Influence of crystal structure on the magnetoresistance of Co/Cr multilayers

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Epitaxial Co/Cr multilayers, and single-crystal Co thin films etc. have been grown on MgO and Al_2O_3 substrates with Cr and Mo as buffer layers by molecular beam epitaxy technique. From the structure and magnetoresistance studies, we have found that the ferromagnetic anisotropy of resistance (AMR) is strongly influenced by the buffer layer, but with negligible effect due to the variation of the structure of Co films. The AMR of Co film on Cr buffer layer is quite small (0.1%); however, the MR of Co/Cr multilayers is almost one order larger than the AMR of Co film on Cr buffer layer. An enhancement factor of 4 for the MR in Co/Cr multilayers by the interface roughness has been observed. This suggests that the effect due to the spin dependent scattering at the interfacial regions of the superlattice is larger than that due to the spin dependent scattering in the ferromagnetic layers for the MR in the Co/Cr multilayer system.

During the past several years, the magnetoresistance (MR) behaviors in many metallic multilayers have become the subject of considerable interest. Large (or giant) MR was first realized in Fe/Cr multilayer system,[1] and has been referred to as GMR. Relatively small MR occurs in the Co/Cr multilayers.[2] The MR in multilayers results from the spin dependent scattering of the conduction electrons which occurs both in the ferromagnetic layers and at the interfacial regions between the ferromagnetic and nonferromagnetic layers. It is quite different from the ferromagnetic anisotropy of resistance (AMR) in ferromagnetic systems, which depends on the direction of the magnetization.[3]

Epitaxial Co/Cr multilayers as well as single-crystal hcp-Co, fcc-Co, and polycrystal Co thin films have been grown on both MgO and Al_2O_3 substracts with Cr and Mo as buffer layers using an Eiko EL-10A molecular beam epitaxy (MBE) system with base pressure of $\sim 2\times10^{-10}$ Torr. Pure elements (99.99%) of Co, Cr, and Mo were evaporated from three independent e-beam evaporators. During deposition of the elements, the growth pressure was controlled at below 5×10^{-9} Torr, and the deposition rate at $\sim$0.1 Å/s. To enable the growth of high-quality samples, polished and epitaxial grade MgO and Al_2O_3 substrates were chemically precleaned and rinsed in an ultrasonic cleaner. They were then outgassed at 900 to 1000 °C for at least 1/2 h under ultra high vacuum in the MBE system. For samples with a Mo (or Co) buffer layer, Mo (or Co) was deposited on the substrates at 900 (or 500) °C. The substrate temperature for all films during evaporating was kept between 300 and 350 °C. The crystallographic structure of the films were examined, throughout all the growth, by reflection high energy electron diffraction (RHEED). The interface roughness and the thickness of the superlattice structures were determined by the x-ray reflectivity analyses.

The magnetic properties of all the samples were studied by using a SQUID magnetometer. The AMR and MR measurements were carried out by the conventional four probe technique.

Before discussing the experimental data of AMR and MR in the CoCr system, we have to clarify their definition. The AMR in ferromagnetic films is defined by $(R_{\parallel}-R_{\perp})/R_0$, where R_0 is the electrical resistance in zero internal magnetic field, and $R_{\parallel}$ and $R_{\perp}$ are the resistances when the saturated magnetization is parallel and perpendicular to the current, respectively.

The MR (or GMR) in multilayers is defined as $[R_{AF}-R_S]/R_S$, where R_S is the electrical resistance at saturated high magnetic field, and the spins in Co layers align in the field direction. R_{AF} is the electrical resistance when the field is removed, the Co layers adjacent to the Cr layer in-between exhibit antiferromagnetic coupling.

Crystal structures of thin films or multilayers may be considerably affected by the choice of buffer layers, substrates, and their orientations. In this study, we chose MgO(100), Al_2O_3 ($1\bar{1}02$), and Al_2O_3(0001) as substrates, and Cr and Mo as buffer layers to study the variation of AMR for Co films. In general, for Co grown on MgO(100) substrate without a buffer layer, an epitaxial fcc-Co film with (100) growth plane can be formed for Co thicknesses larger than about 60 Å, but for Co grown on an Al_2O_3(0001) substrate

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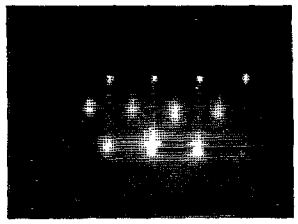
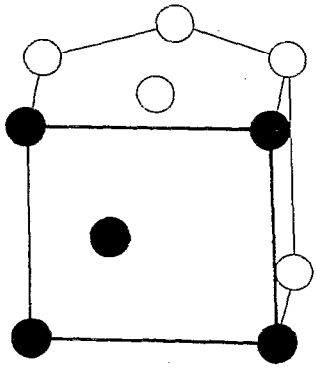
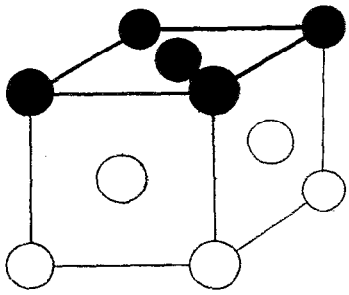




