Control of plasma-CVD SiO$_2$/InAlN interface by ultrathin atomic-layer-deposited Al$_2$O$_3$ interlayer

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To control a plasma-CVD SiO$_2$/InAlN interface, the insertion of an ultrathin Al$_2$O$_3$ interlayer deposited by atomic layer deposition was investigated. The thickness of the Al$_2$O$_3$ interlayer was varied between 0.5 and 2 nm. Compared with the direct deposition of SiO$_2$, the insertion of a 2- or 1-nm-thick Al$_2$O$_3$ interlayer resulted in the similar interface state density $D_{it}$ distribution. However, a significant reduction in $D_{it}$ was achieved by using a 0.5-nm-thick Al$_2$O$_3$ interlayer. Slight oxidation of the InAlN surface under the 0.5-nm-thick Al$_2$O$_3$ interlayer was observed by X-ray photoelectron spectroscopy (XPS). This result indicated that a native oxide/InAlN interface was formed without disorder beneath the 0.5-nm-thick Al$_2$O$_3$ interlayer, while plasma damage was prevented at the interface. On the other hand, the density of positive charges at the interface increased with the Al$_2$O$_3$ interlayer thickness. The generation process of the interface charge is discussed on the basis of XPS data.
1. Introduction

GaN has a wide band gap, a high breakdown field, a high electron saturation velocity, and good thermal conductivity. These characteristics make GaN suitable for use in power electronics devices. In addition, by combining GaN with a III-nitride alloy to form a heterostructure, a high-mobility two-dimensional electron gas (2DEG) can be obtained in a high-electron-mobility transistor (HEMT). AlGaN/GaN HEMTs are actually used as high-power high-frequency devices to miniaturize high-power communication equipment. In$_x$Al$_{1-x}$N ($x = 0.17 – 0.18$) lattice matched to GaN has a large band gap$^1$ and a large spontaneous polarization$^2$. A lattice-matched InAlN/GaN heterostructure can provide a higher 2DEG density than an AlGaN/GaN heterostructure owing to the larger conduction band discontinuity and the larger interface polarization$^3$. In fact, a 2DEG density higher than $2 \times 10^{13}$ cm$^{-2}$ has been obtained and reported$^4$–$^{10}$. Therefore, InAlN/GaN heterostructures are expected to construct HEMTs exhibiting excellent performance. However, the leakage current through an InAlN barrier is high owing to its large polarization, resulting in a high internal electric field that allows electron tunneling$^{11}$, $^{12}$. To solve this problem, the construction of a metal-insulator-semiconductor (MIS) gate structure has been proposed for InAlN/GaN HEMT applications$^{11}$. Thus far, combinations of InAlN and several insulators, including Al$_2$O$_3$$^{11}$, $^{13}$–$^{17}$, ZrO$_2$$^{14}$, $^{18}$, $^{19}$, GdScO$_3$$^{14}$), HfO$_2$$^{18}$, SiO$_2$$^{20}$, plasma oxides$^{21}$–$^{25}$, and thermal oxides$^{26}$, $^{27}$, have been used to construct MIS gate structures to reduce the leakage current. Among these insulators, SiO$_2$ is attractive because it has the largest band gap among the insulators in practical use. Nevertheless, a method of controlling the SiO$_2$/InAlN interface to reduce the interface state density $D_{it}$ has not been established.

