Strain engineering of amorphous-Si thin film interfaces for efficient thermal spin to charge conversion

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Abstract

Interfacial asymmetry in conjunction with strain engineering can provide an alternate pathway to achieve efficient and controllable spin to charge conversion. This hypothesis is experimentally verified using spin-Seebeck effect measurement in case of B-doped amorphous-Si thin film interface. The spin-Seebeck voltage and spin-Hall angle in amorphous-Si is found to be an order of magnitude larger than the corresponding value for Pt thin film spin detector. Further, the spin-Seebeck effect is greatly enhanced in the multilayer heterostructures and it diminishes when the strain effects in the sample are reduced. The inhomogeneous strain induces strong interfacial Rashba-Dresselhaus spin-orbit coupling in the two-dimensional electron gas at the metal-Si interface. The resulting intrinsic inverse spin-Hall effect is the underlying cause of efficient spin to charge conversion, which is of the same order as the topological surface states. This study gives a new direction of research for spin-caloritronics applications using strain engineering and amorphous materials.
Introduction

Spin to charge conversion and its reciprocal are essential for development of spintronics and spin caloritronics devices. Spin to charge conversion in heavy metals and topological insulators primarily arises due to intrinsic spin orbit coupling (SOC). Rashba SOC due to structural inversion asymmetry (SIA) can provide an alternate mechanism since it can be tailored using interfacial properties. Rashba SOC arises due to an electric field perpendicular to the interface. At semiconductor interfaces, the spontaneous polarization under an applied strain gradient, known as flexoelectric effect as shown in Figure 1 (a), will give rise an electric field normal to the interface. While the flexoelectric polarization is very weak in bulk centosymmetric materials but it can be orders of magnitude larger at nanoscale and interfaces[1-3]. For a uniform strain gradient, the flexoelectric field can be written as:

\[ E_l = \frac{\mu_{ijkl} \epsilon_{ijkl}}{\varepsilon} \frac{\partial E}{\partial x_k} \]  

(1)

Where \( \epsilon_{ijkl} \) and \( \varepsilon \) are strain gradient, flexoelectric coefficient and dielectric constant respectively. Hence, the flexoelectric field will give rise to interfacial Rashba SOC[4] at the interface, which can be written as:

\[ H_R \propto (\vec{E} \times \vec{p}) \cdot \vec{\sigma} = E_z (-p_y \sigma_x + p_x \sigma_y) = \frac{\mu_{xzz}}{\varepsilon} \frac{\partial E}{\partial x_z} (-p_y \sigma_x + p_x \sigma_y) \]  

(2)

Where \( \vec{p} \) and \( \vec{\sigma} \) are momentum and spin polarization vectors respectively. Unlike the Rashba SOC in metallic thin film structures, the flexoelectric field driven Rashba SOC can be controlled by changing the strain gradient. Theoretical studies, over the years, have proposed that mechanical strain [4-6] can be used to induce spin splitting[7] in bulk non-centosymmetric semiconductors and 2D-materials. Hence, Dresselhaus type spin splitting may also arise due to normal and shear strain at the interface[7]. This effect will arise in centosymmetric materials only in the presence of strain gradient that breaks the symmetry. The flexoelectric effect will also lead to charge separation and increasing the charge carrier density in two-dimension electron gas system (2DES) at the interface. The combined effect of strain, strain gradient and charge separation in doped semiconductors will lead to an interfacial 2DES with strong Rashba-Dresselhaus SOC (RD-SOC). The resulting spin to charge conversion can be tuned to have efficiency larger than the heavy metals and same order as topological insulator surface states.

The crystalline Si is centosymmetric but amorphous-Si (a-Si) lacks center of inversion, which may exhibit larger flexoelectric polarization as well as strain splitting at nanoscale. Hence,
2DES at the strained a-Si interface may exhibit strong RD-SOC and large spin to charge conversion. This hypothesis is tested using spin-Seebeck effect (SSE) measurement in p-doped a-Si thin film interfaces. The spin to charge conversion in case of a-Si is found to be an order of magnitude larger than that of Pt. This difference disappears when the strain is relaxed, which proves our hypothesis that strain and strain gradient due to residual stresses give rise to 2DES and RD-SOC.

