Large spin-orbit torque efficiency enhanced by magnetic structure of collinear antiferromagnet IrMn

Jing Zhou1, Xiao Wang2, Yaohua Liu3, Jihang Yu1, Huixia Fu4, Liang Liu1, Shaohai Chen1, Jinyu Deng1, Weinan Lin1, Xinyu Shu1, Hering Yao Yoong1, Tao Hong3, Masaaki Matsuda3, Ping Yang1,5, Stefan Adams1, Binghai Yan4, Xiufeng Han2, Jingsheng Chen1,6

Spin-orbit torque (SOT) offers promising approaches to developing energy-efficient memory devices by electric switching of magnetization. Compared to other SOT materials, metallic antiferromagnet (AFM) potentially allows the control of SOT through its magnetic structure. Here, combining the results from neutron diffraction and spin-torque ferromagnetic resonance experiments, we show that the magnetic structure of epitaxially grown L10-IrMn (a collinear AFM) is distinct from the widely presumed bulk one. It consists of twin domains, with the spin axes orienting toward [111] and [−111], respectively. This unconventional magnetic structure is responsible for much larger SOT efficiencies up to 0.60 ± 0.04, compared to 0.083 ± 0.002 for the polycrystalline IrMn. Furthermore, we reveal that this magnetic structure induces a large isotropic bulk contribution and a comparable anisotropic interfacial contribution to the SOT efficiency. Our findings shed light on the critical roles of bulk and interfacial antiferromagnetism to SOT generated by metallic AFM.

INTRODUCTION
Spin-orbit torque (SOT) has been extensively investigated due to its application in the electric switching of magnetization (1–5). Magnetic memory based on SOT is considered to have higher speed and lower energy consumption than spin transfer torque magnetic random access memory (6, 7). In a ferromagnetic/heavy metal (FM/HM) bilayer, SOT arises from the spin Hall effect (SHE) in HM (1, 2) and/or the Rashba-Edelstein effect at the interface (3–5). In the case of SHE, a spin current generated from HM due to spin-orbit interaction is transferred to FM and exerts a spin transfer torque on the magnetic moment. The magnetization switching efficiency therefore depends substantially on the overall efficiency of SOT, which includes the charge-to-spin conversion efficiency and the transmission efficiency of spin current into the adjacent FM layer. To date, evidences of large SOT efficiency have been observed in topological insulators (6, 8, 9), HM (10–14), and metallic antiferromagnet (AFM) (15–18). These materials, in general, have strong bulk and/or interfacial spin-orbit coupling (SOC), which accounts for the intrinsic contribution to SOT.

Studies on SOT of metallic AFM, in addition, emphasize the roles of antiferromagnetism. In L12-IrMn12, the triangular arrangement of magnetic moment within the (111) plane results in considerable spin Hall conductivity (SHC) in the [001] direction, which is associated with a facet-dependent SOT efficiency (15). Zhang et al. (19) demonstrate anisotropic SOT efficiency in L10-IrMn, which is attributed to the changed spin texture associated with different epitaxial growth directions, although the measured spin Hall angles (SHAs) are not notably different from those of polycrystalline IrMn. The interfacial exchange coupling in an AFM/FM bilayer has also been shown to influence the SOT efficiency (15–17, 20). A higher SOT efficiency is observed in IrMn/FM with a larger exchange bias, but the direction of exchange bias appears to have little effect (15, 16). Sometimes, conflicting results on the influence of exchange bias are also reported (16, 20). The measured SOT efficiencies of IrMn are very scattered and range from 0.02 to 0.35 (15–17, 19, 20). These imply multiple, and perhaps conflicting, factors that may affect SOT of IrMn.