For the deposition of the SiO$_2$ layer, plasma-CVD is a frequently used method. In this method, however, surface damage caused by plasma may result in the generation of interface states. We previously reported that the direct deposition of SiO$_2$ on InAlN by plasma CVD resulted in the surface oxidation of InAlN, which indicated that the surface was damaged by the plasma$^{28}$. According to the disorder-induced gap state (DIGS) model, the origin of the interface states is interface disorder$^{29}$. To reduce interface disorder caused by plasma damage, one of the useful methods is to insert an interface control layer. We reported that an ultrathin N$_2$O plasma oxide interlayer reduced $D_{it}$ at the plasma-CVD SiO$_2$/InAlN interface compared with the case of the direct deposition$^{28}$. Therefore, the native oxide/InAlN interface is highly likely an excellent interface as long as the plasma damage during plasma CVD is blocked by the interlayer.
Since the native oxide of InAlN contains an Al oxide component, there is a possibility that the ultrathin native oxide interlayer can be replaced by an ultrathin Al$_2$O$_3$ layer carefully formed by atomic layer deposition (ALD) to control the SiO$_2$/InAlN interface. Although we also attempted to insert a 2-nm-thick ALD Al$_2$O$_3$ interlayer at the SiO$_2$/InAlN interface in our previous work$^{28}$, the reduction in $D_{it}$ was not significant compared with that achieved by the insertion of the plasma native oxide interlayer. However, the thickness of the Al$_2$O$_3$ interlayer has not been optimized. Furthermore, in our previous work$^{28}$, we evaluated $D_{it}$ by applying the Terman method to the measured capacitance–voltage ($C–V$) curves. However, the error caused by the Terman method is relatively large, especially for low $D_{it}$ values.

In this work, we focused on the control of the plasma-CVD SiO$_2$/InAlN interface by inserting an ultrathin Al$_2$O$_3$ interlayer. Samples with 2-, 1-, and 0.5-nm-thick Al$_2$O$_3$ interlayer were fabricated and tested. The parameters related to the band alignment were measured by X-ray photoelectron spectroscopy (XPS). The $C–V$ characteristics were simulated using these parameters to extract the $D_{it}$ distributions from the experimental data. We found that the 0.5-nm-thick Al$_2$O$_3$ interlayer markedly reduced $D_{it}$ at the plasma-CVD SiO$_2$/InAlN interface. The mechanism of the reduction in $D_{it}$ is discussed on the basis of the results of XPS. In addition, the interface charge introduced by the insertion of an Al$_2$O$_3$ interlayer was investigated. The generation process of the interface charge was also studied and is discussed.

2. Experimental

For electrical measurements, MIS diodes, as shown in Fig. 1, were fabricated as follows. A Si-doped InAlN layer was grown by metal organic vapor phase epitaxy (MOVPE) on sapphire substrates via a GaN buffer layer. The carrier concentration, $n$, of the InAlN layer was $2 \times 10^{18}$ cm$^{-3}$ and the thickness was 160 nm, which enabled us to characterize the samples as ordinary MIS diodes without considering the 2DEG that was located much deeper than the depletion layer edge. Chemical treatment was carried out using buffered HF solution (BHF, NH$_4$F:HF=5:1) to remove the surface oxide. A 20-nm-thick SiN layer for protecting the InAlN surface during ohmic annealing was deposited by electron cyclotron resonance plasma CVD at 270°C with 100 W plasma power using a N$_2$ and SiH$_4$/Ar gas mixture. After photolithography and partial removal of SiN by BHF, a Ti (20 nm; bottom)/Al (50 nm)/Ni (20 nm)/Au (50 nm; topmost) ohmic contact was formed as a large pad perforated with a periodic-hole array. Subsequently, the sample was annealed at
850°C for 1 min using the SiN surface protection layer. After annealing, the SiN layer was removed by the BHF solution. The Al₂O₃ interlayer was formed by ALD at 300°C using trimethylaluminum and H₂O. The thickness of the Al₂O₃ interlayer was controlled by setting the deposition cycle number (0.12 nm/cycle). For comparison, a sample without an Al₂O₃ interlayer was also fabricated. SiO₂ was deposited by plasma-enhanced chemical vapor deposition (PECVD) at 300°C with 30 W plasma power using a N₂O and SiH₄ gas mixture. The thickness of the SiO₂ layer was controlled to be 20 ± 1 nm by adjusting the deposition time (6.7 nm/min). Finally, the Ni (20 nm; lower)/Au (50 nm; upper) circular gate electrode was formed in the center of the hole of the ohmic contact pad to complete the MIS structure.

The insulator/InAlN interface was examined by XPS using a sample with an ultrathin insulator layer deposited onto an undoped 30-nm-thick InAlN layer grown by MOVPE on a sapphire substrate via a 2-μm-thick GaN buffer layer. A monochromated Al-Kα X-ray source (1486.6 eV) was used. The charge-up error in binding energy was calibrated by setting the C 1s spectral peak to 285.0 eV.

