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Structural and magnetic properties of epitaxially grown full-Heusler alloy 
Co$_2$MnGe thin films deposited using magnetron sputtering

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Full-Heusler alloy Co$_2$MnGe thin films were epitaxially grown on MgO (001) substrates using magnetron sputtering. The films were deposited at room temperature and subsequently annealed in situ at temperatures ranging from 400 to 600 °C. X-ray pole figure measurements for the annealed films showed (111) peaks with fourfold symmetry, which gives direct evidence that these films are epitaxial and crystallized in the $L2_1$ structure. Furthermore, cross-sectional transmission electron microscope images of a fabricated film indicated that it is single crystalline. The annealed films had sufficiently flat surface morphologies with roughnesses of about 0.26 nm rms at film thicknesses of 45 nm. The saturation magnetization of the annealed films was 4.49 $\mu_B$/f.u. at 10 K, corresponding to about 90% of the Slater–Pauling value for Co$_2$MnGe. © 2006 American Institute of Physics.

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Half-metallic ferromagnets (HMFs)$^1$ are characterized by an energy gap for one spin direction at the Fermi level ($E_F$), leading to a complete spin polarization at $E_F$. This feature of HMFs is highly favorable for ferromagnetic electrodes used in spintronic devices. Magnetic tunnel junctions (MTJs) using Co-based full-Heusler alloy thin films have recently been studied intensively$^2$–$^5$ because of the half-metallic ferromagnetic nature theoretically predicted for some of these alloys$^6$–$^9$ and because of their high Curie temperatures, which are well above room temperature (RT).$^9$ Using these thin films in MTJs requires abrupt, smooth interfaces. Hence, fully epitaxial MTJs are the most promising approach. Co$_2$MnGe (CMG) is one of the Co-based full-Heusler alloys that is theoretically predicted to be half metallic.$^6$–$^7$ Up to now, epitaxial $L2_1$-structured Co$_2$MnGe thin films have been grown on GaAs substrates using molecular beam epitaxy (MBE)$^{10,11}$ or pulsed laser deposition,$^{12}$ and spin injection from CMG into compound semiconductor heterostructures has been studied.$^{11}$ However, MTJs using a CMG thin film have not yet been developed. We recently reported a method of epitaxially growing Co-based full-Heusler alloy Co$_2$Cr$_{0.6}$Fe$_{0.4}$Al (CCFA) thin films on MgO substrates by magnetron sputtering and used it to fabricate fully epitaxial magnetic tunnel junctions with MgO tunnel barriers demonstrating relatively high tunnel magnetoresistance (TMR) ratios of 42% at RT and 74% at 55 K.$^5$ When we take into consideration that the lattice mismatch between CMG (lattice constant $a=0.5743$ nm)$^9$ and MgO (diagonal length $\sqrt{2}\times a_{\text{MgO}}=0.5997$ nm) on a 45° in-plane rotation is relatively small ($\approx -3.6\%$), comparable with that between CCFA and MgO, CMG is also a good candidate to be epitaxially grown on a MgO substrate by magnetron sputtering.

In this article, we report on the preparation of single-crystal CMG thin films epitaxially grown on MgO substrates by magnetron sputtering and describe their structural and magnetic properties. A 10-nm-thick MgO buffer layer was first grown on a MgO (001) single-crystal substrate at 400 °C to reduce or eliminate surface defects and microscopic roughness using electron beam evaporation at a deposition rate of 0.01 nm/s. Then the Co$_2$MnGe film was deposited on the MgO buffer layer using rf magnetron sputtering. These layers were successively deposited in an ultra-high vacuum chamber (with a base pressure of around $2\times 10^{-7}$ Pa) incorporating magnetron sputtering with electron beam evaporation. The sputtering target used for preparing CMG thin films was an off-stoichiometric one. The CMG layers were deposited at a rate of 0.075 nm/s at RT and subsequently annealed in situ at 400–600 °C for 15 min. As a sputtering gas, Ar was used at a pressure of 0.1 Pa. The CMG film composition was analyzed using inductively coupled plasma optical emission spectroscopy for a 100-nm-thick CMG film deposited at RT. The CMG film composition was determined to be Co$_{2.00}$Mn$_{1.05}$Ge$_{1.17}$ with an accuracy of 2%–3% for each element. Hereafter we represent the film composition as Co$_2$MnGe (CMG).