We perceive the above controversies as two related problems that have not been sufficiently addressed. First, the magnetic structure of IrMn is not defined while probing its SOT. Ideally, IrMn can exist in three forms with distinct crystallographic and magnetic structures, namely, the γ, L10, and L12 phases (17, 21). In practice, the strength of SOT of IrMn is substantially lowered due to the averaging effect of randomly oriented crystallites and domains (17). The effects from substrate, such as epitaxial strain, further complicate this. The assumption of magnetic structure based on merely atomic composition or crystal structure is therefore weak. Second, the interfacial contribution to SOT from the IrMn/FM bilayer is not well separated from the bulk contribution. Previous studies frequently use field cooling (after annealing) to tune the magnetic structure of the IrMn/FM bilayer (15, 20). This technique results in the simultaneous changes in the bulk and/or interfacial magnetic orders in AFM, the degree of crystallization in both layers, as well as interfacial mixing, all of which influence the measured SOT efficiency. Moreover, the magnetic damping constant of the FM in the exchange-biased AFM/FM bilayer is usually enhanced (16, 20). This frustrates the initiatives of energy-efficient memory, where a small damping constant is favored (6). Although field-free switching of magnetization using SOT and exchange bias has been realized in IrMn/CoFeB (5), the identification of useful sources of SOTs in IrMn and the effective ways to engineer them remain elusive.

Here, we report the epitaxial growth of the L10-IrMn (001) thin film of well-defined crystal and magnetic structures. Neutron diffraction measurement shows that the antiparallel Mn spin moments in our L10-IrMn tilt away from the crystallographic c axis by 56° ± 8°, which is different from bulk L10-type AFM (17, 22–24). The measured SOT efficiency in L10-IrMn/Ni81Fe19 [also known as permalloy (Py)] reaches up to 0.60 ± 0.04 and displays a fourfold anisotropy. This anisotropy coincides closely with the two types of domains in L10-IrMn,
which align with [111] and [−111], respectively. After breaking the interfacial exchange coupling with a Cu spacer, the SOT efficiency drops substantially to 0.22 ± 0.03, and the fourfold anisotropy vanishes completely. Our results suggest that, in addition to a large bulk component of SOT efficiency in L1₀-IrMn, a comparable interfacial contribution dependent on the magnetic structure exists. By passing electric current along different crystal directions of L1₀-IrMn in a L1₀-IrMn/Pt bilayer, the strength of current-induced SOT can be controlled.

RESULTS
Thin films in the stacking of IrMn/Pt are deposited on KTaO₃ (001) substrates. See more details in Materials and Methods. Figure 1B shows a representative θ-2θ x-ray diffraction (XRD) pattern of the IrMn film. The presence of all peaks in the (001) family and the absence of the (200) peak indicate a high degree of (001) texture. It also rules out the presence of the γ phase, which would show only a (002) peak at 48° due to the face-centered cubic (FCC) crystal structure (21). With Rietveld whole powder pattern fitting of the XRD measurements using the General Structure Analysis System (GSAS) software suite (25), the occupancy of Mn on the Ir site and vice versa is estimated to be (8 ± 2)%, which corresponds to a chemical ordering of \( S_{\text{IrMn}} = 0.84 ± 0.04 \). Figure 1 (C and D) shows the reciprocal space mappings (RSMs) of (103) and (113) planes, respectively. The presence of the (113) peak and the absence of the (103) peak are consistent with the characteristic selection rules of the L1₀ superlattice, where the sum of \( h + k = \text{even} \). The RSMs also rule out the existence of L1₂-IrMn due to the absence of the (103) peak. The lattice constants determined from the (113) RSMs are \( a = b = 3.868 ± 0.004 \) Å and \( c = 3.638 ± 0.001 \) Å \((c/a = 0.941)\). The lattice constant \( a \) in our films is slightly larger than the bulk value (3.855 Å), while the lattice constant \( c \) is slightly smaller than the bulk value (3.644 Å) (17). These findings can be attributed to the tensile strain from the lattice mismatch with the KTaO₃ (001) substrate \((a = 3.989 \) Å\), which stretches \( a \) and compresses \( c \). Similar results have been observed in L₁₀-FePt films, where it is also found that an appropriate strain will favor the formation of the L₁₀ structure (26).

