Large Spin Hall Conductivity in Epitaxial Thin Films of Kagome Antiferromagnet Mn₃Sn at Room Temperature

Himanshu Bangar, KachoImtiyaz Ali Khan, Akash Kumar, Niru Chowdhury, Prasanta Kumar Muduli, and Pranaba Kishor Muduli*

Mn₃Sn is a non-collinear antiferromagnetic quantum material that exhibits a magnetic Weyl semimetallic state and has great potential for efficient memory devices. High-quality epitaxial c-plane Mn₃Sn thin films have been grown on a sapphire substrate using a Ru seed layer. Using spin pumping induced inverse spin Hall effect measurements on c-plane epitaxial Mn₃Sn/ Ni₈₀Fe₂₀, spin-diffusion length (λ_Mn₃Sn), and spin Hall conductivity (σ_SH) of Mn₃Sn thin films are measured: λ_Mn₃Sn = 0.42 ± 0.04 nm and σ_SH = −702 h/e Ω⁻¹ cm⁻¹. While λ_Mn₃Sn is consistent with earlier studies, σ_SH is an order of magnitude higher and of the opposite sign. The behavior is explained on the basis of excess Mn, which shifts the Fermi level in these films, leading to the observed behavior. These findings demonstrate a technique for engineering σ_SH of Mn₃Sn films by employing Mn composition for functional spintronic devices.

1. Introduction

Recently, there has been a huge amount of interest in quantum materials for the field of spintronics, which makes use of the electron’s spin degree of freedom. Spin-charge interconversion is important for the application of spintronics. Spin current can be generated from the charge current by a mechanism such as spin Hall effect (SHE), which is typically observed in non-magnetic heavy metals. However, recently non-collinear antiferromagnetic quantum materials have gained attention as potential spin Hall materials. This is driven by: 1) theoretical studies that predict large intrinsic spin Hall conductivity (σ_SH) (1,4) (2) the experimental observation of large anomalous Hall effect (σ_AH) and 3) the observation of un-conventional spin-orbit torque in these materials. Mn₃Sn is one example of non-collinear antiferromagnet that exhibits exotic properties such as anomalous Hall effect, anomalous Nernst effect, and magneto-optic Kerr effect despite having nearly zero magnetization. These exotic properties originate from the Berry curvature associated with Weyl points near the Fermi energy. More recently, new phenomena such as magnetic spin Hall effect chiral domain walls as well as spin-orbit torque induced chiral-spin rotation has been demonstrated in Mn₃Sn. These reports suggest that Mn₃Sn is a promising quantum material for antiferromagnetic spintronics; a rapidly developing field that offers several advantages such as zero stray field, robustness against magnetic field perturbation, and ultrafast THz dynamics.

Theoretical works predict an intrinsic σ_SH in Mn₃Sn due to its non-collinear magnetic structure. Only a few works are reported on the measurement of SHE in Mn₃Sn. The σ_SH of polycrystalline Mn₃Sn was recently reported to be σ_SH ~ 47 h/e Ω⁻¹ cm⁻¹ based on non-local spin transport experiments. Yu et al. reported a spin Hall angle (θ_SH) of 0.144 using the spin pumping induced inverse spin Hall effect (ISHE) approach, which is greater than that of Ta. To realize the full potential of Mn₃Sn for spintronics applications, it is important to estimate σ_SH as well as other important parameters like spin mixing conductance (g₁₁), and spin diffusion length (λ_Mn₃Sn) in epitaxial thin films. Recently, Yoon et al. reported a drastic change in transport properties with the crystalline orientation of Mn₃Sn thin films. Theoretical studies also predict strong anisotropic θ_SH for Mn₃Sn. Consequently, it is anticipated that the values of g₁₁, λ_Mn₃Sn, and σ_SH will depend on the orientation of Mn₃Sn thin films.

