Effect of deposition parameters on the CPP-GMR of NiMnSb-based spin-valve structures

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Abstract

We present measurements of current perpendicular magnetoresistance of NiMnSb/Cu/NiMnSb spin-valve structures based upon the predicted half-metallic (100% spin polarized) ferromagnetic alloy NiMnSb. The observed effect of 5–10% is much smaller than the complete spin-valve effect expected from a 100% spin-polarized system. The dependence of the magnetoresistance on film thickness and deposition parameters are explored to understand the factors limiting giant magnetoresistance in these structures. © 1999 Elsevier Science B.V. All rights reserved.

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Energy band calculations [1] indicate that the ferromagnetic Heusler alloy (HA) NiMnSb should be 100% spin polarised at the Fermi level (i.e., half-metallic). Although positron-annihilation experiments confirm the existence of an energy gap for the minority spin electrons at the Fermi level [2], this has not been conclusively verified from the transport properties [3,4]. Tunneling structures in which NiMnSb and $Ni_{0.8}Fe_{0.2}$ layers are separated by thin insulating (I) layers show only 15% spin polarization [5]. Giant magnetoresistance (GMR) studies of HA-based F/N (N = non-magnetic metal) multilayers and spin valves in the usual current in plane (CIP) geometry have, so far, revealed only relatively small MRs, $\sim$ 1–3% [6]. The most direct test of fully-polarized behavior would come from measurements of the current perpendicular to the planes (CPP) MR of HA-based metallic multilayers or spin valves, since CPP transport probes the bulk of the HA- and N-metals as

well as their interfaces. In particular, a fully antiparallel (AP) magnetization alignment of adjacent 100%-polarized ferromagnetic films should give, in principle, insulating behavior at low temperatures. In this paper, we describe in detail measurements of the CPP-MRs of NiMnSb-based spin valves of the form NiMnSb/Cu/NiMnSb/FeMn.

Our CPP-MR measurements used the well-established crossed superconducting strips technique [7,8], with the sample of interest sandwiched between 1500 Å-thick, crossed, 1 mm-wide Nb strips. The samples and Nb strips were deposited in situ, as described elsewhere [9]. This geometry gives a uniform current perpendicular to the plane of the sample. However, the measured specific resistance AR_T (overlap area $A \approx 1 \text{ mm}^2$ times the total resistance R_T) contains, in addition to the intrinsic resistance of the NiMnSb/Cu/NiMnSb stack, the specific resistances of the Nb/NiMnSb, NiMnSb/FeMn and FeMn/Nb interfaces present in the samples, as well as the resistance of the FeMn layer itself. In this study, the specific resistances for NiMnSb/Nb ($4.5 \pm 1 \text{ f}\Omega \text{ m}^2$) and FeMn/Nb ($3.33 \pm 1 \text{ f}\Omega \text{ m}^2$) interfaces, and bulk

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FeMn ($\rho_{\text{FeMn}} = 111 \, \mu\Omega$ cm), were determined from independent measurements of AR_T vs. t for a series of Nb/NiMnSb(t)/Nb and Nb/FeMn(t)/Nb sandwiches. The interface values are comparable to those observed for other magnetic materials such as Nb/permalloy [10] and Nb/Co [11], while the resistivity of FeMn is somewhat larger than that usually observed for sputtered films.

Details on the structural, magnetic and magneto-optical characterization of our sputtered NiMnSb thin films have been given elsewhere [12,13]. The NiMnSb films for the CPP samples described here are different from previous films only in the fact that they are grown on thick Nb buffer layers instead of directly on glass or Si substrates. XRD analysis of 1000 Å-thick NiMnSb films grown on Nb indicates that NiMnSb grows with (2 2 0) and (2 0 0) orientation and that the crystalline quality of such films are at least equivalent to those grown directly on substrates. Additionally, the value of the saturation magnetization, which is very sensitive to crystalline order in NiMnSb, is observed to be unchanged with the addition of a Nb buffer layer. Thus Nb generally appears to be an attractive buffer layer for NiMnSb films, both as a superconductor at low temperatures, or normal metal at higher temperatures.

