Influence of Point Defects on the Properties of Undoped and Ga-Doped ZnO Films Grown by Plasma-Assisted Molecular Beam Epitaxy in an O-Rich Environment

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Undoped and Ga-doped ZnO films are grown on GaN templates by plasma-assisted molecular beam epitaxy for systematic investigation of the influence of native defects on their electrical properties. The distinct electrical properties of the undoped and Ga-doped ZnO films are observed after thermal treatment in a nitrogen and an oxygen ambient. It is found that the undoped ZnO films show improved characteristics when annealed in an oxygen ambient, and the Ga-doped ZnO films exhibit stable characteristics when annealed in a nitrogen ambient. The variation in electrical properties revealed by the photoluminescence measurements can be attributed to the generation and annihilation of native defects, which are dependent on the ambient of treatments. The properties of the annealed undoped ZnO films grown in an O-rich environment are affected mainly by the oxygen vacancies, antisite oxygen and oxygen interstitials, while the annealed Ga-doped ZnO films are dominated by oxygen interstitials.

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The II-VI semiconductor zinc oxide (ZnO), having a direct bandgap of 3.37 eV and a large exciton binding energy of 60 meV, has been extensively studied for applications in light-emitting diodes and lasers.1,2 Recently, it has also attracted a great deal of attention for use in electronic devices, such as thin-film transistors and high electron mobility transistors.3,4 For these applications, low background concentration undoped ZnO films as well as stable doped ZnO films are necessary, depending on their roles in the devices. However, ZnO epilayers with a low background concentration are difficult to achieve because of the formation of donor-like native defects, such as oxygen vacancies (V O ), zinc interstitials (Zn i), and their complexes.5,6 So far, post thermal annealing under selected conditions (i.e., temperature and ambient), is still the most common technique used to improve material quality. It has been reported that the electrical and optical properties of ZnO films are sensitive to the annealing ambient.8,9 However, the spectral assignments of the native defects in ZnO epilayers on GaN have rarely been reported upon and are still controversial.10

In this study, undoped and Ga-doped ZnO films were grown on GaN templates by plasma-assisted molecular beam epitaxy (PA-MBE). The electrical, structural, optical and chemical properties of ZnO epilayers were then investigated with post thermal annealing in different ambient. The generation-annihilation of defects in these films after thermal treatment is clarified accordingly.

Experimental

Undoped and Ga-doped ZnO films were grown on high resistivity GaN templates by PA-MBE. The GaN template used in this study was a 4 um-thick GaN epilayer grown on sapphire by metal-organic chemical vapor deposition. Before the commencement of growth, the templates were cleaned at 700 °C for 20 minutes in the growth chamber. Then, a 20 nm undoped ZnO buffer layer was grown at low temperature (350 °C), and annealed at 700 °C for 10 minutes under a radical oxygen ambient. A nominally undoped or Ga-doped ZnO layer was then grown on the annealed ZnO buffer layer at 600 °C. The growth was carried out with an oxygen flow rate of 1.0 sccm and RF power of 350 W. Oxygen-rich conditions were used for both the undoped and Ga-doped ZnO epilayers. Reflection high energy electron diffraction (RHEED) was employed during the growth and thermal annealing to monitor the surface reconstruction. Typically, a spotty RHEED pattern was observed on the buffer layer before annealing which was reconstructed as a sharp streaky (1 × 1) pattern after annealing. The thickness of the studied ZnO films was about 600 nm. Eventually, the undoped and Ga-doped ZnO films were annealed in a nitrogen or an oxygen ambient for 20 minutes in a furnace heated to the range of 500 °C to 700 °C. The crystalline quality of the ZnO films was evaluated by examining the full width at half maximum (FWHM) of their (002) and (102) X-ray rocking curves (XRC). Hall measurement was carried out using the Van der Pauw method to determine the electrical properties, i.e., carrier concentration and mobility. The PL spectra of the undoped and Ga-doped ZnO layers were measured at room temperature using a 325 nm He-Cd laser as the excitation source. To investigate the effects of thermal annealing on film chemical bonding, X-ray photoelectron spectroscopy (XPS) was performed using Al Kα X-rays with an energy of 1486.7 eV. Before the measurement, a 10 nm-thick film was etched near the surface to exclude contamination of signals from the surface.

