

Disorder-enhanced spin Hall effect in Ru-Pt alloy thin films

Vineetha S. Bheemarasetty ^{1,*}, Xinhao Wang ^{2,‡}, Shreya Shrestha ³, M. Benjamin Jungfleisch ³, Lars Gundlach ^{2,4},
John Q. Xiao ² and Gang Xiao ^{1,†}

¹Department of Physics, *Brown University*, Providence, Rhode Island 02912, USA

²Department of Physics and Astronomy, *University of Delaware*, Newark, Delaware 19716, USA

³Materials Research Science and Engineering Center, *University of Delaware*,
Newark, Delaware 19716, USA

⁴Department of Chemistry and Biochemistry, *University of Delaware*, Newark, Delaware 19716, USA



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The search for material systems exhibiting a large spin Hall effect is central to the development of ultrafast, energy-efficient spintronic technologies. We systematically investigate the spin Hall angle (SHA) in Ru-Pt binary alloys using three complementary techniques: spin-torque magneto-optic Kerr effect, spin-torque ferromagnetic resonance, and time-domain terahertz emission spectroscopy. Within the measured composition range, a maximum SHA of 0.158 is observed in Ru₃₈Pt₆₂, representing a 56% enhancement over Pt while maintaining a resistivity below 70 $\mu\Omega\text{cm}$. Notably, these high SHA values are achieved in as-grown sputtered films across the full composition range without the need for complex fabrication procedures or high-temperature annealing. These results establish Ru-Pt as a scalable model alloy system in which composition-driven disorder, together with gradual structural evolution, governs spin-charge conversion, providing a practical pathway for optimizing materials for spin-orbit torque devices.

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I. MOTIVATION AND INTRODUCTION

Recent perspectives on spin-orbit torques (SOTs) emphasize that further progress toward low-power scalable spintronic devices critically relies on identifying material systems that combine high charge-to-spin conversion efficiency (η) with low resistivity and good integrability [1]. Identifying and developing materials with large η is a central challenge in spintronics research. The efficiency, defined as the ratio of spin-current density j_s to charge-current density j_c : $\eta = j_s/j_c$ [2], is commonly expressed as the spin Hall angle (SHA). A large SHA enables efficient current-driven SOTs, which are essential for low-power magnetic switching, fast operation, and reliable performance in next-generation memory and logic technologies.

While materials such as β -W [3] and Ta [4] have demonstrated large SHAs up to 0.4, their electrical

resistivities (exceeding several 100 $\mu\Omega\text{cm}$) pose critical challenges for device integration. High resistivity not only leads to substantial Joule heating and reduced thermal stability but also limits the achievable current density due to current shunting in bilayer systems. Such limitations stem from the intrinsic nature of the spin Hall effect (SHE), where large SHAs are often correlated with high resistivities due to strong Berry curvature effects in the electronic band structure [5].

In parallel, extrinsic mechanisms of the SHE arising from spin-dependent impurity scattering provide another pathway to tune and enhance the SHA [6]. Two dominant processes have been identified: skew scattering [7] and side-jump scattering [8]. Alloy systems such as CuIr, CuBi, and AuTa have demonstrated that controlled additions of impurities can produce large SHAs while maintaining moderate resistivities [5,9,10]. These examples highlight that alloy engineering offers a promising strategy for optimizing the balance between spin-charge conversion efficiency and conductivity.

From a materials science perspective, the design of effective spin Hall materials requires consideration of not only the spin-orbit coupling (SOC) strength but also structural stability, phase formation, and ease of thin-film growth. SOC scales strongly with atomic number (Z^4), favoring heavier elements such as Pt, a 5d transition metal known for its large intrinsic SHA. Ru, in contrast, is a 4d

*Contact author: vineetha_bheemarasetty@brown.edu

†Contact author: gang_xiao@brown.edu

‡These authors contributed equally to this work.

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nonmagnetic transition metal widely employed in spintronic heterostructures for interlayer exchange coupling, which crystallizes in an hcp phase. When combined, the Ru-Pt binary system offers a unique alloy platform: Pt provides strong SOC and an fcc crystal structure, while Ru contributes structural compatibility, reduced resistivity, and stability across a wide composition range. Of note, while the crystal structure provides an essential electronic context, impurity-induced disorder and scattering are expected to play the dominant role in governing transport and spin Hall responses in such substitutional alloys. The coexistence of fcc and hcp phases in this alloy system allows investigation of how gradual structural evolution modulates spin-charge conversion in the presence of strong alloy disorder.

Several recent experimental studies have examined spin Hall and spin-orbit torque phenomena in Ru-Pt-based systems. Ju *et al.* [11,12] reported enhanced effective spin Hall efficiencies in Pt₈₀Ru₂₀ alloy layers integrated into SOT device stacks, demonstrating dampinglike torque efficiencies up to 0.156 with reduced switching currents. Zhang *et al.* [13] similarly explored compositional tuning in Ru-Pt alloys to optimize SOT signals. In engineered Pt/Ru multilayers, spin-current generation efficiencies as large as 0.28–0.31 have been achieved through multi-interface design and enhanced spin transparency [14]. These studies highlight the promise of Ru incorporation into Pt-based materials for improving spin-charge conversion, but have not provided a comprehensive, composition-resolved mapping of intrinsic SHA trends within homogeneous alloys.

