Orientational dependence of the exchange biasing in Ni-Fe/Mn1-xNix (x=0.11, 0.18, 0.23) bilayers

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Orientational dependence of the exchange biasing in Ni–Fe/$Mn_{1-x}Ni_x$ ($x=0.11, 0.18, 0.23$) bilayers

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The correlation between the exchange anisotropy of ferromagnetic (F)/antiferromagnetic (AF) bilayers and the spin alignment of antiferromagnet at the interface is still under question. In the present study, $Mn_{1-x}Ni_x$ ($x=0.11$,0.18,0.23) alloys, whose spin structure changes with x, were used for Ni–Fe/AF bilayers. The correlation between AF spin alignment at the interface and exchange anisotropy was investigated by controlling the preferred orientation of the bilayers. The crystal orientation of bilayers was changed by adding nitrogen into the sputtering argon gas only during the deposition of the Ni–Fe layer adjacent to the substrate. The preferred orientation was (111) when the nitrogen partial pressure, $P_{N_2}=0\%-1\%$; (100) when $P_{N_2}>1\%$. When $P_{N_2}<1\%$, J_k decreased and H_c increased with increasing P_{N_2}, because of the decrease of the grain size of γ-Mn–Ni. When P_{N_2} exceeds 1%, J_k increased discontinuously, accompanied by the orientational change of the bilayers. © *1999 American Institute of Physics.* [S0021-8979(99)63908-1]

I. INTRODUCTION

The spin alignment of the antiferromagnet at the interface plays an important role in the exchange anisotropy of ferromagnetic (F)/antiferromagnetic (AF) bilayers. Although several studies have been reported, such as micromagnetic calculations for the compensated F/AF interface[1] and experimental investigations for the F/AF bilayers in different preferred orientations,[2–7] the correlation between the spin alignment of the antiferromagnet at the interface and the exchange coupling has still not been clarified.

γ-$Mn_{1-x}Ni_x$ alloys are suitable antiferromagnets to investigate this correlation. The crystal structure is fct for $x<0.13$ and $0.18<x<0.21$, and fcc for $x>0.21$.[8] When $x<0.13(0.18<x<0.21)$, spins align ferromagnetically in a c plane (an a plane) and collinear to a c axis (an a axis).[9] When $x>0.21$, spins form a $\langle 111\rangle$ canted structure similarly to FeMn.[10,11] Therefore, $Mn_{1-x}Ni_x$ films with different crystal orientation present various AF spin alignments at the F/AF interfaces, when they are deposited on the F layer, and this work examines the dependence of the exchange anisotropy on the AF spin alignment.

In the present study, Ni–Fe/$Mn_{1-x}Ni_x$ ($x=0.11$,0.18,0.23) films with (111) or (100) orientation were fabricated and the exchange anisotropy of these films was investigated in connection with their crystal orientation and microstructure.

II. EXPERIMENTAL PROCEDURE

Substrate/Ni–Fe 200 Å/$Mn_{1-x}Ni_x$ 1000 Å/Ni–Fe 300 Å trilayered films were deposited on Si (100) wafers at room temperature by a facing-targets dc sputtering method. The Ni content x of the Mn–Ni layer was either 0.11, 0.18, or 0.23 and was chosen by changing the composition of alloyed targets. A magnetic field of 30 Oe was applied parallel to the substrate surface during deposition. The top Ni–Fe layer was deposited to protect the Mn–Ni layer from oxidation. In this paper we are concerned only with the Ni–Fe layer next to the substrate in the as-deposited films. The argon pressure used during film deposition was 1 mTorr. In order to change the crystal orientation of the Ni–Fe films, up to 7% partial pressure of nitrogen (P_{N_2}) was mixed into argon gas, only during deposition of the Ni–Fe layer on the substrate side. The MH loop was measured by a VSM at room temperature (RT) and a SQUID magnetometer below RT to 10 K. A unidirectional anisotropy constant, J_k was calculated as $M_s d_F H_{ex}$. Here $M_s d_F$ is the saturation magnetization of each F layer per unit area, which was determined from the MH loop. H_{ex} is the exchange coupling field, determined by the shift of the center of MH loops. The coercivity, H_c, of the exchange biased Ni–Fe layer was defined by the half width of a shifted MH loop. Microstructure and crystal orientation of films were examined by x-ray diffraction (XRD) with a Co $K\alpha$ radiation source.

