Effect of a CoFeB layer on the anisotropic magnetoresistance of Ta/CoFeB/MgO/NiFe/MgO/CoFeB/Ta films

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Abstract

The anisotropic magnetoresistance (AMR) and magnetic properties of NiFe films can be remarkably enhanced via CoFeB encapsulation. In the case of an ultrathin NiFe film having a Ta/CoFeB/MgO/NiFe/MgO/CoFeB/Ta structure, the CoFeB/MgO layers suppressed the formation of magnetic dead layers and the interdiffusions and interface reactions between the NiFe and Ta layers. The AMR reached a maximum value of 3.56% at 450 °C. More importantly, a single NiFe(111) peak can be formed resulting in higher AMR values for films having CoFeB encapsulation. This enhanced AMR also originated from the significant specular reflection of electrons owing to the crystalline MgO layer, together with the sharp interfaces with the NiFe layer. These factors together resulted in higher AMR and improved magnetic properties.

Keywords: Anisotropic magnetoresistance (AMR); NiFe; CoFeB; Annealing; X-ray diffraction (XRD); **High-resolution transmission electron microscopy (HRTEM).**

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1. Introduction

In the last few decades, plenty of work has been devoted to anisotropic magnetoresistance (AMR) thin films, which have promising applications as magnetic sensors and detectors. AMR sensors have vast potential for information processing, automation, aerospace, navigation and military applications[1–6]. AMR sensors have attracted much attention owing to their high sensitivity, small volume and excellent reliability characteristics. AMR thin films can be deposited on silicon substrates, being directly embedded in integrated circuits. Thus, an integrated assembly of magnetic sensors and other circuit units can be achieved. AMR sensors are also easily matched with an electrical circuit compared with other sensors such as Hall devices and semiconductor magnetoresistors.

NiFe (Permalloy) is the best material for use in AMR devices due to its large AMR amplitude and magnetic softness. The typical sensing component contains a Ta/NiFe/Ta structure for desired application performance. A thickness of 10 nm or less is desirable for NiFe films to diminish the demagnetisation effect. However, the AMR decreases rapidly with decreased NiFe thickness[7]. For Ta/NiFe (10 nm)/Ta films, the AMR is about 1.0%–2.0%, depending on the experimental conditions. How to best enhance the AMR in NiFe films with good thermal stability has become an important issue. Liu measured an AMR ratio of 2.92% for the Ta/NiFe (15 nm)/Ta trilayer. However, for Ta/Pt/NiFe (15 nm)/Pt/Ta films, the AMR ratio reaches a maximum value of about 3.21% at a Pt layer thickness of 0.15 nm[8]. Ding reported that the AMR reaches a maximum value of 3.10% after 380 °C annealing for Al₂O₃ encapsulated in a 10-nm NiFe film[9]. Wang reported the AMR value to reach a maximum of 3.5% for the Ta/NiFe (15 nm)/Ta trilayer with a substrate temperature of 400 °C[10]

High AMR values and thin thicknesses are both necessary for the practical applications of NiFe films. Permalloy properties can be optimized by appropriate buffer layers and annealing, although the annealing process may lead to the interdiffusions and interface reactions of thin films, thereby damaging their properties [11,12]. In the case of Ta/NiFe/Ta films, magnetic dead layers are present at the Ta/NiFe and NiFe/Ta interfaces, decreasing the effective thickness of the NiFe layer and thus resulting in poor properties [13]. As reported by

Ding, the AMR of NiFe films with Ta/MgO/NiFe/MgO/Ta structure can be remarkably enhanced via MgO encapsulation. The MgO layers suppressed the formation of the magnetic dead layers and prevented the interdiffusions between the NiFe and Ta layers. The higher magnetoresistance (MR) ratio and sensitivity originated from the significant specular reflection of electrons owing to the crystalline MgO layer, together with the sharp interfaces between the MgO/NiFe layers[14]. Bae's group reported that MgO is poly-crystalline when as deposited in CoFeB/MgO/CoFeB structures. After annealing at 340 °C, it exhibits a better crystallised state with fairly smooth interfaces between MgO and CoFeB [15]. All the works described above imply that MgO deposited on CoFeB would significantly enhance the crystallization of the former while facilitating the formation of smooth interfaces. This work was aimed to enhance the crystallization of MgO by inserting a CoFeB layer to increase the AMR and magnetic properties of NiFe thin films. The structure of the film was Ta(5 nm)/CoFeB(*t*)/MgO(3 nm)/NiFe(10 nm)/MgO(3 nm)/CoFeB(*t*)/Ta(3 nm).