Results and Discussion

Figure 1 shows the concentration and mobility results for ZnO films annealed at various temperatures in an O2 or N2 ambient. Since...
there is serious decomposition of the ZnO films above 800°C, the temperature range used in this study is between 500°C and 700°C. The as-deposited undoped ZnO films exhibit a carrier concentration of approximately $3 \times 10^{17}$ cm$^{-3}$ with a mobility of 80 cm$^2$/V·s. Figure 1a reveals that as the undoped ZnO films are annealed in N$_2$, the carrier concentration increases with the annealing temperature. The mobility increases to 100 cm$^2$/V·s with an increase in the annealing temperature up to 600°C but decreases afterwards. The undoped ZnO films that have been annealed in O$_2$ exhibit similar mobility behavior to those annealed in N$_2$, but there is little change in carrier concentration with annealing temperature. Based on these observations, it can be concluded that annealing the ZnO films in an oxygen ambient improves the crystal quality and thus the electron mobility without a severe loss of oxygen. In contrast, annealing in a nitrogen environment produces more oxygen vacancies (V$_O$), which increases the carrier concentration. The dramatic degradation of electron mobility at 700°C and above can be attributed to material decomposition.

The as-deposited Ga-doped ZnO film has a carrier concentration of $5 \times 10^{17}$ cm$^{-3}$ with a mobility of 60 cm$^2$/V·s. Annealing the Ga-doped ZnO film in O$_2$ increases the carrier mobility and reduces the carrier concentration, whereas annealing in N$_2$ causes no significant change in either the carrier concentration or mobility of the film. The highest mobility of 100 cm$^2$/V·s and the lowest carrier concentration of $8 \times 10^{17}$ cm$^{-3}$ are achieved for the sample annealed at 600°C. This result may be postulated as due to either the filling of oxygen vacancy donors by oxygen molecules or the generation of Zn vacancy acceptors by decomposition of Ga atoms. We exclude the incorporation of acceptor-like nitrogen into the ZnO epilayer because no significant reduction in the electron concentration in undoped and Ga-doped ZnO films annealed in nitrogen can be observed. To clarify the effect of the variation in the carrier concentration, further analysis of the structural and optical properties is carried out.

The X-ray diffraction spectra of the undoped ZnO and Ga-doped ZnO films are presented in Figure 2. The peak at 34.56° is identified as the diffraction peak for the GaN template. The as-deposited undoped ZnO film prepared in this study exhibits a (002) X-ray diffraction peak at 2θ = 34.42°. The Ga-doped ZnO diffraction peak shifts to 34.44°, suggesting a decrease in the lattice constant associated with the c-axis. A slightly smaller Ga-O bond length (1.92 Å) than that of Zn-O (1.97 Å) is expected to make the ZnO lattice decrease.12 Observation of the full-width at half-maximum (FWHM) of the X-ray rocking curve (XRC) is a nondestructive method to assess material crystallinity. Heying et al.13 found that the (002) XRC is sensitive to screw and mixed type dislocations, while the (102) XRC is sensitive to all dislocations in GaN films. Clearly, the FWHM of the (102) asymmetric rocking curve can be used as an index for quantifying crystal quality.

The GaN templates used in this work exhibit (002) and (102) XRC FWHMs of 240 and 400 arcseconds, respectively. The FWHMs of the undoped ZnO (002) and (102) rocking curves are 205 and 660 arcseconds, respectively. There is significant broadening of asymmetrical reflection caused by the high density of pure-edge dislocations along the (001) axis, which is estimated to be $1-4 \times 10^{10}$ cm$^{-2}$ based on the linewidth of its rocking curve.14 The as-deposited Ga-doped ZnO films prepared in this work exhibit (002) and (102) XRC linewidths of 220 and 780 arcseconds, respectively. The broadened (002) and (102) XRCs are generally observed in doped crystals and can be attributed to the increase in local strain around the doped atoms or point defects associated with the dopants.15