The structures of the fabricated CMG thin films were characterized using in situ reflection high-energy electron diffraction (RHEED), x-ray Bragg scans, x-ray pole figure measurements (Bruker AXS D8 DISCOVER Hybrid), electron diffraction, and cross-sectional high-resolution transmission electron microscope (HRTEM) observation. The surface morphologies were observed using atomic force microscopy (Digital Instruments). Magnetic properties were measured using a superconducting quantum interference device magnetometer (Quantum Design MPMS) at temperatures from RT to 10 K. For the estimation of magnetization ($M$), the contribution from the MgO substrate was subtracted.

X-ray $\theta$–$2\theta$ Bragg scans for as-deposited 45-nm-thick CMG films showed (004) peaks of CMG with low intensity, while clear (004) peaks with significantly higher intensity...
were observed for 400–600 °C-annealed films. In fact, the (004) peak of the 400 °C annealed film was about ten times more intense than that of the as-deposited film, and the (004) peak intensity further increased as the annealing temperature was increased from 400 to 600 °C. In the sense that the (004) peak intensity of CMG film was significantly increased by post-deposition annealing at 400–600 °C, the crystal structure of CMG film was improved by the annealing.

An x-ray pole figure scan of a CMG film deposited at RT and subsequently annealed at 600 °C is shown in Fig. 1, along with one of the MgO substrate. As can be seen in Fig. 1(a), MgO (111) diffraction peaks with fourfold symmetry with respect to the sample rotation angle, φ, were observed at a tilt angle, χ, of 54.7°. Here, we set the MgO [100] direction to the origin of the φ axis. Figure 1(b) shows the fourfold symmetry of the CMG (111) peaks at χ=54.7°, which gives direct evidence that the 600 °C annealed film is epitaxial and crystallized in the L2₁ structure. Because the φ values for the CMG (111) peaks were shifted by 45° with respect to those of the MgO (111) peaks, the crystallographic relationship was CMG (001)[100] || MgO (001)[110]. This relationship is reasonable in view of the relatively small lattice mismatch (3.6%) between CMG (001) and MgO (001) on a 45° in-plane rotation. In contrast, in the x-ray pole figure scans for the as-deposited CMG thin films, (002) peaks were observed, while (111) peaks were not observed. These results indicate that the as-deposited CMG films were also grown epitaxially but with the B2 structure. By annealing the as-deposited CMG film at temperatures ranging from 400 to 600 °C, the (111) peaks specific to the L2₁ structure appeared and their intensity increased with increasing annealing temperature. Furthermore, microbeam electron diffractograms indicate that the fabricated film annealed at 600 °C had the L2₁ structure with some residual regions of the B2 and A2 structures.

We also confirmed from the in situ RHEED observation along the two azimuths of [100]MgO and [110]MgO that the as-deposited CMG film on the MgO buffer layer had been grown epitaxially. The streak patterns of the CMG films annealed in situ at 600 °C were sharper and more distinct than those of the as-deposited film.

Figure 2 shows a cross-sectional HRTEM lattice image, along the [110] direction of CMG, of a heterostructure consisting of a 50-nm-thick CMG film annealed at 600 °C and a 10-nm-thick MgO buffer layer. It clearly indicates that both the MgO buffer layer and the CMG layer were epitaxially grown and are single-crystalline. Furthermore, as is shown in Fig. 2, the interface between the CMG layer and MgO buffer layer was abrupt and highly smooth, although dislocations formed to relax the lattice mismatch of ~3.6% between CMG and MgO were observed at some parts of the CMG/MgO interface.