The microstructure of the IrMn film is examined using high-resolution transmission electron microscopy (HRTEM), as shown in Fig. 1E. The IrMn film grows epitaxially on the KTaO₃ (001) substrate. The reciprocal lattices at the substrate, film, and interface are similar except for the absence of the (103) peak (RSMs around (113) and (103) planes, respectively). HRTEM image of cross section of L₁₀-IrMn thin film. Diffraction patterns from the substrate, interface, and IrMn are shown on the right.

Fig. 1. Structure of L₁₀-IrMn. (A) Schematic drawing of L₁₀-IrMn unit cell. (B) XRD θ-2θ scan of IrMn along the (001) direction. Dotted lines show the reference peak positions of bulk L₁₀-IrMn, a.u., arbitrary units. (C and D) RSMs around (113) and (103) planes, respectively. (E) HRTEM image of cross section of L₁₀-IrMn thin film. Diffraction patterns from the substrate, interface, and IrMn are shown on the right.

\[ q^2 = 1 - \frac{l^2 + \left(\frac{1}{2}(h^2 + k^2) - l^2\right)\sin^2(\psi_c)}{c^2(h^2 + k^2) + l^2} \] (1)

Here, \( \psi_c \) is the angle between the magnetic moment and the [001] direction. Assuming no magnetic moment on Ir atom and a collinear arrangement of magnetic moment on Mn, \( \psi_c \) can be evaluated from the ratio \( R = \frac{c}{l_m} \). The measured \( R \) is 1.64 ± 0.26, which is equivalent to a 56° ± 8° (Fig. 1A) tilting angle with the \( c \) axis (see Supplementary Materials and Methods). The in-plane orientation of the magnetic moment cannot be determined from neutron diffraction due to the tetragonal symmetry of the L₁₀ phase (22-24). Theoretical calculations on L₁₀-type AFMs predict that the staggered magnetic moments on Mn are either parallel or perpendicular to the [001] direction (22, 23), which correspond to \( R \) being 3.58 or 1.09, respectively, in perfectly ordered L₁₀-IrMn. Thus, taking the error bar in measured \( R \) into account, our L₁₀-IrMn is still considered to exhibit spin canting, which might be possibly due to the small strain induced from the lattice mismatch with the substrate (24). In addition, the existence of small chemical disorder might also influence the alignment of magnetic moments.
The estimated magnetic moment is $3.54 \pm 0.27 \mu_B$ per Mn atom by comparing the integrated peak intensities of the magnetic peaks and the nuclear peaks, which is comparable to previous work (22, 23).