In this work, we present direct measurement of σ_SH in high quality c-plane oriented Mn₃Sn thin films. We grow c-plane oriented Mn₃Sn on Al₂O₃ substrate using Ru as the seed layer. We determine the σ_SH by employing spin pumping driven ISHE measurements on epitaxial Mn₃Sn/ Ni₈₀Fe₂₀ bilayers. In contrast to previously reported investigations, σ_SH of our epitaxial Mn₃Sn thin films is an order of magnitude greater, while λ_Mn₃Sn is comparable. We also report g₁₁ of this system to be (1.54 ± 0.27) x10¹⁹ m⁻², which is also an order of magnitude higher compared to Py/Ta bilayers. The sign of σ_SH is found to be negative.
which can be explained by a shift of the Fermi level caused by a slight excess Mn in our films.

2. Results

Mn$_3$Sn has a hexagonal crystal structure, for which the (0001)-plane or the c-plane has a kagome lattice as shown in Figure 1a. Figure 1b shows the X-ray reflectivity (XRR) measurement of the Al$_2$O$_3$/Ru (7 nm)/Mn$_3$Sn (30 nm)/AlO$_x$ (2 nm) thin film. The thickness of Mn$_3$Sn is extracted from the XRR fit, from which we determine the growth rate to be $\approx 0.48$ Å s$^{-1}$. We obtained a low roughness ($\approx 0.28$ nm) from the fit of the XRR data. The AFM surface morphology also showed a lower root mean square roughness of $\approx 0.15$ nm measured over a scan area of 2 $\mu$m x 2 $\mu$m as presented in Figure 1c. Figure 1d shows the corresponding $2\theta$–$2\theta$ X-ray diffraction (XRD) pattern, showing the (0002)-peaks corresponding to both Mn$_3$Sn and Ru seed layer. We also observe several satellite peaks (thickness fringes), indicating smooth interfaces and uniform film growth, which is also consistent with sharp Kiessig fringes observed in the XRR measurements (Figure 1b). We only observed (0002) and (0004) Mn$_3$Sn peaks, indicating that a c-plane oriented epitaxial Mn$_3$Sn film has been achieved on the Ru seed layer. Similar XRD pattern was recorded in lower thickness of Mn$_3$Sn films (see Figure 1e). To determine the epitaxial relationship, we performed $\varphi$-scans for the peaks: Al$_2$O$_3$ (1123) (violet), Ru (1011) (cyan), and Mn$_3$Sn (20 $\overline{2}$ 1) (wine) as presented in Figure 1f. The $\varphi$-scan clearly shows that a reflection appears periodically every 60° for Al$_2$O$_3$, Ru, and Mn$_3$Sn, indicating that we have obtained non-twinned, highly epitaxial films. The location of the peaks indicates that the epitaxial film is formed with a relationship of Mn$_3$Sn (0001)[20 $\overline{2}$ 0] || Ru (0001)[10 $\overline{1}$ 0] || Al$_2$O$_3$ (0001)[11 $\overline{2}$ 0], which is similar to the work by S. Kurdi, et al.$^{[26]}$

