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ABSTRACT

Epitaxial thin films of fully nonequilibrium hcp-Ru₅₀Mo₅₀(0001) nanoalloys were prepared as a chemically disordered alloy, in which the intrinsic spin Hall effect is expected to be negligible. Structural analyses confirmed the epitaxial growth and atomic scale alloying of the films. In contrast to a tiny torque efficiency (ξ_{DL}) of $\sim 0.4\%$ for Ru₅₀Mo₅₀/CoFeB, the ξ_{DL} for the Ru₅₀Mo₅₀/Ni heterostructure reached $\sim 30\%$ with a long-range relaxation length. The apparent dependence of ξ_{DL} on the ferromagnetic layer can be attributed to the orbital Hall effect (OHE). Interestingly, a smaller ξ_{DL} was observed for Ru/Ni, suggesting that the nonequilibrium Ru₅₀Mo₅₀ enhances its OHE. Furthermore, the enhanced ξ_{DL} is maintained by inserting a Ru layer between the Ru₅₀Mo₅₀ and Ni layers, showing orbital transport through Ru. This finding illustrates potential applications of nonequilibrium nanoalloy films in spin orbitronics and contributes to getting insights into the understanding of the interrelationships between nanostructures and orbital transport properties.

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I. INTRODUCTION

Spin and orbital torques^{1–5} have become a promising technology to realize fast and energy efficient magnetization switching for spintronic devices, such as spin-orbit torque (SOT) magnetoresistive random access memories (MRAMs).^{6–8} When a charge current is applied in the longitudinal direction of a nonmagnetic (NM) thin film, the spin and orbital currents can be generated flowing along the transverse direction due to the spin Hall effect (SHE)⁹ and orbital Hall effect (OHE),^{10–13} respectively. By direct transfer of spin angular momentum carried by the spin current into the moment of an adjacent ferromagnetic (FM) layer, the magnetization manipulation and switching can be achieved. Meanwhile, the orbital current first must be transferred to a spin current in the FM layer and then contributes to the spin torque, acting on the moment of the FM layer, as shown in Fig. 1(a). These interplays of spin/orbital

angular momenta with magnetic moments showcase the potential for advancements in SOT-MRAMs and related technologies.

The large SHE, usually found in 5d heavy metals,^{2,14,15} their doped alloys,^{16,17} as well as topological materials,^{18–20} requires strong spin-orbit interactions (SOIs) and originates from both an intrinsic mechanism, such as the spin Berry curvature from electronic band structures, and an extrinsic mechanism, including skew scattering and the side-jump mechanism.^{9,21} However, large OHEs with a positive sign were predicted even in the materials with small SOIs, such as CuO_x,^{1,5} Si,¹⁰ Ti,^{4,13} Ru,²² Nb,²² and other light metals.²³ The origin of OHE is mainly lying on the intrinsic mechanism, i.e., the orbital texture related to the electronic structures of the materials.²⁴ The spin and orbital torques show the same symmetry and, therefore, are difficult to be distinguished from each other. Recently, it has been reported that these two torques can be separated by comparing the torques in NM/FM systems using FMs with

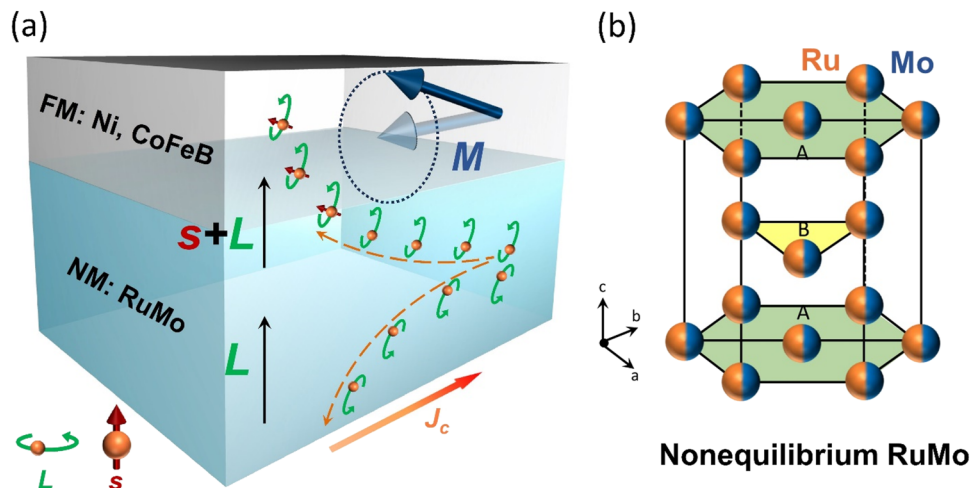


FIG. 1. (a) Illustration of the orbital Hall effect in a normal metal layer and interaction with the magnetic moment of the adjacent FM layer. (b) Schematic of a crystal lattice structure of nonequilibrium Ru₅₀Mo₅₀ (RuMo).

