Colossal magnetoresistive and ferroelectric thin films deposited by excimer laser induced plasma

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Received 15 February 2001; accepted 8 May 2001

Abstract

We prepared the colossal magnetoresistive $\text{La}_{0.8}\text{Sr}_{0.2}\text{MnO}_3$ thin films on single crystal MgO, SrTiO$_3$ and LaAlO$_3$ substrates by KrF excimer pulsed laser deposition technique. The thickness dependence of the structural, electric and magnetic properties of the LSMO films on different substrates is reported. The integration of the ferroelectrics with the colossal magnetoresistance thin film was achieved. The plasma plume ablated from target was investigated by intensified charge coupled device and optical emission spectroscopy. © 2001 Elsevier Science Ltd. All rights reserved.

Keywords: Colossal magnetoresistance; La$_{1-x}$Sr$_x$MnO$_3$; Pulsed laser deposition; Magnetoresistance ratio; Ferroelectric; PbZr$_{0.52}$Ti$_{0.48}$O$_3$; SrBi$_2$Ta$_2$O$_9$

1. Introduction

In recent years, the colossal magnetoresistance (CMR) of perovskite manganite $\text{Re}_{1-x}\text{Ae}_x\text{MnO}_3$ (Re: rare earth elements, Ae: alkaline earth elements) thin films have been extensively investigated for their application to magnetic sensor working at low magnetic field and room temperature. Ferromagnetic–nonmagnetic superlattice have been shown to exhibit magnetoresistance referred to as giant magnetoresistance (GMR). The magnetoresistance of the GMR is anisotropic. However, the CMR shows isotropic magnetoresistivity that is larger than that of the GMR. $\text{Re}_{1-x}\text{Ae}_x\text{MnO}_3$ can be described as $(\text{Re}_{1-x}\text{Ae}_x\text{Mn}_2^{3+}\text{Mn}_3^{4+})\text{O}_3$. The doping of divalent Ae element into the antiferromagnetic insulator $\text{Re}_{1-x}\text{MnO}_3$ induces a transition to a ferromagnetic metallic state. The maximum magnetoresistance (MR) ratio of CMR materials appears near the Curie temperature ($T_C$). The MR ratio is defined as $(\rho_H - \rho_0)/\rho_0$, where $\rho_0$ and $\rho_H$ are the resistivities measured at zero field and at $H$, respectively. The electric and magnetic properties of CMR oxides can be modified by average ionic radius of perovskite A-site [1–3], Mn$^{3+}$/Mn$^{4+}$ ratio (x) [4], oxygen content [5–7], grain size [8], and mechanical strain. $\text{La}_{1-x}\text{Ca}_x\text{MnO}_3$ (LCMO) oxides show very large MR ratio only in high magnetic field of several Tesla and temperature lower than 300 K, which limits their practical utility [9–11]. On the other hand, $\text{La}_{1-x}\text{Sr}_x\text{MnO}_3$ (LSMO), 0.18 < x < 0.3 has $T_C$ above room temperature and is a potential candidate for room temperature device [12–14]. We have prepared LSMO (x = 0.2) thin films using pulsed laser deposition (PLD) [15,16]. PLD method has been used for epitaxial growth of multicomponent oxide films because of its stoichiometric transfer from target to substrate. Recently, there have been several reports on the changes of structural, electrical and magnetic properties of CMR thin films resulting from the strain induced by lattice mismatch between the substrate and the film [17–20]. The LSMO films were deposited on single crystal substrates of MgO, SrTiO$_3$ (STO) and LaAlO$_3$ (LAO). The lattice mismatches of $(\alpha_{\text{LSMO}} - \alpha_{\text{MgO}})/\alpha_{\text{MgO}}$, $(\alpha_{\text{LSMO}} - \alpha_{\text{STO}})/\alpha_{\text{STO}}$ and $(\alpha_{\text{LSMO}} - \alpha_{\text{LAO}})/\alpha_{\text{LAO}}$ are −7.8, −0.5 and +2.4%, respectively. We also prepared ferroelectric PbZr$_{0.52}$Ti$_{0.48}$O$_3$ (PZT) and SrBi$_2$Ta$_2$O$_9$ (SBT)/LSMO heterostructured films [21]. The PZT has high remanent polarization and the SBT has been attractive because of its switching endurance [22,23]. The CMR thin film is of perovskite type and the lattice mismatch between this CMR film and the ferroelectric film is very small. The lattice mismatch is believed to strongly affect near the interface between the film and the substrate. It makes CMR thin film a candidate for electrode of ferroelectric capacitor. The ferroelectric/CMR stuck film
has possibilities for new devices working under electric and magnetic field. During the deposition, the plasma plume of laser ablated LSMO target was observed by intensified charge coupled device (ICCD) and optical emissive spectroscopy (OES). We report on the studies of the thickness dependence of the properties of LSMO films fabricated on different substrates.

