

Polar nano-regions enable large spin Hall conductivity in metallic PtCoO₂

Received: 12 March 2026

Accepted: 18 May 2026

Published online: 12 June 2026

 Check for updates

Guiping Ji ^{1,14}, Yi Zheng ^{2,3,14}, Yuejie Zhang ^{1,14}, Yuhan Liang^{1,4,14}, Renhuai Wei ^{5,14}, Hang Zhou⁵, Yuanyuan Jiang⁵, Yang Zhang⁶, Hao Lin⁶, Hao Bai⁶, Qi Song⁷, Xiaojiang Li⁸, Sankalpa Hazra⁸, Aitian Chen ⁹, Cancheng Xiao¹, Dingsong Jiang¹, Chao Li ², Zewei Huang⁶, Yifeng Du¹⁰, Witold Skowroński ¹¹, Jinming Guo ¹⁰, Peng Li ¹², Xixiang Zhang ⁹, Jianshi Tang ¹, Venkatraman Gopalan ⁸, Darrell G. Schlom ⁷, Luyi Yang ⁶, Wanjun Jiang ⁶, Xuebin Zhu ⁵✉, Yuan-Hua Lin ², Di Yi ^{2,3}✉, Pu Yu ⁶, Dingfu Shao ⁵, Daniel C. Ralph ^{4,13}, Rong Yu ^{2,3} & Tianxiang Nan ¹✉

Spin-source materials for spintronic devices are required to convert charge currents into spin currents efficiently and maintain minimal electrical losses. However, conventional strategies to enhance charge–spin conversion often come at the cost of increasing ohmic dissipation. Here we demonstrate that introducing polar lattice distortions into a highly conductive metal can overcome this trade-off. We report the discovery of polar displacements in PtCoO₂ and show that these displacements enhance charge–spin conversion by two orders of magnitude compared with its non-polar phase. The coexistence of polarity and metallicity yields a room-temperature spin Hall conductivity of $1.6 \times 10^7 \hbar/2e (\Omega \text{ m})^{-1}$. Electron ptychography reveals that the inversion-symmetry breaking originates from local polar nano-regions. Polar PtCoO₂ drives efficient spin–orbit torque switching with substantially reduced switching voltage and power compared with Pt-based control devices, and is compatible with silicon substrates, establishing a strategy for designing highly efficient spin-source materials.

Spintronics, which exploits spin currents to process information^{1–3}, promises next-generation memory and computing applications beyond conventional electronics^{4,5}. In particular, spin–orbit torque (SOT) enables electric currents to manipulate the magnetization of nanoscale magnetic devices^{6,7}, opening up a range of emerging spintronic devices^{4,8–12}. One such application is SOT magnetic random-access memory (MRAM)^{7,13}, in which a charge current passed through a heavy-metal spin-source layer generates a pure spin current or spin accumulation via mechanisms such as the spin Hall effect², orbital Hall effect¹⁴ or the Rashba–Edelstein effects¹⁵. The resulting pure spin current diffuses into an adjacent magnetic bit (typically the free ferromagnetic layer of a magnetic tunnel junction), where it exerts a torque that flips the magnetization. Compared with commercialized spin-transfer-torque MRAM⁵, SOT-MRAM is expected to provide higher switching efficiency, faster operation speed using

spin-source materials, and greater device endurance by decoupling the read and write current paths in a three-terminal device geometry^{16,17}. This combination of non-volatility, high endurance and fast operation makes SOT-MRAM a compelling candidate for next-generation non-volatile memories.

One of the key limiting factors for SOT-MRAM is the spin-source material itself⁵. Its charge–spin conversion efficiency θ_{SH} and resistivity ρ together determine the key device metrics in SOT-MRAM, such as the write energy and memory cell size. Conventional spin-source materials (such as elemental heavy metals like Pt) typically exhibit low resistivity but only moderate charge–spin conversion efficiency^{18–20}. To enhance the charge–spin conversion efficiency, researchers are actively exploring new materials and novel charge–spin conversion mechanisms. In metal-based spin sources, introducing impurities can enhance the spin Hall effect, but this approach inevitably comes at the cost of higher

A full list of affiliations appears at the end of the paper. ✉e-mail: xbzhu@issp.ac.cn; diyi@mail.tsinghua.edu.cn

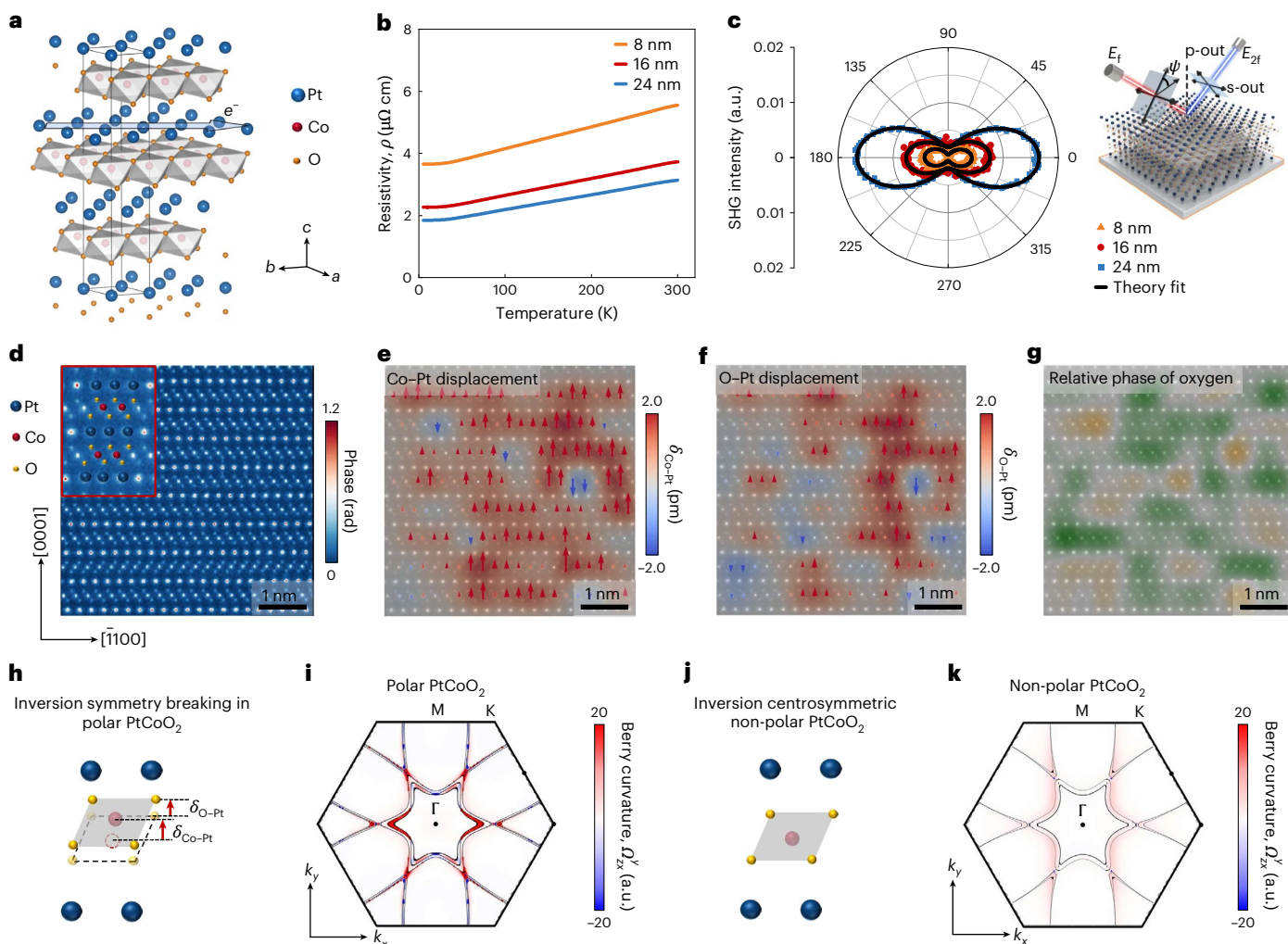


Fig. 1 | Experimental evidence of OOP polarity in PtCoO₂ thin films and DFT calculations. **a**, Schematic of the bulk PtCoO₂ crystal structure, showing alternating conductive Pt layers and insulating CoO₂ layers. **b**, Temperature-dependent resistivity of PtCoO₂ thin films with film thicknesses of 8 nm, 16 nm and 24 nm. **c**, Optical SHG signal E_{2f} of PtCoO₂ thin films as a function of incident polarization angle ψ of the incident beam measured in the p-out configuration. The solid lines are fit based on the functions derived from C_{3v} point symmetry (Methods). The inset illustrates the optical SHG measurement setup: an incident beam of frequency f with linear polarization angle ψ is directed onto the sample surface, and the reflected harmonic beam of frequency $2f$ is detected in either the p-out or s-out polarization. **d**, Ptychographic phase image of a 24-nm PtCoO₂ film. **e**, Spatial map of the Co-Pt displacement $\delta_{\text{Co-Pt}}$, defined as the offset of Co relative to the midpoint of the two nearest Pt atoms, extracted from **d**. **f**, Spatial map of the O-Pt displacement $\delta_{\text{O-Pt}}$, defined as the offset of the centre of two

nearest O ions relative to the midpoint of two nearest Pt atoms, also extracted from **d**. **g**, Distribution of the relative phase of oxygen in the Co-O octahedra, derived from **d**. The green and orange areas represent the relative phase difference between the nearest O in the upper and lower rows of each Co atom (Gaussian filtered with a width of 18 pixels to enhance the visualization). This pattern reveals the O intensity asymmetry within each CoO₂ octahedron. **h**, Schematics of the polar phase in PtCoO₂, showing the Co-Pt and O-Pt displacements. **i**, Momentum-space distribution of spin Berry curvature Ω_{yz}^y in the first Brillouin zone for the PtCoO₂ supercell containing a single oxygen vacancy, derived from the calculated band structures shown in Extended Data Fig. 5b. **j**, Schematics of the atomic structure in non-polar (centrosymmetric) PtCoO₂. **k**, Momentum-space distribution of spin Berry curvature Ω_{yz}^y in the first Brillouin zone for the pristine (centrosymmetric) PtCoO₂ supercell, derived from the calculated band structures shown in Extended Data Fig. 5d.

resistivity^{20–23}. For unconventional materials, topological insulators can produce large spin currents via spin-momentum locked states^{24–26}. Oxide interfaces²⁷ and transition metal oxides²⁸ also exhibit sizable charge–spin conversion that can be tuned by external electric-field²⁹ or crystalline symmetry^{30,31}. However, all of these unconventional materials suffer from high resistivity, meaning the spin-source layer made from them will be highly resistive. In practical devices, this leads to large resistive losses, high write voltages, substantial current shunting into the magnetic layer, slower operation and reliability issues. After over a decade of intensive investigation into spin-source materials, a universal trend has emerged: attempts to increase charge–spin conversion are almost invariably accompanied by higher resistivity^{4,20}. This trade-off is often quantified by the spin Hall conductivity (SHC), a figure of merit that combines charge–spin conversion efficiency and

electrical resistivity, expressed as $\sigma_s = \hbar/2e \frac{\theta_{\text{SH}}}{\rho}$ (where $\hbar$ and e are the reduced Planck's constant and electron charge, respectively).

Here we propose an alternative strategy to enhance the charge–spin conversion, that is, introducing a polar distortion (to break spatial inversion symmetry) in a highly conductive quasi-two-dimensional metal. It is known that breaking inversion symmetry at interfaces and surfaces (in materials with a strong spin–orbit coupling) can produce pronounced charge–spin conversion by lifting the spin degeneracy of electronic bands (via Rashba effects)¹⁵. Recent studies suggested that breaking inversion symmetry in bulk crystals can enhance charge–spin interconversion³², for example, in ferroelectric semiconductors³³ and in locally non-centrosymmetric oxides³⁴. Polar metals that are electrically conductive and lack inversion symmetry³⁵ become a promising material platform for achieving efficient

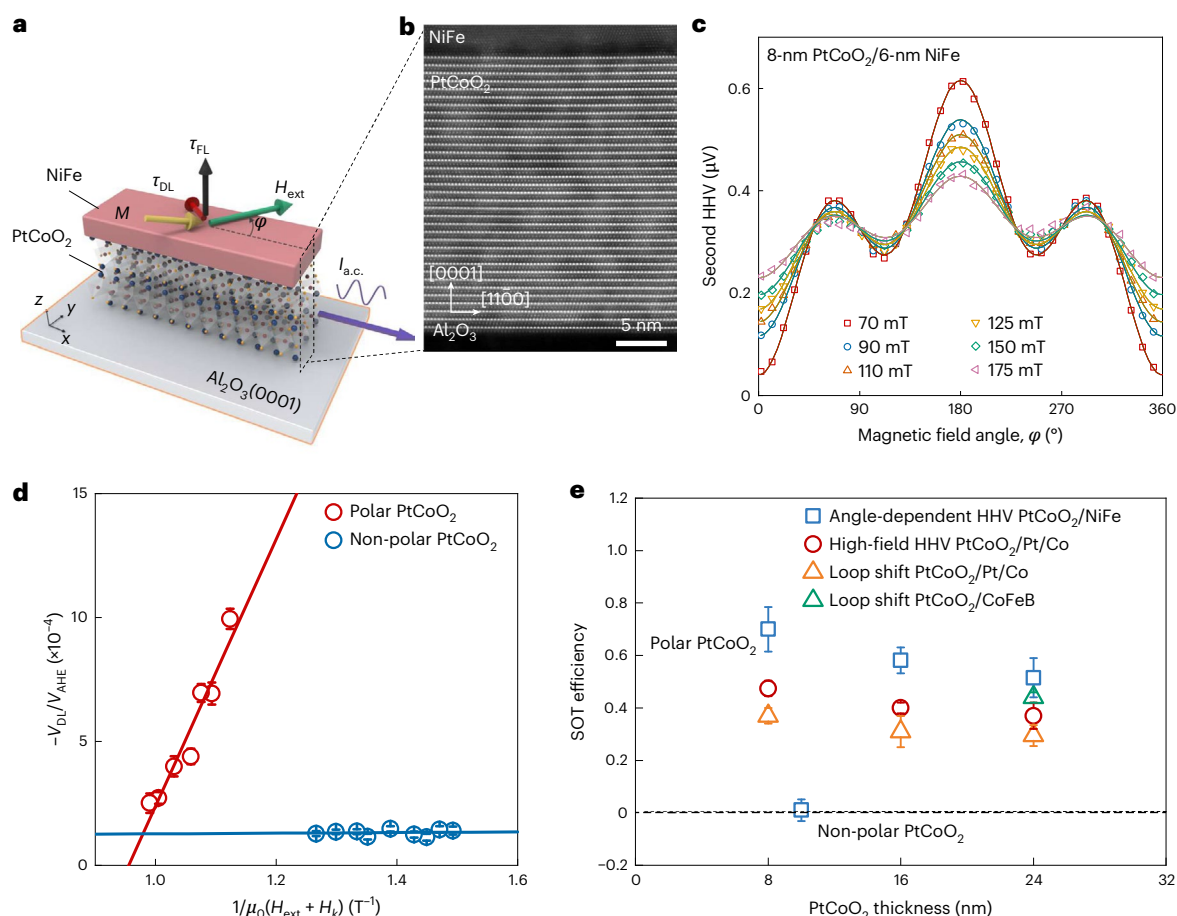


Fig. 2 | Measurements of SOT efficiency in PtCoO₂-based heterostructures.