Taken together, these factors motivate a systematic exploration of the Ru-Pt alloy system as a materials platform for spin Hall physics and SOT applications. By bridging electronic mechanisms with disorder-mediated transport and structure-property relationships, Ru-Pt alloys provide an opportunity to advance the search for efficient, low-resistivity spin-orbit materials.

II. SAMPLE FABRICATION

Ru-Pt films were deposited on thermally oxidized silicon wafers using high-vacuum dc magnetron sputtering at 30 W with an argon pressure of 4 mTorr. The base pressure of the sputtering system was 3×10^{-8} Torr. Various Ru_xPt_{100-x} alloys with compositions ranging from $x = 0$ to $x = 100$ were fabricated using an alternative co-sputtering technique.

In conventional co-sputtering, two or more magnetron targets are operated simultaneously, depositing materials at the same time onto a common substrate to form an alloy or compound film. The final composition is therefore highly sensitive to the dc power of each source, the deposition rate of each constituent material, as well as the sputtering pressure of the process gas. As an alternative, we designed an array of binary targets composed of a single Ru base

target with Pt pieces of various shapes and sizes strategically placed on top (see Supplemental Material, Sec. I [15]). Using this approach, we can simultaneously rotate the substrate while depositing both Ru and Pt atoms uniformly from the same source. Notably, this provides a cost-effective solution for fabricating uniform Ru-Pt films with varying compositions using only two elemental sputtering targets.

The nominal alloy stoichiometry was confirmed on bulk, 70-nm-thick films by energy-dispersive spectroscopy. The film thickness, determined from x-ray reflectivity, indicates a monotonically decreasing deposition rate with increasing Ru composition, from 1.214 Å/s at $x = 38$ to 0.689 Å/s at $x = 100$.

III. STRUCTURAL CHARACTERIZATION

X-ray diffraction (XRD) was used to characterize the crystalline structure of the as-grown Ru_xPt_{100-x} thin films. As expected, pure Pt ($x = 0$) exhibits an fcc structure with a dominant (111) diffraction peak identified using The International Centre for Diffraction Data's Powder Diffraction File database [Fig. 1(a)]. In contrast, pure Ru ($x = 100$) adopts an hcp structure with a pronounced (002) peak. Across the alloy series, the crystal structure evolves smoothly from fcc to hcp with increasing Ru concentration.

This gradual structural transition is reflected in Fig. 1(b), where the diffraction peak systematically shifts from the fcc (111) position toward the hcp (002) position with increasing x . The continuous peak shift indicates a monotonic decrease in the lattice constant as Pt is progressively substituted by the smaller Ru atoms. Well-defined diffraction peaks across all compositions demonstrate that the films remain highly crystalline despite being deposited by magnetron sputtering without postdeposition annealing. It is further shown that the films are thermally stable, as their crystal structure is robust to the high-temperature annealing required for industry applications (see Supplemental Material, Sec. II [15]).

Lattice constants were extracted from the XRD peak positions in Fig. 1(b) using Bragg's law: $n\lambda = 2d_{hkl} \sin \theta$. Here, n is the diffraction order taken to be 1, λ is the characteristic x-ray wavelength, and d_{hkl} is the lattice spacing. For the fcc structure, the cubic lattice constant was obtained by $a_{\text{fcc}} = d_{hkl} \sqrt{h^2 + k^2 + l^2}$, while for the hcp structure the interplanar spacing was related to the lattice constant by

$$\frac{1}{d_{hkl}^2} = \frac{4(h^2 + hk + k^2)}{3a_{\text{hcp}}^2} + \frac{l^2}{c_{\text{hcp}}^2}, \quad (1)$$

where a_{hcp} and c_{hcp} are the two unique hexagonal lattice parameters. The lattice plane spacing between the (111)