different SOIs, such as Ni and Fe.²⁵ The SOIs between the electronic band structures of Ni and Fe are different so that the orbital-to-spin conversions in these two FMs show different strengths.²⁵ To date, orbital Hall torques (OHTs) with a corresponding damping-like torque efficiency (ξ_{DL}) were experimentally reported in Ti/Ni, Ru/Ni, and Nb/Ni heterostructures.^{4,13,22} However, the material dependence of OHE and OHT has not been well understood. Particularly, the previous studies mostly focused on monoatomic metals and were rarely done for disordered alloys, in which the extrinsic mechanisms of SHE, i.e., skew scattering and side-jump, seem to be missing in OHE.

In this work, we fabricated Ru₅₀Mo₅₀(0001) epitaxial thin films, as a chemically disordered alloy, on Al₂O₃(0001) substrates. The SHE in the Ru₅₀Mo₅₀ alloy was expected to vanish,^{26,27} which would be a good platform for the study of OHE and OHT. The torque efficiencies were investigated based on Ru₅₀Mo₅₀/FM heterostructures, where the FM is CoFeB or Ni, as shown in Fig. 1(a). The structural characteristics demonstrated that the Ru₅₀Mo₅₀ alloy was formed in a chemically disordered state of fully nonequilibrium, as illustrated in Fig. 1(b). An enhanced orbital torque efficiency was achieved in Ru₅₀Mo₅₀ films compared to that in pure Ru films.

II. EXPERIMENTAL METHODS

All the thin films were prepared using a high vacuum magnetron co-sputtering technique. The processes were performed under a base pressure of 6×10^{-7} Pa and a deposition pressure of 0.19 Pa in argon gas. The films were grown on sapphire c -plane substrates, i.e., single-crystalline Al₂O₃(0001), at the substrate temperature of 250–300 °C and with *in situ* post-annealing at 750–850 °C for 1 h. The substrates were pre-baked *ex situ* at 1000 °C for an hour in a muffle furnace (AS ONE HPN-ON) for thermal cleaning. The deposition utilized individual Ru and Mo targets, with the sputtering power for each set between 20 and 80 W, which was determined by the desired Ru-to-Mo ratio in the films. The total deposition

rate was about 1.1 Å/s. To characterize the films, a combination of techniques was applied: *in situ* reflection high-energy electron diffraction (RHEED), *ex situ* atomic force microscopy (AFM), and out-of-plane x-ray diffraction (XRD) using Cu K α radiation of wavelength 0.154 nm. Further detailed microstructural characterization was performed using high-resolution high-angle annular dark-field scanning transmission electron microscopy (HAADF-STEM) and energy dispersive x-ray spectroscopy (EDS) on an FEI Titan G2 80–200 ChemiSTEM system. During microfabrication, the films were patterned rectangular-shaped devices (width: 10 μ m, length: 40 μ m) for spin-torque ferromagnetic resonance (ST-FMR) measurements. Electrodes made of a Ta (5 nm)/Au (100 nm) layer were then applied to the structures using sputtering followed by a lift-off process. The ST-FMR measurements were conducted at room temperature, applying an RF power of 15 dBm with a frequency range from 6 to 18 GHz using a signal generator (Keysight E8257D). In addition, an in-plane external magnetic field was swept with magnitude 0–3000 Oe and $\phi = 0$ –360°.

III. RESULTS AND DISCUSSION

A. Structural characterizations

Under optimized deposition conditions, 10-nm-thick single layer Ru and Ru₅₀Mo₅₀ films were grown on sapphire c -plane substrates, as illustrated in Fig. 2(a). Specifically, to achieve the epitaxial growth and the good crystallinity in the Ru₅₀Mo₅₀ film, an ultra-thin Ru of 0.7 nm was utilized as a buffer layer, which is critical to tune the increased lattice mismatch of Ru₅₀Mo₅₀ with the sapphire substrate. The surface structures were then characterized by *in situ* RHEED. Figures 2(b) and 2(c) show the RHEED patterns of the surfaces of the Ru and Ru₅₀Mo₅₀ films, which were taken when the electron beam incident direction was parallel to the Al₂O₃ [10 $\bar{1}$ 0] azimuth. The RHEED patterns show sharp and narrow streaks clearly, indicating the epitaxial growth of both films. AFM was employed to elucidate