2. Experimental details

Fig. 1 shows a schematic diagram of the PLD equipment with OES measurement system. The base pressure in the stainless chamber \((p = 280 \text{ mm})\) was \(10^{-6} \text{ Torr}\). KrF excimer laser (Lambda Physik LPX305icc, pulse duration \(= 25 \text{ ns}\), \(\lambda = 248 \text{ nm}\), maximum energy \(= 850 \text{ mJ}\)) was used to ablate a stoichiometric target of \(\text{La}_{0.6}\text{Sr}_{0.4}\text{MnO}_3\).

The LSMO films were deposited on single crystal substrates of the \(\text{MgO} (a = 4.21 \text{ Å})\), \(\text{STO} (a = 3.89 \text{ Å})\) and \(\text{LAO} (a = 3.79 \text{ Å})\) with a laser energy density of \(2 \text{ J/cm}^2\), repetition rate of 5 Hz, substrate temperature of \(850^\circ \text{C}\), oxygen pressure of 500 mTorr and the distance between the target and the substrate was 40 mm. After the deposition, the film was cooled to room temperature in \(760 \text{ Torr of oxygen}\). During the deposition, the images of laser produced plasma plume were observed by ICCD camera (Princeton Inc. ITE/CCD).

Optical emission from the plasma plume was collected by a quartz lens \((f = 100 \text{ mm})\) and led to monochromator/spectroscopy (ARC SpectralPro-308i) via an optical fiber. The crystallization of the deposited films was studied using X-ray diffractometer (XRD: Rigaku RINT2000/PC) with \(\text{CuK}_\alpha\) radiation. The surface morphologies were observed by atomic force microscopy (AFM: Seiko Instruments Inc. SP3800N). The electric and magnetic properties were measured using four terminal resistance techniques. The ferroelectric PZT was deposited by ablating \(\text{PbZr}_{0.52}\text{Ti}_{0.48}\text{O}_3\) target on the LSMO film. The deposition conditions of PZT films were: laser energy density of \(2 \text{ J/cm}^2\), repetition rate of 5 Hz, substrate temperature of \(550^\circ \text{C}\), oxygen pressure of 100 mTorr and the target-substrate distance of 40 mm. The deposition was followed by annealing in 600 Torr of oxygen at \(400^\circ \text{C}\) for 1 h. The ferroelectric SBT was fabricated in 100 mTorr of oxygen. The substrate temperature was varied from 550 to \(750^\circ \text{C}\) and cooled down to room temperature after the deposition. Surface morphologies of the deposited films were studied by AFM.

3. Results and discussion

3.1. Observation of the plasma plumes from the LSMO target

The behavior of the plasma plume from the LSMO target was observed during the deposition of films. Fig. 2(a) shows ICCD images of the expanding plasma plume in oxygen at a pressure of 500 mTorr. The plasma plume in vacuum observed for comparison is shown in Fig. 2(b). The gate width of 20 ns was kept fixed. Delay time \((t)\) after the ablation was varied using a delay pulse generator (Princeton Inc. PG-200). The plumes observed at \(t = 100 \text{ ns}\) show the same shapes in both oxygen and vacuum. Plasma plumes in