The FWHMs of the (002) and (102) XRCs of the undoped and Ga-doped ZnO epilayers annealed at various temperatures are plotted in the insets to Figure 2. Examination of Figure 2a shows that the (002) XRC linewidth of the undoped ZnO decreases only 5 to 10 arcsec, whereas annealing in N$_2$ causes no significant change in either the carrier concentration or mobility of the film. The highest mobility of 100 cm$^2$/V·s and the lowest carrier concentration of $8 \times 10^{17}$ cm$^{-3}$ are achieved for the sample annealed at 600°C. This result may be postulated as due to either the filling of oxygen vacancy donors by oxygen molecules or the generation of Zn vacancy acceptors by decomposition of Ga atoms. We exclude the incorporation of acceptor-like nitrogen into the ZnO epilayer because no significant reduction in the electron concentration in undoped and Ga-doped ZnO films annealed in nitrogen can be observed. To clarify the effect of the variation in the carrier concentration, further analysis of the structural and optical properties is carried out.

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Figure 2. X-ray diffraction spectra of (a) undoped ZnO and (b) Ga-doped ZnO films grown on GaN/Sapphire templates. The insets represent the FWHMs of the (002) and (102) X-ray rocking curves for samples annealed at various temperatures.
attributed to Zn vacancies ($V_{Zn}$), which have a low formation energy in an O-rich growth condition and act as an acceptor. The as-deposited undoped ZnO sample exhibits n-type characteristics, while the emissions from donor-like defects i.e., oxygen vacancies, as claimed in previous reports are not present in the PL spectrum. This result supports the postulation that the energy required for the formation of oxygen vacancies is too high to appear easily under the O-rich growth condition used. Instead, the n-type conductivity for the as-deposited undoped ZnO might be attributed to unintentional impurities such as hydrogen, which is quite consistent with the activation energy of 26 meV extracted from the temperature-dependent Hall results. After annealing in N$_2$, the deep level emission band moves toward 2.49 eV. In fact, two emissions can be extracted from this emission band. One is the original $V_{Zn}$ emission at 2.34 eV while the other is located at 2.49 eV, which is known to be the donor-like oxygen vacancy ($V_O$). This variation of spectra is consistent with the increase in carrier concentration of this sample. In contrast, the samples treated in the oxygen ambient exhibit very different behavior. Figure 3b shows that the deep level emissions of the undoped ZnO films annealed below 600°C in oxygen are similar. When the annealing temperature exceeds 700°C, the intensity of the deep level emission is suppressed but an emission at 2.28 eV appears, which is reportedly associated with oxygen interstitials ($O_i$). The reduction of the 2.34 eV emission at 700°C is attributed to the transformation of Zn vacancies to oxygen antisites since the latter have a higher formation energy. Because the number of acceptor-like Zn vacancies is reduced, the carrier concentration increases, as depicted in Fig. 1a.

Figure 4 shows the normalized PL spectra of Ga-doped ZnO films measured at room temperature. The near-band edge emission of Ga-doped ZnO is dominated by a bound exciton at 3.26 eV. The FWHMs of the bound exciton emission remain the same after thermal annealing, while the defect-related emissions of the as-deposited Ga-doped ZnO films differ considerably from that of the undoped ZnO films. The as-deposited Ga-doped ZnO exhibits no defect emissions because $V_{Zn}$ is suppressed due to the binding energy of Ga-O being higher than that of Zn-O. Upon annealing at 700°C in nitrogen, a very weak deep-level emission of 2.28 eV associated with oxygen interstitials ($O_i$) appears. This slight change in defect emissions is consistent with little change in carrier concentration after annealing. In contrast, annealing at 600°C in oxygen substantially increases the intensity of the $O_i$ emission. This significant change is reflected by the large reduction in the carrier concentration due to the increase in the number of acceptor-like $O_i$ defects. These observations indicate that the Zn vacancies or Ga desorption postulated above are not major factors, but that the chemisorption of oxygen plays an important role in Ga-doped ZnO films annealed in a nitrogen or oxygen ambient.