**Experimental setup**

The SSE, discovered by Uchida et al.[8,9], is a composite effect of thermal spin current and spin to charge conversion, which produces an electric field given by,

\[ E_{SHE} = -S \sigma \times \nabla T \]  

where, S is spin Seebeck coefficient and \( \sigma \) is the spin polarization vector. In the above equation, \( \sigma \) can be replaced by \( M \) (magnetization), which gives rise to equation for anomalous Nernst effect (ANE). It is noted that both SSE and ANE have identical symmetry behavior, which may lead to false identification of ANE as SSE[10-13]. Recent observation of SSE in sputtered Si[14,15] thin films have challenged the current understanding of spin transport and spin to charge conversion behavior. Bhardwaj et al.[14,15] attributed the spin to charge conversion due to Rashba SO at the Ni_{80}Fe_{20}/a-Si interface. However, the reported studies lack the mechanistic insights into the origin of Rashba-SOC, which led to questioning of experimental observations and attributing the results to ANE. This motivated us to undertake spin transport studies on p-doped a-Si to uncover the mechanistic origin of SSE.

We use the LSSE[16] configuration in this experimental study, where spin current is parallel to temperature gradient[17]. We make the LSSE experimental devices using microelectromechanical system (MEMS) fabrication methods[14,15] (Supplementary Information A) as shown in Figure 1(b). In the LSSE device configuration, the temperature gradient (\( \Delta T_z \)) is generated across the thickness of the thin film sample by passing an electric current (I) through the Pt heater, which is placed on top of the sample and generates a non-local heating effect across the sample thickness as shown in Figure 1 (c). The temperature difference across the thin film sample creates a spin current (\( J_s \)), which then get converted into a charge current (\( J_c \)) and measured as \( V_{2\omega} \) response (being quadratic in heating current) as shown in Figure 1 (b) and (d).

In the current experimental setup, a sputtered 50 nm thick MgO layer is used to electrically isolate the heater from the sample. The sputtered MgO layer on top of sample will have residual
compressive stress, which can be attributed to lattice mismatch between the thin films and difference in thermal expansion\cite{18,19}. Similarly, Pt heater thin film layer will also have large residual compressive stresses (Supplementary Figure S1). The residual compressive stresses in MgO and Pt heater layer will lead to a strain as well as a strain gradient in the sample as shown in the Fig. 1 (d). The strain gradient will cause flexoelectric effect as shown in Fig. 1 (a) and may lead to hypothesized RD-SOC at the interface. It is noted that Pt heater layer is electrically isolated from the sample and will have no contribution towards any spin dependent response.

(Figure 1)

Both SSE and ANE have the same symmetry behavior and careful experimental analysis is required to separate the two. Since the experimental setup has a ferromagnetic thin film in contact with the a-Si layer, ferromagnetic proximity effect may also lead to an additional spin dependent behavior. To decouple the contribution of SSE, ANE, and ferromagnetic proximity effect, we fabricated (Supplementary information- A) four set of LSSE devices having similar experimental structure but different sample layers. The first two set of devices have following thin film structure- (a) Ni$_{80}$Fe$_{20}$ (25 nm)/a-Si (50 nm) and (b) Ni$_{80}$Fe$_{20}$ (25 nm)/a-Si (5 nm). These devices with different thicknesses of a-Si layer (50 nm and 5 nm) will allow us to verify our hypothesis since strain gradient will change with thickness. The third device is fabricated with following sample structure- (c) Pt (3 nm)/Ni$_{80}$Fe$_{20}$ (25 nm). The Pt layer as spin to charge conversion can help us in evaluating the efficiency as well as eliminate ANE as the underlying cause of observed behavior. The order of sample layer is switched because ANE and SSE will have same sign for heating the top surface\cite{20}. And the last sample has the following structure- (d) Ni$_{80}$Fe$_{20}$ (25 nm)/Cu (10 nm)/a-Si (50 nm). We introduced 10 nm Cu between Ni$_{80}$Fe$_{20}$ and a-Si layer to eliminate the contribution of ferromagnetic proximity effect\cite{12,21}. The amorphous nature of sputtered Si layer is verified using high resolution transmission electron microscope (HRTEM) study (Supplementary information-B) for Ni$_{80}$Fe$_{20}$ (25 nm)/a-Si (5 nm) sample as shown in Supplementary Figure S2 (a)-(b). From the HRTEM study, we observe a continuous a-Si thin film layer for a 5 nm a-Si sample. From energy dispersive X-ray spectroscopy (EDS) analysis and elemental mapping (materials and methods), we eliminate any measurable Ni or Fe diffusion in the a-Si layer either as shown in Supplementary Figure S2 (b). Using HRTEM and AFM measurement, we report that the mean roughness for the a-Si samples is $\sim$1.22 nm as shown in Supplementary Figure S3.
Experimental results