![Fig. 1. Schematic layout of pulsed laser deposition and the optical emission spectroscopy set up.](image1)

![Fig. 2. The ICCD images of the plasma plumes ablated from the LSMO target in (a) oxygen of 500 mTorr and (b) vacuum.](image2)
vacuum could be observed up to 5 μs whereas in 500 mTorr of oxygen, it existed up to 10 μs. This is attributed to the slowing down of plume due to collisions of plasma species with ambient gas. Fig. 3 shows the variation of plume front at varying time. The shock front of the plume moves with a driving velocity of about 3 mm/μs. The typical driving velocity of 10 mm/μs plume generated from a superconducting YBa$_2$Cu$_3$O$_{7-x}$ at 100–200 mTorr of oxygen has been reported [24]. The velocity of 3 mm/μs is low, because of the relatively high pressure of 500 mTorr. The OES measurements were done in the wavelength range of 300–800 nm. Fig. 4 shows optical spectra in the wavelength range of 400–500 nm recorded in 500 mTorr oxygen at a distance, d from the target of 10 mm at t = 1000 ns. Many intense lines from emissive species are observed. LaII and SrII lines dominate in the spectra. Since the ejected species are in excited states, the MnO band at 558.2 nm [25] could not be recognized. At the position of the substrate (d = 40 mm), we could find the emissive species having very fast velocity at t = 100 ns by OES as well as ICCD. The emission spectra recorded at 40 mm could only be observed at delay times less than or equal to 100 ns. Independent studies on ablation of Mn tip showed emission lines at 40 mm and delay time of 100 ns. However, lines corre-

Fig. 3. Variation of position of shock front with delay time.

Fig. 4. The optical emission spectra of the plasma plume of LSMO in 500 mTorr oxygen at the d = 10 mm.

Fig. 5. Thickness dependence of the θ–2θ scan of the LSMO/MgO.

3.2. Thickness dependence of the properties for LSMO thin films

The LSMO films prepared on the MgO substrates showed c-axis orientation in spite of the large lattice mismatch of −7.8%. The φ scan around the (103) LSMO peak showed four fold symmetry indicating in-plane orientation. Fig. 5 shows the θ–2θ spectra (LSMO (002)) of the LSMO (100, 300, 600 nm films)/MgO. No shift in 2θ value is observed for different thickness of the film. The c-axis lattice length estimated from 2θ = 46.7° is 3.8 Å which is same as that of bulk. The full width at half maximum (FWHM) of the LSMO (002) θ–2θ scanned peaks for 100, 300 and 600 nm LSMO films are 0.27, 0.25 and 0.24°, respectively. It is well known, according to Scherrer’s formula, the crystalline size of the film is inversely proportional to the FWHM of the θ–2θ peak. Our results show no significant change of crystalline size of different thickness films. In the case of LSMO/STO films, LSMO (002) peaks appear near STO(002) because of very small lattice mismatch of −0.5%. There is no thickness dependence of the 2θ angles at the LSMO (002) peaks for LSMO (60, 150, 240 nm films)/STO films (Fig. 6). However, for the LSMO/LAO, the lattice length of the film is larger than that of the substrate. Fig. 7 shows the thickness dependence of the θ–2θ spectra for the LSMO/LAO films with thickness of 100, 170 and 630 nm. The split peaks at LSMO (002) were observed implying two phases (phase A and phase B) with different lattice parameters in the film. The phase
A appears at $2\theta$ of 45.3°, corresponding $c$-axis lattice length being 4.00 Å, in all the films irrespective of thickness. However, the $2\theta$ angle of the phase B increases with thickness. The $c$-axis lattice lengths estimated from the $2\theta$ angles of the phase A and phase B is shown in Fig. 8. The $c$-axis lattice length of the phase B in the 100 nm LSMO film is close to that of phase A. However, it decreases as the film thickness increases and converges to 3.88 Å for the films with thickness over 630 nm. It suggests that the LSMO film is compressed in the in-plane direction and stretched along the $c$-axis direction by the lattice mismatch of +2.3% near the interface between the film and substrate. Then, the compressive strain will be relaxed with thickness. There are few reports on the Re$_{1-x}$A$_x$MnO$_3$ thin films prepared on the LAO substrates that are compressed near the interface between the film and the substrate [18–20]. It is believed that the strain is induced in the case of in-plane lattice mismatch. Fig. 9 shows the resistivity ($\rho$)–temperature ($T$) curves of the LSMO/MgO films with thickness of 100, 300 and 600 nm. Inset in the figure shows the relationship between the peak resistivity ($\rho_{max}$) and resistivity peak temperature ($T_p$). When the film thickness decreases, $\rho_{max}$ decreases and the $T_p$ increases. The $T_p$ of over 340 K is necessary for the device working at room temperature.