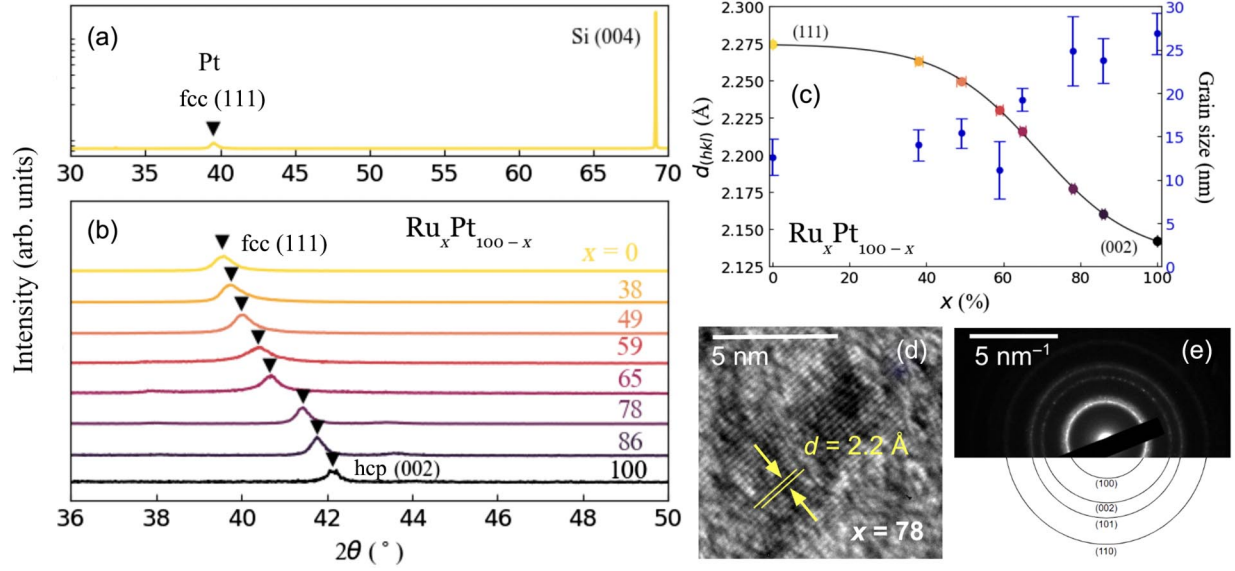


FIG. 1. X-ray characterization of Ru-Pt alloys. (a) Full x-ray $\theta - 2\theta$ scan for pure Pt ($x = 0$) indicating the fcc (111) peak and Si substrate. (b) X-ray scans of alloys with increasing Ru content. Arrows indicate a gradual shift from fcc to hcp phase. (c) Extracted lattice spacing $d_{(hkl)}$ and corresponding grain size from XRD peaks in (b). (d) TEM micrograph of $x = 78$ sputtered film. (e) Selected-area diffraction pattern of the sample in (d) with peaks identified.

family of planes in the fcc structure is equivalent to that between the (002) planes in the hcp structure. Hence, we can directly compare the evolution in the lattice spacings across the series in Fig. 1(c). The smooth reduction in lattice spacing with increasing Ru content is consistent with the binary phase diagram of the alloy.

Average grain sizes were estimated from the XRD peak widths using the Scherrer equation [18]:

$$d = \frac{k\lambda}{\beta_{\text{FWHM}} \cos \theta}, \quad (2)$$

where k is the shape factor taken to be 0.89 for spherical particles, θ is the Bragg angle, and β_{FWHM} is the full width at half maximum of the peak. The calculated average grain sizes vary between 11 and 27 nm across the composition series, denoted by the blue data points in Fig. 1(c). These length scales indicate a polycrystalline microstructure in which impurity scattering and grain-boundary disorder are expected to contribute comparably across compositions.

A transmission electron microscopy (TEM) image of the $x = 78$ alloy [Fig. 1(d)] confirms the lattice spacing extracted from XRD and Eq. (1). TEM image analysis also confirms distinct polycrystalline grains, uniformly distributed films, and no macroscopic clustering (see Supplemental Material, Sec. III [15]). This observation supports the absence of phase segregation or large-scale compositional inhomogeneity. The corresponding selected-area diffraction pattern in Fig. 1(e) illustrates high-intensity reflections characteristic of the expected hcp phase for this particular composition.

IV. TRANSPORT ANALYSIS

Ru_xPt_{100-x} films were patterned into $20 \times 55 \mu\text{m}^2$ Hall-bar structures using photolithography (Heidelberg MLA100) followed by physical ion milling. The longitudinal resistivity, ρ_{xx} , was measured as a function of temperature using a Quantum Design Physical Property Measurement System.

Temperature-dependent measurements are shown in Fig. 2(a). All compositions exhibit metallic behavior, with ρ_{xx} increasing gradually with temperature from 2 to 300 K. As shown in Fig. 2(b), the residual resistivity ratios [RRRs = $\rho_{xx}(300 \text{ K})/\rho_{xx}(2 \text{ K})$] remain below 1.5 for alloys with $38 \leq x \leq 78$, indicating strong scattering in these films. Remarkably, the resistivities remain well below $70 \mu\Omega \text{ cm}$ across the entire composition range and are significantly lower than other high-resistivity alloy systems such as AuTa and TaW, which have demonstrated large SHAs. Low resistivity is beneficial for minimizing energy dissipation in spintronic devices and maintaining efficient spin-current injection.

The 2-K resistivities [inset, Fig. 2(a)] exhibit a characteristic parabolic dependence on the Ru concentration that is consistent with Matthiessen's rule and typical alloy scattering. This confirms that the transport is dominated by substitutional disorder without anomalous scattering contributions.

To clarify the relative roles of intrinsic and extrinsic mechanisms in the Ru-Pt system, it is useful to distinguish between a formal dilute-impurity limit and the higher-disorder solid-solution regime relevant here. In sputtered

