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# Ultrawide Bandgap Gallium Oxide Power Switch

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## EXECUTIVE SUMMARY

This report presents research conducted by the Naval Research Laboratory, Code 6881, on increased power levels in Ga<sub>2</sub>O<sub>3</sub> power switching devices by developing advanced lateral thermally managed, enhancement mode, p-type heterojunction gate (Al<sub>x</sub>Ga<sub>1-x</sub>)<sub>2</sub>O<sub>3</sub>/Ga<sub>2</sub>O<sub>3</sub> heterojunction field effect transistor (HFET), leading to an improved breakdown voltage (2 kV) and DC power (10 kW) at 1 MHz switching frequency. The effective thermal conductivity of the device will be enhanced via plasma-activated wafer bonding to a high thermal conductivity SiC substrate, mitigating self-heating effects in the device by creating a thin Ga<sub>2</sub>O<sub>3</sub>/SiC device platform. Ability to withstand higher drain bias will be achieved via a low-doped homoepitaxial Ga<sub>2</sub>O<sub>3</sub> buffer layer and a top-side slanted field plate. Integrated device design and optimization via simulation for high blocking voltage, high current, and fast switching will be implemented to meet program metrics.

Demonstrate high output power lateral Ga<sub>2</sub>O<sub>3</sub> technology by developing advanced (Al<sub>x</sub>Ga<sub>1-x</sub>)<sub>2</sub>O<sub>3</sub>/Ga<sub>2</sub>O<sub>3</sub> heterojunction field effect transistors (HFET) featuring a thermally managed substrate, enhancement mode operation using a p-type heterojunction gate, and a slanted high field plate. Demonstrate improved breakdown voltage (2 kV) and DC power density (50 W/mm, 5X) at 2 MHz switching frequency (10X). A high effective thermal conductivity Ga<sub>2</sub>O<sub>3</sub>/SiC device platform will be created via plasma-activated wafer bonding to a high thermal conductivity SiC substrate, mitigating self-heating effects in the device that limit output power in wide bandgap semiconductor devices. Advanced delta-doped ternary (Al<sub>x</sub>Ga<sub>1-x</sub>)<sub>2</sub>O<sub>3</sub> alloy will improve channel carrier density and enable high output power. Ability to withstand higher drain bias will be achieved via a low-doped homoepitaxial Ga<sub>2</sub>O<sub>3</sub> buffer layer and a top-side slanted field plate. Enhancement mode operation will be achieved via a novel natural p-type CuI or nanocrystalline diamond material integrated as a heterojunction gate. Device design and optimization via simulation for high blocking voltage, high current, and fast switching will be implemented to meet program metrics.

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# ULTRA-WIDE BANDGAP GALLIUM OXIDE POWER SWITCH

## 1. BACKGROUND

Gallium Oxide ( $\text{Ga}_2\text{O}_3$ ) is an ultra-wide bandgap semiconductor with six known phases with crystalline properties researched for almost six decades. The monoclinic phase of  $\text{Ga}_2\text{O}_3$  ( $\beta\text{-Ga}_2\text{O}_3$ ) is the only thermally stable phase with an ultra-wide bandgap (4.85 eV) and high critical field ( $E_c = 6\text{-}8$  MV/cm), resulting in a nearly ten-fold higher Baliga figure of merit than that of 4H-SiC (BFOM of  $\text{Ga}_2\text{O}_3 = 3444$ , BFOM of 4H-SiC = 300) [1]. Recent major advances in  $\text{Ga}_2\text{O}_3$  melt-growth techniques have resulted in commercial substrate technology, which as of late 2018 has reached the 4-inch wafer scale. This critical milestone in semiconductor technology, combined with its attractive properties, have positioned  $\text{Ga}_2\text{O}_3$  to offer savings in both energy and cost in high-power, high-temperature electronic device applications. In addition, native melt-grown  $\text{Ga}_2\text{O}_3$  substrates have enabled homoepitaxial growth of lattice-matched  $\text{Ga}_2\text{O}_3$  epilayers by a number of methods, including metal organic CVD, molecular beam epitaxy, and halide vapor phase epitaxy, among others. These high quality, low defect density films can be doped n-type over six orders of magnitude ( $10^{14}\text{-}10^{20}$   $\text{cm}^{-3}$ ). Low-doped epitaxial  $\text{Ga}_2\text{O}_3$  can serve as a high voltage blocking buffer layer in a transistor device, and its high critical field significantly reduces the required thickness, thus mitigating extended defect density. Furthermore,  $\text{Ga}_2\text{O}_3$  can readily form ternary alloys with In and Al, and the Al composition in a  $\beta\text{-(Al}_x\text{Ga}_{1-x})_2\text{O}_3$  alloy can readily reach 25 percent without compromising phase control, which is sufficient for a ternary heterostructure to be used as the channel of a field effect transistor. Such an ultra-wide bandgap transistor device has the potential to operate at very high power because of fundamental  $\text{Ga}_2\text{O}_3$  properties, and is not limited by the lack of p-type doping in this material. Simulations have shown that the extremely low thermal conductivity of  $\text{Ga}_2\text{O}_3$  ( $< 30$  W/m-K) can be effectively mitigated by controlled thinning of epitaxial  $\text{Ga}_2\text{O}_3$  and its bonding to a high thermal conductivity substrate such as SiC. NRL has explored a number of fundamental material and device issues under current 6.1 investment and has been a leader in this technology with a number of fundamental contributions to the nascent  $\text{Ga}_2\text{O}_3$  power electronics field. Among NRL's demonstrations are a 2.3 kV breakdown Schottky diode, a 33 A forward current Schottky diode, and a  $\text{Ga}_2\text{O}_3$ -on-diamond transistor by mechanical exfoliation of quasi-2D  $\text{Ga}_2\text{O}_3$  platelets. The last device exhibited over 60 W/mm DC output power density, the highest for any known semiconductor device, with similar thermal boundary resistance to state-of-the-art GaN-on-diamond technology but reduced self-heating effects. NRL has also pioneered advanced characterization of  $\text{Ga}_2\text{O}_3$  materials and devices such as ultraviolet PL imaging and Raman thermography, which will be essential to quantify device performance in the proposed program. NRL has also been a leader in understanding the effect of point defects such as Ga and O vacancies on the conductivity of  $\text{Ga}_2\text{O}_3$ , as their mitigation is critically important to achieving high performance devices. The application space of  $\text{Ga}_2\text{O}_3$  will likely not compete with mm-wave frequency III-Nitride heterostructures as it is a nonpolar material. The low symmetry of the  $\text{Ga}_2\text{O}_3$  monoclinic lattice also limits its intrinsic mobility to below 300  $\text{cm}^2/\text{Vs}$  at room temperature. However, the high critical field of  $\text{Ga}_2\text{O}_3$  makes it a viable candidate for certain classes of power electronics where size, weight, and power (SWaP) are critically important: pulsed power and power distribution in electric ships and aircraft, compact kW power supplies for decoys, UAV starter generator circuits, and compact converters for UUV wireless transfer charging and space robotics, among others.

## 2. TECHNICAL APPROACH

In order to meet program goals and achieve high power operation, the following key features are incorporated into the device structure: 1) a bonded SiC substrate for maintaining low junction temperature, 2) p-type heterojunction gate for normally-off operation, 3) tapered p-gate design for field termination by means of gradual reduction of channel carrier density in the access region, 4) a low-doped homoepitaxial Ga<sub>2</sub>O<sub>3</sub> buffer layer for high voltage blocking. Device design and optimization will be performed by means of drift-diffusion simulation for high power and fast switching. NRL has recently demonstrated plasma-activated process to bond a commercial 2-inch (-201) Ga<sub>2</sub>O<sub>3</sub> substrate to a 2-inch Si wafer for the first time. Similarly, a semi-insulating 4-inch SiC substrate will be bonded to the epitaxial side of a 4-inch (001) Ga<sub>2</sub>O<sub>3</sub> epiwafer. The Ga<sub>2</sub>O<sub>3</sub> epiwafer will be processed using a H<sup>+</sup> implant prior to bonding. A post-bonding thermal treatment will cause spontaneous splitting of the implanted layer (a.k.a., a SmartCut(TM) process), separating the high-quality (001) epilayer bonded to the SiC substrate from the starting Ga<sub>2</sub>O<sub>3</sub> substrate. We note that the (001) and (00-1) faces of the monoclinic lattice are equivalent as only the  $\beta$  angle in the monoclinic lattice is not orthogonal. As a result, upon a post-polish process the (00-1) Ga<sub>2</sub>O<sub>3</sub> surface will be ready for subsequent low-doped Ga<sub>2</sub>O<sub>3</sub> buffer layer epitaxy. A low thermal boundary resistance (TBR) is expected to be obtained using this process. Preliminary TBR characterization and device fabrication has been performed using time domain thermorefectance of exfoliated (100) Ga<sub>2</sub>O<sub>3</sub> flakes onto diamond and SiO<sub>2</sub>/Si surfaces. Thermal characterization will be performed by Raman thermography in the proposed program tasks using the 169 cm<sup>-1</sup> vibrational mode of Ga<sub>2</sub>O<sub>3</sub> as its temperature dependent Raman shift is linear over a wide temperature range. We note that thermorefectance imaging commercial technology would only be able to characterize temperature on the contact pads of any Ga<sub>2</sub>O<sub>3</sub> device, since Ga<sub>2</sub>O<sub>3</sub> is transparent to the sub-bandgap near-UV illumination typically used in these systems. Normally-off operation of the proposed device will be achieved using p-type gate technology leveraging prior NRL 6.2 investment (p-type diamond gated AlGa<sub>0.5</sub>N/GaN HEMTs) and current NRL 6.2 investment demonstrating p-type CuI (Cuprous Iodide) heterojunctions to Ga<sub>2</sub>O<sub>3</sub>. A high quality epitaxial Cu deposited by sputter-beam epitaxy on single-crystalline Ga<sub>2</sub>O<sub>3</sub> is exposed to iodine vapor under controlled temperature, forming a high quality CuI film with free hole carrier density in excess of 10<sup>18</sup> cm<sup>-3</sup>, as measured by Hall effect. Depending on iodination conditions, the Cu vacancies concentration, a shallow acceptor in CuI ( $E_g = 3.1$  eV), has been tailored to demonstrate heterojunction diodes with ideality factors ranging from 1.37 to 2.25. One challenge is the need for passivation of the CuI film as it can be easily oxidized in air even at moderate temperatures. Effective passivation by a metal or high-k dielectric cap, or alternative p-type material integration (p-type GaN or nanocrystalline diamond), will be developed throughout the program. Nanocrystalline diamond growth has been successfully demonstrated on (010) Ga<sub>2</sub>O<sub>3</sub> substrates for the first time at NRL using a 30 nm thick LPCVD SiN barrier layer. Fabrication of effective electric field termination structures will be performed using a tapered p-type gate for reducing charge density and thus spreading the electric field beyond the drain edge of the gate. The p-gate layer will be patterned using grayscale lithography, producing a slanted photoresist profile which will then be transferred to the p-type gate by dry etching. In addition to two-tone photolithography capability, our team has developed 3D-nanolithography expertise to 3D-print slanted photoresist profiles on GaN. While epitaxial Ga<sub>2</sub>O<sub>3</sub> is expected to be unintentionally doped n-type, and n-type conductivity can be controlled by either in-situ Si or Sn doping or ex-situ by Si implantation, p-type doping is not expected to be a realistic option for Ga<sub>2</sub>O<sub>3</sub> due to the self-trapping of holes in this material. However, acceptors can still be useful, as we have recently demonstrated by co-doping HVPE Ga<sub>2</sub>O<sub>3</sub> with N, a deep acceptor in Ga<sub>2</sub>O<sub>3</sub>. The resulting lateral Schottky barrier diodes exhibited 2.38 kV breakdown voltage (3.18 MV/cm) without any additional field-termination structures. Such controllably-low doped buffer layers will be integrated with an (Al<sub>x</sub>Ga<sub>1-x</sub>)<sub>2</sub>O<sub>3</sub> barrier layer in this program, resulting in a two-dimensional electron gas channel for low on-state resistance and high output power. Preliminary (Al<sub>x</sub>Ga<sub>1-x</sub>)<sub>2</sub>O<sub>3</sub> growth experiments have confirmed the phase stability of the ternary barrier layer with an Al concentration of 24 percent, and a total channel carrier concentration in excess of 10<sup>19</sup> cm<sup>-3</sup> measured by capacitance-voltage techniques for an 18 nm thick barrier layer.

