Tunnel magnetoresistance and spin torque switching in MgO-based magnetic tunnel junctions with a Co/Ni multilayer electrode

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We have fabricated MgO-barrier magnetic tunnel junctions with a Co/Ni switching layer to reduce the demagnetizing field via interface anisotropy. With a fcc-(111) oriented Co/Ni multilayer combined with an FeCoB insertion layer, the demagnetizing field is 2 kOe and the tunnel magnetoresistance can be as high as 106%. Room-temperature measurements of spin-torque switching are in good agreement with predictions for a reduced critical current associated with the small demagnetization for antiparallel-to-parallel switching. For parallel-to-antiparallel switching the small demagnetization field causes spatially nonuniform reversal nucleated at the sample ends, with a low energy barrier but a higher switching current. © 2010 American Institute of Physics

MgO-based magnetic tunnel junctions (MTJs) with a large tunneling magnetoresistance (TMR) (Refs. 1–4) whose magnetic orientations can be controlled by spin-torque switching6–8 are promising candidates for magnetic random access memories.9 However, for widespread application it will be necessary to reduce the switching current density while maintaining thermal stability for the magnetic states. One strategy is to employ a magnetic switching layer with large perpendicular magnetic anisotropy, such that the equilibrium magnetization direction is perpendicular to the sample plane.9–11 Here we investigate an alternative strategy of tuning the perpendicular anisotropy of the switching layer to reduce the demagnetization field but to keep the equilibrium orientation of the switching layer in the sample plane. We show that low-demagnetization fcc-(111)-oriented Co/Ni multilayers can be integrated with MgO tunnel junctions to give high TMR, and that the low demagnetization field of the Co/Ni can provide significant reduction in the spin-torque switching current.

For an in-plane magnetized switching layer within a macrospin approximation for the magnetization dynamics, the critical current for spin-torque switching for an MTJ in the absence of thermal fluctuations has the approximate form

\[ I_{c0} = \frac{2e}{h} \frac{\alpha M_s V}{\eta(\theta)} \left( H_{c0} + \frac{H_{eff}}{2} \right) \]

where \( \alpha \) is the damping constant, \( M_s \) is the saturation magnetization of the switching layer, \( V \) is the volume of the layer, \( \eta(\theta) = p/(1+p^2) \) for parallel-to-antiparallel (P-to-AP) switching and \( \eta(\theta) = p/(1-p^2) \) for AP-to-P switching where the spin polarization \( p = \sqrt{TMR/(TMR+2)} \), \( H_{c0} \) is the coercive field in the absence of thermal fluctuations, and \( H_{eff} \) is the effective demagnetizing field. For a uniform transition-metal magnetic film, \( H_{eff} \) is generally determined by the saturation magnetization, \( H_{eff} \approx 4\pi M_s \approx 10 \text{ kOe} \), while \( H_{c0} \) is much smaller, usually \( \approx 100 \text{ Oe} \) as determined by lateral shape anisotropy. However, the thermal stability of the magnetic bit is governed by \( H_{c0} \), and does not depend on \( H_{eff} \) as long as \( H_{c0} < H_{eff} \). This suggests that \( I_{c0} \) may be reduced by using the interface anisotropy of multilayers like Co/Ni to decrease \( H_{eff} \), while leaving \( H_{c0} \) unchanged so as to maintain thermal stability. In previous work, employing a Co/Ni multilayer within all-metal spin valves,16 our group demonstrated a factor of 5 reduction in \( I_{c0} \) relative to control samples, but all-metal spin valves lack the large TMR provided by MgO-based MTJs that is necessary for applications. Incorporating a Co/Ni electrode with an MgO tunnel barrier is nontrivial, since multilayer Co/Ni has an fcc-(111) structure that does not provide the same band matching to MgO(001) employed in high-TMR MTJs with bcc-(001) Fe or FeCoB electrodes.1–4 We report the fabrication of high-TMR MTJs with reduced-demagnetization switching layers consisting of a Co/Ni multilayer together with a thin FeCoB insertion layer contacting the MgO. We characterize the crystal structure of the interface and discuss the TMR and spin-transfer switching characteristics of these junctions.

