MAGNETISATION PROCESS IN Fe/TERFECOHAN/Fe SANDWICH FILMS

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Abstract: Magnetostrictive sandwich films in the type of Fe/Tb($Fe_{0.55}Co_{0.45}$)$_{1.5}$/Fe (denoted as Fe/Terfecohan/Fe) in which a fixed individual TerfecoHan-layer thickness t_{TbFeCo} = 600 nm and variable Fe layer thickness t_{Fe} = 0, 5, 10, 15, 60 and 90 nm (named as S0, S5, S10, S15, S60 and S 90, respectively) have been prepared by mean of a RF-sputtering. The Terfecohan layers in the as-deposited films exhibit a perpendicular magnetic anisotropy, whereas the Fe layers show a parallel one. In this state, however, the coercivity is low ($\mu_o H_C$ ~ 10 mT). Heat treatments at $T_a \leq 350$ °C destroy the perpendicular anisotropy and improve the magnetic softness. At T_a = 450 °C and 500 °C, the perpendicular anisotropy is reestablished in the samples S10, S15 and S20 and the coercivity is enhanced up to 490 mT. These materials are rather favorable for making films as a recording medium. In the samples S40 and S60, the formation of the magnetic domain structure in the interface is observed, leading to magnetic phase transition from the interlayer ferrimagnetic configurations to the interlayer ferromagnetic ones. These findings are discussed in terms on the basic of the characteristic lengths for the surface effects and for the formation of the domain structure.

1. Introduction

The thin film technology is presently in a state of extremely rapid development. One of interests are giant magnetostrictive thin films which directly transform electrical into mechanical energy especially their cost-effective mass production is possible, compatible to microsystem process technologies. Due to their special features, these giant magnetostrictive films are very application potential for microelectromechanical system (MEMS). Research on magnetostrictive films concentrates on developing low field giant magnetostriction, i.e. reducing the necessary driving magnetic fields. Since achieving this aim will allow actuation of microactuators by integrated microcoils or by remote control.

As a tradition, research of giant magnetostrictive thin films has been based on amorphous (a) rare earth - iron alloys, in particular a-$Tb_{0.27}Dy_{0.73}Fe_2$ (known as a-Terfenol-D) [1,2]. Practically, a record magnetostriction of 1020×10^{-6} has however been achieved on the a-Tb($Fe_{0.55}Co_{0.45}$)$_{2.1}$ thin film [3]. The giant magnetostriction observed in these materials was explained by the enhancement of the 4f-3d exchange, leading to the diminishing of the Tb-sperimagnetic cone-angle.

In order to develop further giant magnetostriction at low magnetic fields, Duc *et al.* [2,3] have prepared successfully the *a-TerfecoHan* compound with constituent of $Tb(Fe_{0.55}Co_{0.45})_{1.5}$. This Tb-rich compound exhibits not only a huge saturation magnetostriction $\lambda_s \sim 10^{-3}$ but also a good magnetic sofness: the (parallel) magnetostrictive susceptibility ($\chi_{\lambda//} = \partial\lambda_{//}/\partial(\mu_0 H)$) of 1.8×10^{-2} T^{-1} is achieved at $\mu_0 H$ = 10 mT. A further enhancement of low-field magnetostrictive susceptibility can be obtained in sandwich Fe/Terfecohan/Fe films, which combine layers with a large room-temperature magnetostriction (Terfecohan) and soft magnetic layers with a high magnetization (Fe). Indeed, a very large magnetostrictive susceptibility $\chi_{\lambda//} = 7.7\times10^{-2}$ T^{-1} was observed at rather low field $\mu_0 H$ = 8 mT in annealed sandwiches. This was reported that these Fe layers played an important role in proving the magnetostrictive softness of the Fe/Terfecohan/Fe sandwiches [5].

In this paper, we present the results of our investigation on the dependence of the magnetisation processes and coercivity on the Fe layers as well as heat treatment process. Among them, the formation of the extended domain wall (EDW) at the interface is also reported.

2. Experimental

The Fe/TerfecoHan/Fe sandwiches with a fixed individual TerfecoHan-layer thickness t_{TbFeCo} = 600 nm and a variable Fe layer thickness t_{Fe} = 0, 5, 10, 15, 20, 40, 60 and 90 nm (denoted as the samples S0, S5, S10, S15, S20, S40, S60 and S90, respectively) were prepared by rf-magnetron sputtering. The Terfecohan layer was sputtered under a power of 200 W, while the sputtering power of the Fe-layers was 100 W. A composite target has been used for the Terfecohan layer, a high-purity metal plate for the Fe layers. The substrates were glass microscope cover slips. Both target and sample holder was water-cooled. Samples were annealed in the temperature range from T_a = 250 °C to 500 °C for one hour in a vacuum of 5×10^{-5} mbar.