The $\rho_{max}$ seems to decrease with the increase of $T_p$ and this tendency is similar to that of the La$_{0.75}$Sr$_{0.25}$MnO$_3$/STO [19]. Though thickness dependence of the lattice length and the size of crystalline for the LSMO/MgO were not observed, the surface morphology changed with thickness. The estimated grain sizes of 100, 300 and 600 nm LSMO/MgO films were 70, 120 and 150 nm, respectively. The $\rho_{max}$ increased and $T_p$ shifted to lower temperature with the increase in the grain size of the film. The thickness dependence of the $\rho$–$T$ curves of the LSMO/STO thin films are shown in Fig. 10. The $T_p$ of the films are higher than 350 K owing to small mismatch between LSMO film and STO substrate. However, the LSMO films that have higher $T_p$ than the room temperature are not proper for the device applications. The $T_p$ increases as the film thickness decreases. The relationship between the $\rho_{max}$ and the $T_p$. 

Fig. 6. Thickness dependence of the $\theta$–$2\theta$ scan of the LSMO/STO.

Fig. 7. Thickness dependence of the $\theta$–$2\theta$ scan of the LSMO/LAO.

Fig. 8. Relationship between the $c$-axis lattice length and the film thickness of the LSMO/LAO.

Fig. 9. Thickness dependence of the resistivity ($\rho$)–temperature ($T$) property of the LSMO/MgO. Inset shows the relationship between the peak resistivity ($\rho_{max}$) and resistivity peak temperature ($T_p$).
seems to be the same as that of the LSMO/MgO. Fig. 11 shows the AFM images (500 nm × 500 nm) of the surface morphologies of the LSMO/STO films. We did not observe any change in the c-axis lattice length, but the grain size became larger as the film thickness increased. The grain sizes of the LSMO/STO films with the thickness of 100 and 240 nm are 30 and 50–70 nm, respectively. We noted that the grain size of our LSMO/STO is almost similar to that observed in Ref. [26]. It is established that the grain size may affect the electrical property. The grain size of the LSMO/STO was smaller than that of LSMO/MgO. Fig. 12 shows the $\rho$–$T$ curves of the LSMO/LAO (100, 170, 450, 1200 nm) thin films. Inset in the figure shows the $\rho_{\text{max}}$–$T_p$ characteristics. Unlike the LSMO/MgO and LSMO/STO, the 100 nm LSMO film has relatively low $T_p$ of 324 K and high $\rho_{\text{max}}$ of 37 m$\Omega$ cm. It suggests that the strong compressive strain of 100 nm LSMO (see Fig. 8) reduces the electronic transfer energy. The thickness of 170 nm is a threshold for the LSMO/LAO over which the induced strain is relaxed and shows the conventional transport property. The MR ratio of the LSMO films on the different substrates was measured. The MR ratio increased linearly with the applied magnetic field in the range of 0–0.4 T. The thickness dependence of the MR effect measured in $H$ of 0.4 T at room temperature is shown in Fig. 13. The magnetic field was applied perpendicular to the film. The maximum MR ratios of the LSMO films on MgO, STO and LAO are 7.3 (300 nm), 5.0 (150 nm) and 4.8% (1000 nm), respectively. The room temperature MR ratios of the thick LSMO films are small, because they have $T_p$ lower than room temperature. Thus, $T_p$ is a very important parameter for estimated magnetoresistive property of a CMR thin film. Fig. 14 shows the relationship between the MR ratio and $T_p$. It is clear that the films having $T_p$ at 340–350 K demonstrate large MR effect. We assume that adjusting the $T_p$ of the LSMO above room temperature is easier than that in the case of LCMO. The radius of Sr$^{2+}$ ion (1.27 Å) is larger than that of Ca$^{2+}$ ion (1.06 Å). Thereby, a tolerance factor of LSMO defined as $(t_A + r_O)/\sqrt{(t_B + r_O)}$, where $r_A$, $r_B$, $r_O$ represent the radii of A, B, O ions of ABO$_3$ type perovskite
[27], is very close to a perfect size matching of 1.00. This means that the octahedral MnO₆ is stabilized and the Mn–O–Mn bond angle is close to 180°. Hwang et al. reported that $T_p$ above room temperature is obtained when the average $r_A$ is larger than about 1.22 Å [28].

