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Formation of low resistance Ti/Al-based ohmic contacts on (11–22) semipolar n-type GaN

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ABSTRACT

The electrical properties of Ti/Al-based ohmic contacts to (0001) c-plane and (11–22) semipolar n-type GaN were investigated as a function of annealing temperature. The electrical properties of both c-plane Ti/Al/Au and semi-polar Ti/(Ta)/Al/Au contacts became improved upon annealing at 600 °C for 1 min. The specific contact resistances of the 600 °C-annealed c-plane Ti/Al/Au, semi-polar Ti/Al/Au, and semipolar Ti/Ta/Al/Au contacts were 3.2×10^{-4}, 1.5×10^{-4}, and 4.8×10^{-5} Ωcm^2, respectively. The X-ray photoemission spectroscopy (XPS) results showed that the Ga *2p* core level for the c-plane samples experienced a smaller shift toward the conduction band than that for the semipolar samples. The XPS depth profile results exhibited that the semipolar samples contained more interfacial oxygen than the c-plane samples. Oxygen was present in the form of Al-oxide at the interface region, but as oxygen atoms in the GaN surface region. On the basis of the electrical and XPS results, the annealing-induced improvement of the electrical properties was described in terms of the formation of donor-like defects.

1. Introduction

Currently most of commercially available GaN-based light-emitting diodes (LEDs) are grown on (0001) c-plane (polar) sapphire substrates [1]. It is well known that these polar LEDs suffer from the quantum-confined Stark effect (QCSE) [2] induced by large polarization-induced internal electric fields, resulting in a separation between the electron and hole wave functions in the quantum wells (QWs) and so reducing radiative recombination efficiency [3–5]. Thus, to enhance the internal quantum efficiency by eliminating or minimizing the Stark effect, GaN-based LEDs were grown on non-polar or semipolar substrates, e.g., (11–20), (1–100) and (11–22) plane substrates [5–8]. For example, Chakraborty et al. [6], investigating the DC and pulsed performance of InGaN/GaN multiple-QWs LEDs grown on non-polar free-standing m-plane GaN substrates, showed that packaged LEDs (300×300 μm^2) (a wavelength of 452 nm) yielded a CW output power as high as 0.6 mW at a current of 20 mA, corresponding to an external quantum efficiency of 1.09%. Pulsed power measurement on the lamps produced an output power of 23.5 mW at 1 A for a duty cycle of 0.25%. Scholz et al. [8], investigating semipolar LED structures on the side-facets of triangular GaN stripes grown by selective area epitaxy, demonstrated the significant reduction in the internal electric field (i.e., QCSE) by means of spectroscopic methods. Moreover, for next-generation lighting applications, the external quantum efficiency (EQE) of LEDs ought to be further improved. One of the ways of maximizing the EQE is to increase current injection by forming low resistance ohmic contacts and thereby to increase radiative recombination efficiency [9–13]. For c-plane *n*-GaN, Ti/Al-based metal contacts have been widely investigated because Ti/Al-based schemes easily formed low-resistance ohmic contact when annealed above 600 °C [14–16]. On the other hand, Ti/Al-based contacts on semipolar *n*-GaN planes showed different electrical behavior from that on polar *n*-GaN [17]. Thus, it is important to develop low-resistance and thermally stable *n*-contact schemes to semipolar *n*-GaN in order to fabricate the Stark effect-minimized GaN-based LEDs. In this study, the electrical properties of Ti/Al/Au and Ti/Ta/Al/Au contacts to (0001) c-plane and (11–22) semipolar *n*-GaN were investigated as a function of annealing temperature. X-ray photoemission spectroscopy examination was performed to understand interface characteristics and the improved electrical characteristics.