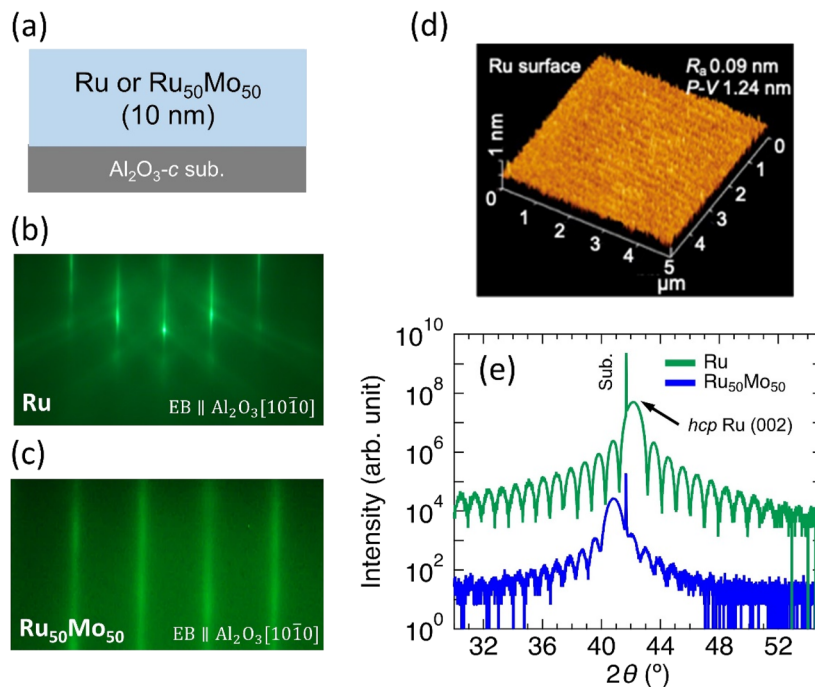


FIG. 2. Structural properties of the 10-nm-thick Ru and $\text{Ru}_{50}\text{Mo}_{50}$ films. (a) Film stacks. (b) and (c) RHEED patterns of the Ru film and the $\text{Ru}_{50}\text{Mo}_{50}$ film, respectively. (d) Typical AFM image of the surface of the Ru film. (e) Out-of-plane XRD patterns for the Ru and $\text{Ru}_{50}\text{Mo}_{50}$ thin films.

the surface morphology of the $\text{Ru}_{50}\text{Mo}_{50}$ and Ru films, with a representative Ru surface image shown in Fig. 2(d). The films exhibited an impressive smooth surface, evidenced by an average roughness of less than 0.1 nm and a peak-to-valley value of ~ 1.24 nm. The presence of terraces in the morphology image further underscored the flat, well-textured nature of the surface. The crystalline structure of the films was further confirmed by the out-of-plane XRD patterns shown in Fig. 2(e). The XRD patterns for the Ru film exhibit a singular, well-defined peak at $2\theta = 42.16^\circ$, corresponding to the hcp Ru(002) lattice plane, affirming the film's single-crystalline nature. In contrast, the XRD pattern for the $\text{Ru}_{50}\text{Mo}_{50}$ film reveals a leftward shift of the peak of Ru(002) to 40.81° , which implies the lattice expansion consequent to Mo doping. The quality of the film growth was further indicated by Laue oscillations, which were observable in the XRD patterns for both types of films. These oscillations are typically the sign of uniform film thickness and the smooth interface between the film and the substrate.

To elucidate the structural features of the fabricated films, HAADF-STEM was utilized for a detailed examination of a 10-nm-thick $\text{Ru}_{50}\text{Mo}_{50}$ film, which was topped with a MgO capping layer. This analysis was conducted along the $[11\bar{2}0]$ crystallographic direction of the Al_2O_3 substrate. The comprehensive TEM imagery, which is shown in Fig. 3(a), delineates the $\text{Ru}_{50}\text{Mo}_{50}$ film layer between the Al_2O_3 substrate and the MgO cap, revealing interfaces that are consistently flat and sharply defined. Zooming into the atomic scale, the HAADF-STEM image presented in Fig. 3(b) reveals the microstructure of the $\text{Ru}_{50}\text{Mo}_{50}$ film. This image clearly shows an ABAB stacking sequence, indicating the hcp structure, agreeing

well with the XRD measurements. To complement the composition analysis, EDS mapping and depth profiling were conducted, with the results displayed in Figs. 3(c)–3(g). These EDS maps and profiles confirm the stoichiometric balance of the Ru and Mo elements, maintaining an approximate 1:1 ratio. Importantly, Ru and Mo elements show a homogeneous distribution. These results also confirm the absence of elemental interdiffusion into the substrate and the cap. The structural features shown above indicate that the miscible nonequilibrium $\text{Ru}_{50}\text{Mo}_{50}(0001)$ alloy epitaxial thin film was achieved.