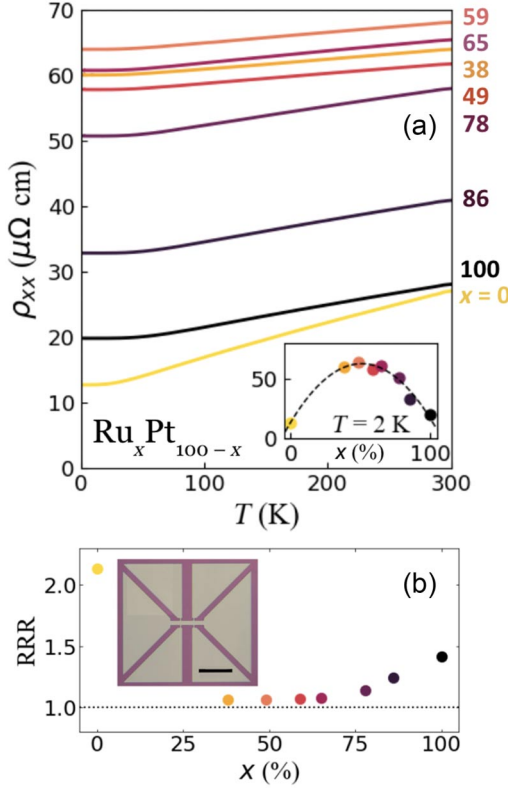


FIG. 2. (a) Temperature-dependent longitudinal resistivity ρ_{xx} for various alloy compositions. Inset: resistivities at 2 K plotted as a function of Ru composition ($x\%$). (b) Composition-dependent residual resistivity ratios. Inset: optical image of representative Hall bar structure used for transport measurements. Scale bar is 150 μm .

Pt films and Pt-based alloys with ρ_{xx} in the range of tens of $\mu\Omega\text{ cm}$, spin Hall transport is well established to be dominated by intrinsic Berry-curvature contributions, with side-jump scattering becoming increasingly relevant with increasing impurity disorder, while skew scattering becomes negligible outside ultraclean conditions [19–21]. To make the scaling argument explicit, the SHA can be written as [19,22]

$$\theta_{\text{SH}} = -\frac{\rho_{\text{SH}}}{\rho_{xx}} \approx (\sigma_{\text{SH}}^{\text{int}} + \sigma_{\text{SH}}^{\text{sj}})\rho_{xx}, \quad (3)$$

where the skew-scattering term is negligible in the moderately dirty regime relevant here and therefore omitted in Eq. (3). This formulation clarifies that the composition dependence of the SHA is governed by the product of the longitudinal resistivity and an effective spin Hall conductivity. Hence, for sputtered alloys in the moderately dirty regime relevant here, the SHA can be approximated as

$$\theta_{\text{SH}} \approx \sigma_{\text{SH}}^{\text{eff}}\rho_{xx}. \quad (4)$$

The increase in SHA from pure Pt to intermediate Ru concentrations reflects an increase in resistivity while

maintaining a large Pt-dominated spin Hall conductivity. The subsequent decrease of the SHA in the Ru-rich regime indicates a reduction in the effective spin Hall conductivity as the electronic structure evolves away from Pt.

The Ru-Pt alloys studied here lie well within this higher-disorder regime. While the crystal structure evolves gradually from predominantly fcc at low Ru concentrations to predominantly hcp at higher Ru concentrations, this evolution does not produce distinct features in either resistivity or SHA. Instead, disorder-driven scattering governs the primary composition dependence, with structural changes providing an electronic-structure context that may modestly modulate intrinsic contributions without acting as the dominant mechanism [23,24].

V. SPIN HALL ANGLE CHARACTERIZATION

To directly probe the SHE in $\text{Ru}_x\text{Pt}_{100-x}$ alloys, we employed three independent techniques: spin-torque magneto-optic Kerr effect (ST-MOKE), spin-torque ferromagnetic resonance (ST-FMR), and time-domain terahertz (THz) emission spectroscopy (TETS). These methods provide complementary and quantitative measurements of the SHA, enabling us to cross validate the extracted SHAs across techniques with differing sensitivities to bulk and interface effects, and to correlate the observed spin-charge conversion efficiency with alloy composition and scattering mechanisms.

A. ST-MOKE measurements

A 780-nm pulsed-laser source was used with an estimated spot size of $2r = 9 \pm 1 \mu\text{m}$, which was used in the fitting procedures [25]. The detailed calibration process is outlined in Ref. [15]. As shown in Fig. 3(a), stacks of $\text{Ru}_x\text{Pt}_{100-x}$ (5 nm)/ $\text{Co}_{40}\text{Fe}_{40}\text{B}_{20}$ (10 nm) were fabricated (hereafter we use CoFeB to represent $\text{Co}_{40}\text{Fe}_{40}\text{B}_{20}$). The yellow and red arrows represent the effective fields from dampinglike (τ_{DL}) and fieldlike (τ_{FL}) torques, respectively. A sinusoidal ac current, I_c , with an amplitude of 20 mA at 1.333 kHz was applied to the sample along the x direction during the measurement. MOKE voltages are measured while the laser position is fixed and the sample is translated along the y direction. To quantify the SHA, we use the self-calibration method [26,27] [Fig. 3(b)]. The MOKE signal from the SOT and Oersted field can be calculated as

$$\Delta V_{\text{Oersted}}(x, y) = \zeta \frac{1}{2\pi r^2} \oint \int dy' dx' \epsilon \frac{h_{\text{Oersted}}(y')}{H_{\text{eff}} + M_{\text{eff}}} \quad (5)$$

and

$$\Delta V_{\text{SOT}}(x, y) = \zeta \frac{1}{2\pi r^2} \oint \int dy' dx' \epsilon \frac{h_{\text{SOT}}(y')}{H_{\text{ext}} + M_{\text{eff}}}, \quad (6)$$

where r is the radius of the laser spot, $\epsilon = e^{-\frac{(x-x')^2 + (y-y')^2}{r^2}}$, and ζ describes the sensitivity of the MOKE response,