### 3. EXPERIMENTAL RESULTS

#### 3.1 Design of Gallium Oxide Modulation Doped Field Effect Transistors

Monoclinic  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> based transistors possess fundamental electronic properties that are advantageous for high power devices.<sup>1</sup> A number of these properties derive directly from the wide band-gap of  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> ( $E_g = 4.8$  eV) including an exceptionally high electric breakdown field (approximately 8 MV/cm).<sup>2</sup> The high breakdown field allows for the design of a lateral transistor with a short channel that can withstand a high critical field. The short channel design lowers the switching loss and, also, lowers the on-resistance, which reduces the conduction loss.<sup>3</sup> Additionally, the short channel allows the high-voltage switch to operate at high frequency ( $> 1$  MHz) and, thus, enables a system design with smaller (size and weight) passive circuit components.<sup>4</sup>

A lateral metal-semiconductor or metal-oxide-semiconductor field effect transistor is typically based on an n-type doped channel. At carrier concentrations greater than  $1 \times 10^{18} \text{ cm}^{-3}$ , the mobility of  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> is limited to approximately  $50 \text{ cm}^2/\text{V}\cdot\text{s}$  due to impurity scattering.<sup>5</sup> In contrast, at a carrier concentration of  $2.5 \times 10^{16} \text{ cm}^{-3}$ , a mobility of  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> of  $184 \text{ cm}^2/\text{V}\cdot\text{s}$  has been reported – primarily limited by polar optical phonon scattering.<sup>6</sup> This provides the motivation for a modulation doped design where donors within the  $(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3$  barrier accumulate in a triangular potential well in the undoped Ga<sub>2</sub>O<sub>3</sub> crystal at the  $(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3 / \text{Ga}_2\text{O}_3$  interface. An  $(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3$  spacer layer physically separates the dopant atoms from the electrons in the conductive channel, which reduces impurity scattering and, thus, improves electron mobility.

With proper design, conduction predominantly occurs via the dense two-dimensional electron gas (2DEG) that forms in the potential well. Nevertheless, a deleterious secondary channel may form at the modulation-doping location in the  $(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3$  barrier. The supplemental to this article provides general design rules for optimal formation of the 2DEG. The supplemental focuses on mitigating three limitations related to current deposition technology involving the aluminum content in  $(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3$ , modulation-doping concentration, and the persistent tail of dopant atoms. Below, it is shown that residual charge in the  $(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3$  barrier manifests as a secondary inflection in the transconductance curve.

This primary aim of this article is to investigate the coupled relation of on-state self-heating and spatially-dependent mobility reduction, and the resultant alteration of the forward bias response. The mobility decreases significantly at elevated temperature and the low-thermal conductivity of  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> results in self-heating of the device at moderate current density. Approximations to the mobility and thermal conductivity of  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> are presented that can be readily implemented in a TCAD simulation environment. It is shown that self-heating and the electric-field dependence will significantly alter the cross-sectional mobility profile. Significant self-heating can result in a negative differential conductance in the drain current response to increasing drain voltage.

##### 3.1.1 Modelling

A series of electro-thermal simulations, via the Silvaco TCAD environment, were conducted for a modulation doped  $(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3 / \text{Ga}_2\text{O}_3$  structure to provide insight into the device current-voltage relationships as well as the cross-sectional profiles of current density, mobility, and temperature.

##### 3.1.2 Thermal Conductivity and Mobility Models

Accurate description of the temperature dependent parameters is necessary for simulation of a transistor based on Ga<sub>2</sub>O<sub>3</sub>, which suffers from self-heating related to its low thermal conductivity. For application to a TCAD simulation, the thermal conductivity model given in Mu et al.<sup>7</sup> was fit to a power-law thermal model given by

$$\kappa(T) = \kappa_{300} \left( \frac{T}{300} \right)^\alpha, \quad (1)$$

with the parameters listed in table 1.

Table 1. Orientation dependent model parameters for thermal conductivity for Ga<sub>2</sub>O<sub>3</sub> and AlGaO<sub>3</sub>, and a heat-sink material comparable to ceramic poly-AlN.

|                                |       | $\kappa_{300}$<br>[W/cm · K] | $\alpha$ |
|--------------------------------|-------|------------------------------|----------|
| Ga <sub>2</sub> O <sub>3</sub> | [100] | 0.111                        | -1.33487 |
|                                | [010] | 0.155                        | -1.25526 |
|                                | [001] | 0.153                        | -1.31247 |
| AlGaO <sub>3</sub>             | [100] | 0.198                        | -1.62331 |
|                                | [010] | 0.25                         | -1.48543 |
|                                | [001] | 0.304                        | -1.61316 |
| Heat-Sink                      |       | 1.8                          | -1.3     |

Ma et al. developed a model of mobility of Ga<sub>2</sub>O<sub>3</sub>, that accounted for several scattering mechanisms.<sup>8</sup> The Ma mobility model, based on the Boltzmann transport equation via the relaxation-time approximation solution, requires two separate integrations. To introduce a temperature and donor dependent model into the device simulation, the Boltzmann model of Ma et al. was fit to the Arora low-field mobility model<sup>9</sup> with a structure in the Silvaco TCAD environment given by

$$\mu_n = \mu_1 \left( \frac{T}{300} \right)^\alpha + \frac{\mu_2 \left( \frac{T}{300} \right)^\beta}{1 + \frac{N}{N_{crit} \left( \frac{T}{300} \right)^\gamma}}, \quad (2)$$

with fitting parameters given by  $\mu_1 = 1.67938 \text{ cm}^2/\text{V}\cdot\text{s}$ ,  $\mu_2 = 264.557 \text{ cm}^2/\text{V}\cdot\text{s}$ ,  $\alpha = -0.116259$ ,  $\beta = -2.29257$ ,  $\gamma = 2.64426$ , and  $N_{crit} = 4.29447 \times 10^{18} \text{ cm}^{-3}$ .

Figure 1(a) displays the calculated mobility of the Arora model compared to the calculated Ma model and the specific scattering mechanisms defined in the Ma model. Figure 1(b) and 1(c) displays the Arora calculated mobility as a function of dopant density and temperature. The Arora model replicates the Ma model, which predicts a transition from polar optical phonon scattering as the limiting mechanism at a low donor level and impurity scattering as the dominant mechanism limiting the mobility, above 200 °K, at a high donor levels.

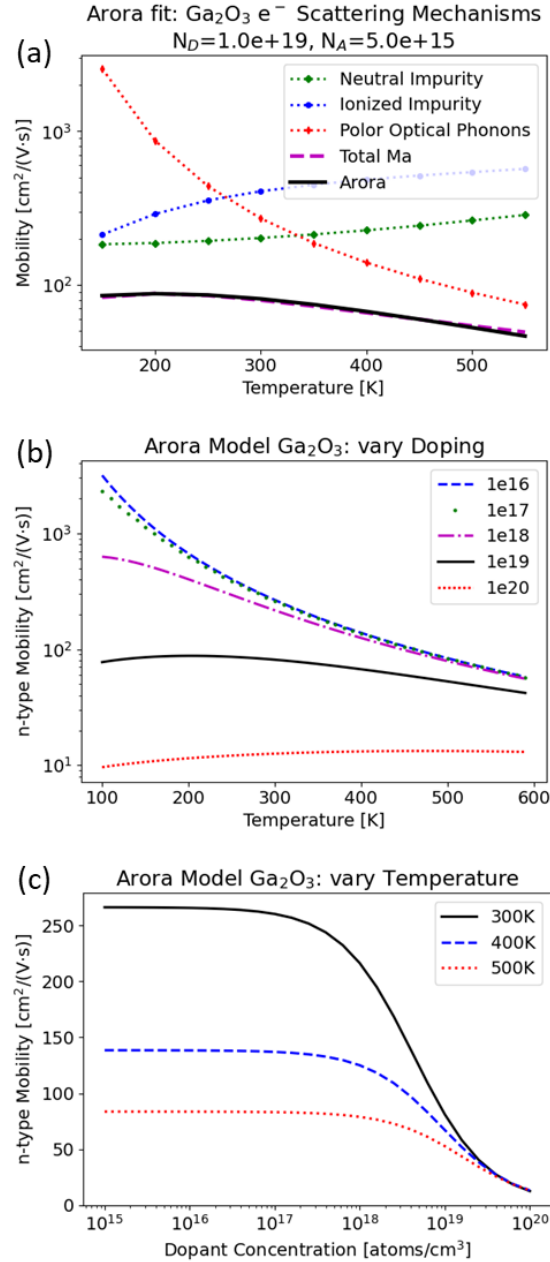


Fig. 1. (a) Comparison of fitted Arora model to Ma mobility model with its relevant scattering mechanisms. Mobility of in  $\beta$ -Ga<sub>2</sub>O<sub>3</sub>. In the Arora model as a function of donor (b) temperature and (c) donor concentration. Color image available in the online version.

### 3.1.3 Device Structure

The base simulation structure consists of 20 nm ALD Al<sub>2</sub>O<sub>3</sub>, 30 nm (Al<sub>0.25</sub>Ga<sub>0.75</sub>)<sub>2</sub>O<sub>3</sub> with a background donor concentration of  $1 \times 10^{17} \text{ cm}^{-3}$  and modulation doped at  $4 \times 10^{19} \text{ cm}^{-3}$  with a Gaussian distribution peaked at 4 nm from the layer bottom and a character length of 0.5 nm, 200 nm Ga<sub>2</sub>O<sub>3</sub> with a background donor concentration of  $4 \times 10^{19} \text{ cm}^{-3}$ , and a 200  $\mu\text{m}$  (201) Ga<sub>2</sub>O<sub>3</sub>:Fe substrate with a thermal boundary resistance of  $0.001 \text{ cm}^2\cdot\text{K}/\text{W}$ . The source and drain contacts are 50  $\mu\text{m}$  in length and composed of 20 nm Ti and 300 nm Au. The gate contact is 5  $\mu\text{m}$  in length and composed of 20 nm Ni and 300 nm Au. The source to gate spacing is 3  $\mu\text{m}$  and the gate to drain spacing is 12  $\mu\text{m}$ . The remaining structure consists of Si<sub>3</sub>N<sub>4</sub>.