2. Materials and methods

Samples with a Ta(5 nm)/CoFeB(*t*)/MgO(3 nm)/NiFe(10 nm)/MgO(3 nm)/CoFeB(*t*)/Ta(3 nm) structure were deposited onto corning glass substrates of 22 × 22 mm² via magnetron sputtering at a base pressure of 3.0×10^{-8} Torr and a sputtering pressure of 3 mTorr of 99.99% Ar gas, with *t* varying from 0 to 3 nm. The substrates were rotated at 40 rpm during deposition to ensure uniform film thickness. A magnetic field of 25 kA/m was applied parallel to the film planes to induce NiFe easy axis during deposition. Ta, NiFe and CoFeB were deposited by DC sputtering, whereas MgO was deposited via RF sputtering. After deposition, the samples were annealed at different temperatures (400 and 450 °C) for 1 h and 2 h in a vacuum furnace without external magnetic field under a pressure of 4×10^{-7} Torr. The MR of the thin films was measured by the four-probe method (probe current = 1 mA). The magnetic properties were measured by a physical properties measuring system (PPMS), and the structure of the samples was characterized by X-ray diffraction (XRD) and the high-resolution transmission electron microscopy (HRTEM).

Fig. 1. The dependence of CoFeB thickness dependence on MR for Ta(5 nm)/CoFeB(t)/MgO(3 nm)/NiFe(10 nm)/MgO(3 nm)/CoFeB(t)/Ta(3 nm) before and after the annealing .

3. Results and discussion

Figure 1 shows MR of Ta(5 nm)/CoFeB(t)/MgO(3 nm)/NiFe(10 nm)/MgO(3 nm)/CoFeB(t)/Ta(3 nm) thin films as a function of the CoFeB thickness at different annealing temperatures. In the case of the as-deposited film, MR remained nearly constant (ca. 1.8%) with the CoFeB thickness. However, it slightly decreased at a CoFeB layer thickness of 2.5 nm. Thus, the CoFeB layer did not have a significant effect on the MR of the as-deposited films. After annealing at 400 and 450 °C at varying times (1 h and 2 h), the MR ratio of the thin films increased significantly. Maximum MR values were achieved for films having CoFeB layers of 1.5 nm in thickness, and a further increase in the CoFeB layer thickness resulted in lower MR. For example, MR ratio peaked at 3.56% for the film annealed at 450 °C, with this value being 30% higher than the maximum MR of the film annealed at 450 °C without CoFeB layer (2.74%). These results also indicate that a proper increase of the annealing duration may result in higher MR ratio values of thin films. In particular, the existence of a CoFeB layer after annealing apparently enhanced the MR values for Ta/CoFeB/MgO/NiFe/MgO/CoFeB/Ta thin films.

According to other reports, maximum MR values have been achieved for

Ta/MgO/NiFe/MgO/Ta films annealed at 400 °C or 450 °C[14]. In our report, the AMR reaches a maximum value in Ta/MgO/NiFe/MgO/Ta films after annealing at 400 °C. However, the films with a CoFeB layer have a higher AMR than films without a CoFeB layer, and they are thermally stable up to 450 °C. When the annealing duration is further increased from 1 h to 2 h, the AMR values show just a slight increase for films annealed at 400 °C. This result shows that the annealing duration has little influence when the duration is extended for longer than 1 hour. In contrast, the MR ratio with the Ta/NiFe/Ta structure decreased dramatically with an increase in annealing temperature.

Fig. 2. MR–H curves of Ta(5 nm)/CoFeB(t)/MgO(3 nm)/NiFe(10 nm)/MgO(3 nm)/CoFeB(t)/Ta(3 nm) thin films with CoFeB thicknesses of 1.5 and 3 nm after annealing at 400 °C for 2 h.

