Spin-orbit torque efficiency in Ta or W/Ta-W/CoFeB junctions

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Abstract
Efforts to improve the spin–orbit torque efficiencies in junctions consisting of nonmagnetic and ferromagnetic layers are current technological interest to construct magnetic memory and logic devices. However, most of the previous studies have relied on pure metallic elements, and thus, the outcomes were marginal. Here, we investigate the spin–orbit torque efficiencies of junctions consisting of Ta-W and CoFeB with Ta and W underlayers, focusing on the thickness and concentration of Ta-W alloy. The damping-like torque change is more significant with the W underlayer than with the Ta underlayer. We found an enhancement in damping-like torque value reaching −23% measured for a sample consisting of 4 nm thick β-phase Ta13.5W86.5 at% layer. This value is substantially more significant than that of the junction with β-W only (−17%).

1. Introduction

It is of current technological interest to develop new memory solutions with low-energy consumption [1–3]. Magnetic random access memory (MRAM) is in the process of mass production for embedded memory application and can be used in other applications such as logic and random number generation [3–6]. Magnetization switching by spin–orbit torque (SOT) can be fast and energy-efficient [7]. In the junctions consisting of nonmagnetic (NM) and ferromagnetic (FM) layers, the SOT is generated in FM layers when the electrical current is injected into the NM layers [8–11]. SOT is composed of damping-like torque (DLT) and field-like torque (FLT): $\text{DLT} \sim \vec{m} \times (\vec{\sigma} \times \vec{m})$, and $\text{FLT} \sim (\vec{\sigma} \times \vec{m})$, where $\vec{m}$ is the magnetization vector and $\vec{\sigma}$ is the direction of accumulated spin. The bulk spin Hall effect generates the spin current when the charge current flows in the NM layer with sizeable spin–orbit coupling [12]. At the interface of NM and FM, when the interfacial Rashba effect generates the effective field due to potential gradients, the spin is accumulated by the interfacial Rashba effect [13]. The magnitude of DLT or FLT can be expressed as torque efficiency $\xi_{\text{DL}}$ or $\xi_{\text{FL}}$.

In the past, most studies focused on junction structures combining an NM with a large spin–orbit coupling with an FM, such as Pt/Co and Ta/CoFeB/MgO [8, 9]. Moreover, several approaches have been conducted to either increase the SOT efficiency and achieve deterministic switching of magnetization [14–21].

Additionally, studies on 3d, 4d metals with weak spin–orbit coupling have been carried out to broaden the understanding of spin–orbitronics [22–25]. Thus, the study on numerous materials serves as appropriate information to reveal the mechanism of SOTs.

Ta and W are adjacent on the periodic table and have excellent torque efficiency among the various materials. Therefore, they are widely used in IC fabrication and are semiconductor process-friendly. However, both Ta and W have two phases ($\alpha$ and $\beta$) with different electrical resistivities.

Table 1 presents a collection of the resistivity and DLT efficiency values from the previous studies measured by various methods, including spin-torque ferromagnetic resonance (ST-FMR), harmonic Hall resistance (HHR), and spin–Hall magnetoresistance (SMR). Besides, a first-principles electronic structure calculation showed that the spin Hall conductivity of the Ta13.5W86.5 alloy maintaining the $\beta$-phase was about 40% higher than that of the pristine $\beta$-W [26]. In our previous study, we measured the damping-like and field-like effective fields of the Ta-W alloys without underlayers [27]. However, the theoretical prediction and experimental results
did not match because the phase of the Ta-W alloy changed from $\beta$ to $\alpha$. Thus, in this study, we deposited the Ta-W alloy on the Ta or W underlayer to see whether we could keep the $\beta$-phase of the Ta-W alloy. The phase of Ta is known to be amorphous when it becomes ultrathin.

Meanwhile, the phase of W is $\beta$ up to 11 nm at the as-deposited state and 7 nm after 300 °C annealing [28].