We use the spin torque ferromagnetic resonance (ST-FMR) technique (10, 15, 20, 27) to evaluate the SOT efficiency of L10-IrMn in a L10-IrMn $(d_{AF})$/Py $(d_t)$ bilayer (see Supplementary Materials and Methods). $d_{AF}$ and $d_t$ are the thicknesses of the IrMn and Py in nanometers, respectively. The stacking is patterned into the microstrip by photolithography and ion milling. A microwave is applied to the microstrip, while an external magnetic field ($H$) in the sample plane is swept at $45^\circ$ with respect to the microstrip, as illustrated in Fig. 3A. While precessing, the magnetic moment in the Py layer experiences a field-like torque ($\tau_{FL}$) and a damping-like torque ($\tau_{DL}$). In the original ST-FMR model on Pt/Py (27), the Oersted field ($h_{oer}$) of microwave is considered to be the only contribution to $\tau_{FL}$, while $\tau_{DL}$ is mainly caused by spin current ($J_s$) converted from charge current ($J_t$) flowing in the Pt (IrMn in our case) layer due to SHE. The precession of the magnetic moment leads to an anisotropic magnetoresistance effect in Py. This, when mixing with the alternating current, produces a rectifying DC voltage $V_{mix}$ (zeroth harmonic) over the microstrip. As shown in Fig. 3C, $V_{mix}$ can be decomposed into a symmetric component (S), which is associated with $\tau_{DL}$, and into an antisymmetric component (A), which corresponds to $\tau_{FL}$. Thus, the SOT efficiency is expressed as $\theta_{DL,m} = \frac{S}{A} = \frac{M_{sat,DL}}{M_{sat,FL}} \sqrt{1 + \frac{H_{res}}{H_{eff}}} (10, 15, 20, 27)$. Here, $M_s$ and $M_{sat}$ are the saturation magnetization and effective in-plane magnetization of the Py layer; $H_{res}$ is the resonant field. In the absence of other sources of SOTs under the drift-diffusion model, the SOT efficiency here represents the overall charge-to-spin conversion efficiency and describes the lower bound of a traditional SHA, considering the losses of spin current density at the interface. As we shall discuss later, when addition-
**Fig. 3. Measurement of SOT efficiency ($\theta_{DL,m}$) from ST-FMR.** (A) Schematics of measurement setup. The moment $m$ in Py follows an elliptical precession route around the direction of $H$. It is influenced by two orthogonal torques $t_{FL}$ and $t_{DL}$. Top right shows the optical image of the device and electrode (dark color). (B) Voltage spectra of L10-IrMn (22)/Py (17) measured from 8 to 12 GHz with nominal input power of 18 dBm. (C) Typical fitting of $V_{mix}$ at 9 GHz. $V_{sym}$ and $V_{asym}$ correspond to the symmetric and antisymmetric components, respectively. (D) Fitting of Kittel equation. (E) $\theta_{DL,m}$ of L10-IrMn, p-IrMn, and Pt. The error bar describes 1 SD over at least five devices.
the resonance. It is worth noting that \( H \) is always at 45° with the microstrip during measurement (Fig. 3A). Thus, the two easy axes of magnetic anisotropy in Py are oriented along \([100]\) (0°) and \([010]\) (90°) of the \(\text{L}_10\)-IrMn lattice, respectively. When a polycrystalline FM is exchange-coupled with a single-crystal AFM with an ordered magnetic structure, a fourfold magnetic anisotropy in FM can be induced by the magnetic anisotropy of AFM (34, 35). More specifically, a collinear AFM with orthogonal domains, such as twinned MnF\(_2\) (110), induces two orthogonal uniaxial anisotropies in the FM film plane. The two uniaxial terms ultimately cancel each other, leaving a fourfold residue at 45° with AFM easy axes (35), as schematically shown in Fig. 6A. The bilayer of \(\text{L}_10\)-IrMn/Py is highly consistent with the above scenario, implying two types of antiferromagnetic domains in \(\text{L}_10\)-IrMn, who in-plane projections align with \([110]\) (45°) and \([-110]\) (135°), respectively (Fig. 6B). Given the measured \(\psi_c = 56° \pm 8°\) and the fact that the angle between \([001]\) and \([111]\) is 54.7°, our \(\text{L}_10\)-IrMn thin films are likely to consist of two domains parallel to \([111]\) and \([-111]\). In addition, the exchange bias field in the bilayer is relatively small since the two values of \(H_{\text{ex}}\) on any measured axis are roughly identical. This is expected based on our fabrication process, where no bias field during growth or post-field annealing is applied. Overall, the results in Fig. 5 confirm an ordered twin-domain spin texture at the \(\text{L}_10\)-IrMn/Py interface, which induces both the anisotropic \(\theta_{\text{DL,}}\) of \(\text{L}_10\)-IrMn and the magnetic anisotropy of Py.

