac{\pi}{2} H_{2} \cos(\psi) - \frac{\pi}{2} H_{3} \cos(\psi) + H_{c}.$$
by variation of the actual integration limits. The magnetic moments per formula unit ion were then calculated as $L_z = 2N_0 q_i / r$ and $S_{\text{eff}} = hN_0 (2p - 3q)/r$, where $N_0 = 6$ is the number of empty $4f$ states of Tm per formula unit (2 per ion) and $S_{\text{eff}} = 2S + J_0$ is the effective spin density that includes the magnetic dipole moment $J_0$.$^{45}$ Assuming that the ratio of $S/T_0$ is constant, we have used literature values for the free ion $S_{\text{free}} / T_0 = (-0.991) / (-0.407)$.$^{45}$ to calculate $S_i$ from the measured $S_{\text{eff}}$ as $S_i = S_{\text{eff}} / (2 + J_0 / S_{\text{free}})$.$^{45}$

Data availability

The data that support the findings of this study are available from the corresponding author upon reasonable request. The source data underlying Fig. 1g, 3c, d, 6 are provided as a Source Data file.

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Author contributions

L.C., C.A.R., and G.S.D.B. conceived the project and planned the experiments; E.R. synthesized and characterized the TmIG and T56C samples; T.F. synthesized and characterized BiYIG films; L.C. and P.R. performed vibrating sample magnetometry; L.C. and P.R. performed X-ray absorption and scattering experiments at the ALBA synchrotron light source. J. Synchrotron Radiat. 23, 1507–1517 (2016).

Competing interests

The authors declare no competing interests.