(a)



(b)

FIG. 1. Typical RHEED patterns of (a) hcp-Co(11$\bar{2}$0) plane viewed along [0001], and (b) fcc-Co(100) plane viewed along [01$\bar{1}$]. The surface with solid circles cutting with the unit cell of both hcp(11$\bar{2}$0) and fcc(100) Co are schematically illustrated below each RHEED pictures.

without a buffer layer, a polycrystal Co film was observed. In addition when we grow a thin buffer layer of Cr(100) about 20 Å on either MgO(100) or Al_2O_3(1$\bar{1}$02), then both RHEED and x-ray diffraction (XRD) studies show an hcp-Co structure with (11$\bar{2}$0) plane parallel to the (100) surface of Cr. For example, Fig. 1 shows the typical RHEED patterns of (a) hcp-Co(11$\bar{2}$0) plane viewed along [0001], and (b) fcc-Co(100) plane viewed along [01$\bar{1}$]. The surfaces with solid circles cutting with unit cell of hcp(11$\bar{2}$0) and fcc(100) are

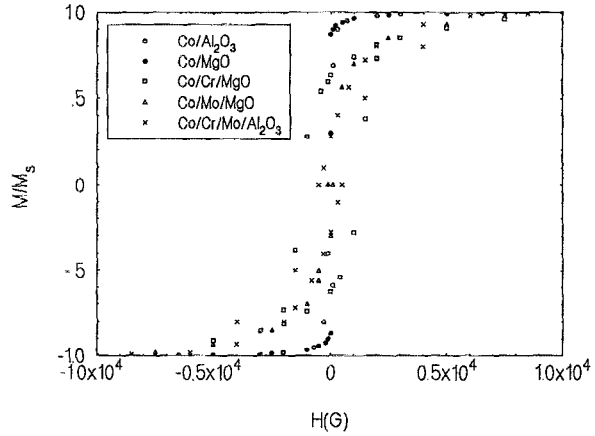


FIG. 2. The normalized magnetization as a function of magnetic field at room temperature for (a) Co/Al_2O_3, (b) Co/MgO, (c) Co/Cr/Mgo, (d) Co/Mo/MgO, and (e) $Co/Cr/Mo/Al_2O_3$.

schematically illustrated below each of the RHEED pictures.

Figure 2 shows the normalized magnetization as a function of the magnetic field at room temperature for 5 thin film samples (Co/Al_2O_3, Co/MgO, Co/Cr/MgO, Co/Mo/MgO, and $Co/Cr/Mo/Al_2O_3$). Generally speaking, the magnetization is saturated after roughly 6 kG for all the samples. Figure 3 presents the normalized difference of resistance as function of the magnetic field at room temperature for 5 thin film samples: (a) polycrystal Co on Al_2O_3 (0001), (b) fcc-Co(100) on MgO(100), (c) hcp-Co(11$\bar{2}$0) on Cr(100) which is on MgO(100), (d) hcp-Co(11$\bar{2}$0) on Mo(100) which is on MgO(100), and (e) hcp-Co(11$\bar{2}$0) on Cr(100) and Mo(100) which is on Al_2O_3(1$\bar{1}$02). One can see that the values of AMR for both polycrystal- and fcc-Co films without a buffer layer are roughly equal to 1.3%. However, the AMRs of all the Co films with either Cr or Mo as a buffer layer are roughly one order of magnitude smaller than that of Co films without a buffer layer. Therefore, we conclude that the effect to the AMR for Co films with different structure is negli-

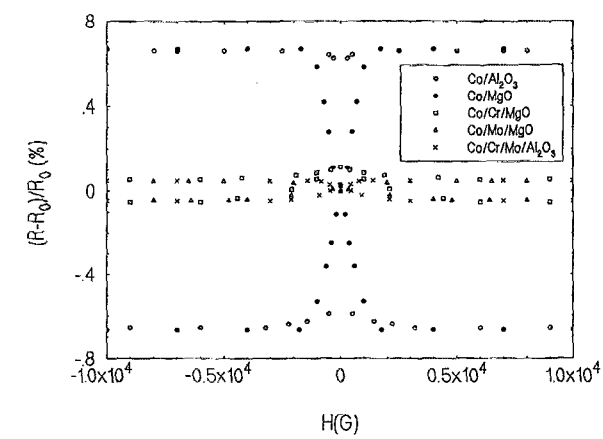


FIG. 3. The normalized difference of electrical resistance between magnetic field and zero field as a function of magnetic field at room temperature for (a) polycrystal Co/Al_2O_3, (b) fcc-Co/MgO, (c) hcp-Co/Cr/MgO, (d) hcp-Co/Mo/MgO, and (e) hcp-Co/Cr/Mo/Al_2O_3.