B. Spin-torque ferromagnetic resonance measurements

The ST-FMR technique¹⁵ was utilized to investigate the SOTs within $\text{Ru}_{50}\text{Mo}_{50}/\text{CoFeB}$ layered structures. An illustration of the ST-FMR experimental configuration is depicted in Fig. 4(a). The film layers were etched into devices with a rectangular shape, incorporating electrodes designed as a coplanar waveguide. The magnetization of the CoFeB layer, denoted by $\mathbf{M}$, is controlled by an in-plane magnetic field $\mathbf{H}$, which is applied at an azimuthal angle ϕ . A radio frequency current, I_{rf} , at frequency f , flows longitudinally, exerting an oscillatory spin torque that induces a precession in $\mathbf{M}$. This precession leads to a fluctuation in the resistance of the device due to the anisotropic magnetoresistance of CoFeB, producing a detectable mixing ST-FMR voltage V that is captured by a lock-in amplifier. Figure 4(b) shows the FMR spectra for the $\text{Ru}_{50}\text{Mo}_{50}/\text{CoFeB}$ structure measured at $\phi = 45^\circ$ and f ranging from 6 to 18 GHz. As

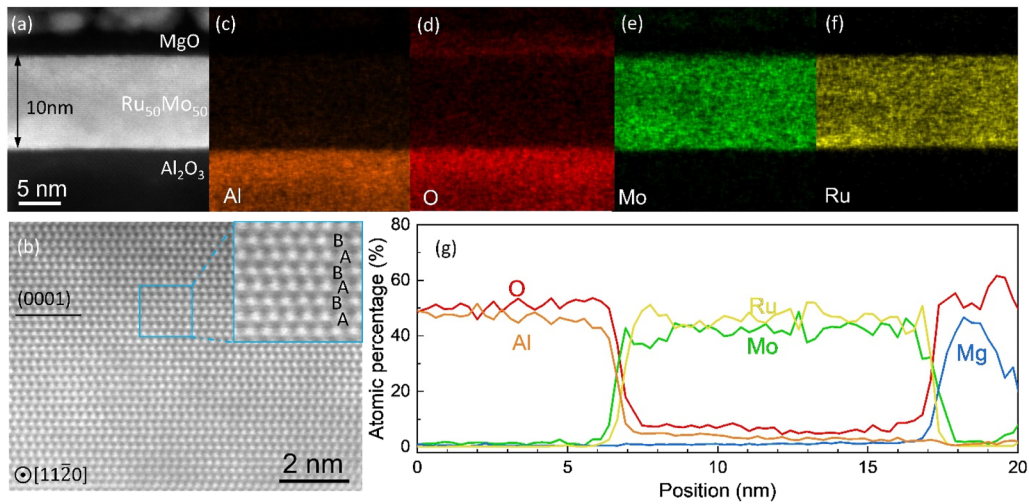


FIG. 3. (a) Cross-sectional HAADF-STEM images for the Al₂O₃ substrate//Ru₅₀Mo₅₀ (10 nm)/MgO cap. (b) Magnified HAADF-STEM image in the Ru₅₀Mo₅₀ layer. (c)–(f) EDS elemental maps for Al, O, Mo, and Ru, respectively. (g) Elemental line profile from the Al₂O₃ substrate to the MgO cap.

the f increases, the resonance field H_{res} correspondingly intensifies. Figure 4(c) illustrates a detailed and representative fit result of the FMR spectrum at $\phi = 45^\circ$ and $f = 14$ GHz, where the symmetric (V_S) and antisymmetric (V_A) contributions of the Lorentzian line shapes from V are separated by²⁸

$$V = V_S \frac{\Delta^2}{(H - H_{\text{res}})^2 + \Delta^2} + V_A \frac{\Delta(H - H_{\text{res}})}{(H - H_{\text{res}})^2 + \Delta^2}, \quad (1)$$