Figure 2 shows the MR–H curves of thin films having CoFeB layers of 1.5 and 3 nm after annealing at 400 °C for 2 h. The maximum MR ratio reached 3.56% for the film with a 1.5-nm CoFeB layer, representing an increase of 30% compared to the film without CoFeB (2.74%). An increase in the CoFeB layer thickness to 3 nm resulted in a lower MR ratio

(2.90%) and an MR–H curve with double peaks indicative of thin films with large coercivities (H_c). The saturation field (H_s) is about 5 Oe for the sample with a 1.5-nm CoFeB layer. In contrast, the sample with a 3-nm CoFeB layer exhibited poor anisotropy, with 15 Oe H_s . These results show that the existence of the CoFeB layer enhanced the MR ratio of the thin films up to a certain thickness value above which this parameter and the magnetic properties and MR ratio of the thin films were negatively affected.

Fig. 3. MR ratio as a function of the CoFeB layer thickness for Ta(5 nm)/CoFeB(t)/Ta(3 nm) thin films annealed at 400 °C for 2 h.

Considering that CoFeB is a type of ferromagnetic material, further measurements are needed to exclude the possible contribution of CoFeB in the overall MR ratio of Ta/CoFeB/MgO/NiFe/MgO/CoFeB/Ta thin films. Figure 3 shows the MR ratio as a function of the CoFeB layer thickness for Ta(5 nm)/CoFeB(t)/Ta(3 nm) thin films annealed at 400 °C for 2 h. The MR ratio increased with the CoFeB thickness and levelled off (0.16%) at CoFeB thicknesses higher than 14 nm. MR ratios of 0.07% and 0.14% were obtained for thicknesses of 3 and 6 nm, respectively. The contribution of the CoFeB layer to the overall MR ratio of the Ta/CoFeB/MgO/NiFe/MgO/CoFeB/Ta thin films was <0.2% and even lower (0.1%) for CoFeB layers with a thickness of 3 nm. The MR ratio for Ta/MgO/CoFeB/Ta structure has been measured. The MR is 0.24~0.40%, which is a slightly higher than that of the

Ta/CoFeB/Ta since the MgO is non-magnetic material. As a result, the direct contribution of CoFeB/MgO layers to the overall MR ratio of the thin films can be considered to be negligible.

Fig. 4. Hysteresis loop of thin films with a 1.5-nm CoFeB insertion layer after annealing at 400 °C for 2 h.

Fig. 5. Hysteresis loop of thin films with a 3-nm CoFeB layer after annealing at 400 °C for 2 h.

The hysteresis loops of the samples after annealing were measured by PPMS to study the

influence of the CoFeB layer on the magnetic properties of Ta/CoFeB/MgO/NiFe/MgO/CoFeB/Ta thin films. Figures 4 and 5 show the hysteresis loops of thin films having 1.5- and 3-nm CoFeB layers after annealing at 400 °C for 2 h. The sample with a 1.5-nm CoFeB layer maintained its soft magnetic and uniaxial anisotropic performance after annealing at 400 °C for 2 h, with H_c being <2 Oe and a saturation field (H_s) being <5 Oe. In contrast, the sample having a 3-nm CoFeB layer showed poor anisotropy after annealing at 400 °C for 2 h, with $H_c = 5$ Oe and H_s being <15 Oe. Notably, two magnetic phases were observed in the hysteresis loop (easy axis) of the latter sample. This may be attributed to the total thickness of the CoFeB layer (6 nm) being of the same order of magnitude as the thickness of the NiFe layer (10 nm), with CoFeB contributing significantly to the overall magnetism of the thin films. CoFeB has different magnetic properties compared with NiFe. Thus, CoFeB forms another magnetic phase with different types of magnetic domains, thereby destroying the uniaxial anisotropy of the material. As a result, H_c , H_s and the uniaxial anisotropy (which are stable at a CoFeB thickness of 1.5 nm) rapidly deteriorated when the CoFeB thickness reached 3 nm, degrading the MR properties and the MR–H curve of the thin films.

Fig. 6. XRD patterns of Ta(5 nm)/CoFeB(t)/MgO(3 nm)/NiFe(10 nm)/MgO(3 nm)/CoFeB(t)/Ta(3 nm) thin films after annealing.