Film structure was investigated with X-ray diffraction (XRD) and high-resolution transmission electron microscope (HRTEM) obtained results were partly reported in [4]. The room temperature magnetic hysteresis loops were measured in magnetic fields applied parallel and perpendicular to the film plane using a vibrating sample magnetometer (VSM) in magnetic fields up to 1.4 T.

3. Experimental results and discussions

Fig. 1 illustrates several typical magnetic hysteresis loops for the as-deposited samples S0, S15 and S60. The analysis shows that the in-plane magnetisation requires a magnetic field higher than 0.35 T to saturate and the remanence is almost zero (e.g. in samples S0, S10 and S15). These features suggest the existence of a perpendicular anisotropy, in addition to the usual shape anisotropy. For sample S60, the magnetisation seems to consist of both perpendicular and parallel magnetic components. Indeed, it is found from Mössbauer studies [5] that, in this sample, the Fe-magnetic moments in

Terfecohan layer orient along the out-of-plane direction, whereas those in the Fe-layers lie in the film plane.

Fig. 2 presents magnetic hysteresis loops for the samples after annealed at temperature T_a = 450 °C. Perpendicular anisotropy was destroyed and in-plane anisotropy was quickly established with the increasing in T_a in the sample with Fe-layer thickness $t_{Fe} \leq 5$ nm (see fig. 3a for samples S0 and S5).

For the samples with 5 nm $< t_{Fe} \leq$ 20 nm (e.g. S10, S15 and S20), however, the in-plane magnetic anisotropy can be formed with annealing at $T_a \leq$ 250 °C only. At higher-temperature annealing, one observes the re-establishment of the out-of-plane magnetic anisotropy (see fig. 2.b) with a rather large coercivity, $\mu_o H_C$ = 490 mT for sample S10.

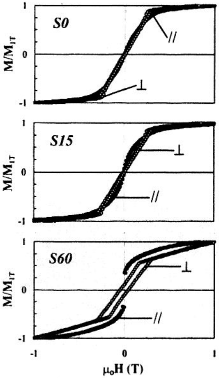

Fig. 1. Magnetic hysteresis loops of the as-deposited samples S0, S15 and S60, see also in [5]

In the case of $t_{Fe} \geq$ 40 nm, again the in-plane magnetic anisotropy is established. In addition, these samples exhibit also a field-induced magnetic transition at $\mu_o H \approx$ 100 mT and 200 mT for samples S40 and S60, respectively (fig. 2c). Such a magnetic behavior is usually observed in sandwich films made by stacking coupled layers with typical thickness around 100 nm [1]. It was associated with the different magnetisation reversal of individual layers, leading to the formation of the so-called extended domain wall (EDW) interfaces that will be described below.

Similar results are obtained for samples annealed at 500 °C.

Fig. 3 shows the variation of in-plane ($\mu_oH_{C//}$) and ($\mu_oH_{C\perp}$) perpendicular coercivity as a function of Fe-layer thickness (t_{Fe}) (see also in the table 1). It can be seen that the coercivity of samples with $t_{Fe} \leq 5$ nm is small and remains almost constant value as compared with as-deposited samples. Meanwhile, the enhancement of μ_oH_C with the increasing in T_a has been observed in samples with Fe-layer thickness range of 10 nm $\leq t_{Fe} \leq$ 20 nm. These findings propose different magnetization reversals for different Fe-layer thickness. It is well known that the coercivity is exclusively governed by the particle size and the stress induced in the sample preparation process. The heat treatment leads to the crystallisation, the growth of crystalline particles and then the enhancement of the coercivity. Meanwhile, the induced stress tends to be released by annealing process, so that it causes decreasing coercivity. Indeed, for samples S0 and S5, the Fe-layer thickness is too thin. In this case, it was indicated by cross section TEM micrographs [5] that a fine particle structure exists in both Terfecohan and Fe layers. For instance, particle size observed in the Terfecohan layer of sample S0 after depositing is of 3-5 nm. After annealing at 450 °C, it becomes 5-7 nm only. This explains small value of the coercivity observed in samples with $t_{Fe} \leq 5$ nm. In the case of 10 nm $\leq t_{Fe} \leq$ 20 nm, the annealing temperature results in the abruptly increasing the μ_oH_C. This may be attributed to the enhancement of the particle size in the Terfecohan layers: in the 450 °C-annealed sample S10, the fine-grain (3-4 nm size) structure still remains in the Fe-layers, while a grain size of the crystalline phase of about 25 nm was formed in the Terfecohan layer [5]. The observed different annealing effects could be attributed to the nucleus formed in Terfecohan/Fe interfaces.