As shown in Fig. 2 a heatsink, with a temperature dependent thermal conductivity similar to ceramic poly-AlN, is optionally added. The heatsink is 50  $\mu\text{m}$  in length and ends 5  $\mu\text{m}$  from the edge of the source metalization. The heatsink is separated by 50 nm of  $\text{Si}_3\text{N}_4$  from the source metalization. The heatsink has a thermal boundary resistance of  $0.001 \text{ cm}^2\cdot\text{K}/\text{W}$  and the  $\text{Ga}_2\text{O}_3\cdot\text{Fe}$  substrate is thermally isolated.

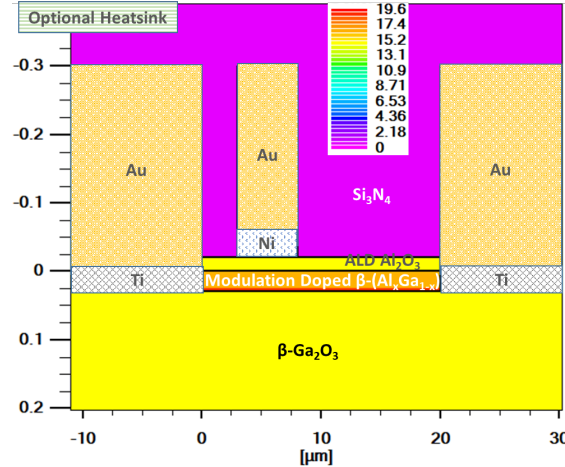


Fig. 2. Components and log donor concentration [ $\text{cm}^{-3}$ ] profile of base simulation structure. The MODFET is designed to have the primary path of current flow in the triangular potential well formed in the first few nanometers in the undoped  $\text{Ga}_2\text{O}_3$  layer below the  $(\text{Al}_{0.25}\text{Ga}_{0.75})_2\text{O}_3 / \text{Ga}_2\text{O}_3$  interface. Color image available in the online version.

### 3.1.4 Results

#### 3.1.4.1 Electro-Thermal Simulation With Thermal Contact To Substrate Base

A simulation was conducted of the device structure, with a thermal contact to the substrate, as described in the previous section. The drain current and peak temperature as a function of forward drain bias are displayed in Fig. 3(a). At a drain bias above 3 V, the temperature of the device rapidly increases. Additionally, above 7 V, a negative differential conductivity is observed. The increase in temperature relative to the increase in power corresponds to a device thermal resistance ( $R_{\text{th}} = \Delta T/P_D$ ) of  $167 \text{ mm}\cdot\text{W}/\text{K}^\circ$  at 10 V (and  $359 \text{ }^\circ\text{K}$ ).

Figure 3(a) displays the drain current and peak temperature as a function of gate bias for drain voltage of 5 V and 10 V. At a drain voltage of 5 V and 10V, the peak device temperature at plateaus at approximately  $325 \text{ }^\circ\text{K}$  and  $385 \text{ }^\circ\text{K}$ , respectively. At a large positive gate bias, the gate minimally impacts the channel, which behaves similar to a resistor. Moving from a drain voltage of 5 V to 10 V corresponds to a 47% increase in drain current but a 132% increase in temperature.

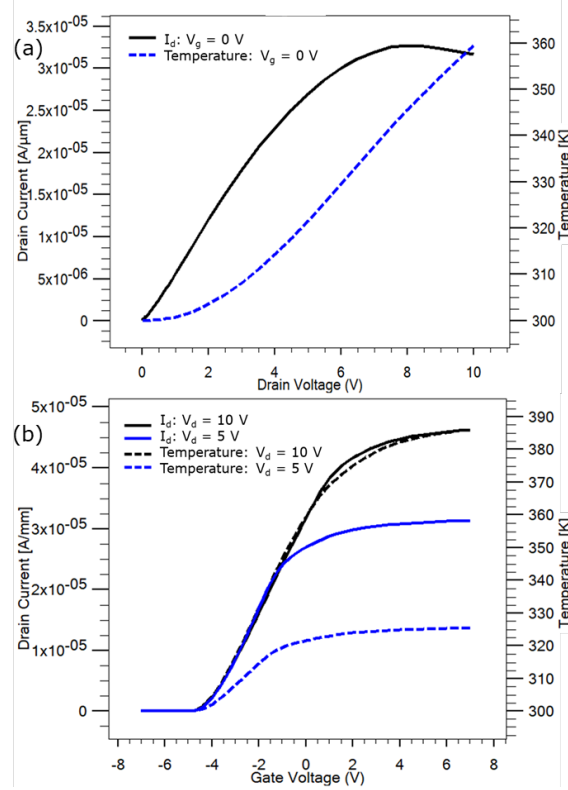


Fig. 3. Drain current and peak temperature as a function of (a) drain voltage at zero gate bias and (b) gate bias at a drain voltage of 5 V and 10 V. The temperature vs. drain bias in (a) displays a slope of 8.2 K°/V for  $V_{ds}$  greater than 5V. Color image available in the online version.

Figure 4 compares the log current density, x-component (direction from source to drain) of the mobility, and temperature at a fixed drain current (of  $V_{ds} = 10$  V). The left-column corresponds to the near complete depletion of the channel (at  $V_g = -4$  V), center-column corresponds to the point of initial depletion (starting at the drain side of the gate edge) in the channel (at  $V_g = 0$  V), and right-column is the transitional point of depletion of the  $(Al_{0.25}Ga_{0.75})_2O_3$  barrier (at  $V_g = 1$  V) in Fig. 4.

In Fig. 4(d) and 4(e), the reduction in mobility at the drain side of the gate edge is caused by the peak in x-component of the electric field as predicted by the high-field dependent mobility model. This causes a local increase in resistance, which similarly causes an increase in Joule heating and a corresponding peak in temperature at the drain side of the gate edge. In Fig. 4(a) (at  $V_g = -4$  V) it is seen that the total current is low, thus, the Joule heating throughout the channel is low and the overall temperature increase in the structure is minimal as observable in Fig. 4(g).

In contrast, in Fig. 4(b) (at  $V_g = 0$  V) a significant current is flowing in the channel, which causes Joule heating along the length of the channel as well as a relatively higher increase in heating at the drain edge of the gate (due to the high electric-field induced mobility reduction). This generated heat sharply increases the temperature at the drain side of the gate edge in the channel (Fig. 4(h)). A temperature gradient exists moving away from the gate edge but the temperature in the surrounding structure is sufficient to cause a appreciable reduction in the mobility in the corresponding region (Fig. 4(e)).

The transitional point of depletion of the  $(Al_{0.25}Ga_{0.75})_2O_3$  barrier (at  $V_g = 0$  V) reveals a small yet discernable secondary current flow in  $(Al_{0.25}Ga_{0.75})_2O_3$  barrier near the location of the modulation doping (Fig. 4(c)). Additionally, at this gate bias, the electric field from the gate has essential no interaction with the primary channel. Hence, the potential well below the  $(Al_{0.25}Ga_{0.75})_2O_3$  /  $Ga_2O_3$  interface effectively acts as a  $Ga_2O_3$  resistor. The manifests as heating along the entire channel that creates a non-localized heating profile (Fig. 4(i)) and a relatively uniform reduction in mobility  $Ga_2O_3$  layer (Fig. 4(f)).

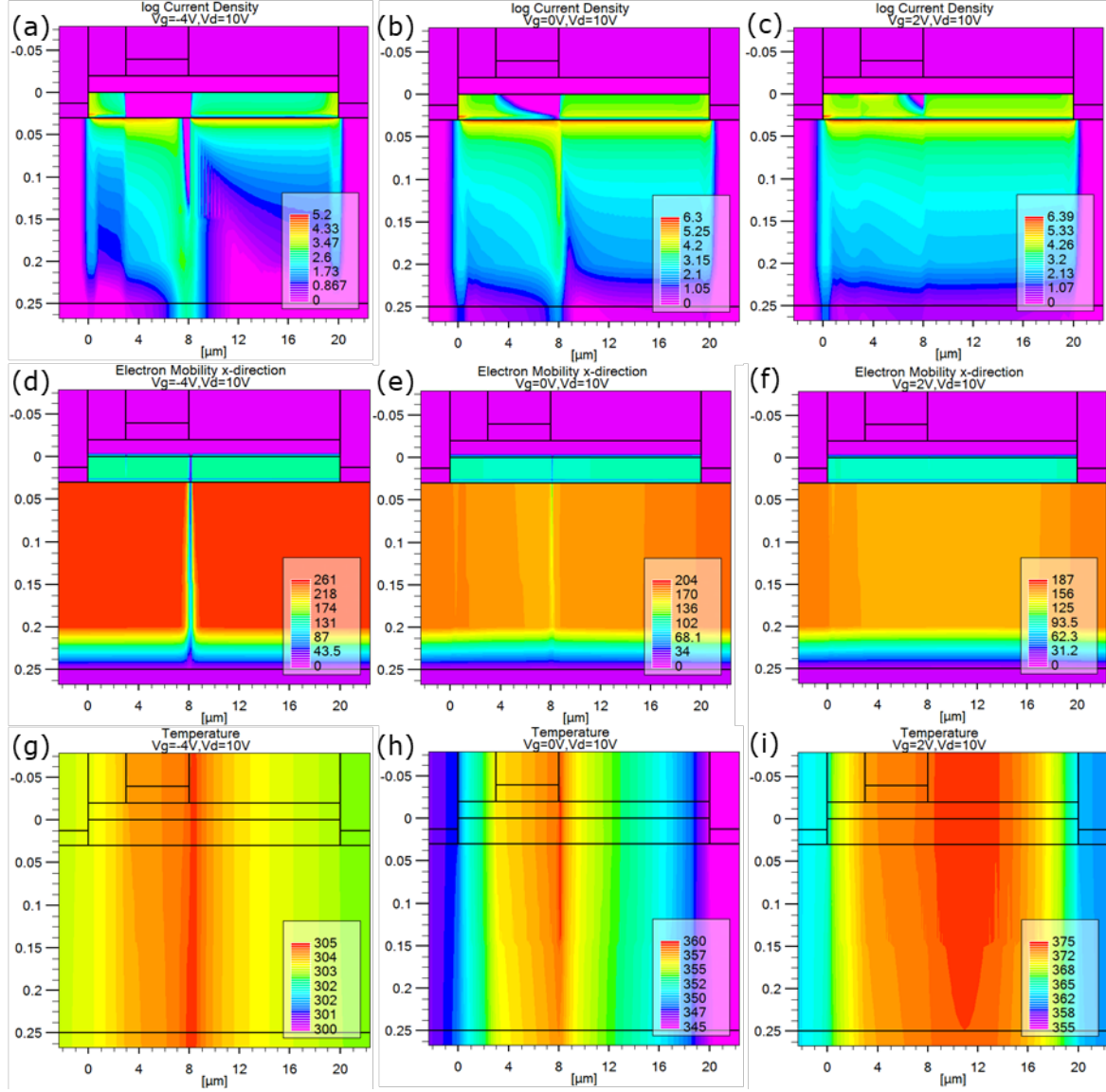


Fig. 4. Cross-sections for  $V_{ds}$  at 10 V with  $V_g$  at -4 V (a, d, and g), with  $V_g$  at 0 V (b, e, and h), and with  $V_g$  at 2 V (c, f, and i) for (a, b, c) log current density [ $A/cm^2$ ], (d, e, f) x-component of electron mobility [ $cm^2/V \cdot s$ ], and (g, h, i) lattice temperature [ $^{\circ}K$ ]. The left-column corresponds to the near complete depletion of the channel (at  $V_g = -4$  V), which greatly reduces (a) current flow; a peak in electric field reduces the (d) mobility at the gate edge, however, the overall current flow is low, which limits Joule heating. The center-column corresponds to the point of initial depletion (starting at the drain side of the gate edge) of the channel (at  $V_g = 0$  V) with (b) significant current flowing along the channel; (e) the sharp reduction in mobility at the drain side of the gate edge is caused by the peak in electric field and the inter-related self-heating effects on (h) temperature. The right-column corresponds to the transitional point of depletion of the  $(Al_{0.25}Ga_{0.75})_2O_3$  barrier (at  $V_g = 1$  V) with (c) large current flowing through the channel with little interaction with the electric field from the gate; (i) a broad heating and temperature profile occurs owing to a relatively uniform Joule heating along the length of the channel. Color image available in the online version.