where Δ is the resonance linewidth. The damping-like SOT is identified as the source of V_S , while V_A is resulted from the combined influence of the Oersted field and field-like torque. By examining the ratios of V_S and V_A , the FMR efficiency ξ_{FMR} is determined. Figures 4(b) and 4(d) show the FMR spectra at varying f from 6 to 18 GHz and the corresponding H_{res} values derived from these spectra. The f dependency of H_{res} in Fig. 4(d) is fitted with the Kittel formula $f = (\gamma/2\pi) [(H_{\text{res}} + H_{\text{ani}})(H_{\text{res}} + 4\pi M_{\text{eff}})]^{1/2}$, where γ is the gyromagnetic ratio, H_{ani} is the in-plane anisotropy field, and $4\pi M_{\text{eff}}$ is the effective demagnetizing field. The azimuthal angle ϕ dependence of the ST-FMR signal was investigated with the f held as a constant at 12 GHz. The variations in the V_S and V_A components with respect to ϕ are depicted in Fig. 4(e). The relationships between V_S and V_A with ϕ are fitted by the functions $V_S = k_S \sin 2\phi \cos \phi$ for V_S and $V_A = k_A \sin 2\phi \cos \phi$ for V_A , where k_S and k_A are the scaling factors of V_S and V_A , respectively. The torque efficiency ξ_{FMR} is then given by²⁸

$$\xi_{\text{FMR}} = \frac{V_S}{V_A} \frac{e\mu_0 M_s t_{\text{FM}} t_{\text{NM}}}{\hbar} \sqrt{\frac{H_{\text{res}} + 4\pi M_{\text{eff}}}{H_{\text{res}} + H_{\text{ani}}}}, \quad (2)$$

where μ_0 is the vacuum permeability, M_s is the saturation magnetization of the CoFeB layer, and t_{FM} and t_{NM} are the thicknesses of the FM (CoFeB) and NM (Ru₅₀Mo₅₀) layers, respectively. Finally, by examining a series of samples varying in CoFeB thickness t_{CoFeB} , the damping-like torque efficiency ξ_{DL} and the field-like torque

efficiency ξ_{FL} can be extrapolated from a linear fit of $1/\xi_{\text{FMR}}$ as a function of $1/t_{\text{CoFeB}}$,²⁹

$$\frac{1}{\xi_{\text{FMR}}} = \frac{1}{\xi_{\text{DL}}} \left(1 + \frac{\hbar}{e\mu_0 M_s t_{\text{FM}} t_{\text{NM}}} \xi_{\text{FL}} \right). \quad (3)$$

$\xi_{\text{DL}} = 0.4\%$ and $\xi_{\text{FL}} = -13.7\%$ were obtained in the Ru₅₀Mo₅₀/CoFeB bilayer structures. The ξ_{DL} was evaluated to be as low as 0.4%, which is as expected due to the tiny spin-orbit interactions in pure Ru and Mo.²⁶ The ξ_{DL} of the Ru₅₀Mo₅₀/CoFeB is comparable to the reported values in Ru/CoFeB²² and Ru/CoFeAl,³⁰ where the spin-orbit effect was responsible for spin torques. The ξ_{FL} is observed to have a substantial value and an inverse sign in relation to the ξ_{DL} , which could be due to the changes in spin-orbital coupling by inversion symmetry breaking at the Ru₅₀Mo₅₀/CoFeB interface³¹ and the complex spin-dependent scattering that occurs at the material interface.^{32,33}

The angular momentum transfer induced ferromagnetic resonance dynamics within Ru/Ni, Ru₅₀Mo₅₀/Ni and Ru₅₀Mo₅₀/Ru/Ni heterostructures have been further explored. Figures 5(a)–5(c) illustrate the examined stack structures. The typical FMR spectra for each heterostructure are presented in Figs. 5(d)–5(f). The extracted V_S component, which originates from the damping-like torque, in the Ru/Ni sample is smaller in magnitude than those in the Ru₅₀Mo₅₀/Ni and Ru₅₀Mo₅₀/Ru/Ni samples. Figures 5(g)–5(i) show the linear fits of $1/\xi_{\text{FMR}}$ against $1/t_{\text{Ni}}$ for the three series of samples, respectively. For the Ru/Ni structure, the ξ_{DL} is determined to be 1.6%, aligning closely with the previously reported value 1.9%,²² which is attributed to the contribution from the OHE. Notably, the ξ_{DL} in the Ru₅₀Mo₅₀/Ni structure is evaluated to be 5.3%, indicating a significant enhancement compared to the ξ_{DL} in Ru₅₀Mo₅₀/CoFeB. This could be understood through the framework of OHT: Orbital currents induced in the NM layer via the OHE are driven into the adjacent FM layer, which is an analog of the generation of spin currents seen in spin torque phenomena. However, the orbital angular