3. Results and discussion

We carried out XPS measurement of the band gap \( E_g \) and the valence band offset \( \Delta E_V \) between plasma CVD SiO₂ and InAlN layers. On the basis of the energy difference between the core peak and the loss peak onset in the O 1s spectrum, \( E_g \) of the oxide layer can be measured\(^{30}\). The measured O 1s energy-loss spectrum for 30-nm-thick SiO₂ is shown in Fig. 2, where \( E_g \) of deposited SiO₂ was determined to be 8.6 ± 0.1 eV. This value is relatively smaller than the reported value for SiO₂\(^{30}\). This is presumably because of SiO₂ film quality, such as defects and content of nitrogen as described later, in addition to the measurement error. On the other hand, on the basis of the angle-resolved XPS for the valence band maximum (VBM) spectrum of the sample with 3-nm-thick SiO₂ on InAlN, we measured \( \Delta E_V \) to be 2.3 ± 0.1 eV, as shown Fig. 3, where \( \theta \) indicates the photoelectron take-off angle. Considering \( E_g \) of InAlN to be 4.6 eV\(^{11}\), the conduction band offset \( \Delta E_C \) was determined to be 1.7 eV. We used these values of \( E_g, \Delta E_V, \) and \( \Delta E_C \) as parameters for the SiO₂/InAlN interface in the simulation. In addition, the Schottky barrier height at the metal/SiO₂ interface was determined to be 3.7 eV, referring to the value measured by internal photoemission spectroscopy for Ni/thermal SiO₂ interface in Ref. 31. For the Al₂O₃/InAlN interface, the previously reported values of \( E_g = 6.7 \) eV for Al₂O₃\(^{32}\), \( \Delta E_V = 1.2 \) eV\(^{16}\), and \( \Delta E_C = 0.9 \) eV\(^{16}\) were used. The band alignment assumed in the simulation is
illustrated in Fig. 4. Here, the dipole at the SiO$_2$/Al$_2$O$_3$ interface was ignored for simplicity.

The $C-V$ curves measured at 1 MHz for the samples without and with 2-, 1-, and 0.5-nm-thick Al$_2$O$_3$ interlayers are plotted by the solid lines in Fig. 5. We carried out a simulation of $C-V$ curves under assumption of a $D_{it}$ distribution by choosing parameters, as described later, to make a best fit to the measured curves using a simulator developed previously$^{[33,34]}$. In Fig. 5, the simulated curves and the ideal curves were also shown by the open circles and the one-dot-chain lines, respectively. Here, the ideal curves were calculated assuming no charge at the interface and shifted to overlap on the measured curves in the deep depletion region, while the measured curves were not shifted. The voltage shift $\Delta V$ for each ideal curve is listed for each sample in Table I with the thicknesses of Al$_2$O$_3$ and SiO$_2$. A slight increase in oxide capacitance $C_{ox}$ of the sample with the 0.5-nm-thick Al$_2$O$_3$ interlayer compared with that of the sample without the Al$_2$O$_3$ interlayer was resulted from the thinner SiO$_2$ layer thickness due to the thickness control error. On the other hand, $\Delta V$ is given by

$$\Delta V = \frac{1}{C_{ox}} (Q_{pol} + Q_f + Q_{it}^{fr}), \quad (1)$$

where $Q_{pol}$ is the density of the polarization charge at the InAlN surface, $Q_f$ is the density of the interfacial fixed charge, and $Q_{it}^{fr}$ is the charge density of ‘frozen’ interface states with long time constants. $Q_f$ may include the contribution of the oxide bulk charge. For the samples with the 0.5-nm-thick Al$_2$O$_3$ interlayer, the maximum capacitance was the closest to $C_{ox}$ and the slope of the $C-V$ curve was the largest among all samples, which indicated the largest and steepest shift of the surface Fermi level $E_{FS}$ to $E_C$ due to a decrease in $D_{it}$.