### 3.1.5 Comparison to Isothermal Simulation

A simulation was conducted of the same device structure except thermal heating is neglected, that is, the device is held at 300 °K. Examination of the mobility cross-section, in Fig. 5(b), reveals a sharp decrease in mobility at the drain side of the gate edge – and a smaller decrease at the edge of the source contacts. This decrease in mobility is due to local peak in electric field.

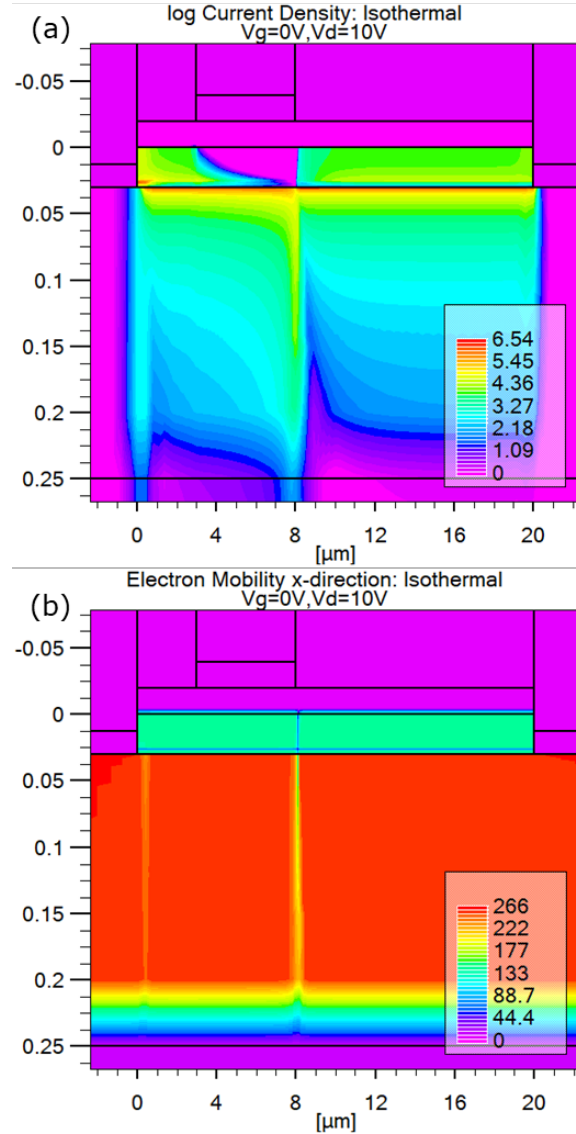


Fig. 5. Cross-sections of iso-thermal simulation at 300 °K for  $V_{ds}$  at 10V for (a) log current density [ $A/cm^2$ ] and (b) x-component of electron mobility [ $cm^2/V \cdot s$ ]. As is common in transistors without field plates, there is a peak in electric field at the drain side of the gate edge, which results in (b) a local reduction in the x-component of electron mobility. Color image available in the online version.

Figure 6 provides a comparison of electro-thermal simulations compared to the equivalent isothermal simulation where the lattice temperature is fixed at 300 °K. The drain current at a drain voltage less than 3 V is similar in the thermal and iso-thermal simulations. This indicates that current density is sufficiently low such that the Joule heat generated can be effectively conducted to and dissipated at the base of the substrate. Additionally, although self-heating is minimal, it is clear the drain voltage is shifting the onset of saturation in drain current – for this particular device geometry and structure.

The drain current and transconductance at a drain voltage of 5 V show a small deviation in the thermal and iso-thermal simulations for a gate bias greater than -1 V. At this transition point the current density is sufficiently high that most but not all the Joule heat generated can be effectively dissipated at the base of the substrate. For a drain voltage of 5 V, an inflection is present in the transconductance curve (Fig. 6(b)) for both the thermal and iso-thermal simulation. This inflection corresponds to the point where the gate voltage is sufficiently positive such that charge, located at modulation doping position in the  $(Al_{0.25}Ga_{0.75})_2O_3$  barrier, is no longer depleted.

As discussed in the previous section, at a drain voltage of 10 V, self-heating can raise the temperature of the lattice, which significantly decreases the electron mobility in and near the channel. This increases the resistance of the device, which is evident in the large difference in the thermal and iso-thermal simulations in Fig. 6.

Figure 6(a) shows a negative differential conductivity for a drain voltage above 7 V with a gate bias of -2 V and 0 V. For this particular device structure, negative differential conductivity is not seen for the isothermal simulations. Here, two mechanisms contribute to the negative differential conductivity; specifically, the sharp reduction in mobility at the drain side of the gate edge caused by the peak in electric field, and the overall reduction in mobility along the entire length of the channel caused by the temperature rise related to self-heating.

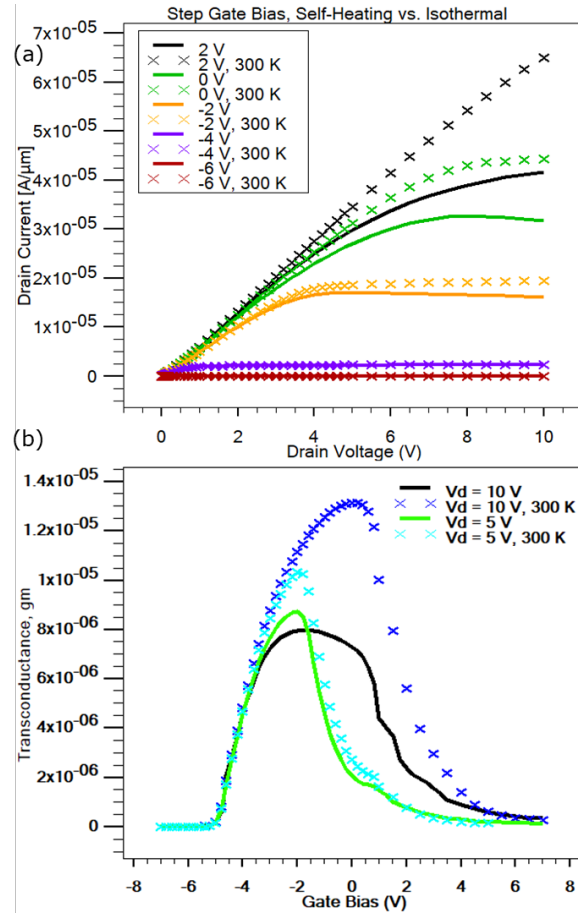


Fig. 6. Comparison of electro-thermal simulation with self-heating vs. iso-thermal simulation at 300 °K. (a) Drain current as a function of drain voltage for series of 2 V steps in gate voltage. (b) Transconductance as a function of gate bias at a drain voltage of 5 V and 10 V. Current saturation due to the high-field mobility reduction effect and global heating mechanisms are inter-related. Still, a significant reduction is observed in drain current owing to the reduction in mobility due to self-heating compared to the iso-thermal simulation. Color image available in the online version.

### 3.1.6 Heat Removal Through the Source Metal

A simple improvement to the thermal design is to place a heatsink material on the surface of the structure near but offset from the source metallization. Most transistor test structures have a large source pad above the source region; that is, the source contact is not accessed via a metal finger. As described above, the device surface is covered with a  $\text{Si}_3\text{N}_4$  layer. A 50nm  $\text{Si}_3\text{N}_4$  layer is between the top of the source metallization and the heatsink material, and the heatsink material is offset  $5\mu\text{m}$  from the edge of the source contact. To further simplify the analysis, no heat exits the bottom of the substrate.

Figure 7(a), shows that the drain current saturates at a drain bias greater than 7V although a slight negative differential conductivity is evident at a drain bias of 10V. Examination of the temperature cross-section, in Fig. 7(b), reveals the expected peak at the drain side of the gate edge – and an overall gradient towards the heat sink located near the source contacts. The increase in temperature relative to the increase in power corresponds to a thermal resistance,  $R_{th}$ , of  $114 \text{ mm}^2\text{W/K}^\circ$  at 10 V (and 327 °K). The thermal resistance (in Fig. 7) for the source metal thermal contact is 41% the thermal resistance for the bottom substrate thermal contact (in Fig. 3).

A similar thermal resistance of the overall device is found in a simulation (not shown) for a device formed on a  $20 \text{ }\mu\text{m}$  substrate with a bottom thermal contact. This result is reasonable as the  $20 \text{ }\mu\text{m}$  distance to the substrate base is similar to the distance from the drain edge of the gate to the source contact. Hence, thinning the substrate is another approach; however, the propensity for  $\beta\text{-Ga}_2\text{O}_3$  to cleave discourages a post-growth wafer thinning approach. Still, several works have demonstrated devices processed on films exfoliated  $\text{Ga}_2\text{O}_3$  films with a thickness on the order of  $10 \text{ }\mu\text{m}$ .<sup>10-13</sup>

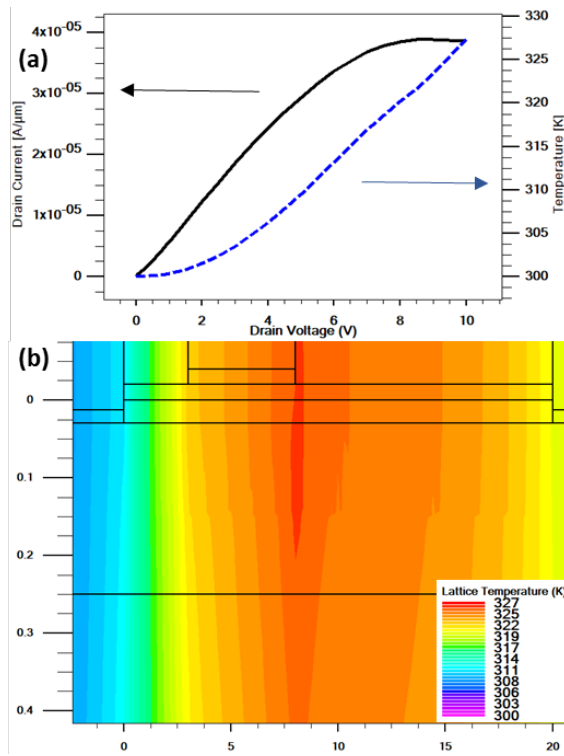


Fig. 7. Simulation with heat extraction via source metallization. (a) Drain current and temperature as a function of drain voltage at zero gate bias. (b) Temperature [ $^\circ\text{K}$ ] cross-sections for  $V_{ds}$  at 10V. The temperature vs. drain bias in (a) displays a slope of  $3.4 \text{ K}^\circ/\text{V}$  for  $V_{ds}$  greater than 5V. Color image available in the online version.