The LSSE measurements are undertaken inside Quantum Design’s physical property measurement system (PPMS). We apply a heating current bias at 5 Hz across the heater and acquire the $V_{2\omega}$ response (being quadratic in heating current) using lock-in amplifier as a function of magnetic field (1500 Oe to -1500 Oe) applied in the y-direction (normal to the temperature gradient). The data is acquired at 300 K as shown in Figure 2 (a). We measure the $V_{2\omega}$ response to be 20.7 $\mu$V, 85 $\mu$V and 165.45 $\mu$V for Pt (3 nm), a-Si (50 nm) and a-Si (5 nm) respectively as shown in Figure 2 (a). The heating current ($I$) in case of Pt sample is 25 mA whereas it is 30 mA in case of a-Si samples. For same heating power ($I^2$), we estimate that the total thermal response will not be more than ~30 $\mu$V in case of Pt sample. The $V_{2\omega}$ response of the Pt sample will have contribution from both ANE and SSE. And the difference between the responses of Pt sample and a-Si samples will arise due to SSE only. The $V_{2\omega}$ response in case of a-Si is significantly larger than Pt: approximately three times larger in case of 50 nm a-Si sample and more than five times larger in case of 5 nm a-Si sample. The thermal resistance and consequently temperature drop across the 50 nm a-Si will larger than 5 nm a-Si. Hence, the measured response in case of 5 nm a-Si sample should not be larger than 50 nm a-Si sample if the response is due to ANE. As stated earlier, the interfacial roughness between 50 nm and 5 nm a-Si samples is not different and will not cause the observed difference in measurements. The difference in response between 5 nm and 50 nm a-Si sample can only arise due to SSE. For 50 nm a-Si sample, the strain gradient and resulting Rashba SOC will be smaller than that in 5 nm a-Si sample. The larger strain gradient and Rashba SOC leads to larger SSE response in 5 nm a-Si sample. For the Pt/ Ni$_{80}$Fe$_{20}$ sample, the ANE and SSE responses are reported to be similar[20]. The magnitude of both ANE and SSE responses is estimated to be ~15 $\mu$V in Pt/ Ni$_{80}$Fe$_{20}$ sample. The corresponding magnitudes of ANE and SSE responses for 5 nm a-Si sample are ~15 $\mu$V and ~150 $\mu$V. Hence, the SSE response in case of a-Si sample is an order of magnitude larger than that of Pt sample.

The intrinsic spin-orbit coupling in Si is negligible as stated earlier and such a large $V_{\text{SSE}}$ should not arise in a-Si layer according to conventional wisdom. It is hypothesized that the ferromagnetic proximity effect may lead to such a behavior. To ascertain the contribution due to ferromagnetic proximity effect, the LSSE measurement is undertaken on Ni$_{80}$Fe$_{20}$ (25 nm)/Cu (10 nm)/a-Si (50 nm) sample is presented in Figure 2 (b). For similar heating power, we measure the $V_{2\omega}$ response to be 95.8 $\mu$V, which is larger than 50 nm a-Si sample without Cu interlayer. The
larger response in this case may be due to additional temperature drop across the sample due to more interfaces. This measurement eliminates ferromagnetic proximity effect and clearly establishes SSE to be the primary mechanism responsible for the observed $V_{2\omega}$ response.

