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Defect-band mediated ferromagnetism in Gd-doped ZnO thin films

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Gd-doped ZnO thin films prepared by pulsed laser deposition with Gd concentrations varying from 0.02–0.45 atomic percent (at. %) showed deposition oxygen pressure controlled ferromagnetism. Thin films prepared with Gd dopant levels (<Gd 0.112 at. %) at low oxygen deposition pressure (<25 mTorr) were ferromagnetic at room temperature. Negative magnetoresistance, electric transport properties showed that the ferromagnetic exchange is mediated by a spin-split defect band formed due to oxygen deficiency related defect complexes. Mott’s theory of variable range hopping conduction confirms the formation of the impurity/defect band near the Fermi level.

I. INTRODUCTION

Diluted magnetic semiconductors (DMS), which are capable of spin injection and detection of selectively polarized spins of either up/down orientation, are suitable for embedding within currently successful semiconductor technologies, thereby achieving multifunctional spintronic devices. ZnO is an excellent candidate for achieving such predicted multifunctionalities due to its inherent properties, such as wide band gap, transparency, and ability to grow large high quality crystals. Previous literature reported that room temperature ferromagnetism (RTFM) was reported for 3d-transition metal doped ZnO. However, the origin of RTFM and optimization of conditions for reproducibility are still equivocal. Dopant segregations and/or defect-induced magnetism, achieving a high Curie temperature and high magnetic moment in doped ZnO, are the most prominent issues.

Rare-earth elements (RE) possess larger magnetic moments and magneto-crystalline anisotropy compared to the 3d TM. In addition, a report on the observation of a colossal magnetic moment (4000 µB) upon Gd-doping in GaN has prompted efforts to investigate the magnetism in Gd-doped ZnO (Gd:ZnO). Earlier theoretical and experimental studies have brought to focus two important prerequisites to observe RTFM in these films by invoking the relation of possible exchange mechanisms with spin splitting of the impurity/defect band.

II. EXPERIMENTS, RESULTS, AND DISCUSSIONS

Gd:ZnO thin films were prepared from ceramic targets with nominal concentrations: 0.05, 0.15, 0.25, 0.375, 0.5, and 1.0 wt. % of Gd2O3, on lattice matched (0.08%) a-plane sapphire substrates by PLD technique (KrF laser with λ = 248 nm, energy = 520 mJ, and substrate temperature = 650 °C). We set the laser energy set (520 mJ per pulse) in the laser module. This energy is decreased by 6%–7% at the target’s position inside the PLD chamber. Therefore, the average effective laser energy on the target was ~485 mJ per pulse. Henceforth, the samples prepared from these targets are referred to as 0.02, 0.06, 0.11, 0.16, 0.22, and 0.45 at. % Gd:ZnO, respectively. Eight sets of thin films were prepared by varying Gd% and oxygen pressure (P0) from 5 to 500 mTorr. Gd concentrations in the thin films were estimated by Rutherford backscattering spectrometry (RBS, Kobelco HRBS-V500). A typical RBS spectrum for 0.11 at. % Gd:ZnO thin film revealed a close estimation of 0.15 at. % Gd, which confirms that a consistent cation transfer ratio was achieved from the ceramic targets to the thin films by the PLD.

X-ray diffraction (XRD) patterns show that all Gd:ZnO thin films were grown along the c-axis with peaks corresponding to the thin films with a smooth and continuous growth of the c-axis.
only to (0002) planes. Typical phi (φ)-scan of the (10–11) planes shows its six-fold symmetry, confirming the epitaxial growth of the Gd:ZnO thin films. The XRD wide 2θ < 20 > 90° scan (fig. not shown), within the instrument resolution limit of ~2 at. %, does not reveal any secondary phases. XRD patterns of the 0.06 at. % Gd:ZnO show (Fig. 1(a)) a gradual peak shift towards higher (2θ) angles (from 34.43° to 34.56°) with increasing Pd from 5 to 100 mTorr. This shift to higher angles with increasing Pd indicates an increase in oxygen incorporation into the growth film. A similar shift is observed with increasing Gd concentration from 0.05 to 0.16 at. % (2θ from 34.43° to 34.59°) in Gd:ZnO thin films (Fig. 1(b)). However, this trend is counterintuitive for the direct substitution of Zn²⁺ (ionic radius 0.74 Å) with Gd³⁺ (ionic radius 0.94 Å). A similar shift in the peak position with increasing Gd% was already observed in Gd-doped ZnO and was attributed to tensile stress and structural defects. The RBS measurements indicated Gd substitution onto Zn tetrahedral sites. Near edge X-ray absorption fine structure (NEXAFS) spectra at the O K-edge and Gd M₅,4-edges (Figs. 1(c) and 1(d)) also indicated substitution of Gd in ZnO.