Our MTJ layer stack was prepared on SiO2/Si(001) wafers by a magnetron sputtering with a base pressure of 10−9 Torr. The layer structure is Ta(3)/[CuNi(20)/Ta(3)]2/Cu(2)/[Co(0.4)/Ni(0.8)]2/Fe60Co20B20(1.1)/MgO(0.3)/Fe60Co20B20(20)/Ta(8)/Pt(30). The numbers in the parentheses are the layer thicknesses in nanometers. The MgO film is formed by rf magnetron sputtering with an oxygen getter driven by sputtered tantalum. The MgO thickness, \( t_s \), was varied from 0.7 to 1.5 nm across the wafer. After the deposition of all layers, the wafers were annealed in a N2 atmosphere at 375 °C for up to 10 min on a sample stage allowing a fast cooling rate of 43 °C/min. Individual tunnel junctions were then patterned using electron-beam lithography and ion-beam etching. X-ray diffraction measurements (not shown) indicate the [Co(0.4)/Ni(0.8)]2 layer is (111)-textured, as required for interface anisotropy in Co/Ni multilayers. The 1.1 nm FeCoB layer is designed to provide a buffer for a lattice matching between fcc-CoNi(111) and MgO(001) while reducing only slightly the mean perpendicular anisotropy. Magnetization measurements show that the equilibrium moment of the [Co(0.4)/Ni(0.8)]2/FeCoB(1.1) film lies in plane with a perpendicular...
saturation field of 2 kOe, which indicates that the demagnetizing field is reduced by about 10 kOe relative to the averaged saturation magnetization of 12 kOe. From ferromagnetic resonance measurements, the Gilbert damping parameter of the [Co(0.4)/Ni(0.8)]2/FeCoB(1.1) film is $\alpha = 0.015 \pm 0.005$.

Figure 1 shows scanning transmission electron microscopy (STEM) images and electron energy loss spectroscopy (EELS) composition maps of the MTJ layer stack after a 3 min anneal, for which we achieved room-temperature TMR ratios as large as 106% [see Fig. 2(a)]. We observe a high degree of crystal coherence extending from the Co/Ni multilayer up through the FeCoB insertion layer to the MgO [Fig. 1(b)]. The crystal lattice at the lower FeCoB/MgO interface seems to be partially matched by introducing dislocations. The fcc-(110) face of the Co/Ni multilayer and (100) face of the MgO are oriented together in the plane of the STEM image [Fig. 1(b)]. From the Fourier transform of the MgO lattice image [Fig. 1(c)], the tunnel barrier has the usual cubic structure with only a compression by 3% in [001] direction. The Fourier transform of the image of the lower FeCoB region [Fig. 1(d)] indicates that the FeCoB insertion layer is crystallized in a strained fcc structure rather than the usual bcc, and has the same orientation as the Co/Ni multilayer. We therefore conclude that the structure at the lower interface of the tunnel barrier is fcc-FeCoB(111)/MgO(001)/[100] rather than the bcc-FeCoB(001)[110]/MgO(001)[100] structure ordinarily used for high-TMR junctions. The upper FeCoB layer is crystallized only near the interface with MgO, in the usual bcc-FeCoB(001)[110]/MgO(001)[100] relationship, most likely due to its greater thickness. This could be reduced by using an antiferromagnetic pinning layer. The EELS image [Fig. 1(e)] shows that the [Co/Ni]$_2$/FeCoB film maintains its layer structure without large-scale intermixing. EELS also reveals that the annealed barrier is Mg(B)O.

The switching properties of a 70 x 220 nm$^2$ device with $R_A = 4.3$ $\Omega$ $\mu$m$^2$ and TMR = 38% are shown in Fig. 2(b). The minor loop as a function of in-plane magnetic field indicates an ~52 Oe coercive field and an average dipole field from the fixed layer of 460 Oe. The hysteresis loop for spin-torque switching as a function of current, for an applied field that cancels the average dipole field, shows quasistatic room-temperature switching currents for AP-to-P and P-to-AP switching of ~0.31 mA and 0.35 mA, respectively. To estimate the effective activation energy $E_a$ and the zero-thermal-fluctuation critical current $I_{c0}$, we performed both current-pulse [Fig. 2(c)] and field-ramp measurements [Fig. 2(d)]. Assuming that current-induced heating effects are negligible, for thermally activated switching the average switching current $\langle I_s \rangle$ and the switching field $\langle H_s \rangle$ measured relative to 460 Oe should take the forms

\begin{equation}
\langle I_s \rangle = I_{c0} \left[ 1 - \frac{k_B T}{E_a} \ln \left( \frac{1}{I_{c0}} \right) \right],
\end{equation}

\begin{equation}
\langle H_s \rangle = H_{c0} \left( 1 - \frac{k_B T}{E_a} \ln \left( \frac{1}{\tau_0 |R_H| \ln 2} \right) \right)^{2/3},
\end{equation}

where $k_B$ is Boltzmann’s constant, $\tau_p$ is the pulse duration, $R_H$ is the ramp rate for field, and $\tau_0$ is the inverse of the attempt frequency which we assume to be $10^{-19}$ s. From the fits to the current-pulse data in Fig. 2(c), we obtain for AP-to-P switching $E_{a,AP-P} = 1.12 \pm 0.07$ eV and $I_{c0,AP-P} = 0.60 \pm 0.02$ mA (corresponding to $5.0 \times 10^9$ A/cm$^2$) and for P-to-AP switch-
For current-driven AP-to-P switching, the critical current density is higher, which simulations suggest is due to a spatially nonuniform, edge-dominated reversal mode, and the switching currents may be further reduced by using switching layers with smaller total magnetic moments and higher tunneling polarizations.

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