Lower ferromagnetic electrodes with little surface roughness must be prepared to fabricate high-quality MTJs. Figure 3(a) shows a typical 3D surface image of a 45-nm-thick CMG film deposited on a 10-nm-thick MgO buffer layer at RT and subsequently annealed at 600 °C. As shown in the figure, the film had sufficiently flat surface morphologies with roughness of about 0.26 nm rms. The annealing temperature dependence of the surface roughness of the 45-nm-thick CMG film is shown in Fig. 3(b). The figure clearly shows that post-deposition annealing improved the surface roughness from that of 0.40 nm rms for as-deposited films, which is consistent with the in situ RHEED observations. For comparison, we also prepared as reference samples CMG films that were directly deposited at a substrate temperature of 400 °C and measured their surface roughness. The rms values of roughness of those films were about 0.72 nm or more. These results clearly indicated that the fabrication procedure of depositing films at RT and subsequently annealing in situ is advantageous for obtaining sufficiently flat surface morphologies in these films.
Figure 4 shows the magnetization curves up to a magnetic field, $H$, of 5000 Oe for 45-nm-thick CMG thin films annealed at temperatures ranging from 400 to 600 °C. The saturation magnetization, $\mu_s$, was extracted from the hysteresis curves. The annealing temperature dependences of $\mu_s$ and coercive force, $H_c$, for 45-nm-thick films at 300 and 10 K are shown in Fig. 5. For as-deposited films, the $\mu_s$ observed at 10 K was 1.71 $\mu_B$/f.u. for the annealed films correspond to about 90% of the theoretically predicted Slater–Pauling value of 5.0 $\mu_B$/f.u. for CMG with the $L_2_1$ structure. Furthermore, the $H_c$ value of approximately 100 Oe at 300 K (~300 Oe at 10 K) for as-deposited films significantly decreased after post-deposition annealing, resulting in $H_c$ values ranging from 3.9 Oe (400 °C annealed) to 29 Oe (600 °C annealed) at 300 K (ranging from 6.4 to 32 Oe at 10 K). The $H_c$ value of 29 Oe observed at 300 K for the 600 °C annealed film is comparable with that of the CMG film (approximately 40 Oe at RT) grown using MBE on a GaAs substrate. The pronounced improvement of the magnetic properties in terms of $\mu_s$ and $H_c$ using postdeposition annealing is probably related to the crystal structure improvement in terms of the increased (004) peak intensity and the appearance of the $L_2_1$ structure. The MgO buffer layer (10 nm)/CMG lower electrode (50 nm)/MgO tunnel barrier (2.5–3.0 nm)/Co$_5$Fe$_{50}$ (CoFe) upper electrode (10 nm), grown on an MgO (001) substrate. The MgO tunnel barrier was deposited by electron beam evaporation at RT. The cross-sectional HRTEM image of a fabricated MTJ layer structure distinctively indicated that all layers, including the CMG lower electrode, the MgO tunnel barrier, and the CoFe upper electrode, were grown epitaxially and single crystalline. The microfabricated MTJs exhibited the strongly temperature-dependent TMR ratios of 14% at RT and 70% at 7 K. The spin-dependent magnetotransport properties of the fabricated MTJs will be reported separately. In summary, single-crystal full-Heusler alloy Co$_2$MnGe thin films were epitaxially grown on MgO (001) substrates using magnetron sputtering. The Co$_2$MnGe thin films were deposited at RT and subsequently annealed in situ at 400–600 °C. X-ray pole figure measurements showed that the annealed films were epitaxial and crystallized in the $L_2_1$ structure. The annealed films had sufficiently flat surface morphologies. The saturation magnetization of the annealed films was 4.49 $\mu_B$/f.u. at 10 K, corresponding to about 90% of the Slater–Pauling value for Co$_2$MnGe. Given these epitaxial Co$_2$MnGe thin films, fully epitaxial magnetic tunnel junctions using a Co$_2$MnGe thin film and a MgO tunnel barrier were fabricated.