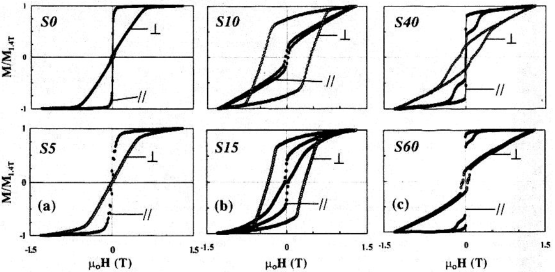

Fig. 2. Magnetic hysteresis loops of the 450 °C-annealed samples S0, S5 (a), S10, S15 (b), S40 and S60 (c)

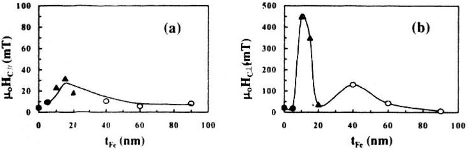

Fig. 3. In-plane (a) and perpendicular (b) coercivity of the 450 °C-annealed samples as a function of the Fe-layer thickness (t_{Fe}). Different symbols indicate different reversal mechanisms

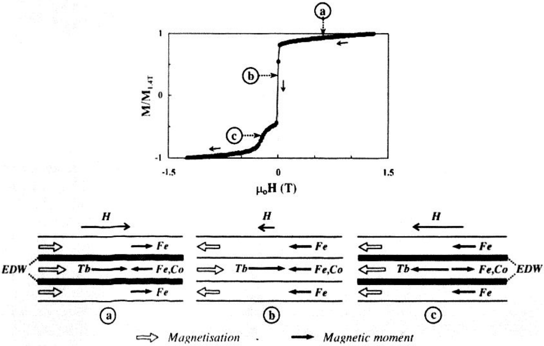

Fig. 4. Magnetisation reversal and EDW formation in sandwich films

The EDW formation is usually observed in sandwich and multiplayer films, in which the individual layer thicknesses are large enough ($t \geq 40$ nm). When reversal takes place in a given layer but not in the adjacent one, a domain wall will be formed at the interface between layers, in order to minimize exchange energy [1]. The magnetisation loop of 450 °C-annealed sample S40 and mechanism of EDW formation can be described as follows (see in fig. 4). Starting from the positive high-field state, where the system is well saturated in the applied field direction, we see that all the (Terfecohan, Fe) magnetisation components in the sample are parallel, but the Fe(Co) moments between layers are antiparallelly coupled. In this case, an EDW is formed at the interfaces (see fig. 4a). As the field is decreased and changes its direction, the Fe-layers (with a small coercivity) is reversed first and the EDWs are suppressed (fig. 4b) because the exchange energy becomes dominate that leads to parallel state of Fe(Co) moments between layers. At negative high-field (fig. 4c), the magnetisation of the Terfecohan layer rotates to the field direction, the magnetic saturation state is established and the EDWs are recreated.

Table 1. The values of in-plane ($\mu_o H_{C//}$) and perpendicular ($\mu_o H_{C\perp}$) coercivity for the 450 °C and 500 °C-annealed samples

Samples	T_a = 450 °C		T_a = 500 °C	
	$\mu_o H_{C//}$ (mT)	$\mu_o H_{C\perp}$ (mT)	$\mu_o H_{C//}$ (mT)	$\mu_o H_{C\perp}$ (mT)
S0	4	20	4.2	18
S5	9.2	18	13.8	2.8
S10	23	450	12	490
S15	32	350	22	420
S20	18.5	37	77	158
S40	10.5	130	11	92
S60	5.6	43	11	74
S90	8.2	5.7	9.2	1.5

4. Concluding remarks

Various magnetic behaviors have been observed in 450 °C-annealed Fe/Terfecohan/Fe sandwich films depending on Fe-layer thickness. The in-plane magnetic anisotropy and low coercivity were obtained in the samples with thin Fe-layer thickness $t_{Fe} \leq 5$ nm. The samples with 5 nm < $t_{Fe} \leq 20$ nm exhibit the out of plane magnetic anisotropy and high coercivity. Finally, in the samples with $t_{Fe} \geq 40$ nm, the formation of the Extended Domain Walls was observed. These phenomena can be attributed not only to magnetically characteristic lengths but also to the microstructure of the Terfecohan layer. It can also be emphasized that different Fe-layer thicknesses cause different driving forces to create nucleus at Terfecohan/Fe interfaces and then to different crystallisation processes.

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