a, Schematic of the angle-dependent HHV measurement setup. An alternating current ($I_{a.c.}$) is applied in the presence of an in-plane magnetic field H_{ext} , causing the magnetization (M) to oscillate due to field-like (τ_{FL}) and damping-like (τ_{DL}) SOT, which induces a second HHV. The first ($V_{xy}^{1\omega}$) and second ($V_{xy}^{2\omega}$) HHVs were measured by rotating the in-plane magnetic field angle (φ). **b**, Cross-sectional HAADF-STEM image of a polar PtCoO₂/NiFe heterostructure. **c**, Measured $V_{xy}^{2\omega}$ as a function of φ for an 8-nm polar PtCoO₂/6-nm NiFe sample at different magnetic field strengths. The solid curves represent fits to the experimental data based on equation (6). **d**, Ratio of the $\cos\varphi$ component of $-V_{xy}^{2\omega}$ and anomalous Hall voltage $-V_{DL}/V_{AHE}$ as a function of $1/\mu_0(H_{ext} + H_k)$. The solid line is a linear fit that yields a damping-like SOT effective field H_{DL} of (0.36 ± 0.04) mT MA⁻¹ cm⁻² for polar PtCoO₂/NiFe and (0.01 ± 0.03) mT MA⁻¹ cm⁻² for non-polar PtCoO₂/NiFe. Error

bars indicate the standard errors of the fitted parameters obtained from the second-harmonic voltage fits in **c**. **e**, Thickness-dependent SOT efficiency for polar PtCoO₂/NiFe heterostructures (obtained from angle-dependent HHV measurements) for polar PtCoO₂/Pt/Co heterostructures (obtained from current-induced hysteresis loop shift and high-field HHV measurements) and polar PtCoO₂/CoFeB heterostructures (obtained from current-induced hysteresis loop shift measurements). For comparison, the SOT efficiency of a 13-nm non-polar PtCoO₂/8-nm NiFe sample is also shown (obtained from the angle-dependent HHV measurements). The error bars indicate the standard errors of the fitted parameters obtained from the linear fits in **d**. All measurements were performed at room temperature on 10 μ m $\times$ 50 μ m Hall-bar devices.

charge–spin conversion. We choose PtCoO₂ as a model system to investigate how polarity influences charge–spin conversion. As shown in Fig. 1a, PtCoO₂ comprises alternating layers of conductive Pt triangular lattice and insulating Co–O octahedra (Fig. 1a). Bulk PtCoO₂ is centrosymmetric (non-polar) and is renowned for its ultralow resistivity, resulting from its highly dispersive conduction band and low electron–phonon coupling^{36–39}. The confinement of charge carriers within the Pt monolayers could allow polar distortions to persist in the insulating CoO₂ layers, as the free carriers cannot effectively screen the out-of-plane (OOP) fields³⁵.

Results

Polar structure of PtCoO₂ thin films

We grew high-quality epitaxial PtCoO₂ thin films on two-inch Al₂O₃(0001) wafers (Extended Data Fig. 1a) using a chemical solution deposition (CSD) method developed previously by us (Methods)⁴⁰. Atomic force microscopy imaging reveals a smooth film surface with roughness below 1 nm (Extended Data Fig. 1b). X-ray diffraction

(Extended Data Fig. 1c–e) confirms the good crystalline quality and epitaxial growth of the films. Temperature-dependent resistivity measurements show metallic behaviour, with ultralow room-temperature resistivity ranging from 3 to 5 $\mu\Omega$ cm for films between 8 nm and 24 nm thick (Fig. 1b), which are approaching the value of the PtCoO₂ bulk single crystals ($\sim 2 \mu\Omega$ cm)⁴¹. For a comparative study, we also prepared a 13-nm PtCoO₂ thin-film sample by molecular-beam epitaxy (MBE; Supplementary Note 1), which showed similar transport properties as the CSD-grown films (Supplementary Note 2).

We investigated inversion symmetry in PtCoO₂ thin films by using optical second-harmonic generation (SHG) measurements (Methods). As shown in Fig. 1c, the CSD-grown films produce a two-fold anisotropic SHG pattern, which indicates broken inversion symmetry with a polar axis oriented along the OOP direction. The increasing signal amplitude with film thickness (Fig. 1c) confirms that the SHG originates from the bulk rather than the surface or interface. No evidence of in-plane polarity is observed (Extended Data Fig. 2). By contrast, MBE-grown samples show no anisotropic SHG signal under the same measurement

conditions (Supplementary Note 3), a signature of a centrosymmetric, non-polar structure.

We directly visualized the polar structure in the CSD-grown PtCoO₂ film using four-dimensional (4D) scanning transmission electron microscopy (STEM) ptychography (Methods), measuring the Pt, Co and O atomic positions at deep-sub-ångström resolution and deep-sub-picometre precision⁴². The ptychography reconstruction of a 24-nm PtCoO₂ film sample yields a series of depth-resolved slices, from which Fig. 1d shows a representative slice revealing distinct polar atomic displacements of both Co ions and oxygen octahedra, shifted off-centre relative to the adjacent Pt layers along the OOP direction with a predominantly upward offset. Spatial maps of Co–Pt ($\delta_{\text{Co–Pt}}$; Fig. 1e) and O–Pt ($\delta_{\text{O–Pt}}$; Fig. 1f) displacements exhibit closely matching nanoscale domains, with smaller O–Pt displacements that produce a small upward shift of Co relative to the octahedral framework (Fig. 1h). Statistical analyses of Co–Pt and O–Pt displacement distributions (Extended Data Fig. 3) show mean values of ~ 2 pm and ~ 1 pm, respectively. These values, averaged over nanoscale polar domains with varying amplitude and orientation of electric dipoles, indicate a small net OOP polarization. Depth-resolved reconstructions (Extended Data Fig. 4) further demonstrate that the polar displacements vary along the beam direction and show correlation over several nanometres, consistent with that derived from the spatial distribution map in the slice. These observations indicate a picture of short-range polar nano-regions throughout the film that, in aggregate, gives rise to the macroscopic inversion-symmetry breaking detected by SHG.

Interestingly, these polar displacement domains correlate strongly with oxygen-vacancy distributions (Fig. 1g), suggesting that although bulk PtCoO₂ is centrosymmetric, polar displacement domains can be induced by localized oxygen vacancies. The absence of inversion-symmetry breaking in MBE-grown samples may result from a lower oxygen-vacancy concentration, imposed by the more strongly oxidizing growth environment (Supplementary Note 1). This mechanism is similar to the formation of polar nano-regions in strontium titanate in which polar domains emerge from local disorder⁴³. The large structural anisotropy of the quasi-two-dimensional delafossite structure makes the lattice more compliant along the *c* axis; therefore, the oxygen vacancies preferentially induce displacements in this direction. Importantly, this polarity from Co and O off-centring does not compromise the low resistivity in PtCoO₂. This observation is consistent with the recently reported sublattice purification mechanism in bulk metallic delafossites, which revealed that impurities are forced to reside in the octahedral layers, whereas leaving the conductive layers ultrapure and electrically decoupled from disorder³⁹.

Density functional theory (DFT) calculations reveal that the oxygen-vacancy-induced polarity affects the electronic band structure (Extended Data Fig. 5). In the supercell model containing oxygen vacancies, the relaxed structure exhibits local OOP polar distortions with Co–Pt and O–Pt displacements of the same order of magnitude as those observed in individual polar nano-regions from ptychography. The local inversion-symmetry breaking produces pronounced band splitting near the Fermi surface in the conduction bands, which enhances the intrinsic spin Berry curvature (following the Kubo formula (equation (4))). The calculated momentum-space distribution of the spin Berry curvature shows that the oxygen-vacancy-induced polar phase exhibits a stronger overall spin Berry curvature (Fig. 1i) than the centrosymmetric phase (Fig. 1k). Consequently, the computed SHC increases by a factor of 3–4 (Extended Data Fig. 5e).

Charge–spin conversion in PtCoO₂

We next measured the charge–spin conversion of our PtCoO₂ films by conducting SOT measurements at room temperature on various PtCoO₂/ferromagnetic heterostructures, including those with Ni₈₀Fe₂₀ (NiFe; in-plane magnetic anisotropy) and Pt/Co multilayers

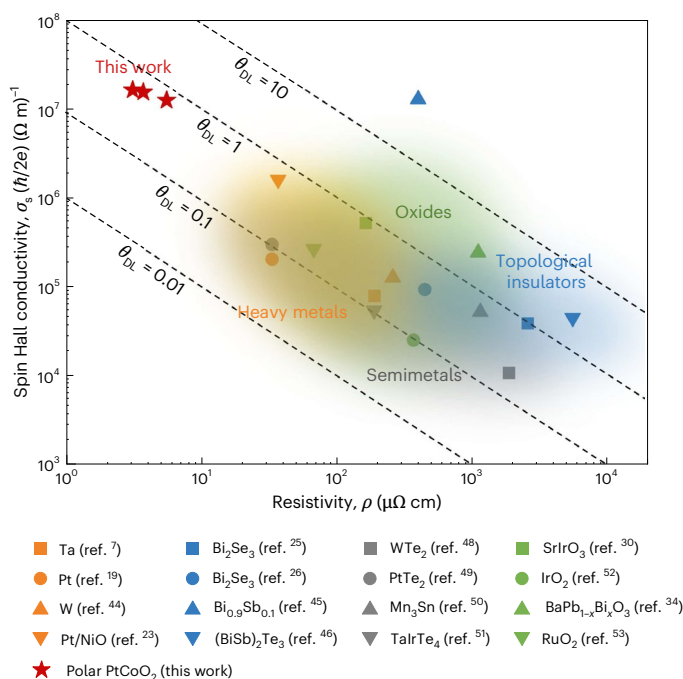


Fig. 3 | Benchmarking SHC in polar PtCoO₂. Comparison of SHC and resistivity of polar PtCoO₂ with those of representative spin-source materials, including heavy metals, topological insulator, topological semimetals and oxides. The dashed lines represent fixed SOT efficiency, which is the product of SHC and resistivity.

(perpendicular magnetic anisotropy). Passing a charge current in the PtCoO₂ layer generates spin currents that diffuse into the adjacent ferromagnetic layer and exert torques on its magnetization. To quantify these torques, we first used angle-dependent harmonic Hall voltage (HHV) measurements (Methods). Figure 2b shows the high-angle annular dark-field (HAADF) image of the PtCoO₂/NiFe heterostructure with a well-defined interface. Figure 2c shows the representative angle-dependent HHV curves for an 8-nm polar PtCoO₂/6-nm NiFe device measured under various field strengths. Figure 2d illustrates the linear fits of the field-dependent damping-like SOT component, from which we extract the damping-like effective field and determine the SOT efficiency θ_{DL} (Methods and Extended Data Fig. 6). The intrinsic charge–spin conversion efficiency is given by $\theta_{\text{SH}} = \theta_{\text{DL}}/T$, where T is the interfacial spin transparency. The lower bound of θ_{SH} can be obtained with $T = 1$, since only a fraction of the spin current can transmit across the interface without reflection or absorption ($T < 0.5$ for typical heavy metal/ferromagnet heterostructures)²⁰.

Remarkably, the 8-nm polar PtCoO₂/6-nm NiFe heterostructure achieves $\theta_{\text{DL}} = 0.70 \pm 0.09$, about an order of magnitude higher than the reported values for conventional Pt-based heterostructures (typically $\theta_{\text{DL}} = -0.05$ – 0.15)²⁰. For reference, our control sample of 8-nm Pt/6-nm NiFe measured under the same condition gives $\theta_{\text{DL}} = 0.11 \pm 0.01$ (Supplementary Note 4). Increasing the PtCoO₂ thickness slightly lowers θ_{DL} (Fig. 2e), probably due to the reduction in resistivity in thicker films, which indicates that the charge–spin conversion in PtCoO₂ is dominated by the intrinsic spin Hall effect. In stark contrast, the heterostructure (13-nm PtCoO₂/8-nm NiFe) with centrosymmetric (non-polar) PtCoO₂ grown by MBE exhibits negligible damping-like SOT signals (Supplementary Note 5), corresponding to a small SOT efficiency ($\theta_{\text{DL}} = 0.02 \pm 0.07$). This enhancement in SOT efficiency for polar PtCoO₂ reveals the link between the charge–spin conversion and polar distortion.