Structural analysis of the films while varying the CoFeB thickness was performed by XRD to gain insight into the reasons behind the enhancement of AMR. Figure 6 shows the XRD patterns of Ta(5 nm)/CoFeB(*t*)/MgO(3 nm)/NiFe(10 nm)/MgO(3 nm)/CoFeB(*t*)/Ta(3 nm) thin films after annealing at 400 °C. No apparent peaks were found in the film without CoFeB. A single NiFe(111) peak was observed for all the films having a CoFeB layer, and the intensity of this peak remained nearly unchanged with the CoFeB thickness. This result indicates that the NiFe texture hardly formed in the Ta/MgO/NiFe/MgO/Ta films even after annealing. In contrast, an intense NiFe(111) peak was observed in the XRD patterns of the Ta/CoFeB/MgO/NiFe/MgO/CoFeB/Ta films after annealing. The intensity of this peak increased with the duration of the annealing process. These results suggest that the NiFe(111) texture easily formed in the Ta/CoFeB/MgO/NiFe/MgO/CoFeB/Ta films, which is crucial in optimizing the MR ratio of thin films.

Ta was used as a buffer layer to induce NiFe with (111) texture. A single NiFe (111) peak can be observed over the entire annealing range in Ta/NiFe/Ta films, and the intensity of the NiFe peak changes very little with increased annealing temperature. However, no NiFe peaks are found for as deposited Ta/MgO/NiFe/MgO/Ta multilayers. When films are annealed at low temperatures, the NiFe(111) peak is weak and changes only slightly with increased annealing temperature. A small but strong NiFe (111) peak occurs when Ta is higher than 400 °[14]. In contrast, an intense NiFe(111) peak is observed in the X-ray powder diffraction (XRD) patterns of thin films containing CoFeB after annealing at 400 °C in Ta/CoFeB/MgO/NiFe/MgO/CoFeB/Ta films. Lee observed high thermal stability and AMR without high-temperature deposition or annealing following the use of a thin NiFeCr seed layer in NiFe films[16]. XRD results show that the NiFeCr seed layer causes the formation of large (111) textured grains in Permalloy film, and that the interface between these two layers is quite smooth. Other research has also reported that the stronger is the NiFe(111) texture, the higher is the AMR value[9].

Fig.7 (A) HRTEM images of the Ta (5 nm)/NiFe (5 nm)/Ta (3 nm)

(B) HRTEM images of the MgO (3 nm)/NiFe (5 nm)/MgO (3 nm)/Ta (3 nm) as deposited (a), enlarged figure of part 1 (b), fast Fourier transforms of part 3, enlarged figure of part 2 (d)

(C) HRTEM images of the MgO (3 nm)/NiFe (5 nm)/MgO (3 nm)/Ta (3 nm) annealed (a), enlarged figure for parts 1 and 2 (b).

(D) HRTEM images of the Ta(5 nm)/CoFeB(1 nm)/MgO(3 nm)/Ta(3 nm) annealed

Fig.7(A) shows HRTEM images of Ta (5 nm)/NiFe (5 nm)/Ta (3 nm). However, neither the Ta/NiFe interface nor the NiFe/Ta interface is obvious. Significant interface diffusions occurred between the NiFe/Ta and Ta/NiFe. Fig.7(B) and 7(C) show HRTEM images of MgO (3 nm)/NiFe (5 nm)/MgO (3 nm)/Ta (3 nm) before and after annealing. In the ‘as-grown’ sample, we see crystalline fringes in the MgO layers only in the small areas shown in Fig. 7(B). After annealing, both the top and bottom MgO layers exhibit a much better crystalline

structure, as shown in Fig. 7(C). However, the CoFeB inserted at the interface of the Ta/MgO and MgO layers exhibits an fcc (100) of MgO textured in Ta (5 nm)/CoFeB (1 nm)/MgO (3 nm)/Ta (3 nm) film in Fig. 7(D). We can observe a smooth and clear CoFeB/MgO interface. Smooth oxide/metal interfaces play a significant role in the scattering of spin-polarised electrons compared to Ta/NiFe/Ta interfaces.