III. RESULTS AND DISCUSSION

A. Structure

Figure 1 shows changes in XRD profiles of trilayers with $x=0.18$ fabricated under the various nitrogen partial pressures. 2θ angles of diffraction lines calculated from the lattice spacing of bulk γ-Mn–Ni[8] are indicated in broken lines; $2\theta=49.5°$ for (111), 57.3° for (002), 58.1° for (200). 2θ angles for Ni–Fe are also shown; $2\theta=51.7°$ for (111), 59.7° for (200). For $P_{N_2}<1\%$, the preferred orientation of trilayered films was confirmed to be (111), because only diffraction lines from γ-Mn–Ni (111) and Ni–Fe (111) planes were observed. For $P_{N_2}>1\%$, only diffraction lines from γ-Mn–Ni (200) or (002) and Ni–Fe (200) planes were observed. These results indicate that the crystal orientation of Mn–Ni grains was changed by tracing that of underlaid Ni–Fe grains, which was changed from (111) to (100) by adding the nitrogen.[12] Similarly, the preferred orientation

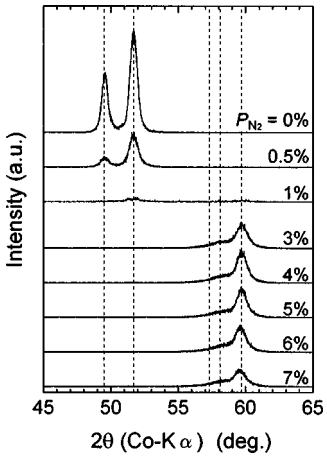

FIG. 1. Changes in XRD profiles of Ni–Fe/$Mn_{0.82}Ni_{0.18}$/Ni–Fe films fabricated under the various partial pressure of nitrogen, P_{N_2}.

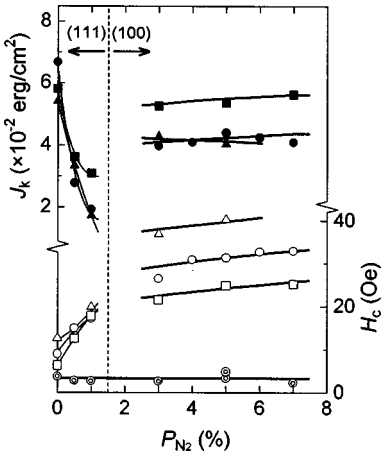

FIG. 3. Changes of unidirectional anisotropy constant, J_k (filled marks), and coercive force, H_c (open marks) of trilayered films measured at room temperature as a function of P_{N_2}. Ni content x of Mn–Ni is 0.11 (triangle), 0.18 (circle), and 0.23 (square). H_c of single layered Ni–Fe film deposited by same method as trilayers is also plotted (double circle).

was also observed to be (111) for $P_{N_2}<1\%$, and (100) for $P_{N_2}>1\%$, in the case of trilayers with $x=0.11$ and 0.23.

Figure 2 shows changes of the grain size, D_{AF} of γ-Mn–Ni estimated from XRD profiles by using Scherrer's formula[13] as a function of P_{N_2}. For $P_{N_2}<1\%$ where trilayers were (111) oriented, D_{AF} decreased rapidly with increasing P_{N_2}, independent of x. Since a Mn–Ni grain grows by epitaxy on a Ni–Fe grain, this decrease is caused by the suppression of grain growth of Ni–Fe by adding the nitrogen in the sputtering gas.[12,14] For $P_{N_2}>1\%$, in (100) orientation, D_{AF} was steady at a small value compared with that for $P_{N_2}<1\%$.

B. Magnetic properties

Figure 3 shows changes of the unidirectional anisotropy constant, J_k, and the coercive force, H_c, of trilayers measured at RT as a function of P_{N_2}. H_c of single layered 200-Å-thick Ni–Fe films deposited by the same method as the trilayers is also shown. First, we consider the orientational dependence of exchange anisotropy of the films with $x=0.18$. For $P_{N_2}<1\%$, in (111) orientation, J_k decreased from 0.067 to 0.019 erg/cm^2, and H_c increased from 9 to 17 Oe with increasing P_{N_2}. In contrast, H_c of single layered Ni–Fe films remained constant. Taking into account the decrease of AF grain size shown in Fig. 2, these changes of J_k and H_c with increasing P_{N_2} are explained. The decrease of AF grain size causes the condition $J>K_{AF}D_{AF}$, in which the AF grains cannot contribute to J_k, but to H_c.[15,16] Here J is the exchange coupling energy per unit area at the F/AF interface and K_{AF} is the magnetic anisotropy constant of AF grains. On the other hand, for $P_{N_2}>1\%$, in (100) orientation, J_k and H_c remained constant independent of P_{N_2}. The important point to note is the discontinuous increase of J_k to 0.04 erg/cm^2, compared to the value when $P_{N_2}=0.5\%-1\%$. This abrupt increase of J_k is not explained simply by the change of AF grain size; 60 Å at $P_{N_2}=3\%-7\%$, which is smaller than that at $P_{N_2}=0.5\%$ (see Fig. 2).

Figure 4 shows temperature dependence of J_k of the films with $x=0.18$ for $P_{N_2}=0\%$, 1%, and 5% when the films were cooled in a magnetic field of 1 kOe along the direction of stray field existing during deposition. For all of the P_{N_2}, J_k gradually increased with decreasing temperature

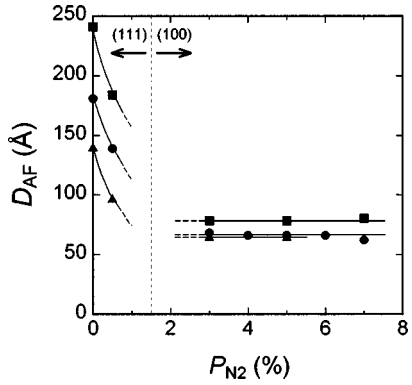

FIG. 2. Changes of grain size of γ-$Mn_{1-x}Ni_x$, D_{AF} estimated from XRD profiles by using Scherrer's formula as a function of P_{N_2}. Ni content x is 0.11 (triangle), 0.18 (circle), and 0.23 (square).