We also measured the SOT efficiency in polar PtCoO₂/[Pt (0.8 nm)/Co (0.5 nm)] $\times$ 2/Pt (1 nm) heterostructures by independent SOT characterization techniques including current-induced hysteresis loop shift

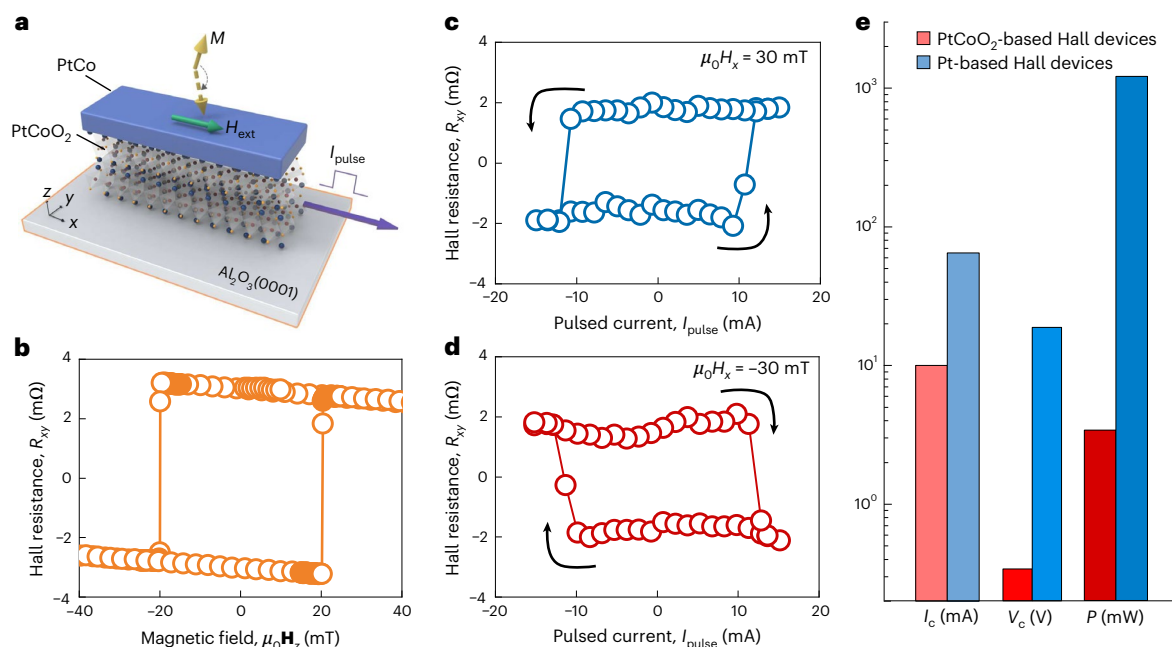


Fig. 4 | SOT-induced magnetization switching in the PtCoO₂/Pt/Co heterostructure. **a**, Schematic of SOT-driven perpendicular magnetization switching. A pulse current (I_{pulse}) is applied in the presence of an in-plane field (H_x). **b**, Hysteresis loop of the anomalous Hall resistance (R_{xy}) for an 8-nm PtCoO₂/Pt (0.8 nm)/Co (0.5 nm)] x 2/Pt (1 nm) heterostructure, measured by sweeping an OOP magnetic field H_z . **c, d**, Current-induced magnetization switching under +30-mT (**c**) and -30-mT (**d**) in-plane fields. The curved arrows indicate the switching polarity. The slight waviness in anomalous Hall resistance reflects small changes in response to current pulses weaker than required for full magnetic reversal,

probably associated with domain nucleation and domain-wall motion. The switching ratio is smaller than that obtained under an OOP magnetic field sweep (**b**), which could arise from domain-wall pinning and current shunting in the Hall voltage arms. **e**, Comparison of the SOT-induced switching critical current I_c , switching critical voltage V_c and switching power P for 8-nm PtCoO₂/Pt (0.8 nm)/Co (0.5 nm)] x 2/Pt (1 nm) versus 5-nm Pt/[Pt (0.8 nm)/Co (0.5 nm)] x 2/Pt (1 nm), measured under identical Hall-bar geometry and experimental conditions. All measurements were performed at room temperature on 20 μm x 100 μm Hall-bar devices.

(Methods and Extended Data Fig. 7a–d) and high-field HHV measurements (Methods and Extended Data Fig. 8). From these measurements, θ_{DL} exhibits a similar PtCoO₂ thickness dependence to that observed in PtCoO₂/NiFe heterostructures (Fig. 2e). The reduced θ_{DL} in these samples compared with PtCoO₂/NiFe heterostructures is probably caused by the decreased interfacial spin transparency¹⁸.

In addition, to demonstrate compatibility with semiconductor platforms, we grew PtCoO₂ thin films directly on Si(001) substrates with a native SiO₂ layer. The films exhibit strong *c*-axis texture (Supplementary Note 6), which preserves both high conductivity (11 $\mu\Omega$ cm at room temperature for a 24-nm film; Supplementary Note 7) and OOP polarity, as confirmed by SHG measurements (Supplementary Note 8). The SOT efficiency in a 24-nm PtCoO₂/20-nm NiFe heterostructure grown on Si is determined to be 0.54 ± 0.04 (Supplementary Note 9), comparable with those of sapphire-based devices. To further assess integration with MRAM-relevant ferromagnets, we developed PtCoO₂/Ta/CoFeB/MgO heterostructures exhibiting robust perpendicular magnetic anisotropy, on both Al₂O₃ and Si substrates. These devices (24-nm PtCoO₂) show SOT efficiencies of 0.44 ± 0.02 on Al₂O₃ (Extended Data Fig. 7e, f) and 0.37 ± 0.01 on Si (Supplementary Note 10), comparable with those of Pt/Co multilayers. The ultrathin Ta interlayer is used to induce perpendicular anisotropy in CoFeB and acts as a spacer rather than a spin-current source (Supplementary Note 11).

The combination of high charge–spin conversion efficiency and ultralow resistivity in polar PtCoO₂ leads to an unprecedentedly high value of SHC, spanning from 0.67×10^7 to 1.6×10^7 $\hbar/2e$ (Ω m)^{−1} at room temperature, with the 24-nm PtCoO₂ film reaching the highest value. This value is higher than those reported for all known spin-source materials, including heavy metals^{7,19,20,23,44}, topological insulators^{25,26,45,46}, topological semimetals^{47–51} and oxides^{30,31,34,52,53}. Figure 3 compares our polar PtCoO₂ (red star) with representative spin-source materials

from each of these categories. In particular, polar PtCoO₂ exhibits a resistivity one to three orders of magnitude lower than these materials.

Efficient SOT magnetization switching

To evaluate the implications of the high SHC in polar PtCoO₂ for SOT switching, we first investigated the power consumption required for magnetization switching in micrometre-scale Hall-bar devices and compared it with that of a conventional spin-source material, Pt, under identical geometry and measurement conditions. We applied current pulses I_{pulse} under a small in-plane magnetic field to induce the deterministic switching of perpendicular magnetization (Fig. 4a). The 8-nm PtCoO₂/Pt (0.8 nm)/Co (0.5 nm)] x 2/Pt (1 nm) heterostructure exhibits robust perpendicular magnetic anisotropy (Fig. 4b). We achieved efficient SOT switching at a critical current I_c of -10 mA corresponding to a current density in the spin-source layer J_c of 5.8 MA cm^{-2} (Fig. 4c, d). The switching polarity reverses on reversing the in-plane field, consistent with the expected behaviour of SOT-induced switching⁵⁴. Complete SOT switching is confirmed by polar magneto-optical Kerr effect microscopy (Extended Data Fig. 9a, b). Furthermore, endurance measurements over 100 consecutive current pulse cycles (Extended Data Fig. 9c) demonstrate stable and reproducible switching. Under identical conditions, the 5-nm Pt/[Pt (0.8 nm)/Co (0.5 nm)] x 2/Pt (1 nm) control device requires a substantially higher current density of 46 MA cm^{-2} for switching (Supplementary Note 12). Owing to the much lower resistivity of PtCoO₂ and larger SOT efficiency, the measured switching voltage ($V_c = I_c R$) and switching power ($P = I_c^2 R$) are correspondingly reduced compared with Pt-based devices (Fig. 4e).

Conclusion

In conclusion, we have achieved over 30 times enhancement in charge–spin conversion in polar PtCoO₂, compared with its non-polar phase.

We further demonstrated PtCoO₂ as a promising spin-source material with notably reduced operation voltage and power for magnetization switching, and that it can be integrated onto silicon substrates.

Our study illustrates that delafossites can retain ultrahigh two-dimensional conductivity as well as accommodate functionalities like polarity, due to their highly defect-tolerant conduction layers³⁹. This approach can unlock a range of exotic states and enable innovative device concepts by merging multiple correlated functionalities in a single material, alongside previously reported surface states³⁶, surface magnetism⁵⁵ and ballistic transport at low temperatures³⁸. This makes metallic delafossites as a compelling platform for spintronics research.

Online content

Any methods, additional references, Nature Portfolio reporting summaries, source data, extended data, supplementary information, acknowledgements, peer review information; details of author contributions and competing interests; and statements of data and code availability are available at <https://doi.org/10.1038/s41563-026-02640-4>.

References

- Fert, A., Ramesh, R., Garcia, V., Casanova, F. & Bibes, M. Electrical control of magnetism by electric field and current-induced torques. *Rev. Mod. Phys.* **96**, 015005 (2024).
- Sinova, J., Valenzuela, S. O., Wunderlich, J., Back, C. H. & Jungwirth, T. Spin Hall effects. *Rev. Mod. Phys.* **87**, 1213–1260 (2015).
- Manchon, A. et al. Current-induced spin-orbit torques in ferromagnetic and antiferromagnetic systems. *Rev. Mod. Phys.* **91**, 035004 (2019).
- Shao, Q. et al. Roadmap of spin-orbit torques. *IEEE Trans. Magn.* **57**, 800439 (2021).
- Worledge, D. C. & Hu, G. Spin-transfer torque magnetoresistive random access memory technology status and future directions. *Nat. Rev. Electr. Eng.* **1**, 730–747 (2024).
- Miron, I. M. et al. Perpendicular switching of a single ferromagnetic layer induced by in-plane current injection. *Nature* **476**, 189–193 (2011).
- Liu, L. et al. Spin-torque switching with the giant spin hall effect of tantalum. *Science* **336**, 555–558 (2012).
- Jiang, W. et al. Blowing magnetic skyrmion bubbles. *Science* **349**, 283–286 (2015).
- Manipatruni, S. et al. Scalable energy-efficient magnetoelectric spin-orbit logic. *Nature* **565**, 35–42 (2019).
- Grollier, J. et al. Neuromorphic spintronics. *Nat. Electron.* **3**, 360–370 (2020).
- Gu, K. et al. Three-dimensional racetrack memory devices designed from freestanding magnetic heterostructures. *Nat. Nanotechnol.* **17**, 1065–1071 (2022).
- Incorvia, J. A. C. et al. Spintronics for achieving system-level energy-efficient logic. *Nat. Rev. Electr. Eng.* **1**, 700–713 (2024).
- Garello, K. et al. Manufacturable 300 mm platform solution for field-free switching SOT-MRAM. In *Proc. Symposium on VLSI Technology* (eds Masahara, M. et al.) T194–T195 (IEEE, 2019).
- Choi, Y.-G. et al. Observation of the orbital Hall effect in a light metal Ti. *Nature* **619**, 52–56 (2023).
- Bihlmayer, G., Noël, P., Vyalikh, D. V., Chulkov, E. V. & Manchon, A. Rashba-like physics in condensed matter. *Nat. Rev. Phys.* **4**, 642–659 (2022).
- Ahmed, I. et al. A comparative study between spin-transfer-torque and spin-Hall-effect switching mechanisms in PMTJ using SPICE. *IEEE J. Explor. Solid-State Comput. Devices Circuits* **3**, 74–82 (2017).
- Grimaldi, E. et al. Single-shot dynamics of spin-orbit torque and spin transfer torque switching in three-terminal magnetic tunnel junctions. *Nat. Nanotechnol.* **15**, 111–117 (2020).
- Zhang, W., Han, W., Jiang, X., Yang, S.-H. & Parkin, S. P. S. Role of transparency of platinum-ferromagnet interfaces in determining the intrinsic magnitude of the spin Hall effect. *Nat. Phys.* **11**, 496–502 (2015).
- Liu, L., Moriyama, T., Ralph, D. C. & Buhrman, R. A. Spin-torque ferromagnetic resonance induced by the spin Hall effect. *Phys. Rev. Lett.* **106**, 036601 (2011).
- Zhu, L., Ralph, D. C. & Buhrman, R. A. Maximizing spin-orbit torque generated by the spin Hall effect of Pt. *Appl. Phys. Rev.* **8**, 031308 (2021).
- Zhu, L., Zhu, L., Sui, M., Ralph, D. C. & Buhrman, R. A. Variation of the giant intrinsic spin Hall conductivity of Pt with carrier lifetime. *Sci. Adv.* **5**, eaav8025 (2019).
- Hu, C.-Y. et al. Toward 100% spin-orbit torque efficiency with high spin-orbital Hall conductivity Pt-Cr alloys. *ACS Appl. Electron. Mater.* **4**, 1099–1108 (2022).
- Zhu, L., Zhu, L. & Buhrman, R. A. Fully spin-transparent magnetic interfaces enabled by the insertion of a thin paramagnetic NiO layer. *Phys. Rev. Lett.* **126**, 107204 (2021).
- Fan, Y. et al. Magnetization switching through giant spin-orbit torque in a magnetically doped topological insulator heterostructure. *Nat. Mater.* **13**, 699–704 (2014).
- Mellnik, A. R. et al. Spin-transfer torque generated by a topological insulator. *Nature* **511**, 449–451 (2014).
- Wang, Y. et al. Topological surface states originated spin-orbit torques in Bi₂Se₃. *Phys. Rev. Lett.* **114**, 257202 (2015).
- Lesne, E. et al. Highly efficient and tunable spin-to-charge conversion through Rashba coupling at oxide interfaces. *Nat. Mater.* **15**, 1261–1266 (2016).
- Trier, F. et al. Oxide spin-orbitronics: spin-charge interconversion and topological spin textures. *Nat. Rev. Mater.* **7**, 258–274 (2022).
- Noël, P. et al. Non-volatile electric control of spin-charge conversion in a SrTiO₃ Rashba system. *Nature* **580**, 483–486 (2020).
- Nan, T. et al. Anisotropic spin-orbit torque generation in epitaxial SrIrO₃ by symmetry design. *Proc. Natl Acad. Sci. USA* **116**, 16186–16191 (2019).
- Liu, L. et al. Current-induced magnetization switching in all-oxide heterostructures. *Nat. Nanotechnol.* **14**, 939–944 (2019).
- Sahu, P., Bhowal, S. & Satpathy, S. Effect of the inversion symmetry breaking on the orbital Hall effect: a model study. *Phys. Rev. B* **103**, 085113 (2021).
- Varotto, S. et al. Room-temperature ferroelectric switching of spin-to-charge conversion in germanium telluride. *Nat. Electron.* **4**, 740–747 (2021).
- Edgeton, A. L. et al. Large spin-orbit torque in bismuthate-based heterostructures. *Nat. Electron.* **6**, 973–980 (2023).
- Bhowal, S. & Spaldin, N. A. Polar metals: principles and prospects. *Annu. Rev. Mater. Res.* **53**, 53–79 (2023).
- Sunko, V. et al. Maximal Rashba-like spin splitting via kinetic-energy-coupled inversion-symmetry breaking. *Nature* **549**, 492–496 (2017).
- Usui, H. et al. Hidden kagome-lattice picture and origin of high conductivity in delafossite PtCoO₂. *Phys. Rev. Mater.* **3**, 045002 (2019).
- Bachmann, M. D. et al. Directional ballistic transport in the two-dimensional metal PdCoO₂. *Nat. Phys.* **18**, 819–824 (2022).
- Zhang, Y. et al. Crystal-chemical origins of the ultrahigh conductivity of metallic delafossites. *Nat. Commun.* **15**, 1399 (2024).
- Wei, R. et al. Solution-processable epitaxial metallic delafossite oxide films. *Adv. Funct. Mater.* **30**, 2002375 (2020).
- Nandi, N. et al. Unconventional magneto-transport in ultrapure PdCoO₂ and PtCoO₂. *npj Quantum Mater.* **3**, 66 (2018).