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2. Experimental procedures

Metalorganic chemical vapor deposition (MOCVD) was used to grow (11–22) semipolar *n*-type GaN on (10–10) m-plane sapphire substrates. In other words, a 4-μm-thick undoped GaN layer was grown on the (10–10) m-plane sapphire substrate, on which a 2-μm-thick *n*-GaN:Si ($n_d = 7.3 \times 10^{18}$ cm^{-3}) layer was grown. Prior to the photolithography and metal deposition, the semipolar GaN samples were treated with a buffered oxide etch (BOE) solution for 1 min, rinsed in de-ionized (DI) water, and then blown dry in a N_2 stream. Ti/Al (10 nm/150 nm) and Ti/Ta/Al (10 nm/10 nm/150 nm) films were deposited on the semipolar *n*-GaN by an RF sputtering system, on which a Au layer (30 nm) was deposited by electron-beam evaporation. For comparison, Ti/Al/Au films were also prepared on BOE-treated c-plane *n*-GaN ($n_d = 4.5 \times 10^{18}$ cm^{-3}). Some of the samples were annealed at 400 and 600 °C for 1 min in a N_2 stream. Circular transfer length method (CTLM) patterns were defined by the standard photolithography technique to measure specific contact resistance. The outer radius of CTLM patterns was fixed at 200 μm and the gap spacing between outer and inner radii was varied from 5 to 40 μm. Current-voltage (*I–V*) measurements were carried out by a high-current source-measuring unit (Keithley 238) to measure the electrical properties. X-ray photoemission spectroscopy (XPS) using K_α X-ray source (1486.6 eV) was performed in an ultrahigh vacuum system to characterize the surface and interface characteristics.

3. Results and discussion

Fig. 1 shows the *I–V* properties of Ti/Al/Au (10 nm/150 nm/30 nm) and Ti/Ta/Al/Au (10 nm/10 nm/150 nm/30 nm) contacts on c-plane and semipolar (11–22) *n*-GaN layers as a function of annealing temperature. Before annealing, the c-plane and semipolar Ti/Al/Au contacts are ohmic with contact resistivities of 7.4×10^{-4}, and 9.4×10^{-4} Ωcm^2, respectively, whereas the semipolar Ti/Ta/Al/Au contacts are non-ohmic. Note that the semipolar samples show resistivity similar to that of the c-plane sample although the former contains more interfacial oxygen. This may be attributed to the fact that the c-plane and semipolar samples contain high electron concentrations of 4.5×10^{18} and 7.3×10^{18} cm^{-3}, respectively, which leads to electron tunneling, so lowering contact resistivity. Furthermore, before annealing, the semipolar Ti/Ta/Al/Au samples show somewhat better electrical property than the semipolar Ti/Al/Au samples, although the Ti layer is in contact with the GaN for both the samples (Fig. 1b and c). Similar phenomena were also observed by other researchers [18,19]. This feature might be attributed to the catalytic effect of the capping layers [19]. Furthermore, since the sputtering deposition of a Ta layer could increase the substrate temperature, this may also affect the formation of surface states and so the electrical properties of the samples [20]. However, a systematic study is needed to clarify the behavior. After annealing at 400 °C, the c-plane Ti/Al/Au contacts are non-ohmic on, while semipolar Ti/Al/Au contacts are ohmic. The semipolar Ti/Ta/Al/Au contacts are also non-ohmic. After annealing at 600 °C, all of the samples show better electrical characteristics than their as-deposited samples. The specific contact resistances of the c-plane Ti/Al/Au contacts, the semipolar Ti/Al/Au contacts, and the semipolar Ti/Ta/Al/Au contacts that were annealed at 600 °C were 3.2×10^{-4}, 1.5×10^{-4}, and 4.8×10^{-5} Ωcm^2, respectively. It is noted that the annealed semipolar samples show lower contact resistivity than the annealed c-plane samples. Table 1 summarizes L_t (transfer length), R_s (sheet resistance), and R_c (specific contact resistance) obtained from the *I–V* relations of the polar and semipolar samples. It is noted that R_c is inversely proportional to L_t.