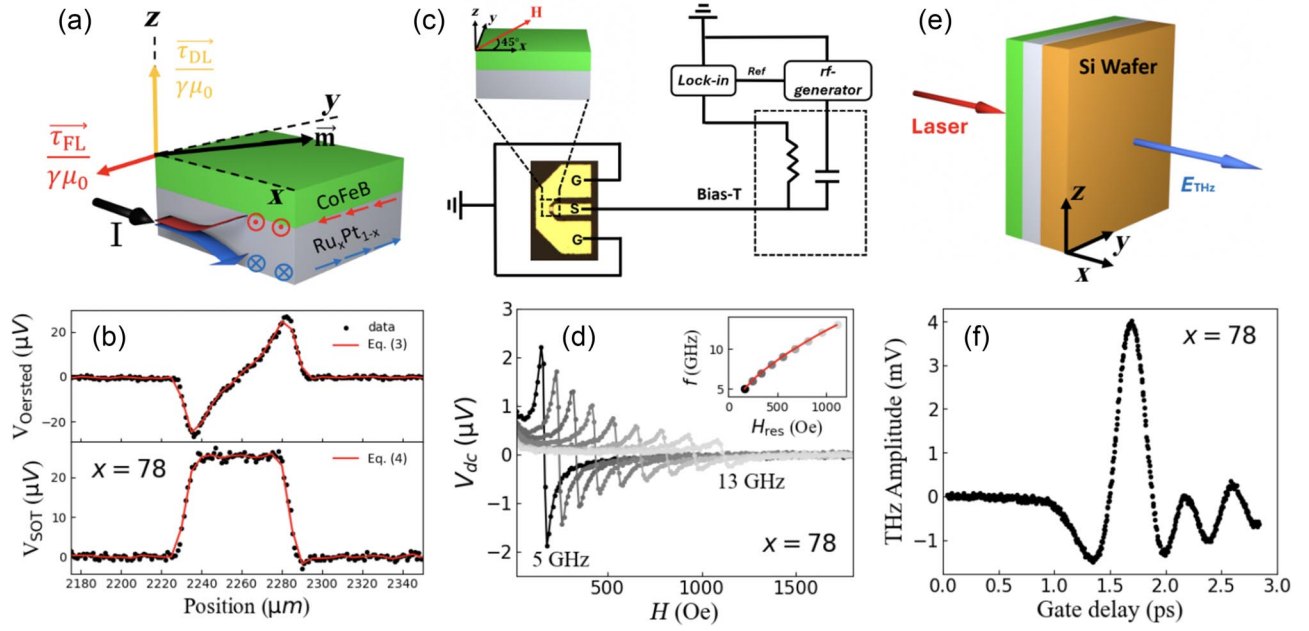


FIG. 3. Spin Hall angle measurement techniques. (a) Schematic of SOT in CoFeB from spin Hall effect in Ru-Pt used for MOKE measurement. The effective fields from dampinglike (τ_{DL}) and fieldlike (τ_{FL}) torques are shown, where γ is the gyromagnetic ratio. Local magnetization reorientation vector $\mathbf{m}$ is denoted. (b) Determination of the magnitude of SOT using self-calibration. Magnitude of SOT is determined by fitting to out-of-plane Oersted field (top) and SOT (bottom). (c) Schematic of ST-FMR measurement setup and device structure with rf excitation and lock-in detection. (d) Frequency-dependent ST-FMR spectra showing resonance shift with external applied magnetic field. Inset shows the Kittel fit. (e) Schematic of THz spectroscopy measurement. (f) THz emission spectrum of the $x = 78$ alloy.

defined as $\zeta = \frac{V_{MOKE}}{\Delta m_z}$. Due to the finite size of the laser spot, the Kerr rotation is calculated as the integral of the local magnetization reorientation, $\mathbf{m}$, weighted by the Gaussian function that describes the spatial distribution of the laser. The magnitude of the SOT-induced reorientation is then calculated as [28]

$$\frac{h_{SOT}}{I_c/2\pi w} = \frac{\Delta V_{SOT}(x, y)/e}{\Delta V_{offset}(x/z)/\oint dy' dx' \ln(\frac{w-y'}{y'})e}, \quad (7)$$

with $w = 50 \mu\text{m}$, the width of the patterned sample strip. The SHA of each Ru_xPt_{100-x} alloy, Θ_{MOKE} , can then be calculated by

$$\Theta_{MOKE} = \frac{2e}{\hbar} \frac{h_{SOT}}{j_{RuPt}} \mu_0 M_s d_{CFB}, \quad (8)$$

where e is the elementary charge, $\hbar$ is the reduced Planck's constant, j_{RuPt} is the current density in the Ru_xPt_{100-x} layer, μ_0 is the vacuum permeability, M_s and d_{CFB} are the saturation magnetization and thickness of the CoFeB layer, respectively. CoFeB has a resistivity of about $100 \mu\Omega\text{cm}$. By considering the thicknesses and resistivities of Ru_xPt_{100-x} and CoFeB within the parallel resistor model, we can estimate the current density in the Ru-Pt layer at different compositions.