Interestingly, oxygen interstitial emissions appear in the spectra of the Ga-doped ZnO films, after they are annealed in nitrogen. To probe this phenomenon further, XPS is used to examine the oxygen binding energy in the ZnO. Figure 5 presents the O 1s peaks in the XPS spectra of the undoped and Ga-doped ZnO films annealed in different atmospheres. Table I presents the relative intensity of each component and the atomic concentration in the ZnO films. The typical O 1s peak can be fitted by three near Gaussian peaks, centered at 530.15 ± 0.15, 531.25 ± 0.2, and 532.40 ± 0.15 eV, which are denoted as $O_{II}$, $O_{I}$, and $O_{III}$, respectively. The $O_I$ energy is attributed to $O^{2-}$ ions in the...
wurtzite structure of a hexagonal Zn$_2^+$ ion array, surrounded by Zn (or the substitution of Ga) atoms.\textsuperscript{21} The binding energy component, O$_{\text{IIi}}$, has been assigned to O$_2^-$ ions in the oxygen deficient regions within the matrix of ZnO.\textsuperscript{22} The binding energy peak, O$_{\text{III}}$, generally related to the O-O bonds, i.e., adsorbed O$_2$, is confirmed by the observation of the enhanced luminescence intensity after annealing in the O$_2$ ambient.\textsuperscript{23}

As the surfaces of ZnO films were sputtered 10 nm to remove the influence of surface contamination, the contribution of surface contamination to the observed O 1s sub-peaks can be excluded. In this study, the O 1s of the undoped and Ga-doped ZnO films can only be fitted to two Gaussian peaks, i.e., O$_3$ and O$_{\text{IIi}}$, corresponding well to the growth in the oxygen-rich environment. The dominance of O$_3$ in the O 1s peak reveals that the majority of bonds in the ZnO are Zn-O bonds. The O$_{\text{III}}$ ratio in the as-deposited undoped ZnO is 7\% and that in the annealed ZnO films is slightly reduced. However, this ratio is 16\% in the as-deposited Ga-doped ZnO with a significant reduction after annealing. Nevertheless, the variation in the O$_{\text{IIi}}$ ratio due to O$_2$ chemisorption cannot explain the changes in the PL spectra and electron concentration observed. Therefore, we postulate that the O-O bonds in ZnO may be associated with the split oxygen interstitials (O$_{\text{Ii}}$), which have been reported to be electrically inactive. However, the oxygen interstitials (O$_{\text{I}}$) revealed to the O-O bonds, i.e., adsorbed O$_2$, is confirmed by the observation of the enhanced luminescence intensity after annealing in the O$_2$ ambient.\textsuperscript{23}

It can thus be concluded that O$_{\text{Ii}}$ forms easily in the Ga-doped ZnO films prepared in an oxygen-rich environment. The generation of acceptor-like O$_{\text{I}}$ is well correlated to the reduction in the electron concentration in the Ga-doped ZnO annealed in oxygen. In contrast, the undoped ZnO, which has a low O$_{\text{Ii}}$ concentration, does not generate many O$_{\text{I}}$ as indicated by the corresponding change in the optical and electrical properties.

Conclusions

High quality ZnO epilayers were prepared using molecular beam epitaxy, after which we systematically investigated the correlation between the electrical and optical properties and the formation of native defects in undoped and Ga-doped ZnO films grown under an O-rich environment. Samples annealed in a nitrogen or oxygen ambient were investigated to explain the generation-annihilation mechanism of their native point defects. The variation of these properties was found to be closely related to the chemisorption and removal of oxygen in the ZnO films. A carrier concentration of 3 $\times$ 10$^{17}$ cm$^{-3}$ and a mobility of 90 cm$^2$/V $\cdot$ s was achieved for undoped ZnO after annealing in oxygen at 600 C. The formation of zinc vacancies (V$_{\text{Zn}}$), oxygen vacancies (V$_{\text{O}}$), and oxygen interstitials (O$_{\text{I}}$) are postulated to explain the variation in electrical and optical properties of the undoped ZnO films annealed under different conditions. For Ga-doped ZnO films, significant variation in carrier concentration was observed after annealing in oxygen but not in nitrogen. This phenomena can be attributed to the presence of split oxygen interstitials O$_{\text{Ii}}$ (split), which decompose to form octahedral oxygen interstitials O$_{\text{oct}}$ after annealing in oxygen.