(Figure 2)

There are two controlling mechanisms in SSE: a) generation of thermal spin current in FM and b) spin to charge conversion due to spin-orbit coupling in NM. Based on experimental measurements, we can state that spin to charge conversion takes place at the a-Si interface layer since conductivity of bulk a-Si is poor as compared to Ni$_{80}$Fe$_{20}$ thin film. To uncover the origin of spin current, we undertake LSSE measurement as a function of large applied magnetic field and temperature. At 5 K, we measured the $V_{2\omega}$ response on the 5 nm a-Si bilayer sample for 10 mA of heating current and applied magnetic field from 14 T to -14 T as shown in Fig. 2 (c). The magnonic spin current can be suppressed at low temperature and with high magnetic field. But we do not observe any reduction in $V_{2\omega}$ response as shown in Figure 2 (c). This suggests that the origin of the spin current is due to spin dependent Seebeck effect (SDSE). This assertion is further supported by the $V_{2\omega}$ response as a function of temperature from 10 K to 300 K with applied magnetic field 1 T, 5 T, 10 T, and 14 T as shown in Fig. 2 (d). The $V_{2\omega}$ response decreases for decreasing temperature, but it does not approach zero at low temperature. This shows that the origin of spin current is most likely electronic, hence SDSE is the underlying cause of SSE reported in this work. The increase in the $V_{2\omega}$ response with increasing magnetic field can be attributed to increase in SDSE due to reduction in electron-magnon scattering at higher magnetic fields. While the origin of spin current is due to SDSE, we propose to call it SSE since the detection is due to spin to charge conversion.

(Figure 3)

In traditional thermoelectrics, multilayer heterostructures are reported to enhance the thermoelectric energy conversion. To test it, we fabricated LSSE experimental device (Supplementary information-A) having a multilayer heterostructure sample consisting of \{Ni$_{80}$Fe$_{20}$ (10 nm) / a-Si (25 nm)\}_3, which will have strain and strain gradient due to MgO and Pt heater layers constraining the sample as shown in Figure 3(a). For this sample, we observe the $V_{2\omega}$ response to be 175.36 $\mu$V for a heating current of 30 mA as shown in Figure 3 (b), which is twice as large as compared to the 50 nm a-Si sample for same heating current shown in Figure 2 (a).
This enhancement occurs due to increase in interfacial temperature drop and spin to charge conversion at multiple interfaces.