### 3.1.7 Discussion

#### 3.1.7.1 Self-Heating And Mobility

Thermal management is an important design consideration in all semiconductor power devices.<sup>14</sup> The simulations in this work show that self-heating owing to the low thermal conductivity of  $\beta\text{-Ga}_2\text{O}_3$  dramatically lowers the mobility in the channel region, which increases the on-state resistance at moderate current densities. The low thermal conductivity of  $\beta\text{-Ga}_2\text{O}_3$  necessitates advanced thermal management solutions for operation at even moderate power levels.<sup>15,16</sup> Candidate solutions include flip-chip designs where heat is extracted globally from the top surface as well as more nuanced modification of the gate region to enhance local heat extraction at or near the gate.<sup>17-19</sup> Naturally, thermal mapping of the device surface or thermal depth probe

directly measures the thermal response of the device.<sup>20</sup> Nevertheless, these measurements can be obscured by the contact structure. Additionally, a thermo-reflectance imaging measurement is complicated by the requirement for illumination greater than the wide bandgap of  $\beta$ -Ga<sub>2</sub>O<sub>3</sub>. As outlined in this paper, understanding the relation of self-heating on the response of the drain current to gate bias and drain voltage provides an additional assessment of the thermal management in the device structure.

### 3.1.7.2 Modulation Doping Considerations Of $\beta$ -Ga<sub>2</sub>O<sub>3</sub> MODFET

The presence of residual charge at the point of modulation doping is undesired as this creates a secondary channel that hampers the ability to modulate the primary channel. This charge at the modulation doping point can be confirmed by an inflection in the transconductance curve. Specifically, as this charge is spatially closer to the gate metal, this inflection occurs at more positive gate bias relative to the primary channel, which is deeper in the structure.

The supplemental to this article presents design rules for the modulation-doped structure with a goal of maximizing the concentration of electrons in the potential well while minimizing a low-mobility secondary channel at the modulation-doped layer. An emphasis is placed on mitigating three limitations related to the deposition of  $\beta$ -Ga<sub>2</sub>O<sub>3</sub>; specifically, the inability to deposit high-quality  $\beta$ -(Al<sub>x</sub>Ga<sub>1-x</sub>)<sub>2</sub>O<sub>3</sub> with an aluminum mole fraction greater than approximately 0.25, the difficulty to deposit a high concentration of dopant atoms without the formation of its native oxide, and the unintentional dopant atom exponential decay tail. Simulations for an aluminum mole fraction of 0.25 show that proper selection of the barrier thickness and modulation-doping level, including accounting for any asymmetric broadening, can yield a dense electron channel confined primarily to the potential well at the (Al<sub>x</sub>Ga<sub>1-x</sub>)<sub>2</sub>O<sub>3</sub> / Ga<sub>2</sub>O<sub>3</sub> interface.

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### 3.2 Experimental determination of critical thickness limitations of (010) $\beta$ -(Al<sub>x</sub>Ga<sub>1-x</sub>)<sub>2</sub>O<sub>3</sub> heteroepitaxial films

#### 3.2.1 Introduction

Interest in ultrawide bandgap semiconductors ( $E_G > 3.4$  eV) has grown rapidly in the electronics community due to enhanced breakdown field strengths and potentially significant improvements in relevant device performance figures of merit. For example,  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> is predicted to have a breakdown field strength of up to 8 MV/cm, offering ~3200X improvement in Baliga's figure of merit (BFOM) as compared to Si.<sup>1</sup> In addition, the incorporation of Al into  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> to form  $\beta$ -(Al<sub>x</sub>Ga<sub>1-x</sub>)<sub>2</sub>O<sub>3</sub> increases the bandgap energy offering further improvement in breakdown field strength.<sup>2</sup> Heteroepitaxial  $\beta$ -(Al<sub>x</sub>Ga<sub>1-x</sub>)<sub>2</sub>O<sub>3</sub> has been demonstrated as a barrier layer in modulation-doped field effect transistors (MODFETs) to confine a two-dimensional electron gas (2DEG) below the heterointerface.<sup>3–12</sup> For such an application, a relatively thin (~20 nm)  $\beta$ -(Al<sub>x</sub>Ga<sub>1-x</sub>)<sub>2</sub>O<sub>3</sub> is generally accepted as sufficient. However, recent work has demonstrated the use of  $\beta$ -(Al<sub>x</sub>Ga<sub>1-x</sub>)<sub>2</sub>O<sub>3</sub> as the active channel layer as first steps towards realizing devices with even greater breakdown fields by taking advantage of the larger bandgap of  $\beta$ -(Al<sub>x</sub>Ga<sub>1-x</sub>)<sub>2</sub>O<sub>3</sub>.<sup>13,14</sup> Furthermore, the tunability of the bandgap energy by varying Al content is also alluring for optoelectronics applications such as deep-UV light emitting diodes (LEDs).<sup>2</sup>

To date,  $\beta$ -phase (Al<sub>x</sub>Ga<sub>1-x</sub>)<sub>2</sub>O<sub>3</sub> has been grown on several orientations of  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> substrates including (100)<sup>15–17</sup>, (010)<sup>18–21</sup>, (001)<sup>22,23</sup>, and ( $\bar{2}01$ ).<sup>24</sup> Of the aforementioned orientations, heteroepitaxial growth of  $\beta$ -(Al<sub>x</sub>Ga<sub>1-x</sub>)<sub>2</sub>O<sub>3</sub> on (010)  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> is most commonly reported, and growth is typically performed via molecular beam epitaxy (MBE)<sup>21,25–28</sup> or metalorganic chemical vapor deposition (MOCVD).<sup>18–20,29,30</sup> For heteroepitaxial growth, strain is induced in the film to accommodate lattice parameter mismatch. Initially, the film grows pseudomorphically and is fully strained as strain forces within the epitaxial layer are lower than those required for chemical bond fracture. However, once a critical film thickness is reached, film stress relaxation is induced via material-specific defect formation such as misfit dislocations or cracking. For example, Matthews and Blakeslee developed a model for critical thickness of GaAs<sub>1-x</sub>P<sub>x</sub> films heteroepitaxially grown on GaAs in which grown-in threading dislocations eventually lead to misfit dislocations for strain relaxation.<sup>31</sup> This model has also been applied to investigate strain relaxation mechanisms in Al<sub>x</sub>Ga<sub>1-x</sub>N/GaN heterostructures<sup>32</sup> and most recently for the (010)  $\beta$ -(Al<sub>x</sub>Ga<sub>1-x</sub>)<sub>2</sub>O<sub>3</sub>/Ga<sub>2</sub>O<sub>3</sub> heterostructure.<sup>26</sup> Strain relaxation via cracking has been theoretically investigated by Mu et al for heteroepitaxial growth of  $\beta$ -(Al<sub>x</sub>Ga<sub>1-x</sub>)<sub>2</sub>O<sub>3</sub> on  $\beta$ -Ga<sub>2</sub>O<sub>3</sub>.<sup>33</sup> In this study, the Griffith criterion was applied to determine the surface energies, which are proportional to the fracture toughness, of various planes in  $\beta$ -(Al<sub>x</sub>Ga<sub>1-x</sub>)<sub>2</sub>O<sub>3</sub>. For growth of  $\beta$ -(Al<sub>x</sub>Ga<sub>1-x</sub>)<sub>2</sub>O<sub>3</sub> on (010)  $\beta$ -Ga<sub>2</sub>O<sub>3</sub>, it was found that the (100)B plane had the lowest unrelaxed surface energy and is therefore the crack plane limiting the critical thickness of these films. While Ohtsuki and Higashiwaki<sup>26</sup> did not report experimental observation of film cracking in their study, the thicknesses of the films investigated was limited to 100 nm, likely because of the intention to use those films as a channel confining back barrier in  $\beta$ -(Al<sub>1-x</sub>Ga<sub>x</sub>)<sub>2</sub>O<sub>3</sub>/Ga<sub>2</sub>O<sub>3</sub> heterostructure field effect transistors. Understanding the critical thickness limitations for  $\beta$ -(Al<sub>x</sub>Ga<sub>1-x</sub>)<sub>2</sub>O<sub>3</sub> heteroepitaxial films as a function of Al content and the associated strain relaxation mechanisms are crucial for the growth of high-quality films for electronics applications.

In this study, we report the growth of a series of  $\beta$ -(Al<sub>x</sub>Ga<sub>1-x</sub>)<sub>2</sub>O<sub>3</sub> films grown on (010) Ga<sub>2</sub>O<sub>3</sub> substrates with varying thicknesses and Al content. The films were then investigated using X-ray diffraction (XRD), atomic force microscopy (AFM), X-ray topography (XRT), and high-resolution transmission electron microscopy (HR-TEM) to study film growth morphology,

film strain, and crack formation and propagation. The experimentally determined critical thicknesses of the  $\beta$ -(Al<sub>x</sub>Ga<sub>1-x</sub>)<sub>2</sub>O<sub>3</sub> films in this study were then compared with the theoretical calculations reported by Mu et al.<sup>33</sup> and experimental results of Bhuiyan et al.<sup>2</sup>

### 3.2.2 Experimental Details

A series of  $\beta$ -(Al<sub>x</sub>Ga<sub>1-x</sub>)<sub>2</sub>O<sub>3</sub> heteroepitaxial films were grown on nominally on-axis Fe-doped (010)  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> semi-insulating substrates via ozone-assisted molecular beam epitaxy<sup>12,21,28,34–36</sup> (O<sub>3</sub>-MBE) at Novel Crystal Technology, Inc. The growth temperature was 555 °C, the temperature of the Ga cell was 760 °C, and the temperature of the Al cell was 885-900 °C. An O<sub>2</sub>/O<sub>3</sub> mixture (O<sub>3</sub> concentration, ~100 g/m<sup>3</sup>; 0.2% N<sub>2</sub> content) was used with a flow rate of 2 standard cubic centimeters per minute (sccm). The  $\beta$ -(Al<sub>x</sub>Ga<sub>1-x</sub>)<sub>2</sub>O<sub>3</sub> films were not intentionally doped during growth. Films were grown with thicknesses varying from 50-400 nm and with Al contents varying from 6.9-19.5% (Table I). XRD was used to determine the Al content according to the following relation:<sup>37</sup>

$$x \cong 0.4727 \times \Delta\theta_{020} \quad (1)$$

where  $x$  is the Al content and  $\Delta\theta_{020}$  is the shift in the (020) diffraction peak. It should be noted that this relationship is only valid for pseudomorphic grown films. Triple-axis asymmetric grazing exit (GE) RSMs were performed using a Jordan Valley (Bruker) D1 diffractometer, with an incident beam mirror to produce a parallel beam, followed by a Si (220) channel cut collimator (Cu K $\alpha$ 1 radiation). The scattered beam optics included a Si (220) channel cut crystal. A Bruker AFM instrument was used to study the effect of film thickness and Al content on the surface morphology of the films. Additionally, AFM was used to investigate the crack morphology on the film surface. Synchrotron X-ray topography measurements were performed at the 1-BM Beamline of the Advanced Photon Source, Argonne National Laboratory with a photon energy of 8.05 keV to image film defects. The FEI Nova 600 Nanolab Dual Beam scanning electron microscope (SEM)/focused ion beam (FIB) was used to prepare cross section TEM samples roughly 100 nm thick using a Ga source and transferred to a TEM grid using a standard lift out procedure. HR-TEM images were taken using an FEI Titan at 300 kV.

