Exchange coupling in nanostructured CoO/NiFe networks

L. Sun,1 Y. Ding,2 C. L. Chien,3 and P. C. Searson1

1Department of Materials Science and Engineering, The Johns Hopkins University, Baltimore, Maryland 21218
2Earth and Planetary Sciences, The Johns Hopkins University, Baltimore, Maryland 21218
3Department of Physics and Astronomy, The Johns Hopkins University, Baltimore, Maryland 21218

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CoO/NiFe(Ni81Fe19) layers were sputtered onto porous alumina templates to form an interconnected network structure. The antiferromagnetic/ferromagnetic coupled nanostructured bilayers exhibit large enhancements of exchange bias $H_E$ and coercivity $H_C$ compared to the continuous AF/FM films with same layer thickness. The temperature dependence of $H_E$ and $H_C$ on the NiFe thickness is reported.

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When an antiferromagnetic (AF)/ferromagnetic (FM) bilayer is cooled below the AF Néel temperature ($T_N$) in the presence of a magnetic field larger than the saturation field of the ferromagnet, the interfacial interactions between the two materials will induce a unidirectional exchange anisotropy which shifts the magnetic hysteresis loop of the ferromagnet from the zero-field axis.1–3 The magnitude of the displacement which shifts the magnetic hysteresis loop of the ferromagnet materials will induce a unidirectional exchange anisotropy.

Enhancements in $H_C$ compared to the continuous AF/FM films with same layer thickness are attributed to domain confinement and pinning in the FM and AF network structures. The as-deposited films have an interconnected network structure, and exhibit significant enhancements in $H_E$ and coercivity $H_C$ compared to uniform exchange coupled thin films. The coercivity of films with different FM layer thicknesses decreases quasilinearly with increasing temperature, and the exchange field is inversely proportional to the FM layer thickness when it is larger than 40 nm. The enhancements in $H_C$ and $H_E$ compared to the continuous films are attributed to domain confinement and pinning in the FM and AF network structures.

To study the thickness dependence of the exchange coupling in the network structure, and allow a comparison to continuous films and uncoupled NiFe networks, three types of samples, were prepared: (i) (001) Si / NiFe wedge (4–78 nm)/CoO (30 nm)/Cu (2 nm), (ii) Al2O3(d = 200 nm)/ NiFe wedge (4–78 nm)/CoO (30 nm)/Cu (2 nm), and (iii) Al2O3(d = 200 nm)/NiFe wedge (4–80 nm)/Cu (2 nm). The NiFe and CoO layers were deposited by dc and rf sputtering, respectively. The 2-nm Cu layer was sputtered to cap the FM/AF bilayer. The continuously varying NiFe layer thickness enabled us to study samples grown under identical deposition conditions, with the thickness being the only variable. The growth rates of the different materials were calibrated by profilometer as well as small angle x-ray diffraction. The alumina templates (Whatman) have an average pore diameter of 200 nm with pseudohexagonal ordering. The average wall thickness at the surface was 40 nm.

Figure 1 shows a plan view scanning electron microscope image of a 30-nm CoO/50-nm NiFe film on an alumina template. The dark areas correspond to the pores, and the bright areas correspond to the FM/AF interconnected network supported by the pore walls. Comparison of the plan view scanning electron microscope images of the network structure with plan view images of an alumina template before deposition reveals no measurable change in pore size. Figure 2 shows plan view transmission electron microscope images of 30-nm CoO thin-film and network structures, illustrating that the grain size in about 7 nm in both cases, significantly smaller than the film thickness and the wall thickness of the network structure.

The magnetic properties of the samples were measured using a vibrating sample magnetometer. During the measurements care was taken to ensure that the applied magnetic field was parallel to the film plane. The samples were first cooled in a 10-kOe external field from room temperature to 80 K, then a series of hysteresis loops were measured while the temperature was increased from 80 to 298 K.

First we discuss the room-temperature magnetic properties of the thin films and network structures. Since the Neél temperature ($T_N$) of CoO is 292 K, there is no exchange coupling between the antiferromagnetic CoO and ferromagnetic NiFe layers. When an antiferromagnetic (AF)/ferromagnetic (FM) bilayer is cooled below the AF Néel temperature ($T_N$) in the presence of a magnetic field larger than the saturation field of the ferromagnet, the interfacial interactions between the two materials will induce a unidirectional exchange anisotropy which shifts the magnetic hysteresis loop of the ferromagnet from the zero-field axis.1–3 The magnitude of the displacement which shifts the magnetic hysteresis loop of the ferromagnet materials will induce a unidirectional exchange anisotropy.
netic NiFe at room temperature, and only the magnetic properties of the NiFe layer are measured. Figure 3 shows the room- netic NiFe at netic NiFe at temperature hysteresis loops for continuous CoO/NiFe thin films and network structures with a NiFe thickness of 30.5 nm. The coercivity of the network structure (164 Oe) is two orders of magnitude larger than the continuous film (<2 Oe). A similar enhancement of $H_C$ has been reported for single FM network structures, and can be attributed to the domain-wall pinning caused by the nonuniform structure.

Figure 3 also shows hysteresis loops for the CoO/NiFe bilayers measured at 100 K after field cooling. The continuous bilayer film, shown in Fig. 3(c), shows a square, shifted loop with squareness ($M_s/M_r$) of 1, a coercivity of 16.2 Oe, and an exchange bias of 53.4 Oe. The corresponding network structure has a squareness of 0.6, a coercivity of 363 Oe, and an exchange bias of 230 Oe.