Magnetization measurements were carried out in a SQUID-vibrating sample magnetometer (SVSM, Quantum Design). The recorded magnetization data were corrected by subtracting the diamagnetic response from the sapphire substrate. Fig. 2 shows the magnetization hysteresis loops of the Gd:ZnO thin films prepared with different Gd concentrations (Figs. 2(a)–2(c)) and at different Pd (Pd = 25 mTorr), as shown in Figs. 2(c) and 2(d). Magnetization loops exhibited hysteresis in the temperature range of 5–380 K, confirming the RTFM nature of these films. The 0.11 at. % Gd:ZnO sample (Pd = 5 mTorr) exhibits the highest coercivity (Hc) and the highest carrier density (nₑ) among all the samples. Fig. 3(a) shows that Hc increases as nₑ increases, suggesting that the carriers play a role in mediating the exchange. Based on the RBS estimation, magnetic moment of 0.11 at. % Gd:ZnO (Pd = 5 mTorr) is estimated to be ~5 μ_B/Gd at 5 K, which is less than the free atomic moment of Gd (7 μ_B). Samples deposited in higher Pd (with the same Gd%) do not exhibit FM (the figure is not shown).

Typical zero-field cooled (ZFC) and field cooled (FC) magnetization measurements (Figs. 2(e)–2(h)) show no transitions corresponding to any secondary phases, such as pure Gd (T_C = 293 K), GdZn or GdZn₂ (T_C ≈ 70 K) that appeared in the ZFC-FC in Gd:ZnO thin films reported elsewhere.
High amount of Gd doping (~2.5 at\%) in ZnO showed segregation of a pure Gd near the film-substrate interface in the transmission electron microscopy studies.\textsuperscript{22} We did not observe such anomalies by using similar experimental analyses in the present case, suggesting that these samples are relatively free from aforementioned secondary phases.

In the present study, only those thin films prepared with Gd\% ≤ 0.11 at.\% and Pd ≤ 25 mTorr were ferromagnetic in their as-deposited condition. On the other hand, the thin films with higher Gd concentration (Gd\% > 0.2) prepared at any Pd\(\text{d}\) were not ferromagnetic. However, upon vacuum annealing (VA) on selected (with high (low) Gd\% and low (high) Pd\(\text{d}\)), non-magnetic thin films become ferromagnetic. Furthermore, undoped ZnO thin films prepared and treated under similar conditions did not show FM.\textsuperscript{21} We propose low Pd\(\text{d}\) or vacuum annealing to introduce more defects related to oxygen deficiency (O\(_{\text{defect}}\)). We therefore believe that the combination of the low Gd concentration and high concentration of oxygen deficiency defects (due to low Pd or vacuum annealing) plays a role in establishing the FM. In addition, we carried out in-situ VA for the 0.22 at.\% and 0.45 at.\% Gd:ZnO thin films during the X-ray magnetic circular dichroism (XMCD, in BL13–1 at SSRL, USA) experiments. Initially the XMCD spectra were collected at room temperature (RT), then the samples were annealed (at 155.69.2.179 On: Mon, 09 Mar 2015 08:06:49

![FIG. 3. (a) Variation of H\(_{\text{c}}\) with carrier density (n\(_e\)) in ferromagnetic Gd:ZnO. Gd at.\% are given next to the symbols. (b) Anisotropic MH loops for 0.02 at.\% Gd:ZnO (Pd\(\text{d}\) = 25 mTorr). (c) Variation of c-parameter (d) M\(_{\text{c}}\), and (e) density of localized states with Gd\% in Gd:ZnO thin films. Pd\(\text{d}\) is given in parenthesis. Open symbols represent VA samples.](Image)

Anisotropic magnetization loops are observed in 0.02 at.\% Gd:ZnO film (Fig. 3(b)), revealing higher M\(_S\) when the field (H) is applied parallel to c-axis relative to the perpendicular to c. As Gd is known to have strong magnetocrystalline anisotropy,\textsuperscript{23} the magnetic anisotropy observed in these samples could arise from the Gd\(^{3+}\) ions due to its 4f electron that cloud experience a tetrahedral crystalline electric field in the lattice and associated spin splitting.\textsuperscript{24} From the variation of c-lattice parameter and M\(_S\) decrease as Gd concentration increases including for annealed samples. We therefore propose that O\(_{\text{defect}}\) is created via low Gd doping at low Pd\(\text{d}\) or subsequent VA to establish a long range FM.