Figure 6 summarizes the $D_{it}$ distributions extracted from the $C-V$ curves by the simulation, which were used to fit the measured curves for samples with various interlayer thicknesses. In the simulation, we assumed the $D_{it}$ distribution based on the DIGS model$^{[29]}$. According to this model, the interface states were originated from interface disorder resulting in a U-shaped $D_{it}$ distribution in the bad gap. It is reported that the $D_{it}$ distributions of III-nitride MIS diodes can be appropriately assumed as$^{[33,34]}$

$$D_{it}(E) = D_{it0} \exp \left( \frac{E-E_{CNL}}{E_{6D,0A}} \right)^{N_{D,A}}, \quad (2)$$
where $D_{it0}$ is the minimum of $D_{it}$, $E_{CNL}$ is the charge neutrality level, $E_{0D}$ and $n_D$ are the curvature factors assumed for the donor-like states below $E_{CNL}$, and $E_{0A}$ and $n_A$ are those for the acceptor-like states above $E_{CNL}$. $E_{CNL}$ was assumed to be located 2.0 eV below the conduction band minimum energy, $E_C$. Based on the $C$–$V$ curves of n-type MIS diode, $E_{0D}$ and $n_D$ cannot be determined because the response of the interface states located below $E_{CNL}$ becomes too slow. Here, $D_{it0}$, $E_{0A}$, and $n_A$ are used as fitting parameters and determined as shown in Table II. $E_{0A}$ and $n_A$ mainly affect the curvature of the $C$–$V$ curves at the accumulation region, while $D_{it0}$ dominantly affects the steepness in the depletion region. Since the difference of 10% in this parameter results in a deviation from the measured curve, the estimation accuracy of $D_{it0}$ is empirically determined to be 10% as a fitting parameter. As shown in Fig. 6, the insertion of the 2- and 1-nm-thick Al$_2$O$_3$ interlayers led to the similar $D_{it}$ distribution as that of the sample without interlayer. Nevertheless, the samples with the 0.5-nm-thick Al$_2$O$_3$ interlayer exhibited markedly reduced $D_{it}$. In the simulation, we also assumed a charge density, $Q_f$, given by

$$Q'_f = Q_{pol} + Q_f$$

(3)

at the insulator/InAlN interface in addition to the $D_{it}$ distribution. The assumed $Q_f (= Q'_f - Q_{pol})$ for each sample is plotted in Fig. 7 as a function of the thickness of the Al$_2$O$_3$ interlayer. $Q_{pol}$ was assumed to be $2.88 \times 10^{13}$ cm$^{-2}$ according to Ref. 2. It can be seen, as a rough trend, that the increase in Al$_2$O$_3$ thickness led to the increase in the assumed positive $Q_f$. The mechanism of this result is discussed later.

To investigate the chemical bonding at the interface, the XPS study was carried out. Figure 8 shows XPS spectra after depositing a 3-nm-thick SiO$_2$ layer onto the samples with a 1-nm-thick Al$_2$O$_3$ interlayer. The O 1s and Al 2p spectra show the existence of an Al$_2$O$_3$ component, which indicates the successful formation of the Al$_2$O$_3$ interlayer. The higher energy component in the N 1s spectrum may be assigned to a small amount of SiON component in the SiO$_2$ layer because this component having the similar intensity ratio to that of the main peak was also observed for a 30-nm-thick SiO$_2$ layer by a separate experiment. Since the shape of the N 1s spectra after Al$_2$O$_3$ deposition was coincident with that before Al$_2$O$_3$ deposition (not shown here), it is unlikely that the Al$_2$O$_3$ layer turned into an AlION layer because of intermixing. Actually, the lower binding energy component at 531 eV in the O 1s spectra in Fig. 8 (a) is assigned to Al$_2$O$_3$. On the other hand, the In 3d spectrum does not show the oxide component, indicating that there was no oxidation.
of the InAlN surface. As plotted in Fig. 9, however, the In 3d spectrum from the sample with a 0.5-nm-thick Al2O3 interlayer after the SiO2 deposition shows an oxide component. To clarify the difference, the In 3d spectra in Figs. 8 (b) and 9 (b) are compared in Fig. 10, where the intensity was normalized and the spectrum of the sample with the 1-nm-thick Al2O3 interlayer was shifted horizontally for overlapping. It can be seen that the spectra for the sample with the 0.5-nm-thick layer is wider. (In O1s spectrum, it is difficult to confirm the existence of the In-oxide component because its intensity should be less than that of the hem of the SiO2 component indicated by a Voigt function, considering the In molar fraction of InAlN and the thinness of the oxidized layer.) Thus, the InAlN surface of the sample with the 0.5-nm-thick Al2O3 interlayer was oxidized during the deposition of SiO2 by PECVD. Slight oxidation during the deposition of SiO2 beneath the sub-nm-thickness Al2O3 interlayer led to the formation of a native oxide/InAlN interface. Plasma damage during SiO2 deposition presumably caused by the stochastic collisions of ions to the InAlN surface can lead to interface disorder by changing connection, length and angle of the chemical bonds at the interface. Considering the electrical measurement results, it is highly likely that the oxidation beneath the Al2O3 interlayer proceeded without causing disorder. Therefore, it is also likely that the native oxide/InAlN interface was protected from plasma damage by the 0.5-nm-thick Al2O3 interlayer as a result of a mechanism where the penetration depth of the oxidizing species was greater than that of the damaging species. This result indicated that the native oxide/InAlN interface has excellent properties if it is formed appropriately so as to prevent plasma damage.