42. Yang, W., Sha, H., Cui, J., Mao, L. & Yu, R. Local-orbital ptychography for ultrahigh-resolution imaging. *Nat. Nanotechnol.* **19**, 612–617 (2024).
43. Lee, D. et al. Emergence of room-temperature ferroelectricity at reduced dimensions. *Science* **349**, 1314–1317 (2015).
44. Pai, C.-F. et al. Spin transfer torque devices utilizing the giant spin Hall effect of tungsten. *Appl. Phys. Lett.* **101**, 122404 (2012).
45. Khang, N. H. D., Ueda, Y. & Hai, P. N. A conductive topological insulator with large spin Hall effect for ultralow power spin–orbit torque switching. *Nat. Mater.* **17**, 808–813 (2018).
46. Wu, H. et al. Room-temperature spin-orbit torque from topological surface states. *Phys. Rev. Lett.* **123**, 207205 (2019).
47. Zhang, Y. et al. Room temperature field-free switching of perpendicular magnetization through spin-orbit torque originating from low-symmetry type II Weyl semimetal. *Sci. Adv.* **9**, eadg9819 (2023).
48. Shi, S. et al. All-electric magnetization switching and Dzyaloshinskii–Moriya interaction in WTe_2 /ferromagnet heterostructures. *Nat. Nanotechnol.* **14**, 945–949 (2019).
49. Xu, H. et al. High spin Hall conductivity in large-area type-II Dirac semimetal PtTe_2 . *Adv. Mater.* **32**, 2000513 (2020).
50. Kondou, K. et al. Giant field-like torque by the out-of-plane magnetic spin Hall effect in a topological antiferromagnet. *Nat. Commun.* **12**, 6491 (2021).
51. Liu, Y. et al. Field-free switching of perpendicular magnetization at room temperature using out-of-plane spins from TaIrTe_4 . *Nat. Electron.* **6**, 732–738 (2023).
52. Patton, M. et al. Symmetry control of unconventional spin–orbit torques in IrO_2 . *Adv. Mater.* **35**, 2301608 (2023).
53. Zhang, Y. et al. Simultaneous high charge-spin conversion efficiency and large spin diffusion length in altermagnetic RuO_2 . *Adv. Funct. Mater.* **34**, 2313332 (2024).
54. Baumgartner, M. et al. Spatially and time-resolved magnetization dynamics driven by spin–orbit torques. *Nat. Nanotechnol.* **12**, 980–986 (2017).
55. Mazzola, F. et al. Itinerant ferromagnetism of the Pd-terminated polar surface of PdCoO_2 . *Proc. Natl Acad. Sci. USA* **115**, 12956–12960 (2018).

Publisher's note Springer Nature remains neutral with regard to jurisdictional claims in published maps and institutional affiliations.

Springer Nature or its licensor (e.g. a society or other partner) holds exclusive rights to this article under a publishing agreement with the author(s) or other rightsholder(s); author self-archiving of the accepted manuscript version of this article is solely governed by the terms of such publishing agreement and applicable law.

© The Author(s), under exclusive licence to Springer Nature Limited 2026

¹School of Integrated Circuits and Beijing National Research Center for Information Science and Technology (BNRist), Tsinghua University, Beijing, China. ²School of Materials Science and Engineering, Tsinghua University, Beijing, China. ³State Key Laboratory of New Ceramics and Fine Processing, Tsinghua University, Beijing, China. ⁴Department of Physics, Cornell University, Ithaca, NY, USA. ⁵Key Laboratory of Materials Physics, Institute of Solid State Physics, HFIPS, Chinese Academy of Sciences, Hefei, China. ⁶Department of Physics, Tsinghua University, Beijing, China. ⁷Department of Materials Science and Engineering, Cornell University, Ithaca, NY, USA. ⁸Department of Materials Science and Engineering, The Pennsylvania State University, University Park, PA, USA. ⁹Physical Science and Engineering Division, King Abdullah University of Science and Technology, Thuwal, Saudi Arabia. ¹⁰School of Materials Science and Engineering, Hubei University, Wuhan, China. ¹¹Institute of Electronics, AGH University of Krakow, Krakow, Poland. ¹²School of Microelectronics, University of Science and Technology of China, Hefei, China. ¹³Kavli Institute for Nanoscale Science, Cornell University, Ithaca, NY, USA. ¹⁴These authors contributed equally: Guiping Ji, Yi Zheng, Yuejie Zhang, Yuhang Liang, Renhui Wei. ✉ e-mail: xbzhu@issp.ac.cn; diy@mail.tsinghua.edu.cn; nantianxiang@mail.tsinghua.edu.cn

Methods

Sample growth, characterization and device fabrication

Thin-film PtCoO₂ samples were grown using the CSD method. The precursor solution of PtCoO₂ was prepared at room temperature using a platinum(II) nitrate solution (Pt(NO₃)₂, Pt of 18.02%) and cobalt(II) nitrite hexahydrate (Co(NO₃)₂·6H₂O) as precursor reagents. Initially, the Pt-based salt was dissolved in 1 ml of nitric acid (HNO₃) and then mixed with 4 ml of 2-methoxyethanol (CH₃OCH₂CH₂OH). After stirring for 8 min, Co(NO₃)₂·6H₂O was added, and the solution underwent 8 h of magnetic stirring to achieve a homogeneous precursor solution. This solution was subsequently filtered through a 0.22-μm aperture filter. The cation concentration of the solution was maintained at 0.2 mol l⁻¹, with a Pt-to-Co molar ratio of 1:1. Thin films of PtCoO₂ were deposited on c-cut Al₂O₃ single crystals (two inches) as substrates. Before thin-film deposition, the sapphire substrates were thoroughly cleaned with acetone, ethanol and demineralized water in an ultrasonic bath for 10 min each. Additionally, Ar-ion plasma cleaning was performed to enhance the surface wettability of the substrates. The precursor solution was spin coated onto the Al₂O₃ substrates at a rotation rate of 4,000 rpm for 30 s under an ambient temperature of 45 °C. The as-coated dried thin films were cured at 120 °C for 1 min, followed by a pyrolysis step at 250 °C for 10 min in a tube furnace. Finally, the pyrolytic films were annealed at 720 °C for 30 min in the ambient atmosphere using a rapid thermal annealing procedure. This entire process was repeated multiple times to achieve the desired thickness of PtCoO₂ thin films. PtCoO₂ thin films were grown on Si(001) wafers (two inches) with a native SiO₂ layer using the CSD method under similar conditions to those used for Al₂O₃ substrates, except that the annealing step was optimized for the silicon substrate. Specifically, the pyrolytic films were annealed at 700 °C for 20 min in the ambient atmosphere using a rapid thermal annealing procedure. The details of the PtCoO₂ thin films grown using the MBE method are provided in Supplementary Note 1⁵⁶.

After the fabrication of the PtCoO₂ thin film, the samples were immediately transferred into a magnetron sputtering chamber with a background vacuum of 2×10^{-8} torr and then annealed at 150 °C for 30 min. Ferromagnetic NiFe, Pt/Co and CoFeB layers were sputtered with an Ar pressure of 3 mtorr at room temperature by d.c. sputtering. The film thickness was characterized by X-ray reflectivity. The saturation magnetization M_s of NiFe, [Pt (0.8 nm)/Co (0.5 nm)] $\times$ 2/Pt (1 nm) and Ta (2 nm)/CoFeB (1.1 nm)/MgO (1 nm) were measured to be -7.95×10^5 A m⁻¹, -4.02×10^5 A m⁻¹ and -1.17×10^6 A m⁻¹, respectively, by using vibrating-sample magnetometry. The epitaxy quality of PtCoO₂ was characterized by a high-resolution X-ray diffractometer. The transport properties were measured in a Quantum Design Physical Property Measurement System. Devices for HHV, loop shift and SOT switching measurements were patterned into Hall bars, by using photolithography followed by Ar-ion beam milling. Then, 5-nm Ti/120-nm Pt electrodes were deposited and defined by the lift-off process.

Optical SHG

For the SHG experiment, a titanium sapphire oscillator system (800-nm central wavelength, 80-MHz repetition rate and 20-fs pulse width) was utilized as the fundamental light. The ultrafast laser beam was modulated by a chopper at 941 Hz to facilitate lock-in detection. All measurements were conducted in the 45° oblique incidence reflection geometry at room temperature. The angle ψ in the polarization-dependent data curve was defined by the incident-beam polarization, which was controlled by a Glan–Taylor polarizer followed by a rotational half-wave plate (with $\psi = 0$ corresponding to the *p*-polarized incident beam). The output second-harmonic light was fixed as either *p* or *s* polarization by another Glan–Taylor polarizer. The fundamental light source was filtered by an 800-nm bandpass filter to eliminate the second harmonics generated by the optics in the front optical path and was focused by a $f = 300$ mm lens before reaching the sample. The output second-harmonic signal passed through three lowpass filters to guarantee the

elimination of fundamental light before being concentrated and collected into a photomultiplier tube. Photons in the photomultiplier tube were converted to a photocurrent, which was amplified by a current preamplifier and ultimately recorded by a lock-in amplifier as a corresponding voltage signal. The polarization-dependent SHG intensities corresponding to the OOP response (*p*-out; Fig. 1c) and the in-plane response (*s*-out; Extended Data Fig. 2) were fitted using the following expressions derived for a crystal with C_{3v} symmetry in the 45° oblique incidence reflection geometry:

$$I_p = \left[\frac{\sqrt{2}}{4} (2d_{15} + d_{31} + d_{33}) \cos^2 \phi - d_{22} \cos \phi \sin \phi + \frac{\sqrt{2}}{2} d_{31} \sin^2 \phi \right]^2, \quad (1)$$

$$I_s = \left[-\frac{1}{2} d_{22} \cos^2 \phi - \sqrt{2} d_{15} \cos \phi \sin \phi + d_{22} \sin^2 \phi \right]^2, \quad (2)$$

where $I_{p,s}(\phi)$ are the SHG intensities for *p* and *s* polarization, and d_{ij} are the elements of the second-order nonlinear susceptibility tensor. Fitting the experimental data yields $d_{33} = 0.34 \pm 0.01$ and $d_{15} = -0.066 \pm 0.001$, with other tensor elements negligible. The dominant d_{33} component indicates a strong OOP polarity.

Electron ptychography

Electron ptychography represents a remarkable advancement in electron microscopy^{57,58}. This technique captures the diffraction intensity at each electron-beam scanning position and uses an iterative process to simultaneously reconstruct the complex functions of both sample and electron beam. Unlike traditional imaging methods such as HAADF and integrated differential phase contrast in STEM, electron ptychography effectively eliminates the influence of lens aberrations, achieving deep-sub-ångström resolution^{42,59}. Additionally, it can simultaneously reconstruct heavy and light elements, whereas avoid artefacts from surface damage layers and misorientation, thereby enhancing the accuracy in the object phase and atomic positions and providing a true representation of the sample's structure. Here we used the adaptive-propagator ptychography method⁶⁰ within the framework of multislice electron ptychography to mitigate the misorientation effect. The HAADF images and 4D-STEM datasets for ptychography reconstructions were acquired using an aberration-corrected electron microscope (JEOL ARM300) operated at 300 kV. The 4D-STEM datasets were captured using a Gatan K3 detector. An underfocus of about 25 nm, scan step size of 0.4 Å, beam current of 15 pA and dwell time of 0.87 ms were used. During the adaptive-propagator ptychography reconstruction, the sample was divided into 30 slices and the slice thickness was about 0.5 nm. The reciprocal-space pixel size is 0.0308 Å⁻¹ and the real-space pixel size is 0.12 Å for ptychographic reconstructions. Potential sources of error such as phase retrieval artefacts, astigmatism and scan distortion were carefully minimized, ensuring that the ptychographic reconstructions provide high-precision and reliable measurements of atomic-scale displacements in PtCoO₂. Detailed validation procedures and reconstruction quality assessments are provided in Supplementary Note 13.