We have reported previously on the incorporation of 100–200 Å-thick NiMnSb layers within CPP-NiMnSb(200 Å)/Cu(t_{Cu})/NiMnSb(100 Å)/FeMn(100 Å) spin valves using sputtering conditions compatible with NiMnSb film growth [12]. The bottom Nb strip was deposited at $T_s < 180°$C while the NiMnSb/Cu/NiMnSb stack was deposited at 250°C. Magnetic measurements suggested that $t_{\text{Cu}} > 120$ Å is required to achieve independent magnetization reversal in the two NiMnSb layers, i.e., two distinct steps in the magnetic hysteresis loops. Although the magnitude of the observed magnetization steps [9] is that expected from the 2 : 1 thickness ratio of the two NiMnSb layers, there are no clearly flat field-independent regions, which suggest that the magnetizations do not achieve a fully antiparallel orientation. For structures with $t_{\text{Cu}} = 150$ Å an intrinsic CPP-MR value of about 10% was observed after removal of the Nb/NiMnSb and FeMn/Nb interface resistances as well as the bulk resistance from the FeMn layer. One possible limitation of the GMR are the significant interdiffusion and rougher interfaces expected when the entire CPP spin valve is fabricated with $T_s \cong 250°$C (high enough to grow atomically ordered NiMnSb layers), which can modify the half-metallic band structure of NiMnSb and also introduce spin-flip scattering centers within the spin-valve structure.

To further clarify the effects of interdiffusion and interface roughness on the spin-valve performance, several samples were fabricated using different deposition temperatures for the bottom Nb strip and Cu spacer. It s expected that variations in T_s(Nb) and T_s(Cu) will
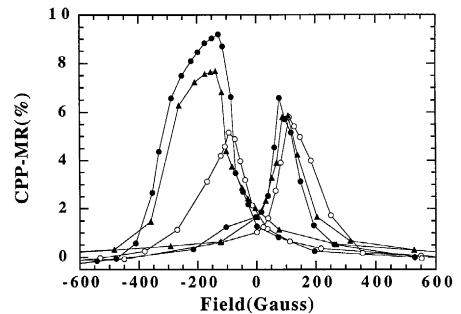

Fig. 1. CPP-MR at 4.2 K for NiMnSb(200 Å)/Cu(150 Å)/NiMnSb(100 Å)/FeMn(100 Å) spin valves with bottom Nb strips deposited at 20°C (open circles), 60°C (filled circles), and 150°C (filled triangles).

modify the Nb/NiMnSb and NiMnSb/Cu interfaces and consequently affect the magnetic and MR properties of our spin valves. Fig. 1 shows the CPP-MR versus field at 4.2 K for NiMnSb(200 Å)/Cu(150 Å)/NiMnSb(100 Å)/FeMn(100 Å) spin valves with bottom Nb strips deposited at 20°C, 60°C and 150°C. The magnetization reversal fields for samples where T_s(Nb) = 60–150°C are close to those observed in earlier samples [9]. Thus, we may assume that for T_s(Nb) = 60–150°C, the NiMnSb film coercivities are almost identical. While the use of lower T_s(Nb) reduces the overall interface roughness, the loss of antiparallel alignment and consequent decrease of MR as T_s(Nb) changes from 60°C to 20°C suggest that the exchange bias from the FeMn is less effective for smoother interfaces. the larger MR ($\sim 9.2\%$) and broader antiparallel state obtained when T_s(Nb) = 60°C indicates that an optimal, moderate T_s(Nb) exists where moderate levels of interdiffusion can be achieved, while retaining the effectiveness of the FeMn layer.