Table I. Sample identification (ID), Al content (%), and thickness of the  $\beta$ -(Al<sub>x</sub>Ga<sub>1-x</sub>)<sub>2</sub>O<sub>3</sub> films grown on (010)  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> substrates. The surfaces of all substrates were treated with chemical mechanical polishing prior to molecular beam epitaxy (MBE) growth of the  $\beta$ -(Al<sub>x</sub>Ga<sub>1-x</sub>)<sub>2</sub>O<sub>3</sub> films.

| <i>Sample ID</i> | <i>Al content [%]</i> | <i>Thickness [nm]</i> | <i>RMS roughness [nm]</i> |
|------------------|-----------------------|-----------------------|---------------------------|
| AL8.8_200        | 8.8                   | 200                   | 0.824                     |
| AL7.7_100        | 7.7                   | 100                   | 0.387                     |
| AL7.5_50         | 7.5                   | 50                    | 0.281                     |
| AL11.5_200       | 11.5                  | 200                   | 0.890                     |
| AL8.5_100        | 8.5                   | 100                   | 0.370                     |
| AL6.9_50         | 6.9                   | 50                    | 0.258                     |
| AL14.3_400       | 14.3                  | 400                   | 1.64                      |
| AL13.6_200       | 13.6                  | 200                   | 0.898                     |
| AL13.9_50        | 13.9                  | 50                    | 0.296                     |
| AL19.5_100       | 19.5                  | 100                   |                           |
| AL18.8_50        | 18.8                  | 50                    |                           |

### 3.2.3 Results and Discussion

The  $\beta$ -( $\text{Al}_x\text{Ga}_{1-x}$ ) $_2\text{O}_3$  films were first characterized using XRD to determine the Al content ( $x$ ). Performing  $\omega$ - $2\theta$  scans of the (020) diffraction peak and applying Eq. 1, the Al content of all 11 films were determined and are given in Table I. As mentioned previously, Eq. 1 is valid for coherently grown films; as will be demonstrated later, despite cracking being present, the films were observed to be fully strained, validating the use of Eq. 1. The measured Al content of the films in Table I ranges from 6.9-19.5%. Exemplary  $\omega$ - $2\theta$  scans of samples AL8.8\_200, AL11.5\_200, and AL13.6\_200 are shown in Fig. 1. All films had similar thickness (200 nm) but increasing Al contents of 8.8%, 11.5%, and 13.6%, respectively. As shown in Fig. 1, as the Al content of the film increases, the  $\beta$ -( $\text{Al}_x\text{Ga}_{1-x}$ ) $_2\text{O}_3$  (020) diffraction peak shifts further away from that of the  $\beta$ - $\text{Ga}_2\text{O}_3$  (020) peak.

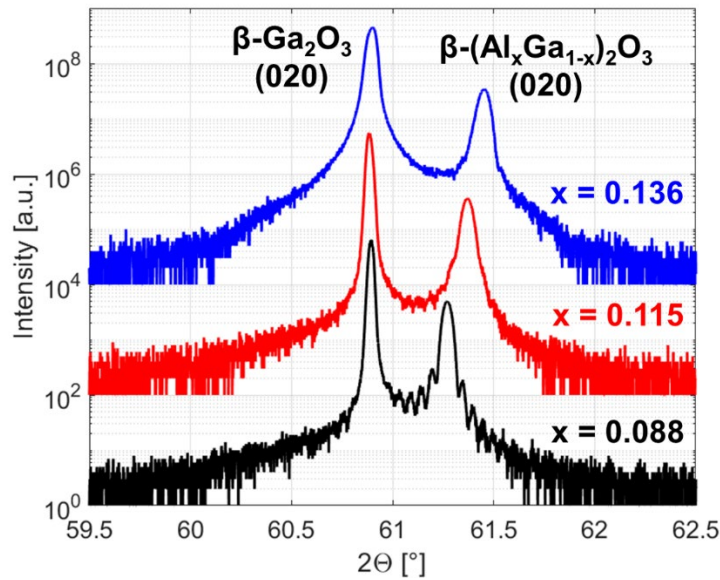


Figure 1. X-ray diffraction (XRD)  $2\theta$  scans of three of the  $\beta$ -( $\text{Al}_x\text{Ga}_{1-x}$ ) $_2\text{O}_3$  films studied (samples AL8.8\_200, AL11.5\_200, and AL13.6\_200 in Table I). The reflection from the (020) plane of the  $\beta$ - $\text{Ga}_2\text{O}_3$  substrates and  $\beta$ -( $\text{Al}_x\text{Ga}_{1-x}$ ) $_2\text{O}_3$  films are shown for  $\beta$ -( $\text{Al}_x\text{Ga}_{1-x}$ ) $_2\text{O}_3$  films with Al contents of 8.8%, 11.5%, and 13.6%.

After determining the Al content of the films, the effect of increasing film thickness and increasing Al content on surface morphology was investigated using AFM. In Fig. 2(a)-(c), films with increasing thickness (50-200 nm) but similar Al content ( $\sim 14\%$ ) were studied. For a film thickness of 50 nm (sample AL13.9\_50, Al content 13.9%), the film was very smooth with rms roughness  $\sim 0.3$  nm. As film thickness was increased to 200 nm (sample AL13.6\_200, Al content 13.6%) and 400 nm (AL14.3\_400, Al content 14.3%), the films became rougher and step-bunching was observed. Step-bunching on the surface of (010)  $\beta$ -( $\text{Al}_x\text{Ga}_{1-x}$ ) $_2\text{O}_3$  films grown via MBE has been previously reported.<sup>38</sup> In Fig. 2(d)-(f), films with increasing Al content (8.8-13.6%) but similar thickness (200 nm) did not show any significant variation in surface morphology. Step-bunching was again observed in all of these films (samples AL8.8\_200, AL11.5\_200, and AL13.6\_200).

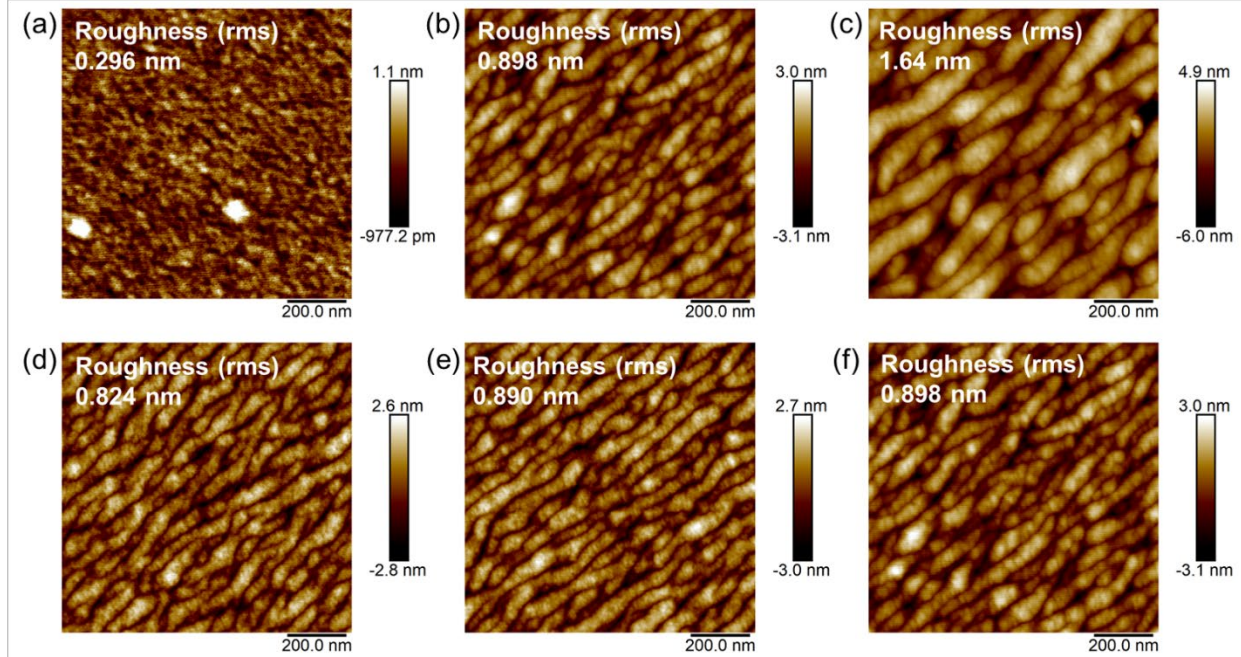


Figure 2. Atomic force microscopy (AFM) scans of the  $\beta$ -( $\text{Al}_x\text{Ga}_{1-x}$ ) $_2\text{O}_3$  films. (a)-(c) AFM of  $\beta$ -( $\text{Al}_x\text{Ga}_{1-x}$ ) $_2\text{O}_3$  films with similar Al content ( $\sim 14\%$ ) and thicknesses of (a) 50 nm, (b) 200 nm, and (c) 400 nm (samples AL13.9\_50, AL13.6\_200, and AL14.3\_400 in Table I). (d)-(f) AFM of  $\beta$ -( $\text{Al}_x\text{Ga}_{1-x}$ ) $_2\text{O}_3$  films with similar thickness (200 nm) and Al content of (d) 8.8%, (e) 11.5%, and (f) 13.6% (samples AL8.8\_200, AL11.5\_200, and AL13.6\_200 in Table I).

To image defects in the films at a larger scale, XRT was used to investigate two films (samples AL8.8\_200, AL13.6\_200) with thicknesses of 200 nm and Al contents of 8.8% and 13.6%. Both films were cracked, but the higher Al content film (13.6%) had a much higher crack density. The cracks in the films were primarily oriented in the [001] direction or (100) plane. Interestingly, as seen in Fig. 3(a), in addition to the cracks in the (100) plane, threading dislocations<sup>39</sup> were observed to be seemingly aligned in the (100) plane as well. In Fig. 3(b), XRT imaging revealed there was also cracking in the [100] direction or (001) plane; however, cracking in this direction was much less prevalent.