In order to determine the composition and its distribution, we used an electron probe microanalyzer (EPMA) and energy dispersive X-ray (EDX) mapping on our samples. The composition of the film is estimated to be Mn$_{1.12\pm0.03}$Sn$_{0.88\pm0.01}$ using EDX mapping and Mn$_{1.2\pm0.02}$Sn$_{0.88\pm0.01}$ by quantitative analysis using EPMA system. We did the composition analysis in 30 and 10 nm Mn$_3$Sn samples, both showed similar results as discussed in Section S1, Supporting Information. Both measurements show the presence of excess Mn, which is known to be essential for the formation of the D0$_{19}$ Mn$_3$Sn.$^{[27]}$ Figure 2a,b provides the elemental distribution maps of Mn and Sn atoms, respectively, showing that the elements are uniformly distributed. Room temperature magnetization ($M$) measurement performed using superconducting quantum interference device (SQUID) on a 30 nm thick Mn$_3$Sn thin film is shown in Figure 2c. It shows a weak magnetic moment of around 2 $\mu$B/Mn at room temperature, which is lower than polycrystalline films, as well as reported results on c-plane Mn$_3$Sn thin films.$^{[2]}$ Magnetization ($M$) measurement at 300 K for Mn$_3$Sn/Ni$_8$Fe$_{20}$ (8 nm) bilayer is shown in the inset of Figure 2c, from which we determine saturation magnetization of Ni$_8$Fe$_{20}$ to be 750 emu cc$^{-1}$. We measured the longitudinal resistivity of Ru (7 nm)/Mn$_3$Sn (30 nm) bilayer using four point probe method to be 570 $\mu$Ω-cm.
Figure 2. Elements distribution maps for a) Mn and b) Sn measured using electron probe microanalyzer (EPMA). c) SQUID magnetization ($M$) curve of Mn$_3$Sn (30 nm) sample. Inset: Magnetization ($M$) curve of Mn$_3$Sn (30 nm)/Ni$_{80}$Fe$_{20}$ (8 nm) bilayer thin film.

Figure 3. a) The measured FMR signal at 4 GHz for samples (Al$_2$O$_3$/Ru/Mn$_3$Sn($t_{Mn3Sn}$)/Ni$_{80}$Fe$_{20}$/AlO$_x$) having different thickness, $t_{Mn3Sn}$ of Mn$_3$Sn near the resonance field ($H_r$). b) Linewidth ($\Delta H$) versus frequency ($f$) (open circle) and their corresponding fit (solid lines) with Equation (1). c) $f$ versus $H_r$ curve for different $t_{Mn3Sn}$ (open circle) and their corresponding fit (solid lines) with Equation (2).

In Figure 3a, we show example FMR spectra of Al$_2$O$_3$/Ru/Mn$_3$Sn($t_{Mn3Sn}$)/Ni$_{80}$Fe$_{20}$/AlO$_x$ for various thicknesses, $t_{Mn3Sn}$ of Mn$_3$Sn measured at 4 GHz. The FMR data is fitted to extract the values of half-width at half maximum or linewidth ($\Delta H$) and resonance field ($H_r$). The $\Delta H$ versus frequency ($f$) dependence of all the samples are plotted in Figure 3b. The measured data is fitted using following equation:

$$\Delta H = \frac{2\pi \alpha_{eff}}{\gamma} + \Delta H_0$$

where $\alpha_{eff}$ is the effective Gilbert damping parameter, $\gamma = 1.85 \times 10^2$ GHz T$^{-1}$ is the gyromagnetic ratio, and $\Delta H_0$ is the inhomogeneous line broadening. The first term on the right hand side is the viscous damping of magnetization motion, while the second term is due to magnetic inhomogeneity and sample imperfections of the FM layer. The slope of the linear fit using Equation (1) is proportional to $\alpha_{eff}$. As can be observed from Figure 3b, the slope of Mn$_3$Sn($t_{Mn3Sn}$)/Ni$_{80}$Fe$_{20}$ (8 nm) sample, which indicates spin current being pumped from Ni$_{80}$Fe$_{20}$ into Mn$_3$Sn. Further, the low values of $\Delta H_0$ ($<3$ Oe) from the fits indicate the high quality of our samples. Figure 3c shows the $f$ dependent $H_r$ (open circles) for various $t_{Mn3Sn}$ with their corresponding fit (solid line) using Kittel’s equation:

$$f = \frac{\gamma}{2\pi} [(H_r + H_k)(H_r + H_k + 4\pi M_{eff})]^{1/2}$$

where $H_k$ is the uniaxial anisotropy field, and $M_{eff}$ is the effective saturation magnetization. The $M_{eff}$ is found to be comparable for all the Mn$_3$Sn($t_{Mn3Sn}$)/Ni$_{80}$Fe$_{20}$ samples and in the range of 740 – 770 emu cc$^{-1}$, which is close to measured saturation magnetization of 750 emu cc$^{-1}$ (estimated from magnetization measurements), indicating negligible perpendicular anisotropy.