In this work, the InAlN layers of the present samples were grown on sapphire substrate. Therefore, an exact determination of the $E_{FS}$ position by XPS is difficult. Nevertheless, since the binding energy was calibrated by the C 1s peak position, the core-level energy at each step of the interface formation can be a rough reference of the charging situation at the InAlN surface. Actually, the core-level spectra exhibited a remarkable change in binding energy. Figure 11 plots the peak binding energy of the InAlN component of the Al 2p spectrum, $E_{Al\,2p\,InAlN}$, at each step of interface formation. Direct deposition of an ultrathin SiO2 layer onto the InAlN surface resulted in an increase of $E_{Al\,2p\,InAlN}$. However, after the deposition of an Al2O3 layer, $E_{Al\,2p\,InAlN}$ decreased. The present result indicates an increase in the amount of negative charge in the vicinity of the surface upon the deposition of the Al2O3 layer onto InAlN. However, $E_{Al\,2p\,InAlN}$ increased after the subsequent SiO2 deposition, which indicated a decrease in or compensation for the negative charge in the vicinity of the surface of the tested samples. Therefore, the
negative charge observed for the ultrathin Al$_2$O$_3$/InAlN structure might have existed at the topmost surface of the Al$_2$O$_3$ layer. These changes were commonly observed for the samples with 1-nm-thick and 0.5-nm-thick Al$_2$O$_3$ interlayer. In addition, the similar changes were observed also in In 3d and N 1s spectra.

Fig. 12 shows the result of the angle-resolved XPS for the O 1s spectrum from the 2-nm-thick Al$_2$O$_3$/InAlN structure. The ratio of the intensity of the higher-energy-component of the O 1s spectrum, which is usually assigned to Al–O–H bonding$^{37,38}$, to that of the Al–O–Al component$^{37,38}$ at the lower binding energy increased as the take-off angle decreased. This result can be explained by the localization of H-terminated oxygen dangling bonds at the surface of the ultrathin Al$_2$O$_3$ layer. This is natural considering that the layer-by-layer growth was interrupted at the surface of the ALD Al$_2$O$_3$ layer. However, statistically, the termination by hydrogen might be absent in part of the dangling bonds. According to previous calculation results$^{39}$, an oxygen dangling bond generates an acceptor-like level near the VBM ($E_V + 0.9\text{eV}$) in $\alpha$-Al$_2$O$_3$, which has a local environment similar to that of amorphous Al$_2$O$_3$. This level can behave as a negative fixed charge, considering the aforementioned $\Delta E_V$ (1.2 eV) between Al$_2$O$_3$ and InAlN. After the subsequent SiO$_2$ deposition, such oxygen dangling bonds were possibly recovered to minimize the negative charge, resulting in the dominance of the residual positive charge at the Al$_2$O$_3$/InAlN interface of the completed MIS diodes.