Fig. 2 exhibits the XPS Ga 2*p* core level spectra obtained from the interface regions of the contact scheme/GaN samples as a function of annealing temperature. During the XPS analysis, the sample surfaces were Ar^+-ion sputtered, and the Ti, Ta, Al, Au, O, N, and Ga photoelectron signals were carefully monitored. The Ga 2*p* core levels were finally collected when only a Ga photoelectron peak (i.e., from Ga–N bonding) was observed. XPS core level peak fittings were performed with a Shirley-type background and Lorentzian–Doniac–Sunsic curves convoluted with a Gaussian profile. For all the samples, the Ga 2*p* core level consists of the Ga–N and Ga–O bonds. Note that the semipolar samples contain more interfacial oxygen than the c-plane samples. This was also confirmed by X-ray fluorescence (XRF) results (not shown here). After annealing at 400 °C, the Ga 2*p* core level for the c-plane and semipolar Ti/Al/Au contacts shifted toward lower energies compared with those for their as-deposited samples. The lower binding energy shift implies that the surface Fermi level moved toward the valence band edge [21]. This increases the band-bending of *n*-GaN and so the Schottky barrier height (SBH). After annealing at 600 °C, the Ga 2*p* core level for all the c-plane and semipolar samples shifted toward higher energies. For instance, the Ga 2*p* core level shifted by 0.06, 0.07, and 0.12 eV for the c-plane Ti/Al/Au, semipolar Ti/Al/Au, and semipolar Ti/Ta/Al/Au samples, respectively. The higher binding energy shift indicates that the surface Fermi level moved toward the conduction band edge [21]. This reduces the band-bending of *n*-GaN and so the SBH. The shift of the Ga 2*p* peak denotes a change in the band bending since the N 1*s* core level spectra show a similar shift [22].

Fig. 3 shows the XPS depth profiles of the contact scheme/GaN

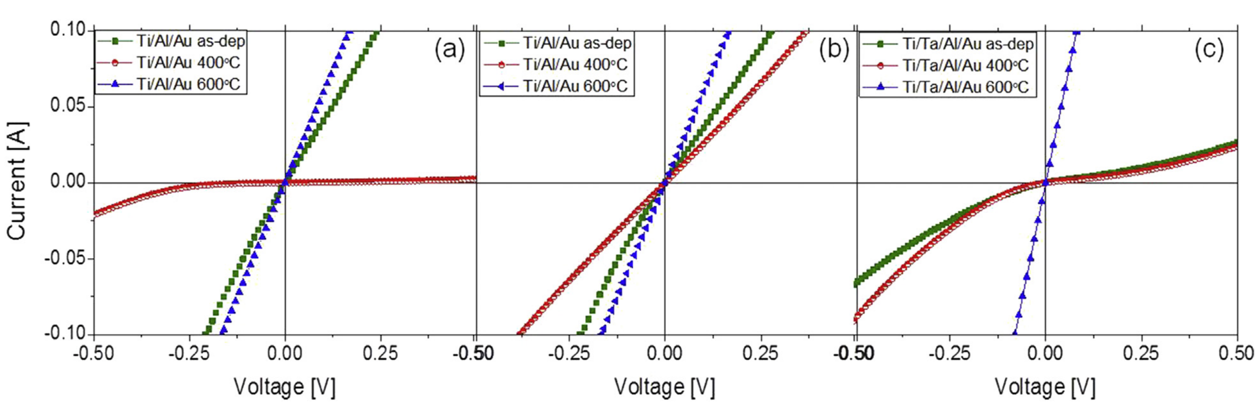

Fig. 1. The *I–V* properties of (a) c-plane Ti/Al/Au (10 nm/150 nm/30 nm), (b) semipolar Ti/Al/Au (10 nm/150 nm/30 nm) and semipolar Ti/Ta/Al/Au (10 nm/10 nm/150 nm/30 nm) contacts as a function of annealing temperature.