### 3.3. Ferroelectric/CMR heterostructured films

The ferroelectric PZT thin films were deposited on the LSMO/MgO. The thickness of the PZT and LSMO films were 400 and 200 nm, respectively. The PZT thin film grew along c-axis direction on the LSMO (00l) thin film. The polarization–electric field hysteresis curve measured by a conventional Sawyer–Tower circuit showed large remanent polarization of 22 μC/cm² and small coercive electric field of 27 kV/cm at the applied voltage of 3.5 V [16], where the LSMO thin film plays a role of bottom electrode. The dielectric constant of the PZT thin film measured by LCZ meter was 860 at frequency of $10^2$ Hz. The SBT thin films have been studied for the fatigue free ferroelectric random access memory. We prepared SBT thin films on LSMO/STO substrate at temperature ($T_s$) varying from 550 to 750°C. The $T_s$ dependence of the XRD measurement in the 2θ range of 25–55° is shown in Fig. 15. The SBT (115) and (200) peaks are observed in the film of low $T_s$ of 550 and 600°C. Above the $T_s$ of 650°C, the c-axis orientation with (00l) peaks appears. The perovskite structure of SBT is not continuous along the c-axis direction, so the SBT film having a or b axis orientation will be needed.

### 4. Conclusions

The LSMO thin films were deposited on the MgO, STO and LAO substrates using the PLD method. During the deposition process, the emissions from LaII and SrII were found to be dominant, while the emissive species did not exist at the substrate position. The $\theta$–2θ spectra of the LSMO films deposited on the MgO and STO did not show the thickness dependence, however, the $T_p$ of the $\rho$–$T$ property increased as the thickness decreased. On the other hand, it was revealed that the LSMO/LAO film has the compressive strained phase in the film and that the strain is relaxed as the thickness of the film increased. The 100 nm LSMO/LAO film that has strong strain showed a low $T_p$. A range of thickness of the LSMO films on the different substrates is found where the films exhibited the high MR ratios at room temperature. The $T_p$ of the LSMO film should be adjusted in the range of 340–350 K. The ferroelectrics/CMR heterostructured films were also prepared. The PZT/LSMO/MgO heterostructured film had excellent ferroelectric properties. The orientations of the SBT films deposited on LSMO/STO changed from (115), (200) to (00l) with increasing $T_s$.

### Acknowledgements

This work was supported by a Grant-in-Aid for Scientific
Research (1998–2000, No. 10045046) from the Ministry of Education, Science, Sports and Culture.

References

[1] M.F. Hundley, J.J. Neumeier, R.H. Heffner, Q.X. Jia, X.D. Wu, J.D. Thompson, Transport and magnetism correlations in thin-film ferromagnetic oxides, J. Appl. Phys. 79 (1996) 4535–4537.

[2] V. Moshraya, I. Khoroshan, A. Sidorenko, P. Petrenko, A. Weidinger, M. Zeitler, B. Rauschenbach, R. Tideck, Preparation of rare-earth manganese-oxide thin films by metalorganic aerosol deposition technique, Appl. Phys. Lett. 74 (1999) 2842–2844.