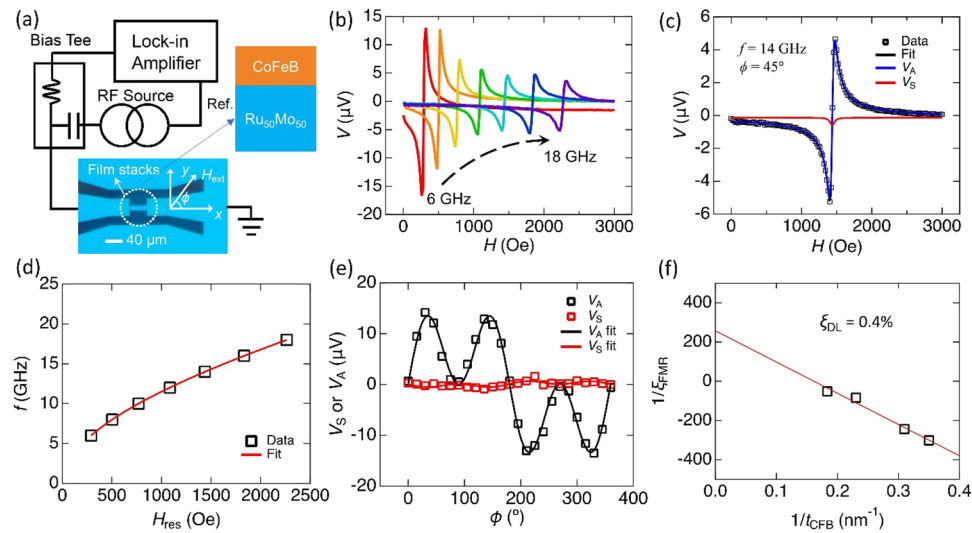


FIG. 4. ST-FMR measurements for $\text{Ru}_{50}\text{Mo}_{50}/\text{CoFeB}$ heterostructures at room temperature. (a) Stack structure, device, and measurement configuration. (b) ST-FMR spectra measured under various frequencies f ranging from 6 to 18 GHz. (c) A typical ST-FMR spectrum of $\text{Ru}_{50}\text{Mo}_{50}/\text{CoFeB}$ with extracted symmetric (V_S) and antisymmetric (V_A) contributions of the Lorentzian line shapes. (d) f dependence of the resonance field H_{res} fitted with the Kittel equation. (e) The in-plane angle ϕ dependence of V_S and V_A . (f) $1/\xi_{\text{FMR}}$ as a function of $1/t_{\text{CoFeB}}$ with a linear fit.

momentum carried by orbital currents does not directly interact with the magnetic moment of the FM. Rather, the orbital currents are converted into spin currents within the FM, which then exert torques on the magnetic moment. This process intimately connects

to the spin-orbit coupling of the FM. Here, Ni is known to be a more efficient medium for the orbital to spin current conversion compared to CoFeB.²² Therefore, the marked deviation in ξ_{DL} between the two systems, $\text{Ru}_{50}\text{Mo}_{50}/\text{CoFeB}$ and $\text{Ru}_{50}\text{Mo}_{50}/\text{Ni}$,