The Atomap Gaussian peak finding function⁶¹ was used to determine the accurate atomic column positions of Pt, Co and O. The mask radii for peak finding of the Pt, Co and O atomic columns are 1.32 Å, 1.32 Å and 1.08 Å, respectively. Details of the phase integration and mask-based analysis for O columns are provided in Extended Data Fig. 10. Relative displacement of Co–Pt is defined as the distance from a Co atomic column to the average position of its nearest-neighbour Pt atoms in the adjacent upper and lower layers, as well as to the average position of its nearest-neighbour O atoms in those layers.

First-principles calculations

First-principles DFT calculations were performed using a plane-wave pseudopotential method with the fully relativistic ultrasoft pseudopotentials implemented in QUANTUM ESPRESSO. The exchange

and correlation effects were treated within the generalized gradient approximation⁶². We used a plane-wave cut-off energy of 40 Ry and a $16 \times 16 \times 16$ k -point mesh in the irreducible Brillouin zone. Spin-orbit coupling was included in all electronic structure calculations. The SHC was calculated using the PAOFlow code based on pseudo-atomic orbitals⁶³. Tight-binding Hamiltonians were constructed from the non-self-consistent DFT calculations on an $8 \times 8 \times 8$ k -point mesh for the SHC calculations. The SHC and Berry curvature were calculated as²

$$\sigma_{ij}^k = \frac{e^2}{\hbar} \int \frac{d^3\mathbf{k}}{(2\pi)^3} \sum_n f_{n\mathbf{k}} \Omega_{n,ij}^k(\mathbf{k}), \quad (3)$$

$$\Omega_{n,ij}^k(\mathbf{k}) = -2im \sum_{n' \neq n} \frac{\langle n\mathbf{k} | j_i^\dagger | n'\mathbf{k} \rangle \langle n'\mathbf{k} | v_j | n\mathbf{k} \rangle}{(E_{n\mathbf{k}} - E_{n'\mathbf{k}})^2}. \quad (4)$$

Here $f_{n\mathbf{k}}$ is the Fermi–Dirac distribution function for band n and wavevector $\mathbf{k}$, $\Omega_{n,ij}^k(\mathbf{k})$ is the spin Berry curvature, $j_i^\dagger = \frac{1}{2}\{v_i, s_k\}$ is the spin-current operator. Also, v_i and s_k are the velocity and spin operators, respectively, and $i, j, k = x, y, z$. To capture the influence of local inversion-symmetry breaking, we constructed fully relaxed $2 \times 2 \times 2$ supercells of PtCoO₂ with and without oxygen vacancies. The pristine supercell represents the centrosymmetric phase, whereas supercells containing one or two oxygen vacancies simulate local polar distortions near vacancy sites. We note that this supercell treatment is an idealized periodic approximation of vacancy disorder, whereas quantitative modelling of realistic, disordered and aperiodic systems remains a challenging open problem. The local Co–Pt and O–Pt displacements and SHC values were extracted for comparison between pristine and oxygen-vacancy-induced polar configurations (Extended Data Fig. 5e).

HHV measurements

An alternating current at 1,137 Hz was applied using a Keithley 6221 current source. The first and second HHV signals were measured simultaneously by two lock-in amplifiers.

When a current I flows along the x axis, the Hall voltage V_{xy} can be expressed as⁶⁴

$$V_{xy} = IR_{xy} = I(R_{\text{AHE}} \cos \theta + R_{\text{PHE}} \sin^2 \theta \sin 2\varphi), \quad (5)$$

where θ and φ are the polar and azimuthal angles of the external magnetic field H_{ext} , respectively; R_{PHE} is the planar Hall resistance; and R_{AHE} is the anomalous Hall resistance. An in-plane external magnetic field $H_{\text{ext}} = H_{\text{ext}}(\varphi, \theta)$ is applied to align the magnetization M to its equilibrium position (φ_0, θ_0) . Under an alternating current $I_{\text{a.c.}} = I_0 \sin \omega t$, the SOT produces an effective field expressed as $H_{\text{eff}} = (\Delta H_x, \Delta H_y, \Delta H_z) \sin \omega t$, which causes the magnetization M to oscillate around the equilibrium position, $(\varphi_0', \theta_0') = (\varphi_0 + \Delta\varphi \sin \omega t, \theta_0 + \Delta\theta \sin \omega t)$. As a result, the second HHV can be written as

$$V_{xy}^{2\omega} = -\frac{1}{2} V_2 = \frac{1}{2} I_0 [(R_{\text{AHE}} \sin \theta_0 - R_{\text{PHE}} \sin 2\theta_0 \sin 2\varphi_0) \Delta\theta - 2R_{\text{PHE}} \sin^2 \theta_0 \cos 2\varphi_0 \Delta\varphi]. \quad (6)$$

Angle-dependent second HHV measurements

In samples for the angle-dependent HHV measurement^{65–67}, the ferromagnetic layer NiFe has an in-plane magnetic anisotropy. The external field $H_{\text{ext}} = H_{\text{ext}}(\varphi, \theta)$ is applied at a constant-magnitude H_{ext} but varying azimuthal angle φ , such that $M(\varphi_0 \approx \varphi, \theta_0 \approx \theta = 90^\circ)$. On the basis of prior analyses correlating $\Delta\varphi$, $\Delta\theta$ and the external field H_{ext} , the second HHV is given by

$$V_{xy}^{2\omega} = -\left[V_{\text{AHE}} \frac{H_{\text{DL}}}{2(H_{\text{ext}} + H_k)} + V_{\text{ANE}} \right] \cos \varphi - V_{\text{PHE}} \frac{H_{\text{FL}}}{H_{\text{ext}}} \cos \varphi \cos 2\varphi, \quad (7)$$

which can be rewritten in the simplified form as

$$V_{xy}^{2\omega} = V_{\text{DL}} \cos \varphi + V_{\text{FL}} \cos \varphi \cos 2\varphi, \quad (8)$$

where

$$V_{\text{DL}} = -V_{\text{AHE}} \frac{H_{\text{DL}}}{2(H_{\text{ext}} + H_k)} - V_{\text{ANE}}, \quad (9)$$

$$V_{\text{FL}} = -V_{\text{PHE}} \frac{H_{\text{FL}}}{H_{\text{ext}}}. \quad (10)$$

Here V_{DL} is the component contributed by τ_{DL} and thermal effect; V_{FL} is the component contributed by the τ_{FL} ; H_k is the OOP magnetic anisotropic field; V_{ANE} is the extra voltage contribution from anomalous Nernst effect; and V_{AHE} and V_{PHE} are the anomalous and planar Hall voltages, respectively. By varying the magnitude of the external field H_{ext} , H_{DL} can be extracted by performing a linear fit of $V_{\text{DL}}/V_{\text{AHE}}$ against $\frac{1}{H_{\text{ext}} + H_k}$. The SOT efficiency can be estimated as $\theta_{\text{DL}} = \frac{2e}{\hbar} \mu_0 M_s t_{\text{FM}} \frac{H_{\text{DL}}}{J_e}$, where t_{FM} is the ferromagnetic layer thickness and J_e is the current density.

High-field second HHV measurements

In the high-field second HHV measurements^{24,64}, the ferromagnetic layer exhibits OOP magnetic anisotropy, and the external field $H_{\text{ext}} = H_{\text{ext}}(\varphi, \theta)$ is applied along a fixed direction. The second HHV is measured as a function of magnitude H_{ext} . In the high-field regime ($H_{\text{ext}} \gg H_k$), the magnetization M is reoriented by the external field from the OOP direction to the in-plane direction. When H_{ext} is applied along the x axis, $\varphi \approx 0^\circ$, and $V_{xy}^{2\omega}$ can be written as

$$V_{xy}^{2\omega} = -V_{\text{AHE}} \frac{H_{\text{DL}}}{2(H_{\text{ext}} - H_k)} + V_{\text{PHE}} \frac{H_{\text{FL}}}{H_{\text{ext}}} + V_{\text{ANE}}. \quad (11)$$

For the PtCoO₂/[Pt (0.8 nm)/Co (0.5 nm)] $\times$ 2/Pt (1 nm) sample, V_{PHE} is about one order of magnitude smaller than V_{AHE} , allowing us to simplify the expression to

$$V_{xy}^{2\omega} \approx -\frac{1}{2} V_{\text{AHE}} \frac{H_{\text{DL}}}{(H_{\text{ext}} - H_k)} + V_{\text{ANE}}. \quad (12)$$

The H_{DL} value can be extracted by varying the intensity of the external field H_{ext} , and linear fitting of the $V_{xy}^{2\omega}/V_{\text{AHE}}$ to $\frac{1}{H_{\text{ext}} - H_k}$. The corresponding SOT efficiency can be estimated as $\theta_{\text{DL}} = \frac{2e}{\hbar} \mu_0 M_s t_{\text{FM}} \frac{H_{\text{DL}}}{J_e}$.

Current-induced hysteresis loop shift measurements

In the loop shift measurement⁶⁸, the ferromagnetic layer exhibits OOP magnetic anisotropy. A d.c. current $I_{\text{d.c.}}$ is applied along the x direction as the Hall resistance R_{xy} is measured. During the measurements, a magnetic field H_x is applied along the x direction and an OOP magnetic field H_z was swept back and forth. Because the presence of H_x tilts the magnetization M towards the x direction, it acquires an x component M_x . A damping-like effective field $H_{\text{DL}} \approx \mathbf{M}_x \times \boldsymbol{\sigma}$ arises, with a z component H_z^{eff} . This H_z^{eff} shifts the anomalous Hall resistance loop, producing an asymmetric coercive field ($|H_{c1}| \neq |H_{c2}|$). The H_z^{eff} value can be estimated via

$$H_z^{\text{eff}} = \frac{|H_{c1}| - |H_{c2}|}{2}. \quad (13)$$

Reversing the $I_{\text{d.c.}}$ or H_x reverses the polarity of H_z^{eff} . By measuring the H_z^{eff} value at different $I_{\text{d.c.}}$ and H_x values, the SOT efficiency θ_{DL} can be obtained using $\theta_{\text{DL}} = \frac{2e}{\hbar} \mu_0 M_s t_{\text{FM}} \frac{H_z^{\text{eff}}}{J_e}$.

Current- and field-induced magnetization switching measurements

The SOT-induced magnetization switching measurements were conducted using Keithley 2900 as the current pulse source. Current pulses with a fixed duration of 500 μs and variable amplitude were applied to the Hall-bar devices. After each current pulse, the device was allowed to relax for 2 s before the anomalous Hall resistance was measured using a Keithley 2182A with a small d.c. probe current of 0.1 mA. An in-plane magnetic field was applied to break symmetry and enable deterministic switching. For magnetic-field-driven switching measurements,

a constant probe current of 0.1 mA was applied as the OOP magnetic field is swept between positive and negative values.

Data availability

The data supporting the findings of this study are available within the Article and its Supplementary Information. Source data are provided with this paper. Additional data are available from the corresponding authors upon request.

References

56. Song, Q. et al. Surface reconstructions and electronic structure of metallic delafossite thin films. *APL Mater.* **12**, 081117 (2024).
57. Yu, R., Sha, H., Cui, J. & Yang, W. Introduction to electron ptychography for materials scientists. *Microstructures* **4**, 2024056 (2024).
58. Miao, J. Computational microscopy with coherent diffractive imaging and ptychography. *Nature* **637**, 281–295 (2025).
59. Chen, Z. et al. Electron ptychography achieves atomic-resolution limits set by lattice vibrations. *Science* **372**, 826–831 (2021).
60. Sha, H., Cui, J. & Yu, R. Deep sub-ångström resolution imaging by electron ptychography with misorientation correction. *Sci. Adv.* **8**, eabn2275 (2022).
61. Nord, M., Vullum, P. E., MacLaren, I., Tybell, T. & Holmestad, R. Atomap: a new software tool for the automated analysis of atomic resolution images using two-dimensional Gaussian fitting. *Adv. Struct. Chem. Imaging* **3**, 9 (2017).
62. Perdew, J. P., Burke, K. & Ernzerhof, M. Generalized gradient approximation made simple. *Phys. Rev. Lett.* **77**, 3865–3868 (1996).
63. Buongiorno Nardelli, M. et al. PAOFLOW: a utility to construct and operate on ab initio Hamiltonians from the projections of electronic wavefunctions on atomic orbital bases, including characterization of topological materials. *Comput. Mater. Sci.* **143**, 462–472 (2018).
64. Hayashi, M., Kim, J., Yamanouchi, M. & Ohno, H. Quantitative characterization of the spin-orbit torque using harmonic Hall voltage measurements. *Phys. Rev. B* **89**, 144425 (2014).
65. Schulz, T. et al. Effective field analysis using the full angular spin-orbit torque magnetometry dependence. *Phys. Rev. B* **95**, 224409 (2017).
66. MacNeill, D. et al. Thickness dependence of spin-orbit torques generated by WTe_2 . *Phys. Rev. B* **96**, 054450 (2017).
67. Avci, C. O. et al. Interplay of spin-orbit torque and thermoelectric effects in ferromagnet/normal-metal bilayers. *Phys. Rev. B* **90**, 224427 (2014).
68. Pai, C.-F., Mann, M., Tan, A. J. & Beach, G. S. D. Determination of spin torque efficiencies in heterostructures with perpendicular magnetic anisotropy. *Phys. Rev. B* **93**, 144409 (2016).

Acknowledgements

We thank C.-B. Eom and A. Manchon for helpful discussions.

Author contributions

T.N. conceived the project. T.N., R.Y., D.C.R., D.S., D.Y., Y.L., X. Zhu, W.J., J.T., X. Zhang. and P.L. supervised the experiments. R.W. and X. Zhu prepared the samples by the CSD method. Q.S. and D.G.S. prepared the samples by the MBE method. Yuejie Zhang, D.J., C.X., A.C. and W.S. fabricated the devices. Yuejie Zhang, G.J. and Y.L. performed the angle-dependent HHV measurements. H.B. and W.J. performed the high-field HHV measurements. Yuejie Zhang and G.J. performed the loop shift measurement. Yuejie Zhang and G.J. performed the SOT-induced switching measurements. G.J. and Yuejie Zhang analysed the HHV, loop shift and SOT-induced switching data. H.L., L.Y., S.H., X.L. and V.G. performed the SHG measurements. Yang Zhang, Y.D., J.G. and P.Y. performed the HAADF-STEM measurements. Yi Zheng and R.Y. performed the electron ptychography measurements and analysed

the data. H.Z., Y.J. and D.S. performed the DFT calculation. C.L. and Z.H. performed the AHE and resistivity measurements. T.N., G.J. and D.Y. wrote the paper. All authors discussed the results and commented on the paper. T.N. directed the research.