To ascertain contribution of strain, we modified the experimental setup by switching the position of the heater and sample as shown in Figure 3 (c). In this second configuration, the sample is no longer constrained by the MgO (insulator) and Pt (heater) layers as a consequence the strain and strain gradient effects will be insignificant. For the new experimental configuration, we fabricated another set of devices with multilayer heterostructure and a reference Pt/Ni$_{80}$Fe$_{20}$ bilayer. We measured the $V_{2\omega}$ response of $\sim+0.35$ $\mu$V for multilayer heterostructure and $\sim-0.7$ $\mu$V for Pt (3 nm)/ Ni$_{80}$Fe$_{20}$ (25 nm) sample in the second configuration. The $V_{2\omega}$ response, in case of Pt (3 nm)/ Ni$_{80}$Fe$_{20}$ (25 nm) sample, is also verified using angular rotation experiment in the plane of the sample (Supplementary Figure S4). The corresponding ANE response for Ni$_{80}$Fe$_{20}$ (25 nm) sample is -0.33 $\mu$V as shown in Supplementary Figure S5. Hence the SSE response for Pt (3 nm)/ Ni$_{80}$Fe$_{20}$ (25 nm) sample is $\sim-0.37$ $\mu$V. For the Pt/ Ni$_{80}$Fe$_{20}$ sample, the ANE and SSE responses are reported to be equal in magnitude[20] and our measurement supports that. The sign of SSE response is also consistent with the reported measurements[20]. This shows that the sign of spin to charge conversion in a-Si is negative and corresponding SSE response should be -0.68 $\mu$V. From the earlier measurement presented in Figure 2 (a) and Figure 3 (b), the $V_{2\omega}$ response for multilayer is expected to be approximately ten times that of Pt sample response. However, the observed response for multilayer sample in second configuration is twice of the Pt sample. The multilayer sample as shown in Figure 3 (c) is not constrained by the MgO and Pt heater layer. Hence, the strain and strain gradient are significantly reduced in the new sample configuration leading to smaller SSE response. It is noted that some residual stresses will always be there in the new configuration. We fabricated another a-Si (50 nm)/Ni$_{80}$Fe$_{20}$ (25 nm) sample on top device and did not observe any SSE response as shown in Supplementary Figure S6. These experiments clearly demonstrate that the combined effect of strain and strain gradient is the underlying cause of the giant SSE observed in Ni$_{80}$Fe$_{20}$/a-Si bilayer and multilayer heterostructures.

Discussion

Sinova et al.[22] predicted intrinsic spin-Hall effect in high mobility semiconductor 2DES with substantial Rashba SOC. The metal/a-Si(p-doped) interface will have a 2DES due to band bending[23,24]. The flexoelectric charge separation will increase the charge carrier density whereas normal strain will enhance the mobility. As a consequence, inhomogeneous strain will
lead to strong RD-SOC coupled interfacial 2DES, similar to the one modeled by Sinova et al.[22]. The resulting intrinsic spin-Hall effect will have spin-Hall angle values orders of magnitude larger than the extrinsic spin-Hall effect from skew scattering and side jump process. The reported values of the spin-Hall angles in extrinsic regime are $10^{-2}$-$10^{-3}$ [25,26], whereas the spin-Hall angle in intrinsic regime is expected to be two orders of magnitude larger[27]. As stated earlier, the spin to charge conversion for a-Si is ten times that of Pt sample. The spin-Hall angle of Pt is $\theta_{SH} = \sim0.055 - 0.1$[28,29] and corresponding spin-Hall angle for a-Si will be $\theta_{SH} = \sim0.55 - 1.0$, which is of the same order as topological insulator surface states[30]. The estimated spin-Hall angle values suggest that intrinsic inverse SHE in Rashba SOC coupled interfacial 2DES may give rise to efficient spin to charge conversion at a-Si interface. Both Ni/Si and Cu/Si interfaces have Ni and Cu at the Si interstitial sites[31,32] with d-p orbital overlap and the resulting Schottky barrier in both cases is also $\sim0.65$ eV[31,33]. This leads to the similar interfacial 2DESs and as a consequence similar SSE response as reported in this study.

It is noted that the spin-Hall angle may not be the ideal way to estimate spin to charge efficiency in case of 2DES. For Rashba SOC coupled 2DES, an alternate method to define the spin to charge conversion efficiency is Rashba-Edelstein length $\lambda_{IEE} = \alpha_R \tau/\hbar$, where $\alpha_R$ and $\tau$ are Rashba parameter and relaxation time, respectively. Using a value of $\lambda_{IEE} = 0.14$ nm for Pt[34], we estimate the corresponding $\lambda_{IEE} = 1.4$ nm for a-Si (5 nm). This value is of the same order as topological surface states in case of strained HgTe [35]($\lambda_{IEE} = 2$) and 2DES at STO/LAO[34]. However, metal interface with highly doped a-Si is not expected to have topological insulator interfacial states. Hence, intrinsic SHE in Rashba SOC coupled 2DES is proposed to be the underlying cause of the observed behavior as proposed by Sinova et al.[22]. However, Ni and Cu impurities at the interface may contribute towards the observed behavior, which needs further interfacial characterization and exploration.