The main differences between our previous and present reports are sample structure and spin–orbit–torque efficiency analysis. The structures of the previous report were the bilayer (Ta/W) and the single alloy (Ta-W).

The present study presents the spin–orbit torque efficiency of Ta-W alloy dependent on the thickness, concentration, and underlayer.

This study observes the change of the $\xi_{DL}$ and $\xi_{FL}$ values with various thicknesses and concentrations of Ta$_{1-x}$W$_x$ alloy with Ta and W underlayers. We varied the thickness of the Ta$_{1-x}$W$_x$ layer in the range of 1 to 4 nm.

We also modified the concentration of the W layer from 0 to 100 at%. It is difficult to obtain the SOT efficiency with the second harmonics measurement depending on the field dependence at the tilted angle in material systems with high planar Hall resistances, like $\beta$-W. Therefore, we measured the SOT efficiency of the junction samples with in-plane magnetic anisotropy [29]. Other researchers estimated SOT efficiency by using the azimuthal angle scanning method [30–32].

### 2. Experimental

We used direct-current and radio-frequency magnetron sputtering with a base pressure below $5 \times 10^{-6}$ Torr to deposit thin films on Si wafers with 300-nm thick amorphous SiO$_2$ layers. The films consisted of SiO$_2$/Ta or W 3/Ta$_{1-x}$W$_x$t/CoFeB 1.5/MgO 1/Ta 2 (nominal thicknesses in nm), where $t$ is in the range of 1–4 nm. The target composition of CoFeB is Co$_{40}$Fe$_{40}$B$_{20}$ (at%). We deposited the upper Ta layer to prevent the oxidation of other layers. All film stacks were annealed in a vacuum ($10^{-6}$ Torr) for 1 h at 300 °C under 0.6 T magnetic field applied horizontally to the sample surface. We measured the magnetic properties and crystal structure of the films by vibrating sample magnetometry (VSM, MicroSense EV9) and grazing-incidence X-ray diffraction (GI-XRD, PANalytical, XPert ProMPD), respectively, at RT.

Figure 1 schematically shows the junction layout for electrical measurements. We patterned the films into Hall bar junctions using photolithography. The width and length of the Hall bar are 5 μm and 35 μm, respectively. We used the second harmonic method to characterize the electrical properties. The injected ac frequency was 13.7 Hz. The external field applied was from 0.7 to 1.7 T. The magnetization aligned to the external field direction because the external field was sufficiently higher than the anisotropy field. Thus, the junction became a single-domain state.

We measured the voltage difference while changing the angle between the AC and external field directions. The lock-in amplifiers can simultaneously collect the first and second harmonic voltages. The first and second harmonic Hall resistance ($R_{1w}$ and $R_{2w}$) can be described as

$$R_{1w} = R_{PHE} \sin 2\varphi$$

$$R_{2w} = \left( \frac{R_{AHE}}{H_{eff} \sum_{\varphi} \cos \varphi} + R_T \cos \varphi \right) + 2R_{PHE}(2 \cos^3 \varphi - \cos \varphi) \frac{H_{FL} + H_{DL}}{H_{ext}}$$

where $R_{PHE}$ and $R_{AHE}$ are the planar and anomalous Hall resistances, $\varphi$ is the azimuthal angle of magnetization, $H_{ext}$ and $H_{eff}$ are the external field and the effective out-of-plane anisotropy field, respectively, $H_{DL}$ and $H_{FL}$ are the DL and FL effective field, respectively, $R_T$ is the second harmonic Hall resistance coefficient due to the thermoelectric voltage, and $H_{DL}$ is the current induced Oersted field.