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An important feature of Gd distinguishing it from other RE atoms is the partially filled 5d and 4f levels, which could activate inter- and intra-ion exchange interactions.\textsuperscript{26} The magnetic coupling mechanism responsible for the RTFM in Gd:ZnO could be: (i) inter-Gd ions (i.e., between neighboring Gd ions) via 5d-5d exchange and intra-Gd ions via 4f-5d exchange, (ii) Gd-Zn (f-s), and (iii) Gd-O (f-p). In addition, s-f exchange can be stronger than f-f or f-p exchange.\textsuperscript{8,27} One of the concerns here is related to the location of the Fermi level (E\(_F\)) and 4f levels in the band, which is required to establish such magnetic exchange. Earlier theoretical studies showed that the 4f levels lie inside the conduction band (CB) (spin-down) and valence (spin-up) bands in Gd:ZnO.\textsuperscript{27} For pure ZnO, E\(_F\) lies in the ZnO band gap and introduction of intrinsic donor defects such as V\(_{\text{Zn}}\)\textsuperscript{28} does not shift (E\(_F\)) into the CB. However, doping ZnO with Gd shifts the E\(_F\) into the CB.\textsuperscript{21,29} Therefore, the location of E\(_F\) inside the CB can assist in establishing the magnetic exchange interaction.
between donor states (defect states or states of impurity-defect complexes). This impurity/defect band formed near the CB can be due to Gd-O\(_{\text{defect}}\) complexes or intrinsic O\(_{\text{defect}}\) donors, which facilitate long range exchange interaction. As the Gd 5\(d\) electrons contribute to the bottom of the CB, 30\( s\-d\) exchange between Gd\(^{3+}\) ions can be mediated by impurity/defect band located close to the CB edge. 30 This defect band has been reported to have spin-split due to the s-\(d\) exchange. 31 In addition, FM originating from the vacancy related defects in the grain boundaries 32 and associated dangling bonds 33 in the present case cannot be completely excluded because spin-split can be due to the result of an exchange interaction between these intrinsic defects. 33 If the spin-splitting is larger than the band width, long range spin-polarization could be realized. 34

The formation of a defect band and its spin-splitting can be envisaged via electrical transport studies. Therefore, we carried out magnetoresistance (MR) and Hall effect experiments for the Gd:ZnO thin films in a physical property measurement system (PPMS, Quantum design). All the samples were ‘\(n\)’ type, with carrier concentrations varying from 8.4 \(\times\) 10\(^{17}\) to 1.9 \(\times\) 10\(^{19}\) cm\(^{-3}\) (Fig. 3(a)). Therefore, it is essential to study the effect of the density of defect states on the magnetic exchange. Temperature dependence of resistivity of all thin films in the present study was typical of a semi-conductor, whereby an exponential increment of resistivity was observed (Fig. 3(b)). At high temperatures (\(T > 180\) K), a thermally activated \(\rho(T) = \rho_0 \exp(\Delta E/k_B T)\) type conduction (fitting not shown) is dominant in all the films. However, at low temperatures, the temperature dependence of resistivity of these thin films followed Mott’s variable range hopping (VRH) due to localization behavior of carriers below a metal insulator transition (MIT). 35 According to VRH theory, localized carriers form an impurity band and the low temperature conductivity is governed by a phonon-assisted hopping of carriers between localized states and the density of localized states at \(E_F\) (\(N(E_F)\)) can be calculated by 35

\[
\rho = \rho_0 \exp \left[ \left( \frac{T_0}{T} \right)^{1/4} \right],
\]

where \(T\) is temperature, \(\rho_0 = (3e^2\nu_{\text{ph}})^{-1}(N(E_F)/8\pi k_B T)^{-1/2}, T_0 = (\nu_{\text{ph}}^2/k_B N(E_F))\), \(\nu_{\text{ph}}\) is phonon frequency at Debye temperature (\(\approx 10^{12}\) s\(^{-1}\)), \(\nu_{\text{ph}}\) is a dimensionless constant (\(\approx 16\)). 36 Fig. 5(a) shows linear \(T^{1/4}\) dependence of \(\ln(\sigma T^{1/2})\) at low T. The best fit (residual sum of squares \(< 0.02\)) to the function shown in Eq. (1) confirms VRH conduction 35 in the Gd:ZnO thin films. With respect to confirming the Mott’s VRH mechanism, the figure of merit should be satisfied: (i) a\(R\) value should be more than the unity (2\(R > 1\)), and (ii) the hopping energy \(W\) should be larger than \(k_B T\). 37 These values are fulfilled for our samples (3.2\( < \alpha R < 2.25\) and 1.86\(< W < 1.29\)), which indicate the formation of an impurity/defect band near \(E_F\). The location of the \(E_F\) within the band of localized carriers is controlled by Gd dopants as shown in a previous theoretical study. 29