Next, we performed the Mn$_3$Sn thickness dependent ISHE measurements to determine the $\theta_{SH}$ of our c-plane Mn$_3$Sn films. Figure 4a shows a representative ISHE signal, $V_{ISHE}$ (open
circular) along with an ISHE signal measured from a Ta/Pt/SiO₂ sample (inset). To eliminate contribution from self-induced ISHE and other rectification effects from Ni₈₀Fe₂₀[32–34] the signal from a reference (Ni₈₀Fe₂₀) sample is subtracted from the measured data (refer to Section S2, Supporting Information, for more details). The sign of signal obtained from Mn₃Sn agrees with Ta, indicating a negative sign of θ₁SH. The ISHE data (V_{ISHE}) was fitted with a combination of symmetric (V_{sym}^{ISHE}) and asymmetric (V_{asym}^{ISHE}) components using equation[35]

\[ V_{ISHE} = V_{sym}^{ISHE} \left( \frac{\Delta H}{2} \right)^2 + V_{asym}^{ISHE} \left( \frac{\Delta H}{2} + (H - H_0) \right)^2 + \frac{2\Delta H (H - H_0)}{2} \]  

where \( H \) is the applied magnetic field. The value of the symmetric part can be taken to be the spin pumping induced ISHE signal \( V_{sym}^{ISHE} \) in our geometry where rectification signals are minimized.[36] This is further supported by the fact that signal shape is entirely symmetric and the signal changes sign on reversal of field polarity, both of which are consistent with a dominant ISHE origin of the signal.[37] Furthermore, in our geometry the magnetic field is perpendicular to the voltage measurement direction. For this condition, the magnetic spin Hall effect vanishes completely (see Figure 3g of ref. [13]). Hence, in the following, we will not consider magnetic spin Hall effect or its inverse in Mn₃Sn. The charge current generated due to ISHE in Mn₃Sn can be written as \( V_{sym}^{ISHE} / R \), where \( R \) is the resistance of the Mn₃Sn/Ni₈₀Fe₂₀ bilayer. We further normalize the \( V_{sym}^{ISHE} / R \) with the width \( w \) of the sample to eliminate any size effect. In order to determine \( \theta_1SH \) and \( \lambda_{Mn₃Sn} \) of Mn₃Sn we plot \( V_{sym}^{ISHE} / Rw \) with \( t_{Mn₃Sn} \) (Figure 4b) and fit it with the following equation[35]

\[ \frac{V_{sym}^{ISHE}}{Rw} = \theta_{1SH} \frac{h}{\sigma_{SH}} \gamma \frac{1}{1 + \sin^2(\lambda_{Mn₃Sn} / 2t_{Mn₃Sn})} \]  