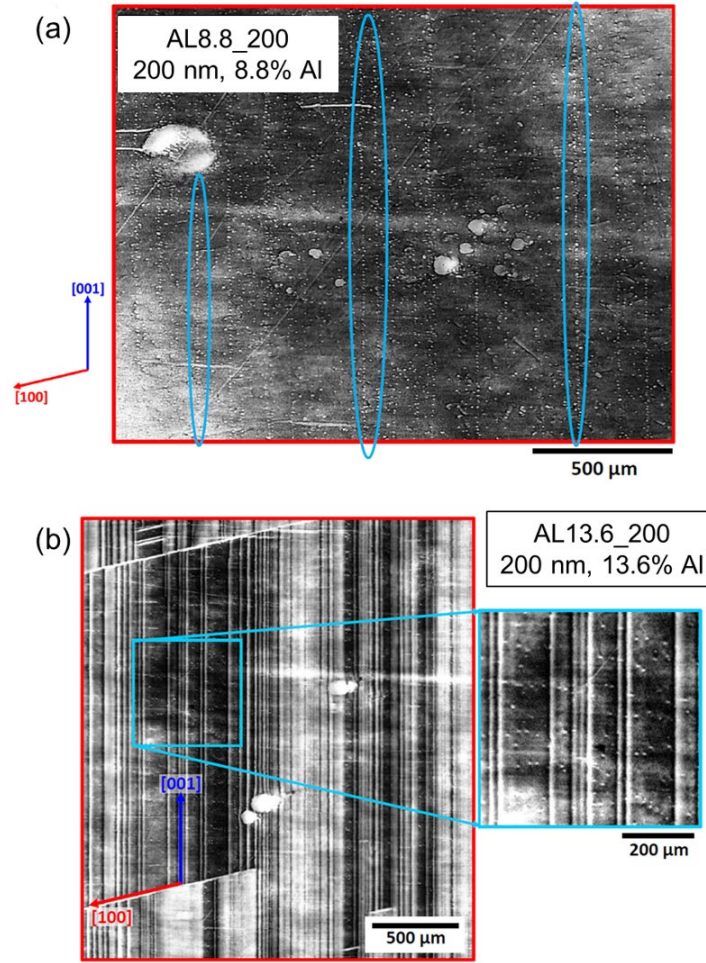


Figure 3. X-ray topography (XRT) of  $\beta$ -( $\text{Al}_x\text{Ga}_{1-x}$ ) $_2\text{O}_3$  films with different crack densities. (a) XRT of threading dislocations aligned along the (100) crack plane for a  $\beta$ -( $\text{Al}_x\text{Ga}_{1-x}$ ) $_2\text{O}_3$  film with Al content of 8.8%. (b) XRT of cracks in the (100) plane of a  $\beta$ -( $\text{Al}_x\text{Ga}_{1-x}$ ) $_2\text{O}_3$  film with Al content of 13.6%. The highlighted region shows the presence of threading dislocations in between cracks on the sample.

Having observed cracking in the  $\beta$ -( $\text{Al}_x\text{Ga}_{1-x}$ ) $_2\text{O}_3$  films, reciprocal space mapping (RSM) was used to assess the strain state (i.e., fully strained, partially relaxed, relaxed) of the films from Fig. 3. Despite the presence of cracking in both films, the films were found to be fully strained, as indicated by the dashed line in Fig. 4. This indicates that the crack formation is not a result of film relaxation. Since the films are still fully strained, the use of XRD and Eq. 1 to determine the Al content of the films is supported.

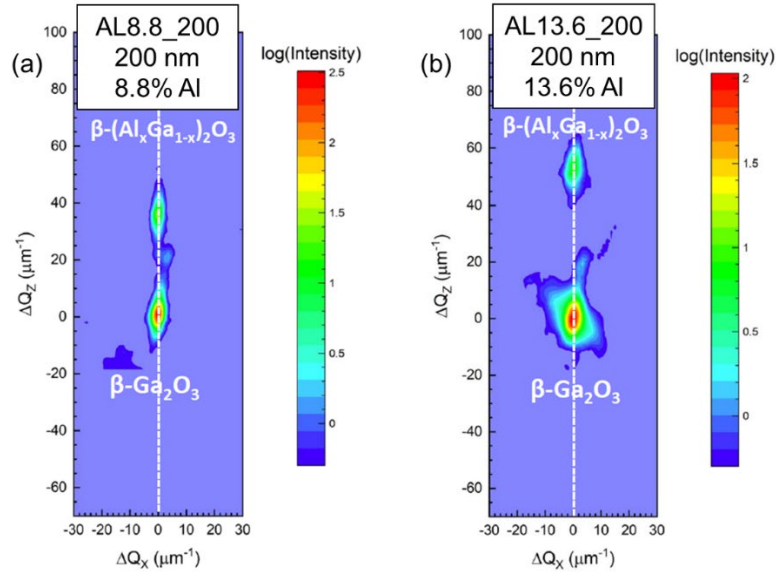


Figure 4. Reciprocal space mapping (RSM) of samples AL8.8\_200 (200 nm thick, 8.8% Al content) and AL13.6\_200 (200 nm thick, 13.6% Al content). Cracking was observed in both films. The dashed line is a visual aid to observe that the  $\beta-(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3$  films are fully strained despite the presence of cracking.

To investigate the crack morphology, formation, and propagation, both AFM and cross-sectional TEM were used. Fig. 5(a) shows a 20  $\mu\text{m}$  AFM scan of the surface of a 400 nm thick film with 14.3% Al content (sample AL14.3\_400). In this scan, cracks in both the [001] and [100] directions are observed. Subsequently, cracks in both directions were probed using AFM (10  $\mu\text{m}$  scans, Figs. 2(b) and 2(d)) and step height profiles perpendicular to the cracks in the [001] and [100] directions are shown in Figs. 2(c) and 2(e), respectively. While the crack in the [001] direction is very narrow (<100 nm), the crack in the [100] direction is much larger (1-2  $\mu\text{m}$ ). Additionally, it can be seen that the crack in the [100] direction appears to terminate (or emanate) at one of the cracks in the [001] direction. Since the cracks in the [001] direction are much more prevalent and the (100) plane has been predicted to be the dominant crack plane,<sup>33</sup> the cross-sectional geometry of these cracks were further studied using TEM.

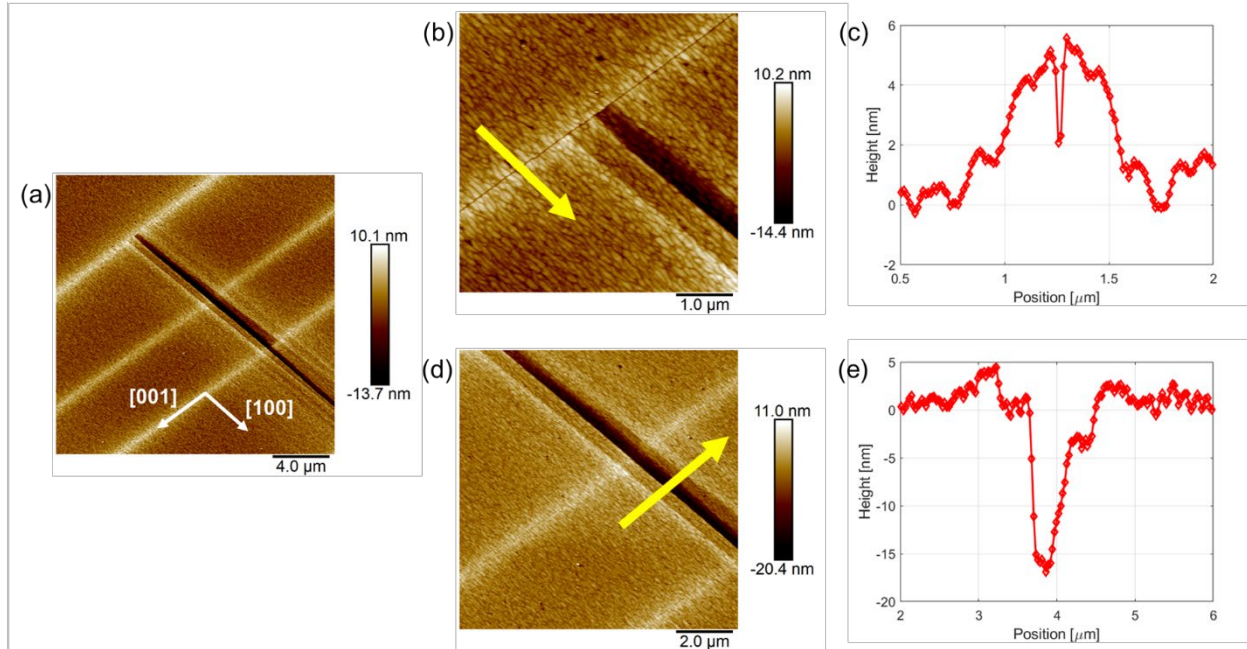


Figure 5. (a),(b),(d) AFM scans and (c),(e) step height profiles of the cracks in the (b),(c) (100) plane/[001] direction and (d),(e) (001) plane/[100] direction on a 400 nm thick  $\beta$ -( $\text{Al}_x\text{Ga}_{1-x}$ ) $_2\text{O}_3$  film with 14.3% Al content (sample AL14.3\_400 in Table I). The arrows in (b),(d) indicate the direction of the step height profiles shown in (c),(e).

Fig. 6(a) shows a bright field TEM (BF-TEM) image of the cross-section of a 200 nm thick film with 13.6% Al content (sample AL13.6\_200). The cracks in the [001] direction can clearly be seen to begin at the surface of the  $\beta$ -( $\text{Al}_x\text{Ga}_{1-x}$ ) $_2\text{O}_3$  film and extend into the  $\beta$ -Ga $_2$ O $_3$  substrate. The crack lengths were approximately 300-400 nm (373 nm in Fig. 6(a)); since the film thickness was only  $\sim$ 200 nm, the cracks extended throughout the entire thickness of the film and extended an additional 100-200 nm in the  $\beta$ -Ga $_2$ O $_3$  substrate. High-resolution TEM images were also taken of the film cross-section as shown in Fig. 6(b). Here, the crack in the  $\beta$ -( $\text{Al}_x\text{Ga}_{1-x}$ ) $_2\text{O}_3$  film is seen to extend laterally in the immediate proximity of the film surface.