In conclusion, we report large thermal spin to charge conversion at strained metal/a-Si interface. The efficiency if spin to charge conversion is an order of magnitude larger than that of Pt and two orders of magnitude larger than other semiconductors. Strain engineering has always been considered for non-centosymmetric materials with large intrinsic spin-orbit coupling. The observation of large SSE in a-Si challenges this inherent assumption in spintronics and spin-caloritronics research. The existing theoretical models has to be modified to account for strain and strain gradient effects. This work may have far reaching effect on the spin transport and topological
materials studies since materials with smaller intrinsic spin-orbit coupling can be incorporated in the new research.

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Figure 1. (a) schematic showing flexoelectric polarization in Si lattice due to strain gradient, (b) schematic showing the experimental setup with temperature gradient, (c) a representative false color scanning electron micrograph showing the experimental device, (d) schematic showing the origin of strain and strain gradient leading to interfacial spin-Seebeck effect,

Figure 2. (a) spin-Seebeck effect measurement for Ni$_{80}$Fe$_{20}$ (25 nm)/a-Si (5 nm), Ni$_{80}$Fe$_{20}$ (25 nm)/a-Si (50 nm) and Pt (3 nm)/Ni$_{80}$Fe$_{20}$ (25 nm) at 300 K, (b) spin-Seebeck effect measurement for Ni$_{80}$Fe$_{20}$ (25 nm)/Cu (10 nm)/a-Si (50 nm) at 300 K, (c) high magnetic field spin-Seebeck effect measurement for Ni$_{80}$Fe$_{20}$ (25 nm)/a-Si (5 nm) at 5 K, and (d) temperature dependent spin-Seebeck effect measurement for Ni$_{80}$Fe$_{20}$ (25 nm)/a-Si (5 nm) at an applied magnetic field ($\mu H_y$) of 1 T, 5 T, 10 T and 14 T from 300 K to 10 K. The fluctuations in the temperature dependent measurements are due to instrumental settings.

Figure 3. (a) schematic showing the spin-Seebeck effect measurement setup and origin of strain gradient in multilayer sample, (b) spin-Seebeck effect measurement for $\{\text{Ni}_{80}\text{Fe}_{20} (10 \text{ nm}) / \text{a-Si (25 nm)}\}_3$ sample at 300 K, (c) schematic showing the spin-Seebeck effect measurement setup with the position heater and sample switched leading to absence of strain and strain gradient, and (d) spin-Seebeck effect measurement for $\{\text{Ni}_{80}\text{Fe}_{20} (10 \text{ nm}) / \text{a-Si (25 nm)}\}_3$ sample and Pt (3 nm)/Ni$_{80}$Fe$_{20}$ (25 nm) sample in new experimental configuration.
Supplementary information- Strain engineering of amorphous-Si thin film interfaces for efficient thermal spin to charge conversion

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A. Device fabrication process

We take a prime Si wafer and deposit 350 nm of thermal silicon oxide using chemical vapor deposition (CVD). Using lift-off photolithography, we then deposit the sample to be studied using the RF sputtering. The sputtering deposition will have substrate conformal thin film coating. Hence, it will have the same interfacial and surface roughness as the underlying layers. The p-Si target used to deposit amorphous-Si layer is Boron-doped with resistivity of 0.005-0.01 Ω-cm. The second lift-off photolithography is carried out to deposit 50 nm MgO to electrically isolate the sample from the heater. The third lift-off photolithography is then used to deposit heater composed of Ti (10 nm)/Pt (100 nm) using e-beam evaporation.