### Table 1. Comparison of $\rho_{DL}$ and $\xi_{DL}$ values with various measurement methods.

| Sample (in nm) | $\rho_{DL}$ (μΩ·cm) | $\xi_{DL}$ | Method | References |
|---------------|----------------------|------------|--------|------------|
| Co$_{40}$Fe$_{40}$B$_{20}$/Ta 8 | 190 | 0.15 ± 0.03 | ST-FMR | [8] |
| W 5.2/CoFeB 2 | 260 | 0.33 ± 0.06 | ST-FMR | [26] |
| W 6/CoFeB 2 | 238 | 0.62 | HHR | [27] |
| W 4/Fe$_{2}$/Co$_{40}$B 1.5 | 185 | 0.40 | HHR | [28] |
| W 7/CoFeB 1.5 | 220 | 0.207 | SMR | [29] |
3. Results and discussion

Figure 2(a) displays the $R_{1w}$ as a function of $\varphi$ with the different external fields. As the external fields are higher than $H_{eff}^{He}$, the magnitude of $R_{PHHE}$ does not change. Figure 2(b) shows the change in $R_{2w}$. As the external field increased, the outline of $R_{2w}$ diminished. We divided equation (2) into $\cos \varphi$ and $(2 \cos^3 \varphi - \cos \varphi)$ contributions. Figures 2(c) and (d) display the $\cos \varphi$ term versus $1/\mu_0 H_{ext}(\mu_0 H_{eff}^{He})$ and the $(2 \cos^3 \varphi - \cos \varphi)$ term versus $1/\mu_0 H_{ext}$. The dotted line expresses the linear fit.
We calculated the \( \frac{1}{\mu_0} \) and \( \frac{H_{\text{ext}}}{H_{\text{ext}} + H_{\text{ex}}} \), respectively. We calculated the HDL and HFL \( \xi_{\text{DL}} + \xi_{\text{FL}} \) from the slopes in the graph of figures 2(c) and 2(d).

As the film thickness became much smaller than the width of the Hall bar, \( H_{\text{ex}} \) was calculated from simplified Ampere's law: \( H_{\text{ex}} = I/2w \), where \( I \) is the charge current in the NM layer and \( w \) is the width of the Hall bar. From the calculated HDL and HFL, the \( \xi_{\text{DL}} \) and \( \xi_{\text{FL}} \) could be described as

\[
\xi_{\text{DL(FL)}} = \frac{2eM_iH_{\text{Ex(DL(FL))}}}{\hbar J}
\]

where \( e \) is the electron charge, \( M_i \) is the saturation magnetization, \( t \) is the thickness of FM, \( \hbar \) is the reduced Planck constant, and \( J \) is the current density in the NM layer.

Figure 3 displays the \( \xi_{\text{DL}} \) and \( \xi_{\text{FL}} \) values using equation (3) depending on the underlayer material (Ta or W). As shown in figure 3(a), the \( \xi_{\text{DL}} \) value of pure Ta was \(-5\%\), and there was no significant change even when the thickness of Ta increased. Unlike pure Ta, the \( \xi_{\text{DL}} \) value of pure W increased as the thickness of W increased, as shown in figure 3(b). The highest \( \xi_{\text{DL}} \) value of pure W was \(-17\%\) at the 7 nm of W. When we deposited the pure W above the Ta underlayer, the \( \xi_{\text{DL}} \) values were about \(-12\%\), where the W thicknesses were 1 and 2 nm. However, the \( \xi_{\text{DL}} \) value was reduced by about 60\% when W thickness was 4 nm. We observed a similar phenomenon when we deposited the pure Ta above the W underlayer. The \( \xi_{\text{DL}} \) values with 1 or 2 nm thick Ta were about 50\% higher than that of 4 nm Ta. Now we are looking at the change according to the concentration of the Ta-W alloy with the Ta underlayer. As the concentration of W increased, the \( \xi_{\text{DL}} \) value also increased, except for the sample with a 4 nm thick Ta-W alloy, as shown in figure 3(a). At 4 nm of the Ta-W alloy thickness, the \( \xi_{\text{DL}} \) value for the sample with 4 nm thick Ta-W alloy was slightly larger than the samples with pure Ta or W layers.
The $\xi_{DL}$ values with 1 or 2 nm thick Ta-W alloy on the W underlayer were about $-10\%$, larger than that on the Ta underlayer. When the thickness of the alloy became 4 nm, the most significant concentration-dependent change occurred. The $\xi_{DL}$ value was $-5\%$ for a sample consisting of Ta$_{80}$W$_{20}$ at%, alloy, but it increased abruptly up to $-23\%$ for a sample with Ta$_{15}$W$_{85}$ at% alloy. We think this abrupt increase is related to the crystal phase transition and will discuss more later. The underlayer material might also have affected the crystallinity of the Ta-W alloy. As shown in figures 3(c) and (d), there was little change in the $\xi_{DL}$ value, about 5% at most concentration and thickness ranges.