B. ST-FMR measurements

The same Ru_xPt_{100-x} (5 nm)/CoFeB (10 nm) stacks were patterned into coplanar waveguide devices with a GSG configuration, impedance matched to approximately 50Ω . During measurement, an rf current, I_{rf} , with a nominal power of +10 dBm was sent through the sample strip along the x axis using a microwave probe [Fig. 3(c)]. During the measurement, we fixed the external magnetic-field angle at 45° to maximize the mixing signal. The measured rectifying voltage could be fitted with Lorentzian and anti-Lorentzian (dispersive) functions [29] expressed as

$$V_{dc} = c \left[V_{sym} \frac{\Delta H^2}{\Delta H^2 + (H_{ext} - H_{res})^2} + V_{asym} \frac{\Delta H (H_{ext} - H_{res})}{\Delta H^2 + (H_{ext} - H_{res})^2} \right], \quad (9)$$

where c is a constant, ΔH is the half linewidth, H_{res} is the resonance field, and H_{ext} is the external magnetic field.

V_{sym} and V_{asym} are proportional to the in-plane torque $\tau_{||}$ (out-of-plane effective field) and the out-of-plane torque $\tau_{\perp}$ (in-plane effective field). It is a reasonable assumption that $\tau_{\perp}$ is dominated by the torque originating from the Oersted field [4,29]. For the conventional case, the SHA of the Ru-Pt heavy metal (HM) can be expressed as

$$\begin{aligned}\Theta_{\text{FMR}} &= \frac{\tau_{\parallel}}{\tau_{\perp}} \frac{e\mu_0 M_s}{\hbar} d_{\text{CFB}} d_{\text{Ru-Pt}} \\ &= \frac{V_{\text{sym}}}{V_{\text{asym}}} \frac{e\mu_0 M_s}{\hbar} d_{\text{CFB}} d_{\text{Ru-Pt}} \sqrt{1 + \frac{M_{\text{eff}}}{H_R}}.\end{aligned}\quad (10)$$

By changing the rf signal frequency, M_{eff} can be extracted by fitting the resonant frequency, f_{res} , as a function of H_{res} to the Kittel formula [inset, Fig. 3(d)] [30]:

$$f_{\text{res}} = \frac{\gamma}{2\pi} \sqrt{H_{\text{res}}(H_{\text{res}} + \mu_0 M_{\text{eff}})}, \quad (11)$$

where γ is the gyromagnetic ratio.

We note that the SHA extracted from ST-FMR represents an effective charge-to-spin conversion efficiency that depends on the magnitude of the spin current injected into the ferromagnet (FM). This injected spin current can be influenced by interface-related parameters such as spin-mixing conductance and spin transparency at the FM/heavy-metal interface, as well as by the spin diffusion length of the RuPt layer.

In RuPt alloys, composition-dependent variations in interfacial electronic structure and spin-diffusion length may therefore modulate how efficiently bulk spin Hall currents are transmitted across the interface. Because all samples share the same FM layer thickness and deposition conditions, these effects act primarily as renormalization factors rather than uncontrolled variables. Notably, these effects modify the absolute magnitude of the extracted SHA while the composition-dependent trends reported here remain robust.

C. THz emission spectroscopy measurements

A 150-mW laser pulse of approximately 40 fs in pulse duration was used with a center wavelength of 800 nm at 10 kHz. The beam diameter was approximately 1 mm, and the laser fluence was 2.23 mJ/cm². The emitted THz wave was measured using electro-optic sampling with a 1-mm-thick (110) ZnTe crystal. The sample was placed in a constant field of 1.4 kOe provided by a pair of permanent magnets. The emitted THz electric field was recorded in the time domain and Fourier transformed for spectral analysis.

Figure 3(e) illustrates the THz emission setup in this experiment. Femtosecond laser pulses illuminate CoFeB and generate a hot, spin-polarized electron distribution that diffuses into the Ru-Pt layer, leading to a spin current in the HM layer. The spin current is then converted into charge currents due to the inverse SHE (ISHE), resulting in THz wave radiation [31]. The polarization of the injected spin current depends on the magnetization direction in CoFeB. Switching the magnetization direction with an applied magnetic field leads to polarity switching in the THz signal. It is worthwhile to note that the amplitude of

the THz signal is not only proportional to SHA but also depends on the resistivity of the HM layer and the light absorption efficiency of the FM layer [32–34].

Using the approach outlined in Ref. [35], we calculate the SHA of each alloy, $\Theta_{\text{THz}}^{\text{RuPt}}$, relative to that of pure Pt, $\Theta_{\text{THz}}^{\text{Pt}}$ by Eq. (12), which accounts for the amplitude of the emitted THz signal $S(\omega)$ obtained from the Fourier-transformed time-domain trace shown in Fig. 3(f), pump fluence absorbance A , spin-current relaxation length λ_r , impedance $Z(\omega)$, and film thickness t :

$$\frac{\Theta_{\text{THz}}^{\text{RuPt}}}{\Theta_{\text{THz}}^{\text{Pt}}} = \frac{S_{\text{THz}}^{\text{RuPt}}(\omega)}{S_{\text{THz}}^{\text{Pt}}(\omega)} \times \frac{A^{\text{Pt}}}{A^{\text{RuPt}}} \times \frac{\lambda_r^{\text{Pt}}}{\lambda_r^{\text{RuPt}}} \times \frac{\tanh(\frac{t_{\text{Pt}}}{\lambda_r^{\text{Pt}}})}{\tanh(\frac{t_{\text{RuPt}}}{\lambda_r^{\text{RuPt}}})} \times \frac{Z^{\text{Pt}}(\omega)}{Z^{\text{RuPt}}(\omega)}. \quad (12)$$

We assume that the detector response function is the same for all samples, as all the samples are measured in the same system, and that the spin-to-photon conversion efficiency is constant. This relative approach minimizes systematic uncertainties associated with absolute calibration and emphasizes composition-dependent trends.