[3] A. Goyal, M. Rajeswari, R. Shreekala, S.E. Lofland, S.M. Bhagat, T. Boettcher, K. Kwon, R. Ramesh, T. Venkatesan, Material characteristics of perovskite manganese oxide thin films for bolometric applications, Appl. Phys. Lett. 71 (1997) 2535–2537.

[4] G.-Q. Gong, C. Camedy, G. Xiao, J.Z. Sun, A. Gupta, W.J. Gallagher, Colossal magnetoresistance of 10,000,000-fold magnitude achieved in the antiferromagnetic phase of La_{1-x}Ca_{x}MnO_{3}, Appl. Phys. Lett. 67 (1995) 1783–1785.

[5] H.L. Ju, J. Gopalakrishnan, J.L. Peng, Qi Li, G.C. Xiong, T. Venkatesan, R.L. Greene, Dependence of giant magnetoresistance on oxygen stoichiometry and magnetization in polycrystalline La_{0.65}Ba_{0.35}MnO_{3}, Phys. Rev. B 51 (1995) 6145–6146.

[6] M. Rajeswari, R. Shreekala, A. Goyal, S.E. Lofland, S.M. Bhagat, K. Ghosh, R.P. Sharma, R.L. Greene, R. Ramesh, T. Venkatesan, Correlation between magnetic homogeneity, oxygen content, and electrical and magnetic properties of perovskite manganese thin films, Appl. Phys. Lett. 73 (1998) 2672–2674.

[7] N. Sengoku, K. Ogawa, Transport properties of La_{1-x}Sr{x}MnO_{3}, thin films in oxygen deficient and excess states, Jpn. J. Appl. Phys. 35 (1996) 5432–5436.

[8] R. Mahesh, R. Mahendiran, A.K. Raychaudhuri, C.N.R. Rao, Effect of particle size on the giant magnetoresistance of La_{0.65}Ca_{0.35}MnO_{3}, Appl. Phys. Lett. 68 (1996) 2291–2293.

[9] E. Gommert, H. Cerva, A. Rucki, R. von Helmont, J. Wecker, C. Kuhn, K. Samwer, Structure and magnetoresistive properties in La-manganese thin films, J. Appl. Phys. 81 (1997) 5496–5498.

[10] R.B. Praus, B. Leibold, G.M. Gross, H.-U. Habermeier, Thickness dependent properties of La_{0.65}Ca_{0.35}MnO_{3} thin films, Appl. Surf. Sci. 138/139 (1999) 40–43.

[11] W. Zhang, X. Wang, I.W. Boyd, Enhanced magnetoresistance behaviour in CeO_{2} buffered LaCaMnO_{3} films on Si, Appl. Surf. Sci. 138/139 (1999) 563–568.

[12] A. Urushibara, Y. Morimoto, T. Arima, A. Asamitsu, G. Kido, Y. Tokura, Insulator-metal transition and giant magnetoresistance in La_{1-x}Sr{x}MnO_{3}, Phys. Rev. B 51 (1995) 14103–14109.

[13] Y. Tokura, Y. Tomioka, H. Kuwahara, A. Asamitsu, Y. Morimoto, M. Kasei, Origins of colossal magnetoresistance in perovskite-type manganese oxides, J. Appl. Phys. 79 (1996) 5288–5291.

[14] H.L. Ju, C. Kwon, Qi Li, R.L. Greene, T. Venkatesan, Giant magnetoresistance in La_{1-x}Sr{x}MnO_{3} films near room temperature, Appl. Phys. Lett. 65 (1994) 2108–2110.

[15] F. Mitsugi, T. Ikegami, K. Ebihara, J. Narayan, A.M. Grishin, The colossal magnetoresistive La–Sr–Mn–O film by pulsed laser deposition, Trans. Mater. Res. Soc. Japan 25 (2000) 357–360.