Additionally, we conducted the current-induced magnetization switching. The switching current is known to be inversely proportional to the damping-like torque efficiency. We made a new sample with the perpendicular magnetic anisotropy in the range of W 60~90 at% with W underlayer. Note that the details in Supplementary material S1 and figure S1 (available online at stacks.iop.org/MRX/8/106102/mmmedia) for the switching experiment. From the switching experiment, we achieved the smallest switching current in the concentration of W 85 at%.

Figure 4 shows the electrical resistivity ($\rho_{xx}$) values of the NM layers and effective magnetic anisotropy energy ($K_{eff}$) values for different alloy thicknesses and concentrations. Note that we calculated the values of $\rho_{xx}$ from the Hall bar measurements and the values of $K_{eff}$ using the area method at out-of-plane magnetic hysteresis loops. Furthermore, since Ta and W resistivity values are phase-dependent, they can predict the phase of the Ta-W alloy. We measured that the resistivity of CoFeB is 168 $\mu\Omega\cdot$cm. The MgO and Ta capping layers were assumed to be insulating. We calculated the resistivity value of the NM layer using the parallel circuit model. The $\rho_{xx}$ values of our samples were in the range of 150–180 $\mu\Omega\cdot$cm for most thicknesses and concentrations, as depicted in figures 4(a) and (b). However, there were cases where the $\rho_{xx}$ value was less than 100 $\mu\Omega\cdot$cm. For example, when the Ta-W alloy was 4 nm thick with the Ta underlayer, the $\rho_{xx}$ values of the W-rich region were less than 100 $\mu\Omega\cdot$cm. Even in the pure W with the Ta underlayer, the resistivity dropped to 67 $\mu\Omega\cdot$cm. Also, when the...
pure W was 2 nm with the Ta underlayer, the resistivity value was about 90 $\mu\Omega\cdot\text{cm}$. With the W underlayer, when the W concentrations of the alloy are 40 and 60 at%, the $\rho_{xx}$ value was less than 100 $\mu\Omega\cdot\text{cm}$. Thus, we thought that the phase of the Ta-W alloy had the $\alpha$-phase at the point where the $\rho_{xx}$ value was lower than 100 $\mu\Omega\cdot\text{cm}$.

Figures 4(c) and (d) display the changes in $K_{\text{eff}}$ values as a function of W concentration for different alloy thicknesses and the choice of the underlayer. The full magnetic hysteresis loops are shown in Supplementary material S2. The $K_{\text{eff}}$ value was negative because a 1.5 nm thick CoFeB had the in-plane magnetic anisotropy. With the Ta underlayer, the calculated $K_{\text{eff}}$ value reached the minimum ($4.5 \times 10^{6}$ erg cm$^{-3}$) at W 85 at% when the alloy thickness was 4 nm. However, as the W concentration decreased, the $K_{\text{eff}}$ reached zero. With the W underlayer, the overall change in $K_{\text{eff}}$ values was similar to that of the Ta underlayer. The minimum value of $K_{\text{eff}}$ was $4.7 \times 10^{6}$ erg cm$^{-3}$ at W 90 at%.

Figure 5 shows the GI-XRD patterns for the samples with 4 nm thick Ta-W alloys with different underlayers, concentrations, and annealing conditions. Despite the presence of a 3 nm thick Ta or W underlayer, XRD intensities were very low when the Ta-W layer thickness was 2 nm (Supplementary material S3). Therefore, the phase of the 3 nm underlayers and the 2 nm Ta-W alloy was amorphous.