Following Ref. [35], we estimate the spin-current relaxation length (λ_r) of Ru-Pt alloys to be approximately 0.475 nm using $\lambda_{\text{MF}}/\sigma_{zz} = 0.3 \times 10^{-15} \Omega\text{m}^2$ and the room-temperature alloy resistivities extracted from Fig. 2(a) at 300 K. We approximate $\lambda_r \approx \lambda_{\text{MF}}$, since at THz frequencies the spin-current relaxation length is comparable to the electron mean free path [36–38]. Similarly for Pt, we obtain $\lambda_s \approx 1.11$ nm, which is in quantitative agreement with previously reported Pt spin-current relaxation lengths at THz frequencies [39].

VI. SUMMARY OF RESULTS

The experimentally determined SHA values are summarized in Fig. 4(a). For plotting purposes, the Ru-Pt SHA values from THz spectroscopy, $\Theta_{\text{THz}}^{\text{RuPt}}$, have been scaled relative to the Pt SHA measured using MOKE to calculate the absolute values.

The MOKE-derived SHA peaks at $x = 38$ with a value of 0.117, which is 36% higher than the 0.086 value measured for pure Pt under the same conditions, while ST-FMR confirms the trend with an even larger SHA of 0.158 at the same composition. Based on these data, we conclude that the Ru₃₈Pt₆₂ alloy composition has the highest SHA within the composition window studied in this work. However, given the monotonic increase in SHA from 100% to 38% Ru, the true maximum may lie between 1% and 37% Ru concentration. Because of the limitations of the co-sputtering target geometry, compositions in this range were not accessible. The composition-dependent SHA exhibits a skewed parabolic trend with the maximum occurring for Pt-rich compositions. This is consistent with previous studies on Pt-based alloys where maxima were observed for Pt concentrations between 60% and

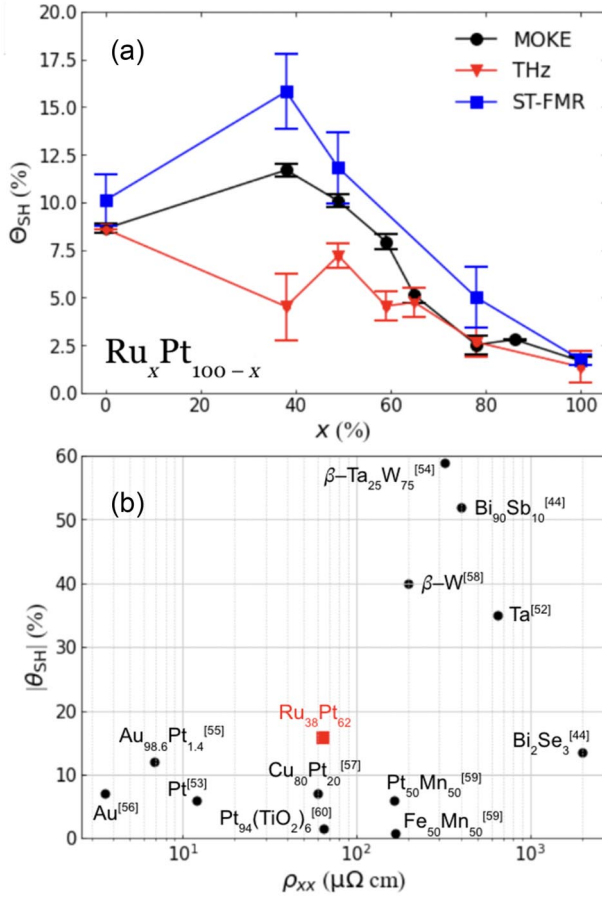


FIG. 4. Comparison of spin Hall angles. (a) Summarized data of measured SHA for each alloy composition using three independent techniques. (b) Comparison of the largest SHA measured within the composition range of this work ($Ru_{38}Pt_{62}$) to other elements and alloys of interest. Resistivity values gathered at 300 K.

90% [20]. Endpoint comparison alone indicates that pure Ru has a much smaller SHA than pure Pt, despite comparable transport scales. This strongly suggests a reduced effective spin Hall conductivity in the Ru-rich limit.

Although the Ru-Pt system undergoes a gradual structural evolution from fcc to hcp with increasing Ru concentration, this transition does not produce a distinct signature in either the resistivity or the SHA. Instead, it corresponds to disorder-driven scattering that governs the overall composition dependence, with crystal structure playing a secondary role in setting the absolute magnitude of the spin Hall response.