To compare the exchange coupled and uncoupled network structures, NiFe single-layer and CoO/NiFe bilayer network structures with the same FM thickness were studied. Figure 4 shows the temperature dependence of the coercivity for CoO/NiFe and NiFe network structures with a NiFe thickness of 30.5 nm. The coercive fields of the two samples are the same at room temperature, and both increase quasilinearly with decreasing temperature. However, the coercivity of the CoO/NiFe bilayer increases more quickly than the uncoupled NiFe film, and at 80 K the coercivity of the exchange-coupled network is twice that of the single layer. This indicates that the exchange coupling is not only the source of $H_E$, but also has a significant influence on $H_C$.

The in-plane angular dependence of $H_E$ and $H_C$ for the 30-nm CoO/30.5-nm NiFe network structure was measured at 100 K. As shown in Fig. 5(a), the exchange field of the network structure exhibits a unidirectional symmetry, and shows a simple $\cos \theta$ angular dependence, where $\theta$ is the angle between the external field direction and the exchange anisotropic axis introduced by the cooling field. Since the network structure can also be viewed as an array of interconnected nanowires with random orientations, the network structure is macroscopically magnetically isotropic in the film plane apart from the exchange coupling. The simple $\cos \theta$ dependence of the exchange field is quite different from continuous crystalline FM/AF (Ref. 6) thin films, where $\cos \theta$ and higher order terms are significant. This $\cos \theta$ dependence is comparable to results obtained from continuous amorphous FM/AF thin films, where the crystalline anisotropy of the ferromagnetic layer has also been suppressed.

The angular dependence of the coercivity in the coupled CoO/NiFe network structure, shown in Fig. 5(b), reveals a uniaxial symmetry. However, the magnitude of the coercivity only varies between 340 and 370 Oe. In contrast to continuous amorphous and crystalline FM/AF bilayer thin films, where the intrinsic coercivity of the FM layer is small, the angular dependence of $H_C$ induced by the exchange coupling is more significant. In our case, the ferromagnetic coercivity due to the network structure and the interaction between the FM/AF layers are already very large; hence the angular dependence due to the exchange anisotropy induced by field cooling has a relatively small influence.

Figure 6 shows the temperature dependence of $H_C$ and $H_E$ for nanostructured networks with different ferromagnetic layer thickness (6–90 nm), illustrating that both the coercivity and exchange bias increase monotonically with decreasing temperature. For FM/AF systems, the exchange bias is inversely related to the thickness ($t_{FM}$) of the FM layer according to

$$H_E = \Delta \sigma / M_{FM} t_{FM},$$

where $M_{FM}$ is the saturation magnetization and $\Delta \sigma$ is the anisotropic energy per unit area of the interface. Figure 7 shows the dependence of $H_E$ on $1/t_{FM}$ for thin films and network structures. For thin films at 100 K we obtain $\Delta \sigma = 0.19 \text{ erg cm}^{-2}$, in good agreement with values of 0.10–0.28 reported for exchange bias with polycrystalline CoO AF thin films at 150 K. In the network structure, the linear relationship between $H_E$ and $t_{FM}$ only holds for $t_{FM} > 40$ nm, where we obtain $\Delta \sigma = 1.1 \text{ erg cm}^{-2}$ at 100 K, almost an order of magnitude larger than for the continuous thin films. Since the average grain size for both thin films...
and network structures is the same, the enhancement of the exchange bias in the network structure cannot be attributed to the microstructure. The enhancement in $H_E$ and $H_C$ is thus likely a result of domain confinement in the network structure.

For exchange-coupled bilayer thin films, the linear relationship between $H_E$ and $1/t_{FM}$ is observed for thin films with the FM thickness is as small as 3 nm. The deviation from linear behavior in the network structure when $t_{FM} < 40$ nm is due to finite-size effects, since the ligaments in

![Figure 3](image3)

**FIG. 3.** Magnetic hysteresis loops of a CoO (30 nm)/NiFe (30.5 nm) bilayer. (a) Continuous thin film measured at room temperature, $H_C = 2$ Oe. (b) Continuous thin film measured at 100 K after field cooling, $H_C = 16.2$ Oe and $H_E = 53.4$ Oe. (c) Network structure measured at room temperature, $H_C = 164$ Oe. (d) Network structure measured at 100 K after field cooling, $H_C = 363$ Oe and $H_E = 230$ Oe.

![Figure 4](image4)

**FIG. 4.** Temperature dependence of the coercivity of (a) 30.5-nm NiFe network structure and (b) 30-nm CoO/30.5-nm NiFe bilayer network structure.

![Figure 5](image5)

**FIG. 5.** Angular dependence of (a) the exchange field $H_E$ and (b) the coercivity $H_C$ of an exchange-coupled CoO (30 nm)/NiFe (30.5 nm) bilayer network structure. The dashed line in (a) is a fit corresponding to $H_E(\theta=0)\cos \theta$. 

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the network structure are about 200 nm long and 40 nm wide. As shown in Fig. 6, the nanostructured networks with thin FM layers have a stronger temperature dependence than those with thicker FM layers. At room temperature, the coercivity of the network structures exhibits a maximum at a thickness of 18 nm. Magnetic susceptibility measurements conducted on superconducting quantum interference devices under zero-field-cooled and field-cooled conditions also show that superparamagnetism becomes important when the NiFe layer is thinner than 20 nm. The fact that the superparamagnetic blocking temperature has a broad distribution indicates that the network structure breaks into small regions of different size on decreasing the film thickness.

In summary, using nanoporous templates, we have fabricated large area interconnected network structures with an average lateral size of 40 nm. The artificial antiferromagnetic/ferromagnetic nanostructures exhibit significant enhancements in coercivity and exchange-bias field. The $H_c$ increase can be attributed to the pinning effects of the pores, which impedes the domain wall movement in the FM, so that the magnetic reversal is realized by spin rotation. Due to the confinement of AF domain size and pinning effects, a significant enhancement in $H_E$ has been observed.

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