[16] F. Mitsugi, T. Ikegami, K. Ebihara, A.M. Grishin, J. Narayan, Preparation of the colossal magnetoresistive La–Sr–Mn–O thin film and application to the ferroelectric Pb–Zr–Ti–O/La–Sr–Mn–O heterostructure using pulsed laser deposition, Trans. IEEE Japan 120-A (2000) 1032–1037.

[17] J.Z. Sun, D.W. Abraham, R.A. Rao, C.B. Eom, Thickness-dependent magnetotransport in ultrathin manganite films, Appl. Phys. Lett. 74 (1999) 3017–3019.

[18] W. Prellier, A. Biswas, M. Rajeswari, T. Venkatesan, R.L. Greene, Effect of substrate-induced strain on the charge-ordering transition in Nd_{0.5}Sr_{0.5}MnO_{3} thin films, Appl. Phys. Lett. 75 (1999) 397–399.

[19] S.I. Khartsev, P. Johnsson, A.M. Grishin, Colossal magnetoresistance in ultrathin epitaxial La_{0.7}Sr_{0.3}MnO_{3} films, J. Appl. Phys. 87 (2000) 2394–2399.

[20] A.M. Haghiri-Gosnet, J. Wolfman, B. Mercier, Ch. Simon, P. Lecoeur, M. Korzenksi, R. Desfeux, G. Baldinuzzi, Microstructure and magnetic properties of strained La_{0.7}Sr_{0.3}MnO_{3} thin films, J. Appl. Phys. 88 (2000) 4257–4264.

[21] F. Mitsugi, Y. Yamagata, T. Ikegami, K. Ebihara, J. Narayan, A.M. Grishin, Ferroelectric and colossal magnetoresistive properties of a Pb(0.5)Zr(0.5)Ti(0.5)O(0.5)/La_{1-x}Sr{x}MnO_{3} heterostructure film, Jpn. J. Appl. Phys. 39 (2000) 5418–5421.

[22] K. Ishikawa, H. Funakubo, Electrical properties of (001)- and (110)-oriented epitaxial SrBi_{2}Ta_{2}O_{7} thin films prepared by metalorganic chemical vapor deposition, Appl. Phys. Lett. 75 (1999) 1970–1972.

[23] H.N. Lee, S. Senz, N.D. Zakharov, C. Harnagea, A. Pignolet, D. Hesse, U. Gosede, Growth and characterization of non-c-oriented epitaxial ferroelectric SrBi_{2}Ta_{2}O_{7} films on buffered Si(100), Appl. Phys. Lett. 77 (2000) 3260–3262.

[24] C.-S. Huang, I.-N. Lin, J.Y. Lee, T.-Y. Tseng, Growth behavior of Y_{1-x}Ba_{x}Cu_{2}O_{y}, superconducting thin films using laser ablation technique, Jpn. J. Appl. Phys. 33 (1994) 4058–4065.

[25] P. Lecoeur, A. Gupta, P.R. Duncombe, G.Q. Gong, G. Xiao, Emission studies of the gas-phase oxidation of Mn during pulsed laser deposition of manganese oxides in O_{2} and N_{2}O atmospheres, J. Appl. Phys. 80 (1996) 513–517.

[26] C. Kwon, M.C. Robson, K.-C. Kim, J.Y. Gu, S.E. Lofland, S.M. Bhagat, Z. Trajanovic, M. Rajeswari, T. Venkatesan, A.R. Kratz, R.D. Gomez, R. Ramesh, Stress-induced effects in epitaxial (La_{0.8}Sr_{0.2})MnO_{3} films, J. Magn. Magn. Mater. 172 (1997) 229–236.

[27] S. Jin, M. McCormack, T.H. Tiefel, R. Ramesh, Colossal magnetoresistance in La–Ca–Mn–O ferromagnetic thin films, J. Appl. Phys. 76 (1994) 6929–6933.

[28] H.Y. Hwang, S.-W. Cheong, P.G. Radaelli, M. Marezio, B. Batlogg, Lattice effects on the magnetoresistance in doped LaMnO_{3}, Phys. Rev. Lett. 75 (1995) 914–917.