Fig. 2 shows the CPP-MR versus field at 4.2 K for spin valves similar to those displayed above. For this set of samples T_s(Nb) was kept at 150°C while T_s(Cu) was chosen to be 250, 150 and 80°C. After deposition of the 200 Å-thick NiMnSb layer at 250°C, the Cu spacer was either deposited immediately at 250°C or after the sample holder was cooled down to 150 or 80°C. Typical cool-down times to bring the substrate from 250°C to 150°C and 80°C are 25 and 70 min, respectively. Although, overall, the observed CPP-MR values are smaller than those of previous sets of samples, the trend shown in Fig. 2 is still revealing. In particular, the largest CPP-MR values are obtained when T_s(Cu) = 150°C. It appears that T_s(Cu) = 250°C is high enough to produce excessive interdiffusion and consequently reduce the GMR value, while the long time necessary to cool down the substrate from 250°C to 80°C may lead to oxidation and/or impurity incorporation on the NiMnSb surface. Such impurities are centers of spin-flip electron scattering and consequently decrease the GMR.
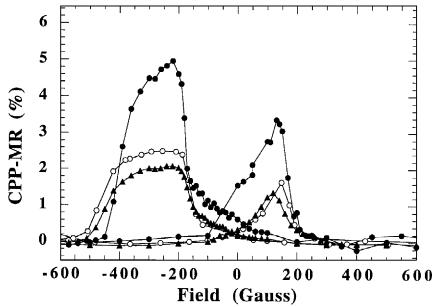
Fig. 2. CPP-MR at 4.2 K for NiMnSb(200 Å)/Cu(150 Å)/NiMnSb(100 Å)/FeMn(100 Å) spin valves with Cu spacers deposited at 250°C (open circles), 150°C (filled circles), and 80°C (filled triangles).
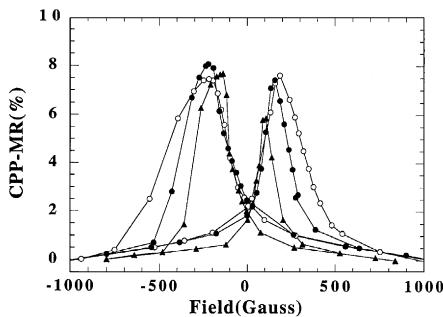

Fig. 3. CPP-MR at 4.2 K for NiMnSb($2X$)/Cu(150 Å)/NiMnSb(X)/FeMn(X) spin-valves with $X \equiv 25$ Å (open circles), 50 Å (filled circles), and 100 Å (filled triangles).

Finally, the dependence of the CPP-MR on the NiMnSb layer thickness was investigated. CPP spin valves of the form NiMnSb($2X$)/Cu(150 Å)/NiMnSb(X)/FeMn(X), where $X = 25, 50$ and 100 Å, were fabricated under the same conditions of those samples reported initially. Fig. 3 shows an increase of the NiMnSb film coercivity as t_{NiMnSb} decreases, but only slight changes in the CPP-MR value are observed. This suggest that the spin diffusion length λ_{sf} in our NiMnSb films is at least 100–200 Å, since a much smaller value would result in a gradual decrease of the measured CPP-MR with increasing NiMnSb thickness due to the extra resistance of the NiMnSb regions not contributing to the GMR. Of course in the case of 100% spin polarization, the antiparallel state of the spin valve should be insulating regardless of the value of λ_{sf}. Therefore, the limited GMR observed here must be explained by the lack of complete polarization, in the NiMnSb as a whole and/or due to spin mixing at disordered NiMnSb/Cu interfaces.

In conclusion, we have shown that various deposition parameters strongly affect the CPP-GMR of NiMnSb/Cu/NiMnSb spin valves. In particular, the deposition temperatures for the Nb buffer layer and the Cu spacer have a significant influence on the measured GMR, presumably due to variations in atomic disorder at interfaces and/or within the NiMnSb layers. The intrinsic CPP-GMR remains small, however (< 10%), which implies that further improvements in the crystalline quality of NiMnSb and NiMnSb/Cu interfaces must be realized. The use of epitaxial Nb layers, for example, with subsequent growth of epitaxial NiMnSb, may be a fruitful approach for further study of this system.

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