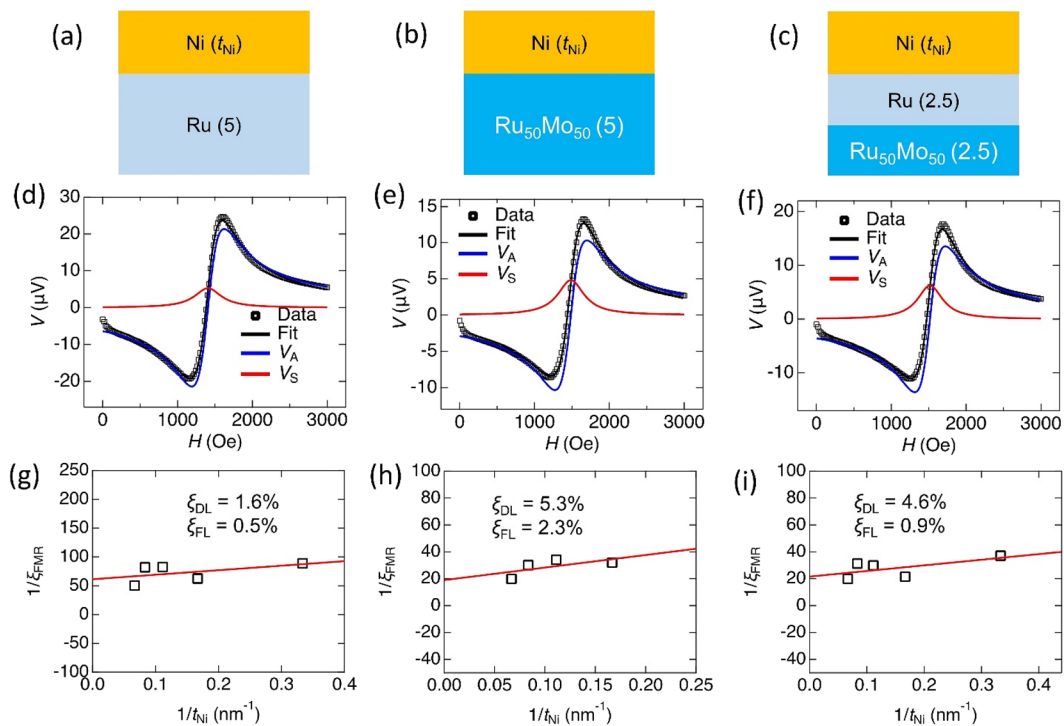


FIG. 5. Illustration of film stacks: (a) Ru/Ni , (b) $\text{Ru}_{50}\text{Mo}_{50}/\text{Ni}$, and (c) $\text{Ru}_{50}\text{Mo}_{50}/\text{Ru}/\text{Ni}$. Representative ST-FMR spectra ($\phi = 45^\circ$ and $f = 12$ GHz) for (d) $\text{Ru}(5 \text{ nm})/\text{Ni}(12 \text{ nm})$, (e) $\text{Ru}_{50}\text{Mo}_{50}(5 \text{ nm})/\text{Ni}(12 \text{ nm})$, and (f) $\text{Ru}_{50}\text{Mo}_{50}(5 \text{ nm})/\text{Ru}(12 \text{ nm})$. $1/\xi_{\text{FMR}}$ against $1/t_{\text{Ni}}$ with linear fits for (g) Ru/Ni , (h) $\text{Ru}_{50}\text{Mo}_{50}/\text{Ni}$, and (i) $\text{Ru}_{50}\text{Mo}_{50}/\text{Ru}/\text{Ni}$.

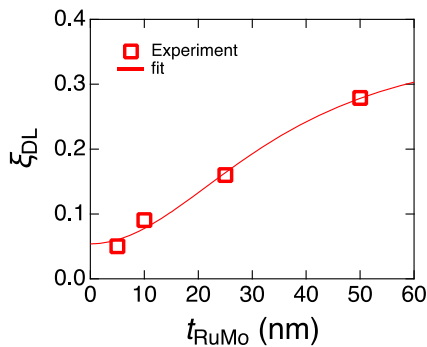


FIG. 6. ξ_{DL} as a function of t_{RuMo} measured in $\text{Ru}_{50}\text{Mo}_{50}/\text{Ni}$ heterostructures. The data are fitted by Equation (4).

serves as evidence for the presence of OHT. Furthermore, the ξ_{DL} in a $\text{Ru}_{50}\text{Mo}_{50}/\text{Ru}/\text{Ni}$ trilayer configuration was investigated, which is depicted in Fig. 5(i). The ξ_{DL} in this trilayer structure was found to be 4.6%, still exhibiting a significant improvement over the Ru/Ni system, although slightly lower than that observed in the $\text{Ru}_{50}\text{Mo}_{50}/\text{Ni}$ bilayer. The maintained ξ_{DL} indicates that the Ru could be rather transparent for orbital angular momenta, allowing for the efficient transfer of orbital currents. In addition, the ξ_{FL} was observed to be relatively small in the three samples, which may be due to the transparency of orbit current transport in the heterostructures.

To gain a deeper understanding of orbital transport phenomena in $\text{Ru}_{50}\text{Mo}_{50}$ alloys, the $\text{Ru}_{50}\text{Mo}_{50}$ thickness (t_{RuMo}) dependence of ξ_{DL} within $\text{Ru}_{50}\text{Mo}_{50}/\text{Ni}$ heterostructures was investigated, where t_{RuMo} ranges from 5 to 50 nm. Figure 6 shows ξ_{DL} as a function of t_{RuMo} . An increase in ξ_{DL} is observed with increasing t_{RuMo} , aligning with tendencies previously reported in Ti/Ni ^{4,13} and W/Ni ¹⁵ systems. The orbital torque efficiency can be extracted by the following equation:^{4,22}

$$\xi_{\text{DL}} = \xi_{\text{OH}} \left(1 - \text{sech} \frac{t_{\text{RuMo}}}{\lambda_{\text{L}}} \right) + \xi_0, \quad (4)$$

where ξ_{OH} is the effective orbital Hall efficiency, λ_{L} is the orbital relaxation length, and ξ_0 is the t_{RuMo} -independent component of ξ_{DL} . From the fitting, we obtain $\xi_{\text{OH}} = 30.0\% \pm 6.0\%$, $\lambda_{\text{L}} = 24.3 \pm 6.8$ nm, and $\xi_0 = 5.4\% \pm 1.5\%$.