### DISCUSSION

The large \(\theta_{\text{DL,}}\) in our \(\text{L}_10\)-IrMn is rationalized from the following perspectives. We evaluate the intrinsic SHC by ab initio Berry phase calculations based on the band structure and wave functions of the pure \(\text{L}_10\)-IrMn single crystal without any disorders (see Supplementary Materials and Methods). The intrinsic SHC depends on the electric current direction relative to the spin direction. It would exhibit a twofold in-plane anisotropy only for the case of a single domain (fig. S7). Vector sum of twin domains with orthogonal in-plane projections results in an isotropic bulk contribution to \(\theta_{\text{DL,}}\). This is consistent with Fig. 5F. The gap between \(\theta_{\text{DL,}}\) of \(\text{L}_10\)-IrMn and p-IrMn is likely to be a result of differed crystal and magnetic structures. In p-IrMn, both the crystallites and magnetic domains are randomly oriented, leading to an averaged SHC substantially smaller than the case of single crystal and single domain (17). The measured SHC (>2500 \(\hbar e S/cm\)) (Supplementary Materials and Methods), though much larger than the simulated intrinsic one (<200 \(\hbar e S/cm\)), is in the same order of magnitude as the experimentally obtained SHC in other works (15). A previous study has observed a similar gap between measured and simulated SHC, which is rationalized by the extrinsic contribution to SHE (15). Since our \(\text{L}_10\)-IrMn is not fully chemically ordered with a small amount of Ir atoms wrongly occupying Mn position and vice versa, the small chemical disorder may be treated as impurities, which enhance the skew scattering (6, 13, 14) and thus an extrinsic contribution to \(\theta_{\text{DL,}}\).
As reported by numerous authors, the electrical resistivity of the spin Hall layer can affect the SOT efficiency $\theta_{DL,m}$ (6, 11, 12). We measure the electrical resistivity of single-layer L1$_0$-IrMn and the electrical resistance of the bilayer microstrip. Both show little dependence on the device orientation (fig. S8).

Therefore, even in a highly ordered L1$_0$-IrMn with well-defined crystal ($s_{IrMn}$) and magnetic ($\psi_r$) structures, $\theta_{DL,m}$ cannot be explained fully on a bulk basis. We attempt to consider the anisotropic interfacial contribution to $\theta_{DL,m}$ in the drift-diffusion model plus interfacial spin-mixing conductance (10, 28, 29, 36). That is, an isotropic spin current (and thus isotropic $\tau_{DL}$ as in Fig. 5F) is generated from bulk L1$_0$-IrMn due to SHE and scattered anisotropically at the interface. The transparency due to electronic band mismatch (10, 36), spin memory loss due to disorder (36, 37) and SOC (10, 37), and spurious voltage signal due to spin pumping and inverse spin Hall effect (ISHE) are likely to be independent of device orientation. The spin-dependent scattering, such as those analogous to the spin Hall magnetoresistance (SMR) effect (38), exhibits a fourfold ($\cos^4\theta$) dependence with the spin direction, and would be averaged because of twin-domain spin texture at the interface. Furthermore, all contributing factors above, except for ISHE, result in smaller $\theta_{DL,m}$ instead of an enhancement observed in this work.

The incompatibility of the model above implies the possibility of SOTs generated at the interface. As discussed by Kim et al. (28) and Haney et al. (29), considering a three-dimensional (3D) Rashba model with strong interfacial SOC, large $\tau_{DL}$ and $\tau_{IS}$ of comparable magnitudes can be generated from the interface. In the context of collinear AFM, the interface can be considered in a 2D Rashba model where inversion asymmetry exists in each sublattice, leading to a net Neel-order SOT of $\tau_{DL}$ nature (39). These torques have a strong dependence on the spin axis direction, which points to the possibility of the four-fold anisotropic $\theta_{DL,m}$ observed in this work. Since $\theta_{DL,m}$ only reflects the fraction of $\tau_{DL}/\tau_{IS}$, $\theta_{DL,m} \neq 0.60 \pm 0.04$ does not necessarily indicate more $\tau_{DL}$ per unit electric current compared to $\theta_{DL,m} = 0.22 \pm 0.03$. Therefore, the technical significance of an observed large SOT efficiency in exchange-coupled AFM/FM bilayers should be treated with care for future experimental design and application.