As shown in figure 5(a), we could not find any $\beta$-phase peaks. For both as-deposited and 300 °C annealed samples, the XRD patterns were almost identical with $\alpha$-phases only. Note that the annealed samples on Ta underlayers with W 20 and 40 at% were amorphous, and those with 60 and 85 at% were $\alpha$-phase, as shown in figure S4(a). As the concentration of W decreased, the peaks of the $\alpha$-W slightly shifted to smaller angles. The peak shift in the XRD patterns means that the interplanar distance of the Ta-W alloy increases according to Bragg’s law. Since the atomic radius of Ta is larger than that of W, it is natural that the interplanar distance increased as the concentration of Ta increased.

Figure 5(b) shows characteristic patterns of $\beta$-phase for most samples with the W underlayers, meaning that the $\beta$-phase developed better on the W underlayer. For the annealed sample containing Ta-W alloy layer with W 80 at%, the intensities of the $\beta$-W patterns reduced, and the $\alpha$-W patterns developed. Note that the annealed samples consisting of Ta-W alloy layers with W 20 at% were amorphous, while those with W 40 and 60% were $\alpha$-phase, as shown in figure S4(b). To supplement the XRD data, we conducted a transmission electron
microscopy (TEM) study. Figure S5 in Supplementary material shows the high-resolution cross-sectional TEM image of the sample with W 20 at%. This image also suggests that the alloy phase was amorphous. The failure to form the β-W structure is presumed to be the difference in the microstructure of the underlayers or the accumulated residual stress. The XRD analysis suggests that the concentration and underlayer material can control the crystal phase transition.

We can describe that the thickness and composition-dependent DLT behavior, particularly for samples shown in figure 3(b), is related to the crystal phase of the Ta-W alloy. A previous study reported the intrinsic spin Hall conductivity (SHC) values of α-W, β-W, and β-Ta12.5W87.5 by first-principles electronic structure calculations [26]. In particular, it was found that a shift of the Fermi level in β-Ta12.5W87.5 synergistically enhanced SHC and reduced the longitudinal conductivity. Furthermore, a previous study reported the SHC of various materials with A15 structure type (β-W structure type) using density functional theory [33]. The authors also calculated the largest SHC from the W2Ta. Both theoretical reports have in common that the SHC of Ta-W alloy increased from that of pristine-W when Ta was inserted while maintaining the β-W structure. Also, other studies reported the torque efficiency of W95Ta15 and W91.5Ta8.5 alloy by using ST-FMR [34, 35]. Thus, our results are in close agreement with previous reports.

Furthermore, our results expanded the range of W concentration which maintained the β-W phase and high ξDL value. From the materials science and spintronics viewpoint, determining the efficiency of the SOTs in various sample structures and measurements is important for understanding the origin of SOT. Therefore, we think the sample consisting of a 4 nm thick Ta15W85 layer on W underlayer was thick enough to develop the β-phase resulting in an abrupt increase in the ξDL value.

4. Conclusions

This study investigated the SOT changes in junctions consisting of Ta-W alloy layers with various thicknesses and concentrations with either the Ta or W underlayer. We performed magnetic, electrical, and crystal phase characterizations. With the W underlayer, the DLT values significantly changed with the thickness and concentration of the Ta-W alloy. Unlike the DLT values, the FLT values changed slightly. The DLT values changed significantly due to the phase transition of the Ta-W alloy. We found the largest DLT value of −23% in the sample with a 4 nm thick β-phase Ta15W85 at% alloy on the W underlayer, about 1.3 times larger than the value of −17% obtained from the obtained value of the sample with β-W layer only. When the α-phase appeared in the Ta-W alloy, the DLT value decreased. Therefore, we can control the crystal phase of the Ta-W alloy by choosing the right underlayer and acquiring the DLT values larger than that of pure β-W.

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Data availability statement

All data that support the findings of this study are included within the article (and any supplementary files).

Conflicts of interest

The authors declare no competing financial interests.

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