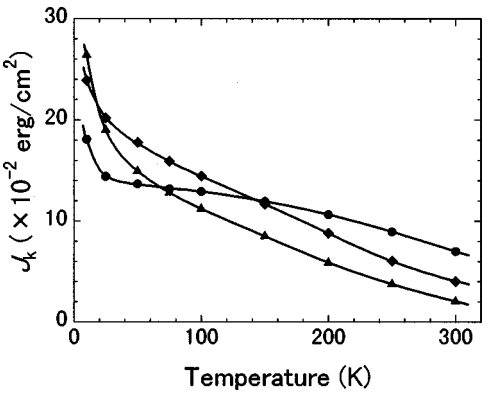

FIG. 4. Temperature dependence of J_k of Ni–Fe/$Mn_{0.82}Ni_{0.18}$/Ni–Fe films with $P_{N_2}=0\%$ (circle), 1% (triangle), and 5% (diamond).

below RT to 50 K, and rapidly increased below 50 K. The rapid increases of J_k below 50 K are mainly contributed by the α-Mn phase ($T_N = 95$ K) coexisting at the interface.[16] The gradual increases of J_k below RT to 50 K are contributed by the γ-Mn–Ni phase.[16] AF spins of a small γ-Mn–Ni grain are flipped when F spins of Ni–Fe layer are reversed by an external magnetic field under finite temperature, because the magnetic anisotropy energy of the small γ-Mn–Ni grain is not enough to adhere AF spins to the anisotropy axis unidirectionally, overcoming the thermal agitation energy.[17] When temperature is decreased, thermal agitation energy becomes small, and small AF grains turn to contribute to J_k. A large increase of J_k with decreasing temperature corresponds to the large population of small grains in the distribution of γ-Mn–Ni grain sizes.[16] Judging from the temperature dependence of J_k below RT to 50 K of the films in Fig. 4, the population of small γ-Mn–Ni grains is larger in the film with $P_{N_2} = 1\%$ than that with $P_{N_2} = 0\%$ and does not differ much from that with $P_{N_2} = 5\%$. These results correspond to the changes of D_{AF} shown in Fig. 2.

Based on the above discussion, we can safely say that 50 K is the temperature where all of the γ-Mn–Ni grains, satisfying the condition $J < K_{AF} D_{AF}$, contribute to J_k because of the small thermal agitation. J_k at 50 K of the films with $P_{N_2} = 5\%$ was 0.18 erg/cm^2 while those of the films with $P_{N_2} = 0\%-1\%$ were 0.14–0.15 erg/cm^2. It means that the exchange anisotropy of Ni–Fe/$Mn_{0.82}Ni_{0.18}$ bilayers increased by changing the crystal orientation from (111) to (100).

Returning to Fig. 3, we will discuss the changes of J_k corresponding to different Ni content x. The difference of x is expected to bring about the difference of AF spin alignment at the interface and to change J_k values. However, J_k of the films with different x were nearly the same value when $P_{N_2} = 0\%$, and changed similarly with increasing P_{N_2}: J_k gradually decreased with increasing P_{N_2} up to 1%, and then discontinuously increased to respective values dependently of x at $P_{N_2} = 3\%$. Taking into account previous discussion, J_k values cannot be compared directly among the films with different x, because the AF grain size of the films was different at each P_{N_2} (Fig. 2). Unfortunately at present, we cannot know exact alignments of AF spins at the interface of {100} oriented films with $x = 0.18$ or 0.11 because of the broad diffraction lines from Mn–Ni {200} planes. Nevertheless, we can conclude experimentally that the exchange anisotropy of Ni–Fe/Mn–Ni bilayers increased by changing crystal orientation from (111) to {100}, in spite of the Ni content, x, reflecting the AF spin alignment at the interface.

IV. SUMMARY

The correlation between AF spin alignment at the interface and the exchange anisotropy was investigated by means of controlling the preferred orientation of Ni–Fe/$Mn_{1-x}Ni_x$ ($x = 0.11$, 0.18, and 0.23) bilayers. We found that: (1) the crystal orientation of bilayers was changed by the addition of nitrogen into the sputtering argon gas only during the deposition of Ni–Fe layer adjacent to the substrate; (2) the preferred orientation was (111) for $P_{N_2} = 0\%-1\%$ and (100) for $P_{N_2} > 1\%$; (3) for $P_{N_2} < 1\%$, J_k decreased and H_c increased with increasing P_{N_2}, because of the decrease of the grain size of γ-Mn–Ni; (4) for $P_{N_2} > 1\%$, J_k increased discontinuously, accompanied by the orientational change of the bilayers.

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