Our work also provides new insights into one of the seemingly contradictory results in the past. We notice in (16) that the exchange bias field is induced by an external field during film deposition. In contrast, the exchange bias field (20) is produced by field cooling (after annealing). These two techniques result in starkly different IrMn/Py interfaces. Annealing would suppress the interfacial SOT by reducing the inversion asymmetry, as discussed by Garello et al. (3). In addition, interfacial disorder enhanced by annealing would also reduce the bulk contribution. Thus, Saglam et al. (20) observed a slight reduction in SOT efficiency after Py is magnetically coupled with IrMn, whereas Tshitoyan et al. (16) reported a huge increment similar to our case.

In addition, the dependence of current-induced SOT on the crystal direction has been reported in WTe$_2$, where the authors control the magnitude of a novel damping-like SOT via the broken crystal symmetry of WTe$_2$ (40). The anisotropic SOT efficiency in our work, though of a different physical origin, shares a similar phenomenological behavior and technical significance.

We are grateful toward one reviewer for suggesting a harmonic Hall voltage measurement as an independent method of evaluating the SOT in L1$_0$-IrMn/Py. The results are consistent with the findings of ST-FMR (see Supplementary Materials and Methods).

In summary, we have demonstrated the critical roles of crystal and magnetic structures in determining the SOT efficiency of L1$_0$-IrMn, which were insufficiently addressed and sometimes ignored previously. The L1$_0$-IrMn thin films in this work have high chemical order and a twin-domain magnetic structure different from bulk L1$_0$-type AFM. These account for the large SOT efficiency of more than $0.22 \pm 0.03$ in bulk L1$_0$-IrMn, in contrast to the small $\theta_{DL,m} \pm 0.083 \pm 0.005$ in p-IrMn. By measuring the in-plane angle dependence of the ST-FMR response, we find a substantial enhancement of SOT efficiency up to $0.60 \pm 0.04$ due to the spin texture of the L1$_0$-IrMn/Py interface. We stress that the fourfold anisotropies in $\theta_{DL,m}$ and the magnetic anisotropy of Py, though they highly

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**Fig. 5.** In-plane angle dependence of SOT efficiency ($\theta_{DL,m}$) and resonance condition ($H_{res}$). (A) Schematic illustration of device orientation. The blue rectangle illustrates that multiple devices are patterned from the same continuous film. (B, D, and F) Normalized $\theta_{DL,m}$ at 9 GHz of p-IrMn (22)/Py (13), L1$_0$-IrMn (22)/Py (13), and L1$_0$-IrMn (22)/Cu (0.5)/Py (13), respectively. (C, E, and G) Normalized resonant fields ($H_{res}$) at 9 GHz of devices in (B), (D), and (F) respectively. The angle refers to the orientation of the microstrip in the film plane relative to the [100] direction for L1$_0$-IrMn samples and an arbitrary axis for the p-IrMn sample.
resemble each other, are not causal. Both are considered the results of L1₀-IrMn’s magnetic structure.

**MATERIALS AND METHOD**

**Thin-film deposition**

Thin films were deposited by DC magnetron sputtering with a base pressure of less than 2 × 10⁻⁸ torr. In an attempt to produce high-quality L1₀-IrMn and to avoid problems with post-annealing, we deposited IrMn at elevated temperature. We tried a few substrates of small lattice mismatch, among which L1₀-IrMn of high chemical order and sufficiently smooth surface was produced only when the KTaO₃ substrate was used. Because of island growth at high temperature, L1₀-IrMn of less than 10 nm became discontinuous with high roughness and large electrical resistivity. Therefore, the thickness of our L1₀-IrMn was larger than past reports. In addition, considerable loss of Mn was observed at temperatures higher than 500°C, probably due to its high vapor pressure. Thus, our L1₀-IrMn was produced by first baking the KTaO₃ (001) substrate at 720°C for 30 min and then co-sputtering 5.08-cm Ir₄₀Mn₆₀ targets at the same temperature. Polycrystalline FCC IrMn as well as polycrystalline Py and Pt were deposited in situ at room temperature. All polycrystalline samples were deposited on the SiO₂ substrate. Because of strong exchange coupling between Py and L1₀-IrMn, Py of less than 10 nm exhibited a substantially larger coercive field and linewidth. The former introduced measurement artifacts in ST-FMR for smaller frequencies; the latter reduced the signal-to-noise ratio. Therefore, Py in this work was also thicker than past reports. A 2-nm SiO₂ layer was deposited by radio frequency (rf) magnetron sputtering on all samples as the protective layer. In this work, the nominal thicknesses of films were calibrated using an atomic force microscope. The real thicknesses were obtained from HRTEM, which were typically around 85% of the nominal ones. Thus, a nominal 25 nm became 22 nm and a nominal 15 nm became 13 nm, etc.

