Current-induced realignment of magnetic domains in nanostructured Cu/Co multilayer pillars

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We have developed a nanofabrication technique to facilitate current-perpendicular-to-plane (CPP) transport measurements on magnetic multilayer pillar structures with diameters as narrow as 100 nm—a size scale at which the reversal of individual domains within the ferromagnetic layers may be detected. When large currents are passed through such pillars, the Oersted field produced by the current can affect the orientation of the magnetic moments of the layers. In pillars ranging from 250 to 500 nm, a stack of alternating hard and soft ferromagnetic layers can controllably be switched between high and low resistance states via this mechanism. © 2000 American Institute of Physics.

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Since the discovery of giant magnetoresistance (GMR), a number of methods have been devised to allow current-perpendicular-to-plane (CPP) measurements to be performed on these structures. Though more difficult than conventional current-in-plane (CIP) measurements, CPP transport can give larger magnetoresistance values, and is a much cleaner probe of the interfacial scattering that gives rise to the GMR effect. The first such measurements relied on superconducting electrodes, but such experiments are limited in the temperature and current density ($J$) ranges that may be studied. By using microfabrication techniques, it is possible to increase the stack resistance such that superconducting voltmeter detection is no longer required for CPP measurements.

In this letter, we describe a versatile process that allows CPP transport studies of virtually any multilayer material system to be performed with device diameters approaching 100 nm. We have used these techniques to study Cu/Co multilayers, and, in our narrowest devices, have clearly observed discrete changes in the resistance $R$ due to the reorientation of individual magnetic domains. In this CPP geometry, the strong Oersted field created by high current densities can switch the magnetic layers into vortex magnetization states, the helicity of which depends on the direction of the current flow. By alternating thin (soft) and thick (hard) Co layers, we have created a device geometry in which the Oersted fields produced by current pulses of the appropriate polarity can controllably switch the relative alignment and, as a consequence, $R$, of adjacent ferromagnetic layers.

We begin fabrication by sputtering 1200 Å Cu/the desired multilayer/600 Å Au onto an oxidized Si substrate. The 1200 Å Cu serves as the bottom electrode, while the Au capping layer prevents oxidation and allows subsequent electrical contacting to the top of the stack. Electron beam lithography, thermal evaporation, and liftoff are used to pattern Cr dots on the Au surface. These dots serve as the mask during the ion milling step that defines the pillars. Figure 1 is a scanning electron microscope image of pillars following milling for a multilayer consisting of 15 repetitions of 12 Å Co/19 Å Cu (second antiferromagnetic coupling maximum). The Cr mask was 100 nm in diameter and has been almost completely eroded during the milling process. The combined effects of lateral erosion and redeposition create the “gumdrop” shape of the pillars in Fig. 1. As a result, near the base of the pillar where the multilayer is situated, we estimate the diameter to be ∼130 nm. Although we have fabricated devices as narrow as 100 nm from thinner multilayer films, redeposition prevents lateral dimensions much below this from being achieved with this technique. Following ion milling, the pillars are planarized with polyimide (cured at 220 °C in an inert atmosphere). Photolithography and O$_2$ reactive ion etching are used to etch the polyimide until the Au cap of the pillar is uncovered. A final photolithography/Cu deposition/liftoff step is used to define the top electrode. To minimize contact resistance, it is essential to clean the pillar surface with a short ion mill immediately prior to the Cu deposition. Although the present work focuses on magnetic multilayers, this fabrication technique should allow CPP measurements to be made on a wide variety of material systems.

Figure 2(a) shows room temperature CPP magnetoresistance data from a single pillar similar to those shown in Fig. 1(a). Although the measured magnetoresistance is only about 20 at room temperature (43 at 4.2 K), it would be signifi-
cantly larger if the spreading resistance and contact resistance were subtracted.\textsuperscript{6} Figure 2(b) is a room temperature CIP measurement for a multilayer film of the identical (Co 12 Å/Cu 19 Å)\textsubscript{15} composition (but without the bottom Cu layer and top Au layer). Unlike the unpatterned film, the pillar magnetoresistance shows discrete steps. We attribute each step to the flipping of an individual domain. Because the pillar diameter is so narrow, each domain represents a significant fraction of the area of a layer, if not the entire layer, making each switching event readily observable in the GMR signal. In fact, we are approaching the dimensions at which single domain behavior has been observed in thin film ferromagnetic disks.\textsuperscript{7} The saturation field (1.8 kOe) of the pillar samples is also much larger than that of the unpatterned film. Although the exchange coupling through the Cu spacer layer should be comparable in the two cases, magnetostatic edge charges enhance the antiferromagnetic coupling in the narrow pillar.\textsuperscript{8}