where \( h_{RF} \) is the RF field generated due to the RF current of frequency \( f = \alpha / 2 \pi \) flowing through the co-planar waveguide, \( 4\pi M_s \) is the saturation magnetization, \( h \) is the reduced Planck’s constant, and \( \epsilon \) is the electronic filling factor. From the fitting, we obtain the values of \( \lambda_{Mn₃Sn} \) and \( \theta_{1SH} \) to be \( 0.42 \pm 0.04 \) nm and \( -0.40 \pm 0.03 \) respectively. The value of \( g_{1SH} \) is determined from the enhancement of damping using \( g_{1SH} = \Delta \alpha_{eff} / (4\pi M_s \lambda_{Mn₃Sn} / (8\pi \mu_0)) \). The value is found to be \( g_{1SH} = 1.54 \pm 0.27 \times 10^{-13} \) m² s⁻¹ at room temperature. This value of \( g_{1SH} \) is comparable to the values reported for other systems such as Mn₃Ga/CoFeB[38] and Bi₂Se₃/Ni₈₀Fe₂₀[39] and one order of magnitude higher than Ta/Ni₈₀Fe₂₀ bilayer.[40] The value of \( \lambda_{Mn₃Sn} \) is comparable to the value reported for polycrystalline Mn₃Sn.[21] The measured \( \theta_{1SH} \) in our case is found to be rather large, especially considering the fact that Equation (4) does not include interface transparency, which is less than one. Thus \( -0.40 \pm 0.03 \) represents a lower limit of \( \theta_{1SH} \). Our value of \( \theta_{1SH} \) is greater than other non-collinear antiferromagnets, such as \( \theta_{1SH} \approx 0.31 \) for polycrystalline Mn₃Ga[38] and \( \theta_{1SH} \approx 0.35 \) for (001)-oriented IrMn₁[19]. In addition, the observed value of \( \theta_{1SH} \) is considerably higher than the polycrystalline Mn₃Sn value of \( \theta_{1SH} \approx 0.05 \), which was reported by one of the authors of this article using a non-local spin transport technique.[21] Our value is also more than twice that of Yu et al.[22] where a value of \( \theta_{1SH} \approx 0.18 \) was reported for polycrystalline Mn₁₃Sn thin films interfaced with yttrium iron garnet. A higher value of \( \theta_{1SH} \) in c-plane Mn₃Sn is in agreement with a previous report in which a much larger \( \theta_{1SH} \) of 0.35 is reported for (001)-oriented single crystalline IrMn₁ films compared to polycrystalline films.[19] Additionally, we believe an ultra-smooth, highly crystalline Mn₃Sn film helps in reducing the loss that may occur due to roughness which results in a relatively large efficiency of spin-to-charge conversion.

In order to compare our results with theoretical calculations[31] of SHE in Mn₃Sn, we determine \( \sigma_{SH} \) using: \( \sigma_{SH} = \sigma_\gamma \times \frac{\hbar}{e} \), where \( \sigma \) is the charge conductivity of Ru/Mn₃Sn bilayer, which was found to be \( 1754 \, \Omega^{-1} \cdot \text{cm}^{-1} \), from four-point probe measurements. We found \( \sigma_{SH} \) to be \( -702 \, \text{h/e} \cdot \Omega^{-1} \cdot \text{cm}^{-2} \). Zhang et al.[36] have predicted an intrinsic SHE in Mn₃Sn that arises due to the non-collinear magnetic structure. They predicted \( \sigma_{SH} \approx 90 \, \text{h/e} \cdot \Omega^{-1} \cdot \text{cm}^{-2} \) at the Fermi level. Our value is significantly larger and has the opposite sign, which can be explained if we assume the Fermi level is shifted in our thin films due to slightly higher Mn content. In fact, the band structure is found to be dominated by Mn-d orbitals near the Fermi level[40] and hence, an excess 3% Mn concentration (Mn₁₂,Sn₈₈) in our samples can induce electron doping leading to a shift in the Fermi level. Such shift in Fermi level due to excess Mn is already reported in Mn₃Sn both by first principle calculations as well as by angle-resolved photoemission spectroscopy measurements.[40] Based on the measured resistivity and a scattering time of 62.3 fs from Ref. [41], we calculated a shift of about 0.09 eV (with respect to stoichiometric Mn₃Sn with no excess Mn) which can easily lead to a sign change in \( \sigma_{SH} \), as shown in the inset of Figure 4b. For a quantitative agreement of \( \sigma_{SH} \) with experiment, the shift in the Fermi level needs to be around 0.2 eV. The experimental shift in the Fermi level might as well be larger than that we calculate with very simple assumptions.