**Device fabrication**

Devices for ST-FMR measurement were fabricated using a two-step method. First, the microstrip with a width of 10 to 30 μm and a length of 40 to 60 μm was patterned. We verified that different dimensions here did not affect the measured SHA. They only affected the resistance of the channel. Most of the data collected in this work were obtained using 30 μm × 50 μm devices. Second, the SiO₂ layer on the terminals of the microstrip was removed by etching before depositing the electrodes, which were typically Ti (5 nm)/Cu (100 nm). The electrode was deliberately bent in the clockwise direction, as shown in Figs. 3A and 5A, to match the position of rf probe in our system.

**ST-FMR measurement**

A Rohde & Schwarz SMB 100A signal generator was used to generate microwave with a frequency of 8 to 12 GHz and a nominal power of 18 dBm. We verified that power in the range of 10 to 18 dBm did not affect the measured SHA but only the signal-to-noise ratio. The ST-FMR measurement was modulated with a sine function to improve the signal-to-noise ratio. A nonmagnetic GSG three-terminal probe manufactured by GGB Industries Inc. was used to make electrical contact with the electrodes. The probe was carefully rotated to ensure homogeneous contact. The modulated voltage output (Vmix) was collected using a Stanford Research SR830 lock-in amplifier.

**Material characterization**

XRD (θ-2θ scan) and RSM were performed at room temperature in Singapore Synchrotron Light Source (SSLS) with an x-ray wavelength of 1.541 Å (Cu-Kα). Using Rutherford backscattering spectrometry, we found that epitaxial IrMn of a wide composition window (45 to 60 at % Mn) exhibited “L1₀-like” peaks from a simple θ-2θ scan. Therefore, we relied on RSM instead of composition to determine the phase and quality of IrMn, which were also verified by the magnetic property from neutron diffraction. HRTEM was performed using a JEOL 2010F field-emission microscope at a voltage of 200 kV. The sample was prepared by focused ion beam (FIB) technology through FEI versa 3D dual beam system and was next attached on an Omniprobe lift-out grid. Neutron diffraction was performed at room temperature in the 0–θ mode with a neutron wavelength of 4.045 Å at the CG–4C at High Flux Isotope Reactor at Oak Ridge National Laboratory. The counts in Fig. 2 were fitted using a flat background plus one or two Gaussian functions, depending on whether there was an obvious contribution from the substrate.

**SUPPLEMENTARY MATERIALS**

Supplementary material for this article is available at http://advances.sciencemag.org/cgi/content/full/5/5/eaau6696/DC1

**Supplementary Materials and Methods**

Fig. S1. M-H loop of L1₀-IrMn/Py.

Fig. S2. Typical ST-FMR results of L1₀-IrMn (22)/Py (13) from 5 GHz to 12 GHz.
REFERENCES AND NOTES


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Author contributions: J.C. conceived the study and supervised the experiment. J.Z. built the ST-FMR measurement system, deposited thin-film samples, and performed ST-FMR measurement. X.W., J.Y., and H.Y.Y. fabricated microdevices. Y.L., T.H., and M.M. performed neutron diffraction experiment. H.F. and B.Y. carried out ab initio calculation. J.D. and P.Y. measured XRD and RSM. L.L. and W.L. measured electrical conductivity. S.C. deposited part of the thin-film samples. S.A. performed GSAS refinement. X.S. performed measurement on HRTEM. J.Z., B.Y., X.H., and J.C. wrote the manuscript. All authors helped to revise the manuscript and contributed to the final version.

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