Funding

T.N. discloses support for the research of this work from the National Key R&D Program of China (grant number 2024YFB3614102) and the National Natural Science Foundation of China (grant numbers 52025024 and 52250418). D.Y. discloses support for the research of this work from the National Natural Science Foundation of China (grant numbers 52250418 and 52025024), and the Basic Science Center Project of the National Natural Science Foundation of China (grant number 52388201). D.S. discloses support for the research of this work from the National Natural Science Foundation of China (grant number 52250418). P.Y. discloses support for the research of this work from the Basic Science Center Project of the National Natural Science Foundation of China (grant number 52388201) and the National Natural Science Foundation of China (grant number 52025024). R.W. discloses support for the research of this work from the National Natural Science Foundation of China (grant number 12274410) and the National Key R&D Program of China (grant number 2024YFB3614102). G.J. discloses support for the research of this work from the National Key R&D Program of China (grant number 2024YFB3614102). L.Y. discloses support for the research of this work from the National Natural Science Foundation of China (grant numbers 12361141826, 12421004 and 12074212) and the Beijing Natural Science Foundation (grant number Z240006). Y.L. discloses support for the research of this work from the Basic Science Center Project of the National Natural Science Foundation of China (grant number 52388201). W.J. discloses support for the research of this work from the Basic Science Center Project of the National Natural Science Foundation of China (grant number 52388201). D.C.R. discloses support for the research of this work from the U.S. National Science Foundation (grant number DMR-2104268). W.S. discloses support for the research of this work from the National Science Centre (Poland) (grant number 2021/40/Q/ST5/00209). Q.S. and D.G.S. disclose support for the research of this work from the US Department of Energy (grant number DE-SC0026002) and the Gordon and Betty Moore Foundation's EPIQS Initiative (grant numbers GBMF3850 and GBMF9073). S.H., X.L. and S.V. disclose support for the research of this work from the US Department of Energy (grant number DE-SC 0012375).

Competing interests

T.N. and Yuejie Zhang are inventors on CN patent CN116847719B and US patent application US20250062060A1. The other authors declare no competing interests.

Additional information

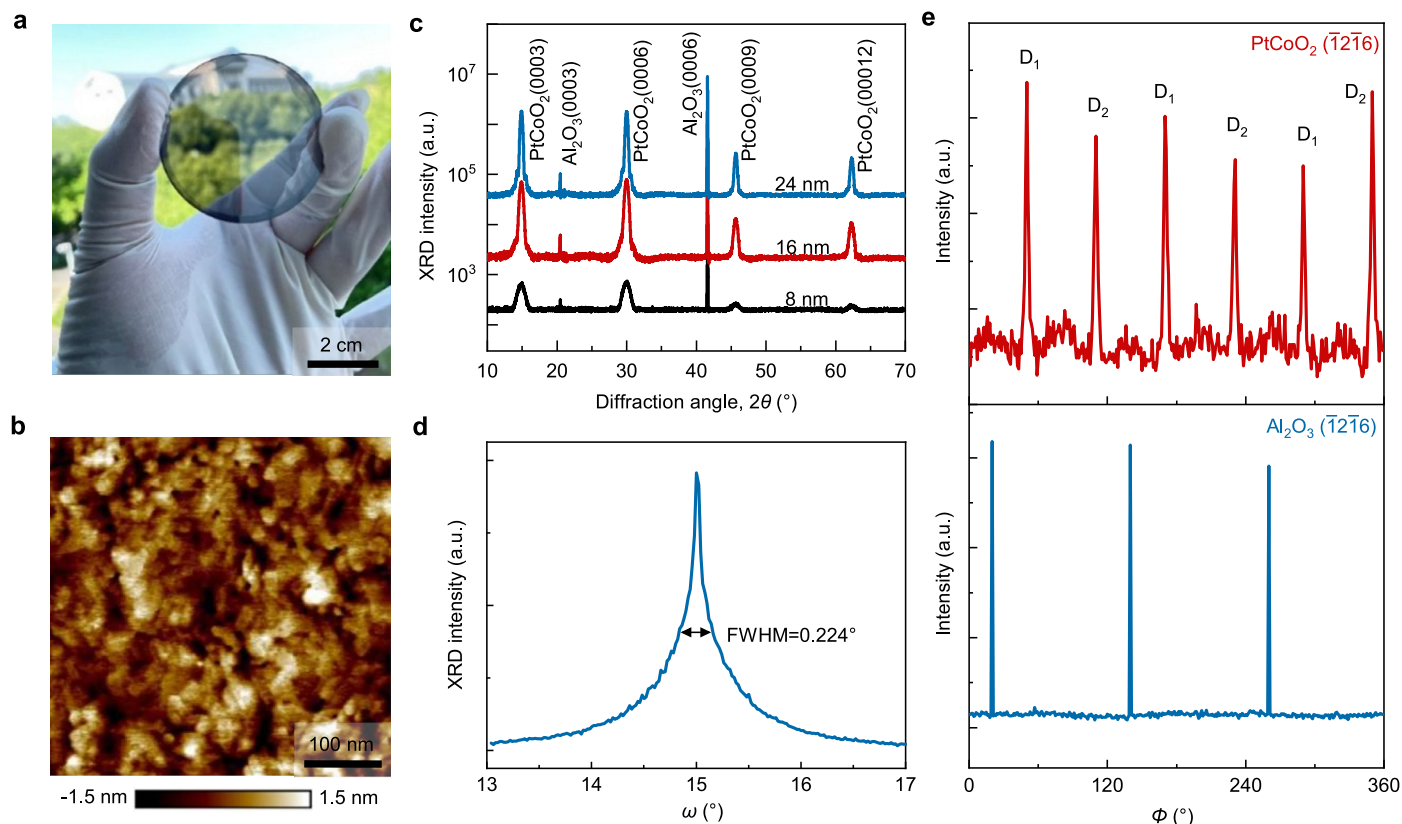
Extended data is available for this paper at <https://doi.org/10.1038/s41563-026-02640-4>.

Supplementary information The online version contains supplementary material available at <https://doi.org/10.1038/s41563-026-02640-4>.

Correspondence and requests for materials should be addressed to Xuebin Zhu, Di Yi or Tianxiang Nan.

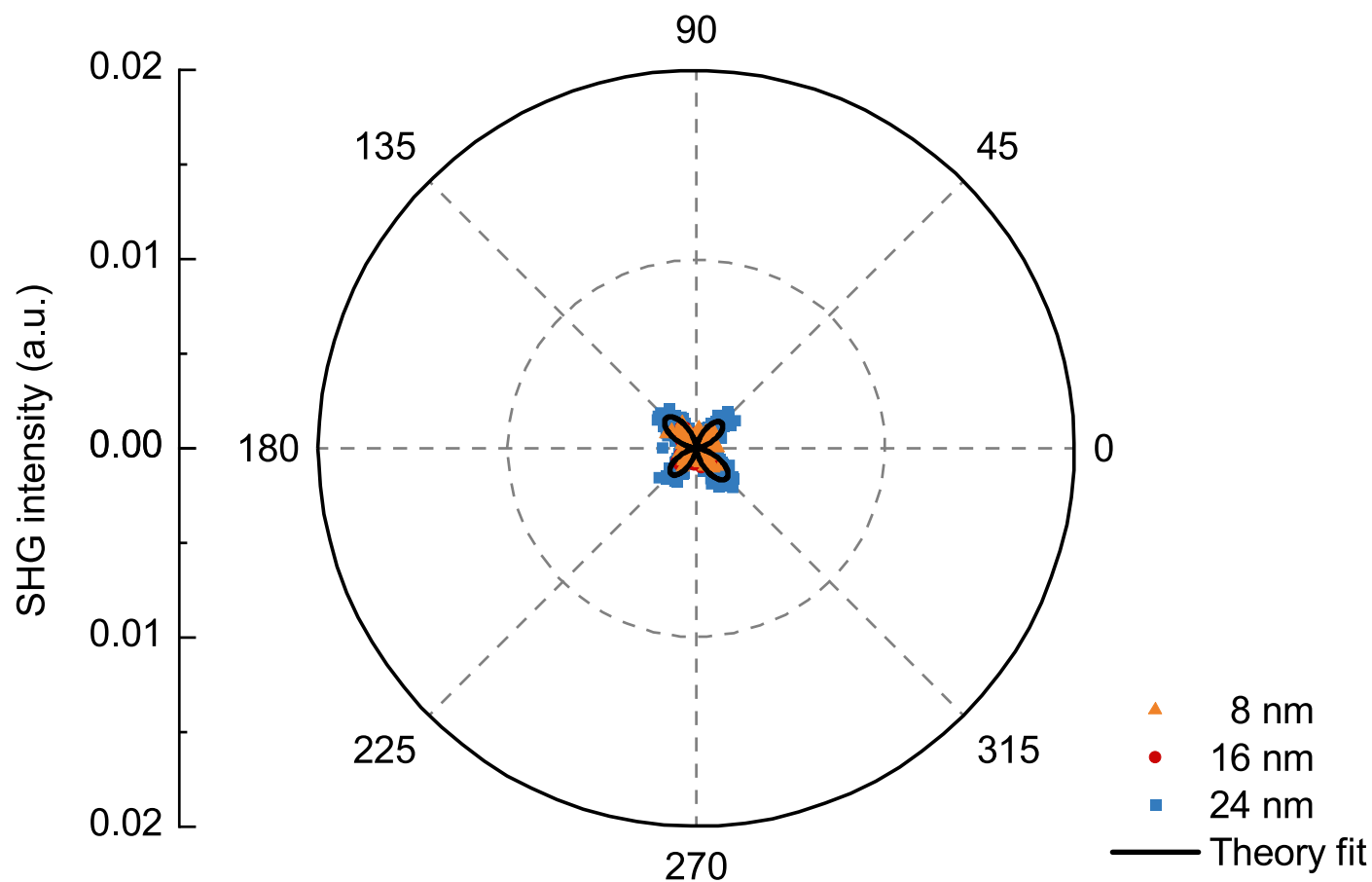
Peer review information *Nature Materials* thanks the anonymous reviewers for their contribution to the peer review of this work. Peer reviewer reports are available.

Reprints and permissions information is available at www.nature.com/reprints.



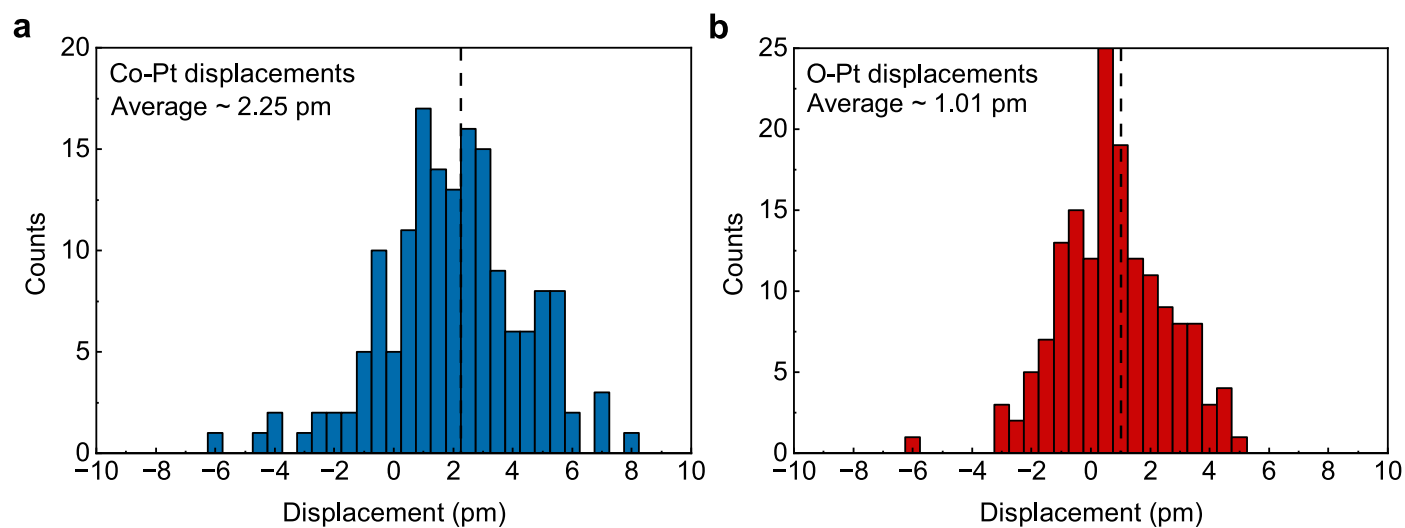
Extended Data Fig. 1 | Morphology and structural characterization of CSD-grown PtCoO₂ thin films. **a**, Optical image of PtCoO₂ on a 2-inch Al₂O₃(0001) wafer. **b**, Atomic force microscopy (AFM) image of the 8 nm PtCoO₂ thin film. The root-mean-square (RMS) surface roughness is ≈ 0.63 nm. The scale bar is 100 nm. **c**, Out-of-plane 2θ - ω X-ray diffraction patterns for PtCoO₂ thin films of 8, 16 and 24 nm thicknesses. **d**, Rocking curve around the (0006)

diffraction peak of the 8 nm PtCoO₂ thin film, showing a full-width-at-half-maximum (FWHM) of approximately 0.224°. **e**, Azimuthal ϕ -scan of the (112)6 diffraction peaks from the 8 nm PtCoO₂ thin film, confirming the epitaxial relationship with the Al₂O₃ substrate. Two structural domains, labelled as D₁ and D₂, are observed.