Although it is desirable to confirm the existence of the hydroxide component also in Al 2p spectra, the separation of oxide and hydroxide components in Al 2p spectra is difficult by the reason as follows. The topmost Al atoms of the Al$_2$O$_3$ layer before SiO$_2$ deposition were possibly connected as HO–Al(–O–Al)$_2$, and those in bulk Al$_2$O$_3$ were connected as Al(–O–Al)$_3$. Therefore, the binding energy of these components in Al 2p spectrum should be close each other due to the similar local chemistry$^{38}$. Actually, the Al 2p binding energy is reported to be very similar between Al$_2$O$_3$ and hydroxides$^{40}$. Even in Ref. 38 where the native oxides on the Al metal surface was investigated by XPS in detail, components from Al atoms in Al$_2$O$_3$ and AlO(OH) were not separated in the Al 2p spectra.

The increase in positive charge with the Al$_2$O$_3$ interlayer thickness shown in Fig. 7 for the completed MIS diodes might have existed in the bulk Al$_2$O$_3$ because the interface quality affected on $Q_{it}^r$ that was separated from $Q_f$ in the simulation. On the basis of a separate experiment, $Q_f$ was $2.54\times10^{13}\text{cm}^{-2}$ for a metal/Al$_2$O$_3$ (20 nm)/InAlN MIS diode prepared for comparison. Although the similar generation of the positive interface charge
has been reported for an Al$_2$O$_3$/GaN interface$^{41)}$, $Q_f$ was $2.27 \times 10^{13}$ cm$^{-2}$. Therefore, $2.88 \times 10^{13}$ cm$^{-2}$ of $Q_f$ observed for the MIS diode with the 2-nm-thick Al$_2$O$_3$ interlayer was higher than these values, which indicates that the volume density of the bulk positive charge in the Al$_2$O$_3$ interlayer should have been higher than the thick Al$_2$O$_3$ layer. This may be a phenomenon that was induced by the plasma irradiation. In Ref. 39, the possibility that the origin of the positive charge is aluminum interstitials was discussed because they generate donor-like defects near the conduction band of $\alpha$-Al$_2$O$_3$. However, $E_g$ of $\alpha$-Al$_2$O$_3$ was calculated to be 9.2 eV$^{39)}$, which is different from that (6.7 eV) measured for amorphous ALD Al$_2$O$_3^{32)}$. Therefore, the dominant defect in the Al$_2$O$_3$ interlayer is unclear at this stage. Recently, post-metallization annealing in air has been found to be an efficient means of reducing $D_{it}$ and the positive charge owing to the Al$_2$O$_3$ bulk defects in the metal/ALD Al$_2$O$_3$/GaN MIS structure$^{42)}$. It is worth finding a method of reducing the positive charge in the Al$_2$O$_3$ interlayer, which is future work.

4. Conclusions

The control of plasma-CVD SiO$_2$/InAlN interfaces by using an ALD Al$_2$O$_3$ interlayer was investigated. The thickness of the Al$_2$O$_3$ interlayer was set to 2, 1, and 0.5 nm. Compared with the direct deposition of SiO$_2$ onto an InAlN surface, the insertion of the 2- or 1-nm-thick Al$_2$O$_3$ interlayer resulted in the similar $D_{it}$ distribution. However, a significant reduction in $D_{it}$ was achieved by using the 0.5-nm-thick Al$_2$O$_3$ interlayer. It is highly likely that the oxidation beneath the 0.5-nm-thick Al$_2$O$_3$ interlayer proceeded without causing disorder because the native oxide/InAlN interface was protected from plasma damage, presumably because the penetration depth of the oxidizing plasma species was greater than that of the damaging species. That is, the formation of a plasma native oxide/InAlN interface without damage led to the reduction in $D_{it}$. On the other hand, the Al$_2$O$_3$ interlayer was found to introduce positive charge at the SiO$_2$/InAlN interface. The generation process of the interface charge was also studied and discussed.

Acknowledgement

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References


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Figure Captions

Fig. 1. Structure of tested MIS diodes.

Fig. 2. XPS O 1s energy-loss spectra for the thick plasma-CVD SiO$_2$ layer.

Fig. 3. Take-off angle dependence of XPS VBM spectrum for 3-nm-thick SiO$_2$/InAlN sample.

Fig. 4. Band alignment assumed in the simulation of $C-V$ characteristics.