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References

1. A. Tsuchiaki, A. Ohtomo, T. Onuma, M. Ohtani, T. Makino, M. Sumiya, K. Ohtani, S. F. Chichibu, S. Fukue, Y. Segawa, H. Ohno, H. Koinuma, and M. Kawasaki, Nat. Mater., 4, 42 (2004).
2. Z. K. Tang, G. K. L. Wong, P. Yu, M. Kawasaki, A. Ohtomo, H. Koinuma, and Y. Segawa, Appl. Phys. Lett., 72, 3270 (1998).
3. R. L. Hoffman, B. J. Norris, and J. F. Wagner, Appl. Phys. Lett., 82, 733 (2003).
4. K. Nomura, H. Hoitra, A. Takagi, T. Kamiya, M. Hirano, and H. Hosono, Nature, 432, 488 (2004).
5. K. Koike, I. Nakashima, K. Hashimoto, S. Sasa, M. Inoue, and M. Yano, Appl. Phys. Lett., 87, 112106 (2005).
6. A. Janotti and C. G. Van de Walle, Phys. Rev. B, 76, 165602 (2007).
7. U. Ozgur, V. I. Alivov, C. Liu, A. Teke, M. A. Reshefchikov, S. Dogan, V. Avrutin, S. J. Cho, and H. Morkoc, J. Appl. Phys., 98, 041301 (2005).
8. B. Du Ahm, S. Hoon Oh, C. Hee Lee, G. Hee Kim, H. Jae Kim, and S. Yeol Lee, J. Cryst. Growth, 309, 128 (2007).
9. N. Y. Garces, N. C. Giles, L. E. Halliburton, G. Cantwell, D. B. Eason, D. C. Reynolds, and D. C. Look, Appl. Phys. Lett., 80, 1334 (2002).
10. H. S. Kang, J. S. Kang, J. W. Kim, and S. Y. Lee, J. Appl. Phys., 95, 1246 (2004).
11. X. J. Wang, I. A. Buyanova, W. M. Chen, C. F. Pan, and C. W. Tu. Physica B, 401, 413 (2007).
12. C. Kittel, Introduction to Solid State Physics, 6th ed, New York: Wiley 1986.
13. B. Heying, X. H. Wu, S. Keller, Y. Li, D. Kapolnek, B. P. Keller, S. P. DenBaars, and J. S. Speck, Appl. Phys. Lett., 68, 643 (1996).
14. S.-K. Hong, H.-J. Ko, Y. Chen, and T. Yao, J. Cryst. Growth, 209, 537 (2000).
15. H. J. Ko, Y. F. Chen, S. K. Hong, H. Wenisch, T. Yao, and D. C. Look, Appl. Phys. Lett., 77, 3761 (2000).
16. L. Wang, Y. Pu, W. Fang, J. Dai, C. Zheng, C. Mo, C. Xiong, and F. Jiang, Thin Solid Films, 491, 323 (2005).
17. Y. F. Chen, D. M. Bagnall, H. J. Koh, K. T. Park, K. Hiraga, Z. Q. Zhu, and T. Yao, J. Appl. Phys., 84, 3912 (1998).
18. S. Cox, E. Davis, S. Cottrell, P. King, J. Lord, H. Alberto, R. Vilao, J. Piroto Duarte, N. Ayres de Campos, A. Weidinger, R. Lichti, and S. Irvine, Phys. Rev. Lett., 86, 2601 (2001).
19. F. Leiter, H. Alves, D. Pfisterer, N. G. Romanov, D. M. Hofmann, and B. K. Meyer, Physica B, 340, 201 (2003).
20. C. H. Ahn, Y. Y. Kim, D. C. Kim, S. K. Mohanta, and H. K. Cho, J. Appl. Phys., 105, 013502 (2009).
21. L. Kameswarra Rao and V. Vinni, Appl. Phys. Lett., 63, 608 (1993).
22. J. C. C. Fan and J. B. Goodenough, J. Appl. Phys., 48, 3524 (1977).
23. S. Major, S. Kumar, M. Bhatnagar, and K. L. Chopra, Appl. Phys. Lett., 49, 394 (1986).