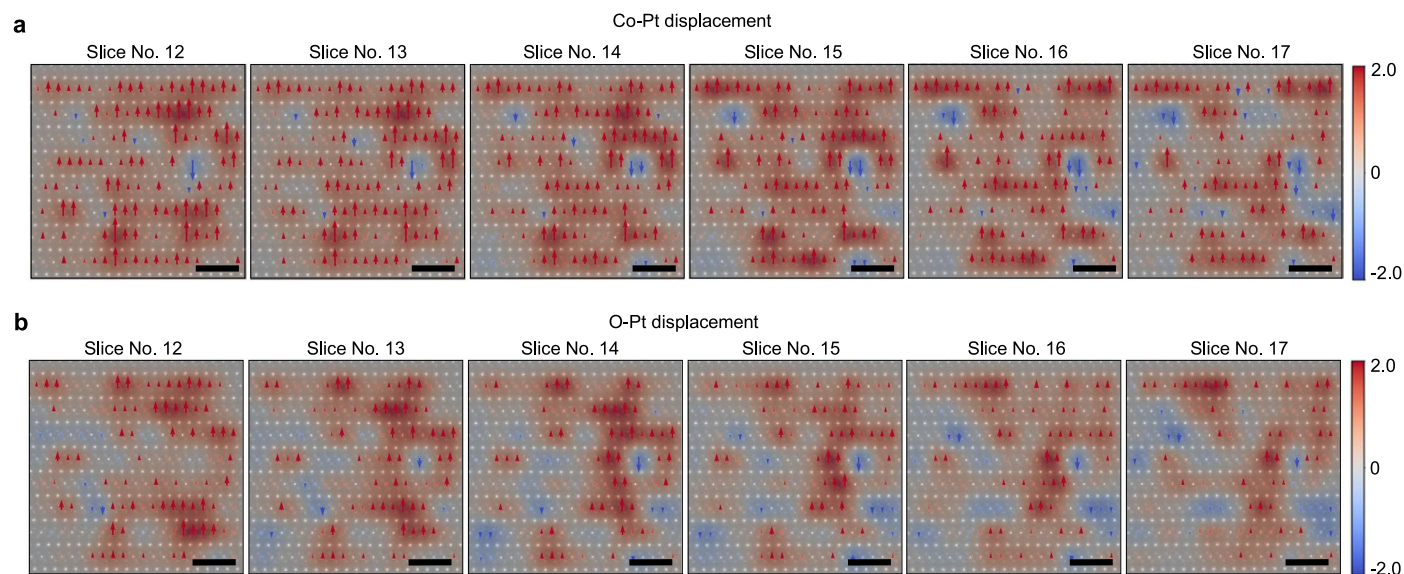


Extended Data Fig. 2 | S-out SHG signals for CSD-grown PtCoO₂ films. Optical second-harmonic generation (SHG) signal E_{2y} of PtCoO₂ thin films as a function of incident polarization angle ψ of the incident beam measured in the s-out configuration. Note that they are plotted in the same intensity scale as for the

p-out configuration (Fig. 1c in the main text). All s-out signals are relatively small compared to the p-out signals, indicating minimal in-plane (IP) polarity. The solid lines are fitting based on the functions derived from C_{3v} point symmetry (see Methods).

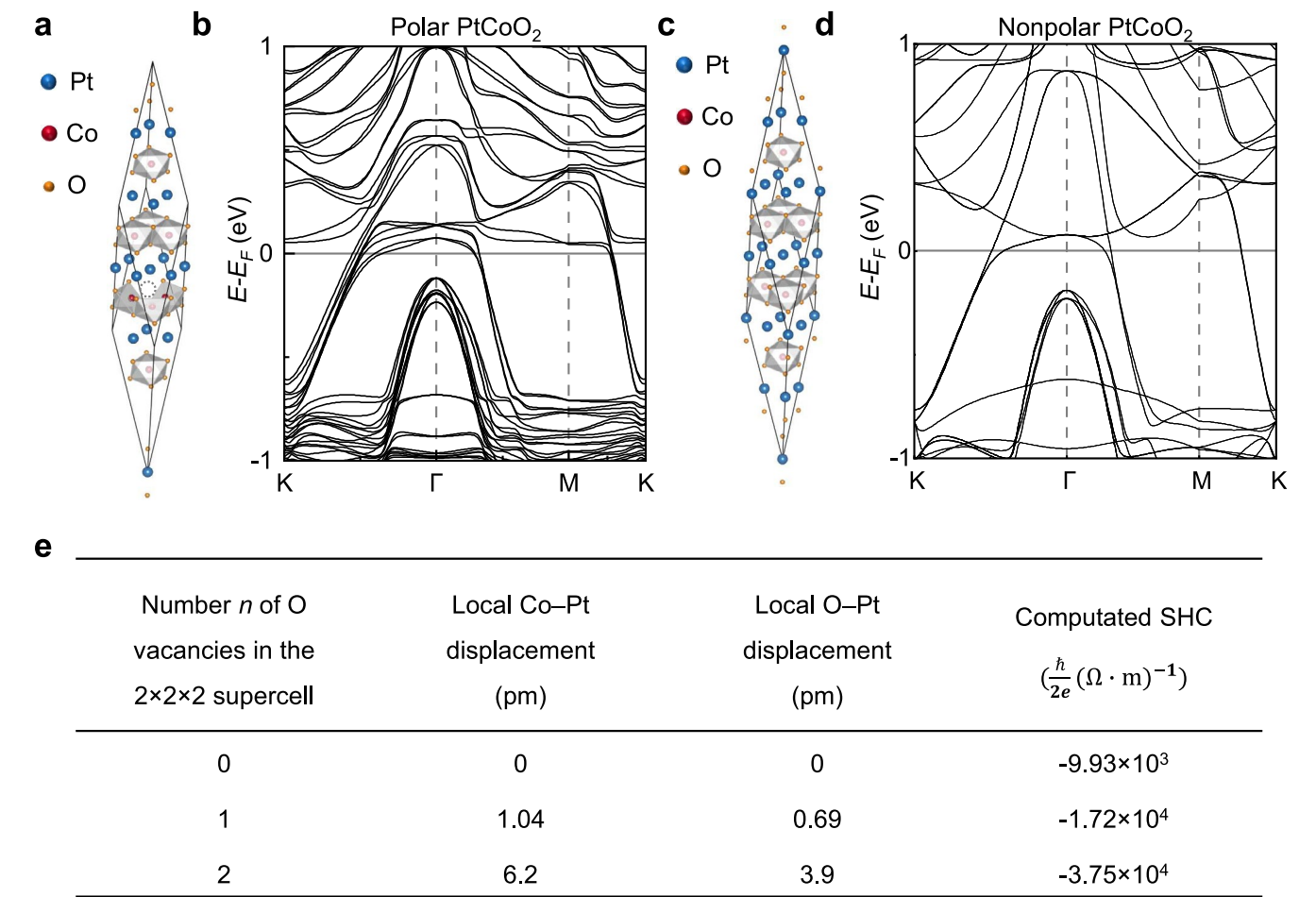


Extended Data Fig. 3 | Statistical analysis of the polar displacement distributions. a, b, Histograms of the (a) Co-Pt and (b) O-Pt displacements.



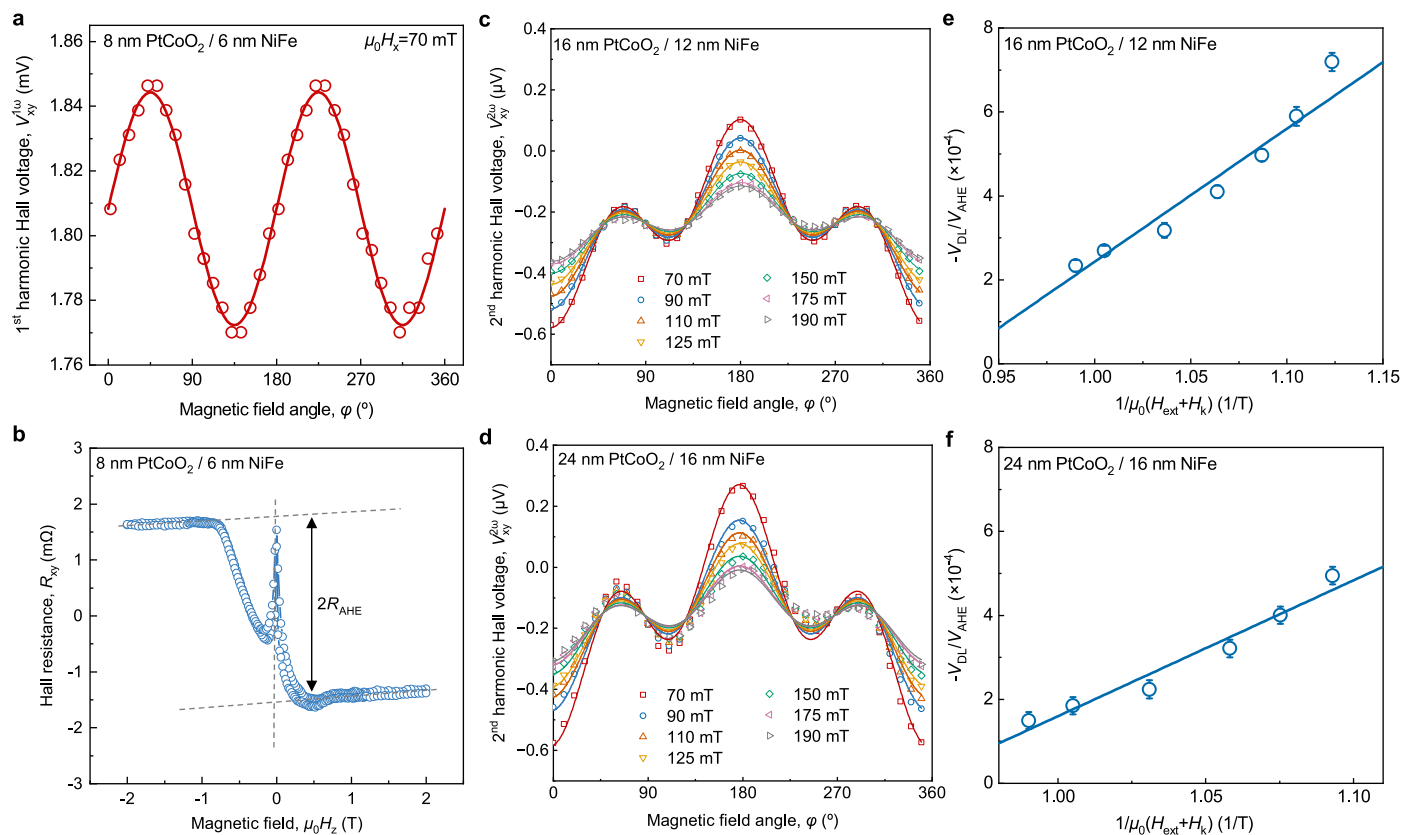
Extended Data Fig. 4 | Depth-resolved ptychographic reconstruction.
a, Co–Pt displacements and **b**, O–Pt displacements from slices 12th to the 17th of a CSD-grown PtCoO₂ specimen, where a total 30 slices in depth are reconstructed.

To exclude possible artifact signals near the specimen surface, only the middle layers were analyzed for displacement mapping. All scale bars correspond to 1 nm.



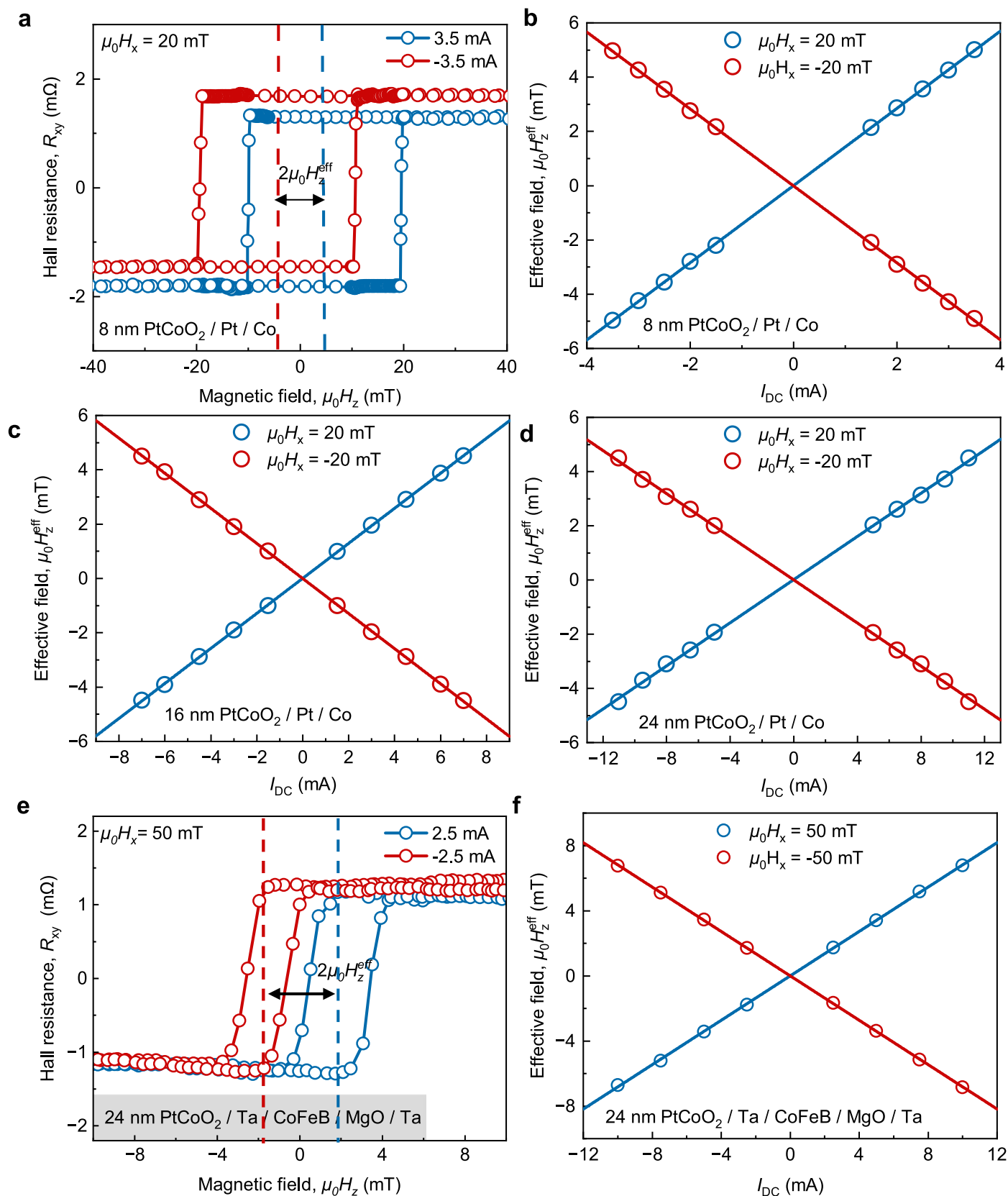
Extended Data Fig. 5 | Density functional theory (DFT) calculations. **a**, Schematic of a 2 × 2 × 2 PtCoO₂ supercell containing a single oxygen vacancy (grey dashed circle). **b**, Calculated band structure corresponding to the oxygen-vacancy supercell shown in **a**, exhibiting band splitting near the Fermi level. **c**, Schematic of a pristine 2 × 2 × 2 PtCoO₂ supercell without vacancies.

d, Calculated band structure for the pristine supercell shown in **c**. **e**, Summary of DFT results showing the number of oxygen vacancies in the 2 × 2 × 2 supercell, local Co–Pt and O–Pt displacements near the vacancy in the relaxed structure, and the corresponding computed spin Hall conductivity magnitude ($|\sigma_{2x}^y|$).