**FIG. 5.** (a) \(T^{-1/4}\) variation of conductivity showing VRH conduction \(\rho_a\) is given within parenthesis; open symbols represent post deposition VA samples. MR fit for (b) 0.067 at. % and (c) 0.11 at. % Gd:ZnO thin films.

comparison of the density of localized states \(N(E_F)\) (deduced from the fitting shown in Fig. 5(a) and Eq. (1)), c-lattice parameter, and \(M_g\) (Figs. 5(a) and 5(b)) shows a monotonic decrement with increasing Gd, thus confirming the relation between the impurity/defect band and the observed FM.

Typical MR for the Gd:ZnO (with 0.06 at. % and 0.11 at. % Gd:ZnO) thin films prepared at \(P_a = 25\) mTorr are shown in Figs. 5(b) and 5(c), respectively. A large negative MR was observed for the 0.11 at. % Gd:ZnO (\(P_a = 25\)mTorr) and a positive MR component was also found to coexist with negative components above 100 K (Fig. 5(b)). Negative MR in the doped semiconductors is generally attributed to the presence of localized magnetic moments, 39,40 impurity band conduction, or degenerate semiconducting states (when \(E_F\) is located inside the CB). 41 In order to understand the MR behavior, we fit MR data using an empirical model proposed by Khosla-Fischer 42 given by

\[
\frac{(R_{H} - R_{H=0})}{R_{H=0}} = -a^{2} \ln(1 + b^{2}B^{2}) + \frac{(c^{2}B^{2})}{1 + d^{2}B^{2}},
\]

where \(a = (A_1 J_\delta [S(S + 1) + \langle M^{2}\rangle])\) and \(b = (1 + 4S^{2} \pi^{2} (\frac{24\mu_B}{8})^{4} (\frac{\mu_B}{k_B T})^{2})\) are negative and positive scattering parameters, respectively. 43 \(B\) is magnetic field, \(c\) is a conductivity-dependent parameter, \(d\) is carrier-mobility-dependent parameter, \(A_1\) is spin-scattering amplitude, \(J\) denotes as an exchange parameter, \(\delta\) corresponds to density of states at \(E_F\), \(S\) is spin of the localized moment, \(\langle M\rangle\) refers to an average effective magnetic moment, \(g\) is Landé g-factor, \(\mu_B\) denotes as Bohr magneton, and \(x_1\) is a constant (0.1 to 10). Parameters ‘\(c\)’ and ‘\(d\)’ in Eq. (2) are dependent on the conductivity and mobility of the electrons under the influence of Lorentz force. The first term in Eq. (2) indicates negative MR, 44,45 which accounts for the spin-dependent scattering of localized electrons in the impurity band and weaken as the field-induced spin-polarization progresses. Therefore, negative MR shows the presence of an impurity band in these films. 44,45 The second term accounts primarily for the positive MR, which is related to two-band conduction (via CB and impurity-O\(_{\text{defect}}\) band). 42
The negative scattering is found to be dominant in all the thin films below 50 K, which is an indication of spontaneous spin-polarization of localized spins at the isolated paramagnetic ions46 (Gd³⁺ in the present case). Using Eq. (2) and the expression for ‘a,’ it can be established that the magnitude of negative MR is proportional to the square of the average effective magnetic moment of the localized spins. Therefore, we attribute the large negative MR (14%) and higher M, found in 0.22 at. % Gd:ZnO (Pₐ = 25 mTorr) thin film to the largest spin-splitting of the impurity-O_{defect} band states. Therefore, we conceive that appropriate concentration of Gd carriers (nₐ), and oxygen deficient defects are requisite conditions for strong ferromagnetic coupling in Gd:ZnO thin films.

In summary, we found that Gd (<0.11 at. %) doped ZnO thin films prepared by PLD at low oxygen pressures exhibited ferromagnetic behavior at RT. Gd doping is essential for moving E_F inside the CB, which may cause spin-splitting in the band. Based on the negative magnetoresistance and low temperature hopping conduction studies, we conclude that a spin-split defect band mediates the FM exchange formed due to either O_{defect} or defect-Gd complexes. Our study confirms that appropriate concentration of Gd, defects associated with oxygen deficiency, and localized carriers are prerequisites to achieving RTFM in Gd-doped ZnO.

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10See supplementary material at http://dx.doi.org/10.1063/1.2490558 for the spectral yield of RBS measurements for Gd doped ZnO (Figure S1) and the typical XRD-phi scan for Gd doped ZnO thin films (Figure S2).
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