HRTEM images of CoFeB/MgO/CoFeB as-deposited and after annealing were reported by Bae et al. The bottom CoFeB was found to be poly-crystalline in the as-deposited state, crystallizing after annealing at 340 °C. The barrier interface of the MgO barrier became sharper and smoother after annealing. HRTEM was also used to study Ta/CoFeB/MgO thin films before and after annealing at 300 °C, and a MgO layer was found to texture along the [001] direction [15]. HRTEM revealed a smooth, uniform and highly crystalline MgO barrier in CoFeB/MgO/CoFeB/Ru/CoFe/PtMn multilayers [17], and similar results were obtained by other researchers [18–21]. Magnetic tunnel junctions based on a single crystalline MgO barrier with a high tunnel magnetoresistance (TMR) ratio have been extensively studied with respect to the CoFeB/MgO/CoFeB structure[22-23]. The high TMR ratio is due to coherent spin-dependent tunnelling across the single crystalline MgO barrier. The sharp interfaces across the MgO layer and its single crystallinity likely play much greater roles in specular reflection. The higher AMR ratio and sensitivity come from the significant specular reflection of electrons at both interfaces due to the crystalline MgO layer together with the sharp interfaces of the Ta/MgO/NiFe/MgO/Ta. All the work described above concluded that MgO deposited on CoFeB significantly enhanced the crystallization of MgO and the formation of smooth interfaces. A MgO layer with good crystallinity and (100) orientation could significantly reduce the defects. In other words, the specular reflection of the electrons can be significantly enhanced, resulting in higher MR values. Moreover, the lattice of the MgO and NiFe layers matched well in some areas, with the MgO/NiFe and NiFe/MgO interfaces becoming significantly smoother after annealing [14]. The sharp interface played a critical role in the specular reflections of electrons compared to the as-grown sample. Based on the above results, it is reasonable to conclude that CoFeB significantly enhanced the crystallization of MgO. This crystalline MgO layer,

together with its sharp interface with the NiFe layer, resulted in an enhanced specular reflection of electrons, finally resulting in significantly higher MR ratios for Ta/CoFeB/MgO/NiFe/MgO/CoFeB/Ta multilayers.

In Ta/NiFe/Ta films, magnetic dead layers occur at the Ta/NiFe and NiFe/Ta interfaces, which decrease the effective thickness of the NiFe layer [13]. As reported by Ding, the AMR of NiFe films with a Ta/MgO/NiFe/MgO/Ta structure can be significantly enhanced by the MgO layer. The MgO layers suppress the formation of magnetic dead layers and prevent interdiffusions between the NiFe and Ta layers [14]. In this letter, we report our results from a study of the effect of CoFeB insertion layers in Ta/CoFeB/MgO/NiFe/MgO/CoFeB/Ta. We found the CoFeB/MgO layers to effectively eliminate the formation of interfacial magnetic dead layer and to suppress the thermally driven interdiffusion between Ta and NiFe.

4. Conclusions

Ta/CoFeB/MgO/NiFe/MgO/CoFeB/Ta thin films were prepared and annealed. CoFeB layers with a thickness of 1.5 nm in thin films containing a 10-nm NiFe layer significantly enhanced the AMR after annealing at 400 °C for 2 h (i.e. the MR ratio increased from 2.74% to 3.56%). The hysteresis loops revealed fine soft magnetism and uniaxial anisotropy in the thin films, with H_c and H_s reaching 2 and 5 Oe, respectively. The CoFeB/MgO layers suppressed the formation of the magnetic dead layers and increased the effective thickness of NiFe. More importantly, the single NiFe(111) peak observed in the XRD profiles of thin films containing CoFeB was indicative of higher AMR values. XRD revealed that the NiFe(111) texture easily formed in the Ta/CoFeB/MgO/NiFe/MgO/CoFeB/Ta films, resulting in thin films with enhanced MR and magnetic properties. CoFeB significantly enhanced the crystallization of MgO, and this crystalline MgO layer, together with its sharp interface with the NiFe layer, resulted in a significant specular reflection of electrons. All these factors together resulted in higher AMR values for the studied thin films.

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