### 3. Conclusion

In conclusion, we have demonstrated epitaxial growth of c-plane oriented non-collinear antiferromagnet Mn₃Sn with Ru as a seed layer on Al₂O₃ substrates. We have investigated ISHE in the c-plane Mn₃Sn/ Ni₈₀Fe₂₀ system. Through Mn₃Sn thickness dependent ISHE measurements, we determine key parameters like \( \theta_{1SH} \) and \( \lambda_{Mn₃Sn} \) for c-plane Mn₃Sn. We found a large \( \sigma_{SH} \) of \(-702 \, \text{h/e} \cdot \Omega^{-1} \cdot \text{cm}^{-2} \), which is higher than other non-collinear antiferromagnets reported till date. The results are important for spin-orbit torque-based spintronic devices utilizing non-collinear antiferromagnets.

### 4. Experimental Section

**Sample Preparation:** In order to grow the c-plane Mn₃Sn, a c-plane sapphire (c-Al₂O₃) substrate and a 7 nm thick Ru-seed layer was chosen. The films were grown using AJA Orion 8 magnetron sputtering system with a base pressure, better than 5 × 10⁻⁸ Torr. First, the Ru seed layer was deposited at 400 °C and annealed at the same temperature for 10 min. Then the sample was allowed to cool to 100 °C, at which Mn₃Sn was deposited by co-sputtering Mn and Sn targets. The Mn was deposited at a growth rate of 0.37 Å s⁻¹ by applying 60 W DC power while the Sn was deposited at a growth rate of 0.20 Å s⁻¹ by applying 40 W RF power. After the
deposition of Mn$_3$Sn layer, the sample was annealed in situ at 600 °C for an hour. Subsequently, the sample was allowed to cool to room temperature, after which Ni$_{90}$Fe$_{20}$ (8 nm)/Al (2 nm) was deposited without breaking the vacuum. The Al layer was used as the capping layer and was fully oxidized after exposure to air. During the deposition, the sample holder was rotated along its normal axis to ensure a uniform composition and thickness. Field modulation technique (EPMA-1720HT) EPMAsystem was used to determinethecomposition. Imaging Asylum Research Probes (AC240TS-R3) cantilevers. The SHIMADZU determined by atomic force microscopy (AFM) scans (Asylum Research, 2018, 20, 073028).

Sample Characterization: The structural characterization was performed using high resolution X-ray diffraction (HR-XRD), while the growth rate and interfacial roughness were determined via X-ray reflectivity (XRR) measurements using a PANanalytical X-ray diffractometer equipped with Cu K$_\alpha$ (λ = 1.5406 Å) source. The morphology and surface roughness was determined by atomic force microscopy (AFM) scans (Asylum Research, www.advancedsciencenews.com www.advquantumtech.com)

FMR and ISHE Measurements: A broadband ferromagnetic resonance (FMR) spectroscopy technique was used to characterize the magnetization dynamics of Mn$_3$Sn/Ni$_{90}$Fe$_{20}$ bilayers. Field modulation technique with lock-in based detection was employed to enhance the sensitivity of FMR measurements. The excitation radio frequency (RF) was varied from 3 to 8 GHz. To detect the ISHE voltage, copper pads were pasted beneath the inverted sample following the method used in the earlier work by Kumar et al.[24]

Supporting Information

Supporting Information is available from the Wiley Online Library or from the author.

Acknowledgements

The partial support from the Ministry of Human Resource Development, Government of India under the IMPRINT program (Grant no: 7519 and 7058), the Department of Electronics and Information Technology (Deity), Science & Engineering research board, Government of India (SERB File no. CRG/2018/001012), Joint Advanced Technology Centre at IIT Delhi, and Department of Science and Technology under the Nanomission program (grant no: SR/NM/NT – 1041/2016(G)) are gratefully acknowledged. H.B. gratefully acknowledges the financial support from the Council of Scientific and Industrial Research (CSIR), Government of India. Partial support from I-HUB Quantum Technology Foundation, India is gratefully acknowledged.

Conflict of Interest

The authors declare no conflict of interest.