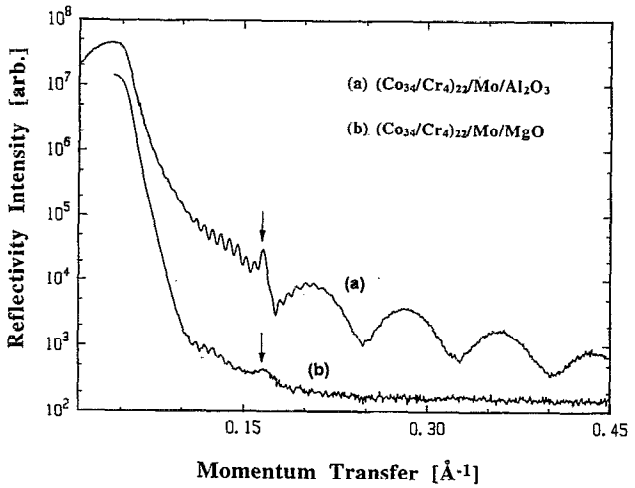


FIG. 4. X-ray reflectivity measurement of the samples (a) $(Co/Cr)_{22}/Mo/Al_2O_3$, and (b) $(Co/Cr)_{22}/Mo/MgO$. The arrow indicates the peak due to the period of the superlattice. The oscillation fringes pattern is due to the interference of the x-ray reflection between the two interfaces of Mo buffer layer and the interfaces of the total growth thickness.

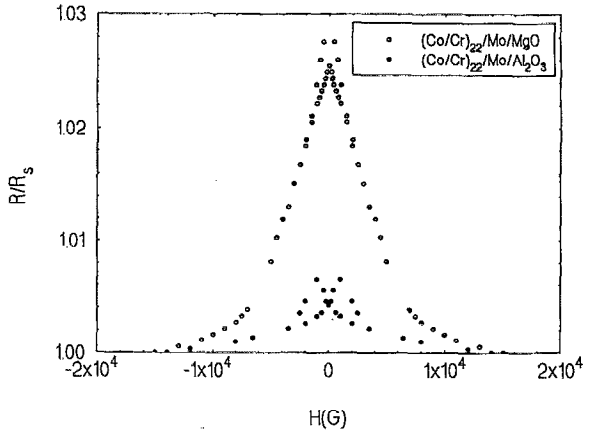


FIG. 5. The Normalized resistivity as a function of applied magnetic field at 10 K for (a) $(Co/Cr)_{22}/Mo/Al_2O_3$, and (b) $(Co/Cr)_{22}/Mo/MgO$ superlattices.

gible, if it is compared with the variation due to the addition of buffer layers of Cr and Mo. However, the exact mechanism of this reduction in AMR is not clear at present. It is noted that the buffer layers are not thick enough to shunt enough current to explain this reduction.

For Co/Cr multilayer samples, we have selected MgO(100), and $Al_2O_3(1\bar{1}02)$ as substrates. The thickness of each layer is varied from 4 to 30 Å for Cr, and from 20 to 40 Å for Co. We chose Cr as the first layer to form the multilayer structures. For samples without a Mo buffer layer, we found that the multilayers we made always had polycrystal structure if the thickness of the first Cr layer was less than 20 Å. This result tells us that it is difficult to grow epitaxial Co/Cr multilayers on either MgO or Al_2O_3 with Cr thickness less than 20 Å and Co thickness less than 60 Å. Therefore, we selected Mo as a buffer layer (about 100 Å) on both MgO and Al_2O_3 substrates to study the epitaxial behavior of the Co/Cr multilayer system. From the RHEED and XRD studies, both hcp-Co and bcc-Cr layers were identified for all the multilayer samples on either MgO or Al_2O_3 substrates. For the sake of comparison, Co/Cr multilayer samples with either MgO or Al_2O_3 as substrate were epitaxially grown side by side under the same batch of crystal growth process. Any difference between these two samples should be due to the different substrate only; e.g., the interface roughness of the multilayers is one of the important factors. For explanation, Fig. 4 shows result of the x-ray reflectivity measurement on two samples: (a) $(Co_{34}\ Å/Cr_4\ Å)_{22}/Mo_{88}\ Å/Al_2O_3$, and (b) $(Co_{34}\ Å/Cr_4\ Å)_{22}/Mo_{88}\ Å/MgO$. We can readily see that the reflectivity intensity drops more rapidly for the sample grown on MgO substrate. The reflectivity formula originally derived by Parratt[4] was used for the calculation of intensity reflected from a multiple-layer film on a substrate. Interfacial inhomogeneity due to roughness was included by adding effective Debye–Waller factors to each of the layers.[5,6] From this analysis the interface roughness of the sample grown on MgO is roughly 8 times larger than that on Al_2O_3.

The normalized electrical resistance as a function of applied field at 10 K for the above two superlattice samples are shown in Fig. 5. The MR, i.e., $(R-R_s)/R_s$, is roughly about 2.72%, and 0.65% for $(Co/Cr)_{22}/Mo/MgO$ and $(Co/Cr)_{22}/Mo/Al_2O_3$, respectively. This tells us that, roughly speaking, the MR is enhanced by a factor of 4 in the Co/Cr multilayers by the interface roughness. It is suggested that the effect by the spin dependent scattering at the interfacial regions of the superlattice is larger than that due to the spin dependent scattering in the ferromagnetic layers for the magnetoresistance in the Co/Cr multilayer system.

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