It is noted that the OHE was typically studied in materials consisting of a single element.^{4,13,22,23,27} Here, a large OHT is observed in the chemically disordered $\text{Ru}_{50}\text{Mo}_{50}$. One of the contributions to the enhancement of the OHT could be the intrinsic origin of OHE. The modification of the electronic band structures in $\text{Ru}_{50}\text{Mo}_{50}$ increases the orbital hybridization of the tangential component and the radial component of the orbits, resulting in a large imbalance of the orbital angular momentum accumulation.²⁴ Another contribution may be related to the extrinsic mechanisms. Theoretically, the extrinsic OHE, including skew scattering and side-jump, was only considered in hole-doped silicon by Bernevig *et al.*¹⁰ Recently, a theoretical prediction based on quantum dynamical formulations has suggested that in the presence of short-range disorder of the system, extrinsic effects associated with the Fermi surface can provide a large portion of the OHE.³⁴ In the $\text{Ru}_{50}\text{Mo}_{50}$ sample, the orbital relaxation length λ_{L} is comparable to those in titanium and tungsten,^{4,13}

which may indicate the predominance of intrinsic contributions. In binary compounds, a stronger OHE might be realized than in single-element materials because the orbitals of distinct characters from different atomic elements share similar energies, leading to a strong orbital hybridization. The introduction of additional elements through doping can bring more scattering centers, and the extrinsic scattering processes may also contribute to the OHE in the chemically disordered compounds. The discovery of a relatively large orbit torque efficiency in the disordered $\text{Ru}_{50}\text{Mo}_{50}$ alloy thin films suggests the possibility of broadening the range of highly efficient orbital Hall materials, as well as contributes to further refining the theoretical origins of OHE.

IV. CONCLUSIONS

We fabricated epitaxial thin films of fully nonequilibrium hcp- $\text{Ru}_{50}\text{Mo}_{50}(0001)$ nanoalloy thin films. Comprehensive structural analyses confirmed both the epitaxial growth and atomic-scale alloying of the chemically disordered films. The $\text{Ru}_{50}\text{Mo}_{50}/\text{Ni}$ heterostructure shows a significant ξ_{DL} of $\sim 30\%$ compared to the tiny ξ_{DL} of 0.4% observed in the $\text{Ru}_{50}\text{Mo}_{50}/\text{CoFeB}$ bilayers. The dependence of ξ_{DL} on the FM layer was notably attributed to the OHE. Intriguingly, the Ru/Ni sample exhibited a smaller ξ_{DL} , indicating an enhancement of OHE in the nonequilibrium $\text{Ru}_{50}\text{Mo}_{50}$ alloy thin films. Moreover, maintaining the large ξ_{DL} by inserting a Ru layer between $\text{Ru}_{50}\text{Mo}_{50}$ and Ni layers indicated orbital transport through the Ru. The results show the potential of nonequilibrium alloy thin films for a large OHE and provide inspiration for the correlation between disordered nanostructures and orbital transport.

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AUTHOR DECLARATIONS

Conflict of Interest

The authors have no conflicts to disclose.

Author Contributions

Ke Tang: Data curation (equal); Formal analysis (equal); Investigation (equal); Methodology (equal); Writing – original draft (equal); Writing – review & editing (equal). **Cong He:** Investigation (equal). **Zhenchao Wen:** Conceptualization (equal); Funding acquisition (equal); Investigation (supporting); Methodology (equal); Supervision (equal); Writing – original draft (equal); Writing – review & editing (equal). **Hiroaki Sukegawa:** Conceptualization (equal); Funding acquisition (equal); Project administration (equal); Supervision (equal); Writing – review & editing (supporting). **Tadakatsu Ohkubo:** Resources (equal); Writing – review & editing (supporting). **Yukio Nozaki:** Conceptualization (equal);

Funding acquisition (equal); Project administration (equal); Writing – review & editing (supporting). **Seiji Mitani**: Conceptualization (equal); Funding acquisition (equal); Project administration (equal); Resources (equal); Supervision (equal); Writing – review & editing (supporting).

DATA AVAILABILITY

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

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