Data Availability Statement

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

Keywords

ferromagnetic resonance, inverse spin Hall effect, Mn$_3$Sn, non-collinear antiferromagnet, spin pumping, Weyl semimetals

Received: September 4, 2022
Revised: October 10, 2022
Published online: November 14, 2022

[1] W. Han, Y. Otani, S. Maekawa, npj Quantum Mater. 2018, 3, 27.
[2] I. Žutić, J. Fabian, S. D. Sarma, Rev. Mod. Phys. 2004, 76, 233.
[3] Y. Zhang, J. Železný, Y. Sun, J. Van Den Brink, B. Yan, New J. Phys. 2018, 20, 073028.
[4] O. Busch, B. Göbel, I. Mertig, Phys. Rev. B 2021, 104, 184423.
[5] S. Nakatsuji, N. Kiyohara, T. Higo, Nature 2015, 527, 212.
[6] K. Kondou, H. Chen, T. Tomita, M. Ikhlas, T. Higo, A. H. MacDonald, S. Nakatsuji, Y. Otani, Nat. Commun. 2021, 12, 6491.
[7] Y. Deng, R. Li, X. Liu, J. Alloys Compd. 2021, 874, 159910.
[8] Y. You, X. Chen, X. Zhou, Y. Gu, R. Zhang, F. Pan, C. Song, Adv. Electron. Mater. 2019, 5, 1800818.
[9] M. Ikhlas, T. Tomita, T. Koretsune, M.-T. Suzuki, D. Nishio-Hamane, R. Arita, Y. Otani, S. Nakatsuji, Nat. Phys. 2017, 13, 1085.
[10] T. Higo, H. Man, D. B. Gopman, L. Wu, T. Koretsune, O. M. van’t Erve, Y. P. Kabanov, D. Rees, Y. Li, M.-T. Suzuki, S. Patankar, M. Ikhlas, C. L. Chien, R. Arita, R. D. Shull, J. Orenstein, S. Nakatsuji, Nat. Photonics 2018, 12, 73.
[11] H. Chen, Q. Niu, A. H. MacDonald, Phys. Rev. Lett. 2014, 112, 017205.
[12] J. Küberl, C. Felser, EPL 2014, 108, 67001.
[13] M. Kimata, H. Chen, K. Kondou, S. Sugimoto, P. K. Muduli, M. Ikhlas, Y. Omori, T. Tomita, A. MacDonald, S. Nakatsuji, Y. Otani, Nature 2019, 565, 627.
[14] X. Li, C. Collignon, L. Xu, H. Zuo, A. Cavanna, U. Gennser, D. Mailly, B. Fauquè, L. Balents, Z. Zhu, K. Behnia, Nat. Commun. 2019, 10, 3021.
[15] Y. Takeuchi, Y. Yamane, J.-Y. Yoon, R. Itoh, B. Jinmai, S. Kanai, J. Ieda, S. Fukami, H. Ohno, Nat. Mater. 2021, 20, 1364.
[16] T. Jungwirth, X. Marti, P. Wadley, J. Wunderlich, Nat. Nanotechnol. 2016, 11, 231.
[17] V. Baltz, A. Manchon, M. Tsoi, T. Moriyama, T. Ono, Y. Tserkovnyak, Rev. Mod. Phys. 2018, 90, 015005.
[18] Y. Deng, X. Liu, Y. Chen, Z. Du, N. Jiang, C. Shen, E. Zhang, H. Zheng, H.-Z. Lu, K. Wang, Natl. Sci. Rev. 2022, nwac154.
[19] A. Markou, J. Taylor, A. Kalache, P. Werner, S. Parkin, C. Felser, Phys. Rev. Mater. 2018, 2, 051001.
[20] Y. Zhang, Y. Sun, H. Yang, J. Železný, S. P. P. Parkin, C. Felser, B. Yan, Phys. Rev. B 2017, 95, 075128.