For the second set of devices with switched sample and heater positions, we first deposited Ti (10 nm) / Pt (100 nm) on a Si wafer with predisposition of thermal silicon oxide (650 nm) using CVD. We then sputter 50 nm MgO using RF sputtering for electrical isolation. We fabricated two set of devices having the \{Ni_{80}Fe_{20} (10 nm) / p-Si (25 nm)\}_{x=3} trilayer heterostructure sample and Ni_{80}Fe_{20} (25 nm)/Pt (3 nm) sample on top of the MgO.
Supplementary Figure S1. (a)-(b) The optical images showing Pt thin film heater layer peeled off (delaminated) due to residual stresses for two devices. The residual stress in Pt heater and MgO layer is proposed to be the cause of strain and strain gradient in the underlying sample.

**B. Sample characterization**

**TEM sample preparation**- TEM lamellae were prepared from the layered sample following established procedures with a DualBeam scanning electron microscope and FIB instrument using Ga ion source (Quanta 200i 3D, ThermoFisher Scientific). First, a strap of 5 µm thick protective Carbon layer was deposited over a region of interest using the ion beam. Subsequently approximately 80 nm thin lamella of was cut and polished at 30 kV and attached to a TEM grid using in-situ Omniprobe manipulator. To reduce surface amorphization and Gallium implantation final milling at 5 kV and 0.5 nA was used to thin the sample further.

**S/TEM imaging and analysis**- TEM and STEM imaging was performed at 300 kV accelerating voltage in a ThermoFisher Scientific Titan Themis 300 instrument, fitted with X-FEG electron source, 3 lens condenser system and S-Twin objective lens. High-resolution TEM images were recorded at resolution of 2048x2048 pixels with a FEI CETA-16M CMOS digital camera with beam convergence semi-angle of about 0.08 mrad. STEM images were recorded with Fischione Instruments Inc. Model 3000 High Angle Annular Dark Field (HAADF) Detector with probe current of 150 pA, frame size of 2048x2048, dwell time of 15 µsec/pixel, and camera length of 245 mm. Energy dispersive X-ray Spectroscopy (EDS) analyzes and elemental mapping were obtained in the STEM at 300 kV, utilizing ThermoFisher Scientific SuperX system equipped with 4x30mm² window-less SDD detectors symmetrically surrounding the specimen with a total
collection angle of 0.68 sr, by scanning the thin foil specimens. Elemental mapping was performed with an electron beam probe current of 550 pA at 1024 x1024 frame resolution.

Supplementary Figure S2. (a) high resolution transmission electron micrograph showing the layered structure of the experimental specimen and (b) energy dispersive X-ray spectroscopy elemental map showing the thin film layers and interfaces.

Atomic force microscope (AFM) characterization of surface roughness- The surface roughness of the bilayer sample directly reflects the underlying interfacial roughness. The interfacial roughness cannot be more than the surface roughness since the sputter coating is conformal. The AFM measurements are carried out on samples having 50 nm a-Si and 5 nm a-Si layers as shown in Supplementary Figure S2.
Supplementary Figure S3. The AFM measurements at the surface of Ni$_{80}$Fe$_{20}$ layer in (a) 50 nm and (b) 5 nm a-Si bilayer samples. The mean roughness of both samples is $\sim$1.2 nm.

C. Angular rotation in yx-plane for Pt (3 nm)/ Ni$_{80}$Fe$_{20}$ (25 nm) sample
Supplementary Figure S4. The angular rotation in yx-plane for Pt (3 nm)/ Ni$_{80}$Fe$_{20}$ (25 nm) sample showing contribution from SSE and planar Nernst effect (PNE) contributions.

$V_{2,0}$ (µV)

$\phi_{yx}$

ANE+SSE: 0.71 µV

PNE: -0.208 µV
D. **ANE response** for Ni$_{80}$Fe$_{20}$ (25 nm) thin film sample on top configuration

Supplementary Figure S5. The ANE response in case of Ni$_{80}$Fe$_{20}$ (25 nm) sample on top configuration.
E. Characterization of a-Si (50 nm)/Ni$_{80}$Fe$_{20}$ (25 nm) sample

Supplementary Figure S6. The $V_{2\omega}$ response of a a-Si (50 nm)/Ni$_{80}$Fe$_{20}$ (25 nm) sample on top of heater configuration showing absence of any measurable SSE response.