These SHA values exhibit slight technique-dependent variations due to differences in device geometry, interface sensitivity, and analysis methods. SHAs extracted using ST-FMR measurements are typically higher than those measured by MOKE. This can be explained by artifact effects including spin-pumping, longitudinal spin Seebeck effect, ISHE, and thermoelectrical voltages via the ordinary or anomalous Nernst effects [28,40–47]. All these

mechanisms can produce symmetric Lorentzian dc signals that mimic spin-torque rectification [48–50]. If these contributions become sufficiently strong, they can obscure or distort the accurate determination of τ_{DL} . Moreover, both genuine spin-torque rectification signals and spurious contributions from spin-pumping and thermoelectric effects exhibit the same angular dependence under an in-plane magnetic field, following a $\sin(2\theta)$ behavior, where θ is the angle between the field and the current direction [48,51,52]. This angular similarity makes disentangling artifact contributions particularly challenging in quantitative SOT studies.

Similarly, the phenomenological model for THz generation in spintronic emitters assumes spin-to-charge conversion solely in the HM layer due to SOC. While the spin-to-charge conversion in the HM layer also contributes to the THz emission, it is smaller than the charge pumping in the FM itself [53,54]. Contributions from FM-HM interface charge pumping can reduce the apparent SHA, which explains the lower THz-derived SHA values observed in this work [35,55].

In sputtered Ru-Pt alloys, the composition dependence of the SHA is governed by the interplay between longitudinal resistivity and effective spin Hall conductivity, while the differences between experimental techniques arise from their distinct transfer functions and interface sensitivities.

Although the average lattice spacing and crystal structure vary weakly between $x = 0$ and 40% Ru, transport properties such as RRR are highly sensitive to impurity scattering and chemical disorder, which are not captured by average structural metrics. The notable reduction in RRR over this composition range therefore reflects enhanced electron scattering from substitutional Ru atoms rather than a change in long-range crystal symmetry.

Figure 4(b) places the $Ru_{38}Pt_{62}$ alloy in context with other well-studied heavy-metal systems with regards to resistivity and SHA [48,56–64]. $Ru_{38}Pt_{62}$ achieves a favorable combination of moderate SHA with low resistivity, distinguishing it from commonly cited systems such as β -W and Ta. This underscores the advantage of $Ru_{38}Pt_{62}$ as a low-resistivity alloy that retains a sizable effective spin Hall conductivity.

VII. CONCLUSIONS

A systematic mapping of the SHA versus composition in Ru-Pt alloys using complementary techniques (ST-MOKE, ST-FMR, THz spectroscopy) has not been reported previously. Within the measured composition range, the largest SHA of 0.158 is obtained for the 38% Ru alloy with a resistivity below $70 \mu\Omega$ cm, making it intriguing and pertinent to SOT applications. This enhancement coincides with a gradual structural evolution from hcp to fcc; however, the dominant factor governing the SHA is impurity-induced disorder and scattering, with crystal

structure playing a secondary, modulatory role. This highlights an indirect structure-property correlation mediated through disorder rather than long-range symmetry alone.

Benchmarking against prior Pt-Ru studies, our result is comparable to, and in some cases exceeds, the effective SHA values reported for Pt₈₀Ru₂₀ in SOT stacks (approximately equal to 0.13–0.16) when normalized for resistivity and measurement technique [11,12]. Engineered Pt_{100-x}Ru_x/[CoNi]/Ru multilayers with optimized spin transparency can exhibit even larger effective SOT efficiencies (approximately equal to 0.28 – 0.31) [14], reflecting the strong role of periodic interface design and spin transparency in enhancing spin-current generation beyond bulk alloy limits.

Unlike multilayer architectures, homogeneous alloys lack periodic interfaces but offer a cleaner metric of bulk spin-charge conversion, allowing interpretation in terms of intrinsic and disorder-mediated mechanisms. Our systematic, composition-resolved study presents cross-technique validation and detailed structural characterization (XRD, TEM, high-temperature annealing) to provide a unified picture of how alloy composition, disorder, and structural evolution influence spin Hall transport, complementing and extending the alloy and multilayer studies previously reported.

From a practical perspective, the demonstration of a large SHA in sputtered films without resorting to molecular-beam-epitaxy growth or postdeposition annealing highlights the technological appeal and scalability of this system. Although the concept of impurity tuning is well understood in alloys such as AuTa and PtIr, Ru-Pt uniquely bridges the gap between 4d and 5d metals and provides a rare binary example where controlled disorder arising from composition-driven phase evolution acts as a tuning parameter. This places the Ru-Pt alloy as a promising, scalable alternative to β -W, Ta, or even topological insulators for spin-orbit torque devices.

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V. S. B. conceptualized the research, prepared the samples, performed structural characterization, analyzed the

data, and drafted the manuscript. X. W. and J. Q. X. conducted SHA measurements using MOKE and ST-FMR, analyzed the SHA mechanisms, processed data, and drafted the manuscript. S. S., M. B. J., and L. G. performed TDTs measurements. G. X. supervised the project, co-conceptualized the study, contributed to manuscript preparation, and secured funding. Special thanks are extended to Nawsher Jahan Parvez for assistance with X-ray photoelectron spectroscopy measurements.

DATA AVAILABILITY

The data that support the findings of this article are openly available [65].

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