Fig. 5. $C-V$ characteristics. The solid lines are measured curves. The one-dot-chain lines are ideal curves. The open circles show the simulated curves assuming interface states to fit the measured curves. (a) Sample without Al$_2$O$_3$ interlayer. (b) Sample with 2-nm-thick Al$_2$O$_3$ layer. (c) Sample with 1-nm-thick Al$_2$O$_3$ layer. (d) Sample with 0.5-nm-thick Al$_2$O$_3$ layer.

Fig. 6. Evaluated $D_{it}$ distributions.

Fig. 7. Assumed $Q_f$ as a function of the thickness of the Al$_2$O$_3$ interlayer.

Fig. 8. XPS spectra for SiO$_2$ (3 nm)/Al$_2$O$_3$ (1 nm)/InAlN. (a) O 1s, (b) In 3d, (c) N 1s, and (d) Al 2p.

Fig. 9. XPS spectra for SiO$_2$ (3 nm)/Al$_2$O$_3$ (0.5 nm)/InAlN. (a) O 1s, (b) In 3d, (c) N 1s, and (d) Al 2p.

Fig. 10. Comparison of XPS In 3d spectra between samples with 1-nm-thick and 0.5-nm-thick Al$_2$O$_3$ interlayer.

Fig. 11. Binding energy of InAlN component in Al 2p spectra measured by XPS at each step of interface formation.

Fig. 12. Take-off angle dependence of O 1s spectrum for 2-nm-thick Al$_2$O$_3$ on InAlN.
Table I.  $\text{Al}_2\text{O}_3$ and $\text{SiO}_2$ thicknesses and ideal curve shift $\Delta V$ for each sample.

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<th>Thickness of $\text{Al}_2\text{O}_3$ interlayer [nm]</th>
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<th>$\Delta V$ [V]</th>
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Table II.  Parameters relating to $D_{it}$ distribution used in simulation of $C$–$V$ curve for each sample.

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<th>Thickness of $\text{Al}_2\text{O}_3$ interlayer [nm]</th>
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<th>$n_A$</th>
<th>$E_{0.4}$ [eV]</th>
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</table>
Fig. 1

Fig. 2.
Fig. 3. 

![Graph showing intensity versus binding energy for two different angles θ = 15° and θ = 90°.](image)

Intensity

Binding energy [eV]

- θ = 15° (x3)
- θ = 90°

ΔEC = 0.8 eV

ΔEV = 1.2 eV

SiO2

Eg = 8.6 eV

ΔEv = 1.1 eV

Al2O3

Eg = 6.7 eV

ΔEv = 1.2 eV

InAlN

Eg = 4.6 eV

ΔEC = 0.9 eV

Fig. 4.
Fig. 5.
Fig. 6.

Fig. 7.
Fig. 8.

(a) O 1s

(b) In 3d$_{5/2}$

(c) N 1s

(d) Al 2p

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Fig. 9.

(a) O 1s binding energy [eV] for SiO$_2$ and Al$_2$O$_3$. 

(b) In 3d$_{5/2}$ binding energy [eV] for InAlN and oxide.

(c) N 1s binding energy [eV] for InAlN.

(d) Al 2p binding energy [eV] for InAlN and Al$_2$O$_3$. 

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Intensity vs. Binding energy [eV] graphs showing the peak locations and intensities for O 1s, In 3d$_{5/2}$, N 1s, and Al 2p in the samples SiO$_2$, Al$_2$O$_3$, and InAlN.
Fig. 10.

![Intensity vs. Binding energy plot for In 3d$_{5/2}$ after SiO$_2$ deposition. The plot shows the intensity at different Al$_2$O$_3$ thicknesses: 0.5 nm and 1 nm.](image1)

Fig. 11.

![Graph showing $E_{\text{Al}_{2}\text{O}_3}$ vs. $E_{\text{InAlN}}$ for different Al$_2$O$_3$ thicknesses: 0 nm, 0.5 nm, and 1 nm.](image2)
Fig. 12.

![Graph showing normalized intensity vs. binding energy for O 1s at different angles: θ = 15°, 45°, 75°. Peaks indicate Al-O-H and Al-O-Al bonds.]