A remarkable feature of the pillars is the large current that can be applied without damaging the devices. Currents as large as 80 mA ($J=6 \times 10^8$ A/cm$^2$) have been applied without any signs of annealing or electromigration. Figure 3 shows the resistance of a similar 15 repetition, second maximum pillar as a function of $I$, the current applied. When, as in the bottom trace in Fig. 3, an external field larger than the saturation field is applied in the plane of the layers, a gradual rise in $R$ is observed with increasing $I$ due to electronmagnon scattering and heating effects. Although such effects are doubtless present in the zero field trace, $R$ actually decreases with increasing $I$. The probable explanation for this is that the antiferromagnetic alignment that is present in zero field is being gradually suppressed by the current. Associated with the current moving vertically through the pillar, there is a circular Oersted field that is strongest at the edge of the pillar. This field can create vortex states in the magnetization of the layers, the helicity of which depend on the direction of the current flow. Due to the short height of the pillar, the field is not uniform along the length of the pillar, and the fields produced by currents flowing in the top and bottom electrodes may also be significant. Ignoring the effects of such electrode currents, we estimate that a 30 mA current produces a 700 Oe field at the edge of a 130 nm diameter Co layer located midway between the electrodes. Considering how an external field $H_{eo}$ of 700 Oe is sufficient to partially align the magnetization of the layers [Fig. 2(a)], it would be surprising if the Oersted field from 30 mA could not partially align the Co layers in our multilayer pillars.

For the zero field trace in Fig. 3, between +15 and −15 mA, there are a number of discrete resistance jumps. Although these may reflect domain reorientation due to the Oersted fields produced by the currents, we have performed experiments with other multilayer configurations that indicate that they are more likely domain switching events created by the spin-transfer phenomenon present in CPP measurements.\textsuperscript{9} For pillars narrower than 200 nm, spin-transfer effects associated with vertical currents are predicted to dominate the effects of Oersted fields due to those same currents.\textsuperscript{10}

Excepting the small jumps in $R$ due to spin transfer effects, the $R$ vs $I$ loops in Fig. 3 are nonhysteretic. This is because all the ferromagnetic layers have essentially the same coercive field, and the strong interlayer coupling returns the stack to its antiferromagnetically aligned state as the current is reduced. However, current hysteresis with potential memory applications becomes possible in pillars with ferromagnetic layers of different coercive fields separated by Cu spacer layers thick enough to suppress interlayer exchange coupling.\textsuperscript{11} Figure 4 shows $R$ vs $I$ hysteresis loops for such devices. The stacks consist of Cu 1200 Å/(Co 12 Å/Cu 40 Å/Co 60 Å/Cu 40 Å)$\times_3$/Au 600 Å. Since they are much thinner, the 12 Å Co layers have a smaller coercive field than the 60 Å Co layers. Of course even greater distinctions in the coercivities of the ferromagnetic layers could be achieved by using a harder magnetic material for the thick layers. Although edge coupling is still present, it is strongly suppressed once the vortex states are formed. Also, to minimize the effects of spin transfer, these are larger diameter samples: 250 nm in Fig. 4(a), and 500 Å in Fig. 4(b).

Examining the zero field $R$ vs $I$ hysteresis loop in Fig. 4(a), we begin with the device in a low $R$, with the helicities of the current pillar. This field can create vortex states in the magnetization of the layers, the helicity of which depend on the direction of

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Aside from slight variations in the fine structure, the curve is very reproducible, and traces over itself as the hysteresis loop is repeated. Figure 4(c) illustrates the principle of how such a device can be employed for static memory application. First, initialize the handedness of all layers with a current pulse larger than $70 \, \text{mA}$. The device can then be repeatedly switched into a high $R$ state with a $-16 \, \text{mA}$ pulse that rotates the soft but not the hard layer, and rotated back into a low $R$ state with a $+16 \, \text{mA}$ pulse.

In Fig. 4(a), we also plot hysteresis loops taken with in-plane $H_{ex}$ of 500, 1000, and 1500 Oe. By 1 kOe, the applied field is making it difficult to achieve any antiparallel alignment, while by 1.5 kOe, the creation of opposite helicity states is almost completely suppressed. This is not surprising since we estimate that the Oersted field at the edge of the 250 nm pillars is only about 250 Oe at 30 mA. The Oersted fields are weaker still at the edge of a 500 nm pillar, data from which are shown in Fig. 4(b). Consequently, a smaller $H_{ex}$ is sufficient to eliminate hysteretic effects. Also note that for the zero field trace in Fig. 4(b), the resistance maxima occur close to $\pm 50 \, \text{mA}$, due to the weakening of the average Oersted field for a given current in larger area pillars.

In summary, we have demonstrated how a combination of e-beam lithography and ion milling can be used to fabricate $\sim 100 \, \text{nm}$ diameter Cu/Co pillars capable of sustaining current densities in excess of $5 \times 10^5 \, \text{A/cm}^2$. Using the GMR effect as a probe of the relative alignment of the Co layers, we have shown that the Oersted field created by a current passing through the pillar can create vortex magnetization states within the Co layers. This Oersted field can be used to alter the relative magnetization alignments of adjacent layers in pillars containing alternating hard and soft magnetic layers, which allows the resistance of such a stack to be controlled by current pulses of the appropriate polarity.

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