Extended Data Fig. 6 | Angle-dependent second-harmonic Hall measurement on PtCoO₂/NiFe heterostructures. **a**, Angular dependence of the first-HHV for the 8 nm PtCoO₂/6 nm NiFe heterostructure. **b**, Hall resistance of the 8 nm PtCoO₂/6 nm NiFe heterostructure measured under an out-of-plane magnetic field sweep. **c,d** Angular dependence of the second-HHV for (c) 16 nm PtCoO₂/12 nm NiFe and (d) 24 nm PtCoO₂/16 nm NiFe heterostructures at various external magnetic fields H_{ext} . The solid lines are fits to $V_{xy}^{2\omega} = V_{\text{DL}} \cos \varphi + V_{\text{FL}} \cos \varphi \cos 2\varphi$. **e,f** Extracted normalized $-V_{\text{DL}}/V_{\text{AHE}}$ plotted

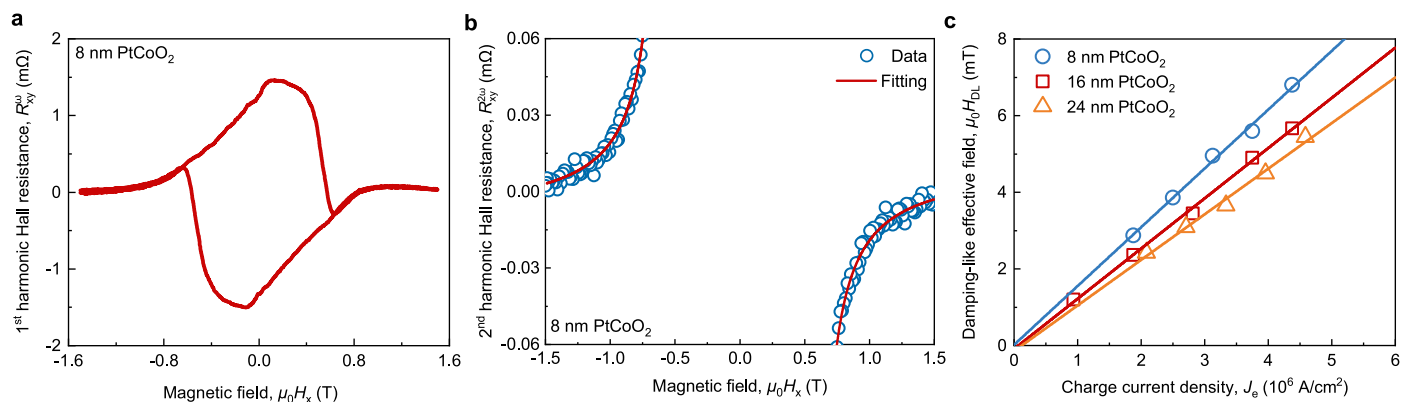
against $\frac{1}{\mu_0(H_{\text{ext}} + H_k)}$, with linear fits (solid lines), for the (e) 16 nm PtCoO₂/12 nm NiFe and (f) 24 nm PtCoO₂/16 nm NiFe heterostructures. The corresponding damping-like effective fields H_{DL} are (0.21 ± 0.02) mT per MA·cm⁻² and (0.11 ± 0.01) mT per MA·cm⁻² from e and f, respectively. In e and f, data points represent fit values obtained from the angular second-HHV fits in c and d, respectively; error bars indicate standard errors of the fitted V_{DL} values and represent fitting uncertainty.



Extended Data Fig. 7 | See next page for caption.

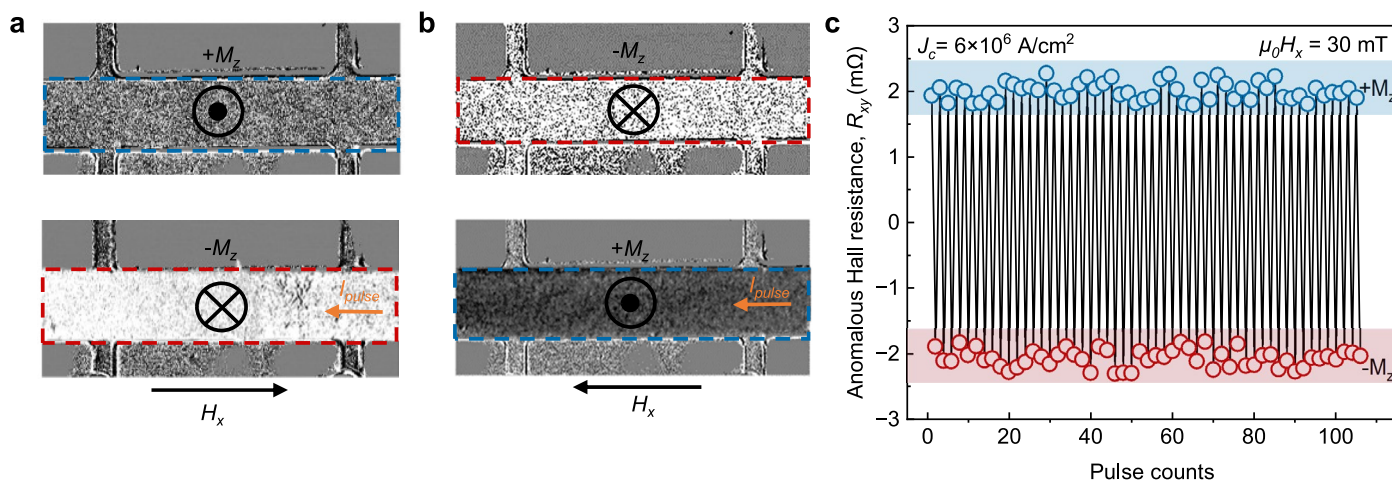
Extended Data Fig. 7 | Current-induced hysteresis loop shift measurements of PtCoO₂/Pt/Co heterostructures. **a**, Representative anomalous Hall resistance loop shift for the 8 nm PtCoO₂/ [Pt (0.8 nm)/Co (0.5 nm)]×2 / Pt (1 nm) heterostructure under a dc current $I_{\text{DC}} = \pm 3.5$ mA and in-plane magnetic fields $\mu_0 H_x = \pm 20$ mT. **b–d**, Extracted effective fields $\mu_0 H_z^{\text{eff}}$ plotted as a function of I_{DC} for the **(b)** 8 nm, **(c)** 16 nm, and **(d)** 24 nm PtCoO₂/ [Pt (0.8 nm)/Co (0.5 nm)]×2 / Pt (1 nm) heterostructures. All measurements were taken with $\mu_0 H_x = \pm 20$ mT. Solid lines are linear fits, yielding damping-like effective fields of (0.84 ± 0.07)

mT per MA·cm⁻², (0.71 ± 0.14) mT per MA·cm⁻² and (0.67 ± 0.09) mT per MA·cm⁻² for **b**, **c**, and **d**, respectively. **e**, Representative anomalous Hall resistance loop shift for the 24 nm PtCoO₂/ 2 nm Ta / 1.1 nm CoFeB / 1 nm MgO / 3 nm Ta heterostructure under a dc current $I_{\text{DC}} = \pm 2.5$ mA and in-plane magnetic fields $\mu_0 H_x = \pm 50$ mT. **f**, Extracted effective fields $\mu_0 H_z^{\text{eff}}$ plotted as a function of I_{DC} for the 24 nm PtCoO₂/ 2 nm Ta / 1.1 nm CoFeB / 1 nm MgO / 3 nm Ta heterostructure. Solid lines are linear fits, yielding damping-like effective fields of (2.31 ± 0.03) mT per MA·cm⁻².



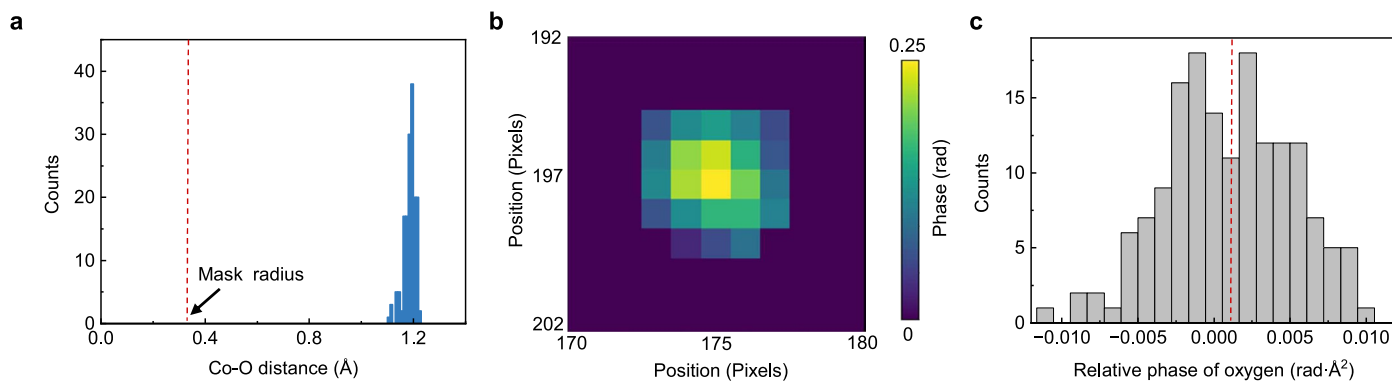
Extended Data Fig. 8 | High-field harmonic Hall measurements of PtCoO₂/Pt/Co heterostructures. **a**, First-harmonic Hall resistance $R_{xy}^{1\omega}$ for the 8 nm PtCoO₂/ [Pt (0.8 nm)/Co (0.5 nm)] $\times$ 2/Pt (1 nm) heterostructure measured under an in-plane magnetic field sweep (along the current axis). **b**, Second-harmonic Hall resistance $R_{xy}^{2\omega}$ of the 8 nm PtCoO₂/ [Pt (0.8 nm)/Co (0.5 nm)] $\times$ 2/Pt (1 nm) heterostructure. Solid line is a fit to $V_{xy}^{2\omega} \approx -\frac{1}{2} V_{AHE} \frac{H_{DL}}{(H_{ext} - H_k)} + V_{ANE}$.

c, Extracted damping-like effective field $\mu_0 H_{DL}$ plotted as a function of current densities J_c for heterostructures with different PtCoO₂ thicknesses. Solid lines are linear fits, yielding (1.1 ± 0.1) mT per MA $\cdot$ cm $^{-2}$, (0.91 ± 0.05) mT per MA $\cdot$ cm $^{-2}$ and (0.84 ± 0.11) mT per MA $\cdot$ cm $^{-2}$ for the 8 nm, 16 nm, and 24 nm PtCoO₂/ [Pt (0.8 nm)/Co (0.5 nm)] $\times$ 2/Pt (1 nm), respectively.



Extended Data Fig. 9 | SOT-induced magnetization switching probed by polar MOKE microscopy and endurance of SOT switching cycles. Polar MOKE images of an 8 nm PtCoO₂/[Pt (0.8 nm)/Co (0.5 nm)] $\times$ 2/Pt (1 nm) heterostructure grown on Al₂O₃ substrate. **a**, Initialization to $+M_z$ (top panel) followed by switching to $-M_z$ after applying a pulsed current under an in-plane field of 30 mT (bottom panel). **b**, Initialization to $-M_z$ (top panel) followed by switching to $+M_z$

after applying a pulsed current under a reversed in-plane field (bottom panel). The contrast change between black and white regions presents magnetization reversal between $\pm M_z$ states. **c**, Anomalous Hall resistance measured during successive positive and negative current pulse cycles in an 8 nm PtCoO₂/[Pt (0.8 nm)/Co (0.5 nm)] $\times$ 2/Pt (1 nm) heterostructure grown on Al₂O₃ substrate.



Extended Data Fig. 10 | Electron ptychography analysis. **a**, Histogram of the Co–O distance. As indicated by the red line, the O mask radius is approximately 1/3 of the Co–O distance. This ensures a pure O signal in the ptychographic intensity analysis, given the narrow Co and O atomic widths in the ptychographic image (Fig. 1d). **b**, Illustration of the product of the mask and the original phase map at the position (170–180, 190–202 pixels) for a single oxygen column. The

integrated phase is defined as the sum of the phase values of the pixels within the mask region. **c**, Statistical distribution of the relative phase of oxygen in Fig. 1d, which is calculated by taking the difference in the integrated phases between two adjacent oxygen atoms for each Co atom and multiplying by the pixel area (0.0144 Å^2). The red dashed line marks the average value of the distribution (-0.0012 rad·Å^2).