[21] P. K. Muduli, T. Higo, T. Nishikawa, D. Qu, H. Isshiki, K. Kondou, D. Nishio-Hamane, S. Nakatsuji, Y. Otani, Phys. Rev. B 2019, 99, 184425.
[22] T. Yu, H. Wu, H. He, C. Guo, C. Fang, P. Zhang, K. L. Wong, S. Xu, X. Han, K. L. Wang, APL Mater. 2021, 9, 041111.
[23] J. Yoon, Y. Takeuchi, R. Itoh, S. Kanai, S. Fukami, H. Ohno, Appl. Phys. Express 2019, 13, 013001.
[24] A. Kumar, R. Bansal, S. Chaudhary, P. K. Muduli, Phys. Rev. B 2018, 98, 104403.
[25] P. Deoran, H. Yang, Appl. Phys. Lett. 2013, 103, 232408.
[26] S. Kurdi, P. Zilskes, X. Xu, M. Frentrup, M. Vickers, Y. Sakuraba, G. Reiss, Z. Barber, J. Koo, J. Appl. Phys. 2020, 127, 165302.
[27] T. Higo, D. Qu, Y. Li, C. Chien, Y. Otani, S. Nakatsuji, Appl. Phys. Lett. 2018, 113, 202402.
[28] T. Ikeda, M. Tsunoda, M. Oogane, S. Oh, T. Morita, Y. Ando, AIP Adv. 2020, 10, 015310.
[29] Z. Celinski, K. Urquhart, B. Heinrich, J. Magn. Magn. Mater. 1997, 166, 6.
[30] M. Farle, Rep. Prog. Phys. 1998, 61, 755.
[31] C. Kittel, Phys. Rev. 1948, 73, 155.
[32] O. Gladi, L. Frangou, A. Hallal, R. L. Seeger, P. Noël, G. Forestier, S. Augert, M. Rubio-Roy, P. Warin, L. Vila, S. Wimmer, H. Ebert, S. Auffret, M. Chishiev, V. Balz, Phys. Rev. B 2019, 100, 174409.
[33] A. Conca, B. Heinz, M. Schweizer, S. Keller, E. T. Papaioannou, B. Hillebrands, Phys. Rev. B 2017, 95, 174426.
[34] H. Bangar, A. Kumar, N. Chowdhury, R. Mudgal, P. Gupta, R. S. Yadav, S. Das, P. K. Muduli, ACS Appl. Mater. Interfaces 2022, 14, 41598.
[35] P. Deorani, J. Son, K. Banerjee, N. Koirala, M. Brahlek, S. Oh, H. Yang, Phys. Rev. B 2014, 90, 094403.
[36] J. Lustikova, Y. Shiomi, E. Saitoh, Phys. Rev. B 2015, 92, 224436.
[37] O. Mosendz, J. Pearson, F. Fradin, G. Bauer, S. Bader, A. Hoffmann, Phys. Rev. Lett. 2010, 104, 046601.
[38] B. B. Singh, K. Roy, J. A. Chelvane, S. Bedanta, Phys. Rev. B 2020, 102, 174444.
[39] W. Zhang, W. Han, S.-H. Yang, Y. Sun, Y. Zhang, B. Yan, S. S. Parkin, Sci. Adv. 2016, 2, e1600759.
[40] K. Kuroda, T. Tomita, M.-T. Suzuki, C. Bareille, A. Nugroho, P. Goswami, M. Ochi, M. Ikhlas, M. Nakayama, S. Akebi, R. Noguchi, R. Ishii, N. Inami, K. Ono, H. Kumigashira, A. Varykhalov, T. Muro, T. Koretsune, R. Arita, S. Shin, T. Kondo, S. Nakatsuji, Nat. Mater. 2017, 16, 1090.
[41] B. Cheng, Y. Wang, D. Barbalas, T. Higo, S. Nakatsuji, N. Armitage, Appl. Phys. Lett. 2019, 115, 012405.
[42] A. Kumar, N. Pandey, D. Kumar, M. Gupta, S. Chaudhary, P. K. Muduli, Phys. B 2019, 570, 254.