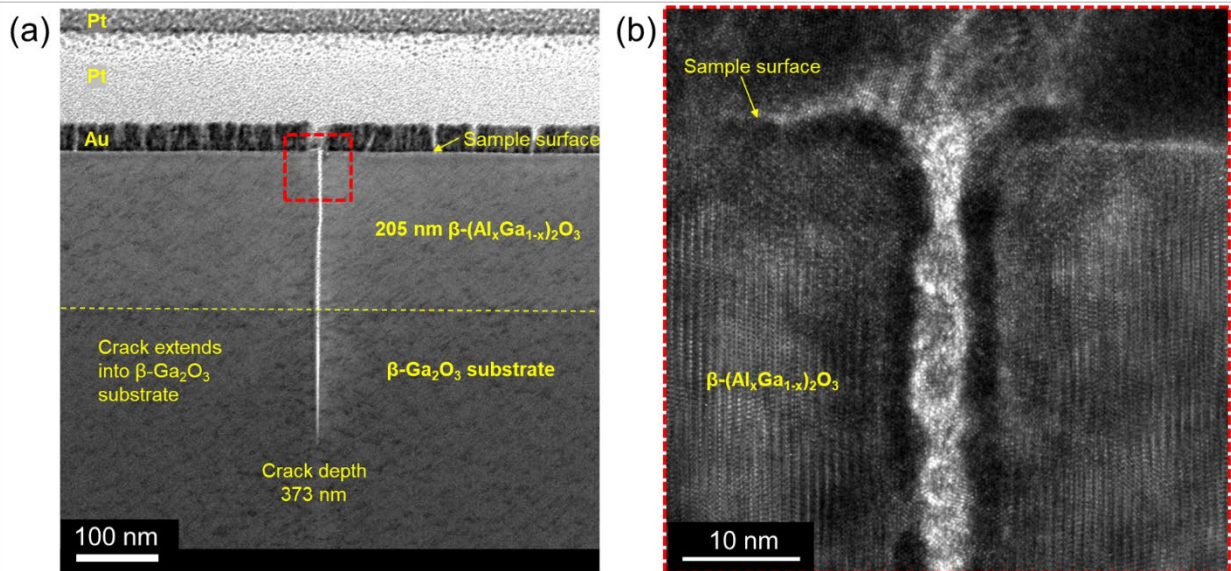


Figure 6. a) BF-STEM image of a representative crack in a  $\beta$ -( $\text{Al}_x\text{Ga}_{1-x}$ ) $_2\text{O}_3$  film (200 nm thick, 13.6% Al content, sample AL13.6\_200). b) HR-TEM at the surface of the crack (the region enclosed by the red dashed box in (a)).

Finally, the experimental determination of critical thickness limitations of  $\beta-(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3$  films reported in this study were compared with the theoretical calculations reported by Mu et al.<sup>33</sup> based on the Griffith criterion. It should be noted that Mu et al reported critical thickness as a function of Al content for (100), (010), and (001) oriented films; as the films in this study were grown on (010)  $\beta\text{-Ga}_2\text{O}_3$ , accordingly, the theoretical critical thickness limitations for (010) oriented films are shown. As shown in Fig. 7, there is very good agreement between the experimentally observed and theoretically calculated critical thickness limitations. Additionally, the theoretical calculations reported by Mu et al were based on cracking of the (100)B plane. As demonstrated in this study, it was experimentally observed that the (100) plane was the most prevalent crack plane.

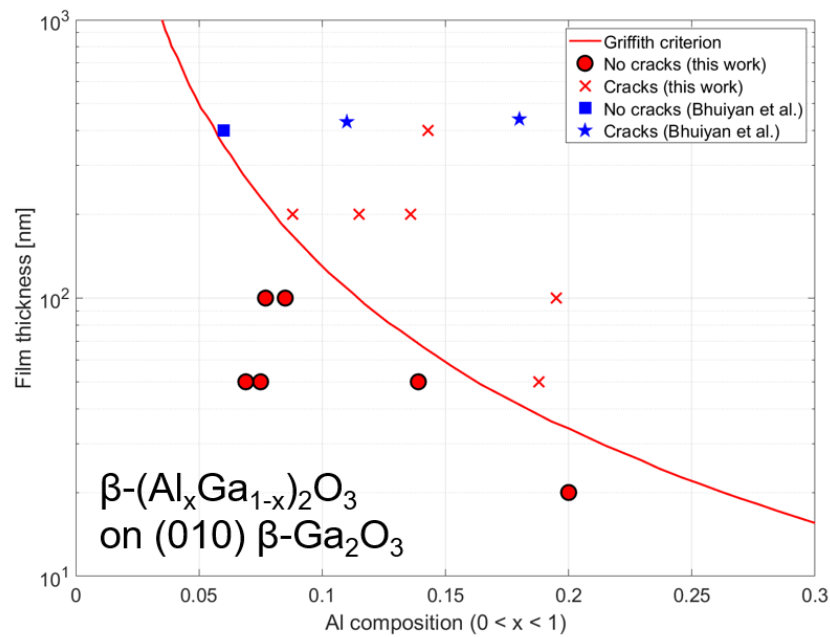


Figure 7. Film thickness vs. Al content for the  $\beta-(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3$  films studied. The experimental data (circle, x) show whether the  $\beta-(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3$  films had no cracks (circle) or were cracked (x). Also shown are theoretical calculations for the critical thickness of  $\beta-(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3$  films grown on (010)  $\beta\text{-Ga}_2\text{O}_3$  from Mu et al.<sup>33</sup> based on the Griffith criterion and experimental results from Bhuiyan et al.<sup>2</sup>

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### 3.3 Delta-Doped $\beta$ -( $\text{Al}_x\text{Ga}_{1-x}$ ) $_2\text{O}_3/\text{Ga}_2\text{O}_3$ Heterostructure Field-Effect Transistors by Ozone Molecular Beam Epitaxy

#### 3.3.1 Introduction

The promise of  $\text{Ga}_2\text{O}_3$  to deliver high power, high efficiency, low cost power electronics has advanced significantly in recent years after breakthroughs in bulk crystal and epitaxial film growth have led to unprecedented improvement in size and quality of this material.<sup>1-3</sup> Its melt growth substrate technology is unique among all wide and ultra-wide bandgap semiconductors and similar to that of silicon. In addition,  $\text{Ga}_2\text{O}_3$  can also form ternary alloys; in particular, the Al alloy enables heterostructure devices by taking advantage of the ultra-wide bandgap range of about 4.8 eV for  $\beta$ - $\text{Ga}_2\text{O}_3$  up to 7.24 eV for monoclinic  $\text{Al}_2\text{O}_3$ .<sup>4-5</sup> In the case of alloying with Al during homoepitaxial growth, the monoclinic phase can be preserved within the relevant Al mole fraction of up to about 30 percent to form pseudomorphic  $\beta$ -( $\text{Al}_x\text{Ga}_{1-x}$ ) $_2\text{O}_3/\text{Ga}_2\text{O}_3$  (henceforth referred to as ( $\text{AlGa}$ ) $_2\text{O}_3/\text{Ga}_2\text{O}_3$ ) heterostructures suitable for device applications.<sup>6-12</sup> Growth of ( $\text{AlGa}$ ) $_2\text{O}_3/\text{Ga}_2\text{O}_3$  heterostructures has been recently demonstrated both by plasma assisted molecular beam epitaxy (PAMBE), metal-oxide catalyzed epitaxy (MOCATAXY), pulsed-laser deposition (PLD), and most recently by metal-organic chemical vapor deposition (MOCVD).<sup>13-18</sup> This excellent progress over only a few years has rapidly expanded the potential applications for  $\text{Ga}_2\text{O}_3$ . Even in the absence of spontaneous polarization in the monoclinic phase of  $\text{Ga}_2\text{O}_3$ , a two-dimensional electron gas (2DEG) can be formed via the delta doping approach, similar to early  $\text{AlGaAs}/\text{GaAs}$  heterostructures.<sup>19,20</sup> Delta doping of compound semiconductor device heterostructures is advantageous for multiple reasons, among which are the improved performance through device scaling, high transconductance, high electron channel density, improved breakdown voltage due to reduced electric field under the gate, and reduced short channel effects.<sup>21</sup> This work builds upon the early ozone-assisted MBE ( $\text{O}_3$ -MBE) growth and device demonstrations by Higashiwaki et al. and Sasaki et. al, who in 2012 were the first researchers to focus the attention of device scientists to  $\text{Ga}_2\text{O}_3$ .<sup>22,23</sup> At the time, epitaxial growth of  $\beta$ - $\text{Ga}_2\text{O}_3$  via HVPE, PAMBE, and MOCVD was yet to be demonstrated. We aim to extend these early results and demonstrate ( $\text{Al}_x\text{Ga}_{1-x}$ ) $_2\text{O}_3/\text{Ga}_2\text{O}_3$  heterostructure field-effect transistors grown via  $\text{O}_3$ -MBE for the first time.

#### 3.3.2 Experimental Details of Epitaxial Growth

Three samples of ( $\text{AlGa}$ ) $_2\text{O}_3/\text{Ga}_2\text{O}_3$  were grown on commercial Fe-doped edge-defined, film-fed grown (EFG) (010)  $\beta$ - $\text{Ga}_2\text{O}_3$  substrates produced by Novel Crystal Technology, Inc. An  $\text{O}_2/\text{O}_3$  mixture with  $\text{O}_3$  concentration of about 100 g/m<sup>3</sup>, flow rate of 2 standard cubic cm per minute (sccm), and a 0.2%  $\text{N}_2$  content was introduced in the vicinity of the sample surface via a nozzle, resulting in a growth rate comparable with the ~120 nm/h reported by Sasaki et al. for (010) oriented  $\text{Ga}_2\text{O}_3$ .<sup>23</sup> The growth temperature in the reactor was 555 °C, with a Ga cell temperature of 760 °C, Si cell temperature of 860-880 °C, and Al cell temperature of 890-900 °C, as summarized in Table I. Following a 125 nm thick unintentionally n-type  $\text{Ga}_2\text{O}_3$  buffer layer growth, an approximately 3 nm thick ( $\text{Al}_x\text{Ga}_{1-x}$ ) $_2\text{O}_3$  spacer layer was grown to reduce carrier scattering in the channel, followed by an approximately 0.25 nm thick Si delta-doped ( $\text{Al}_x\text{Ga}_{1-x}$ ) $_2\text{O}_3$  layer, capped by an approximately 23-25 nm thick ( $\text{Al}_x\text{Ga}_{1-x}$ ) $_2\text{O}_3$  barrier layer. The growth rates for the  $\text{Ga}_2\text{O}_3$  buffer layer and the ( $\text{Al}_x\text{Ga}_{1-x}$ ) $_2\text{O}_3$  layers was about 107 nm/hr and 120 nm/hr, respectively.

TABLE I. Temperature of dopant and Al cell during growth, Aluminum composition ratio,  $(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3$  thickness, ECV carrier concentration, Hall measurement results from samples A, B, and C obtained at room temperature after mesa etch and Ohmic contact deposition, as well as Hall data from samples A and B measured after ALD  $\text{Al}_2\text{O}_3$  deposition.

|                                                                                                           | Sample A | Sample B                                                  | Sample C                                                 |
|-----------------------------------------------------------------------------------------------------------|----------|-----------------------------------------------------------|----------------------------------------------------------|
| Si Knudsen cell temperature ( $^{\circ}\text{C}$ )                                                        | 860      | 880                                                       | 870                                                      |
| Al Knudsen cell temperature ( $^{\circ}\text{C}$ )                                                        | 900      | 895                                                       | 890                                                      |
| Al composition ratio, $x$ (%)                                                                             | 23.1     | 21.2                                                      | 14.2                                                     |
| $(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3$ thickness (nm)                                                 | 28.1     | 28.1                                                      | 26.0                                                     |
| ECV-measured free carrier concentration, $N_D - N_A$ ( $\times 10^{19} \text{ cm}^{-3}$ )                 | 0.79     | Primary peak: 4.0<br>Secondary peak: 1.6                  | Primary peak: 3.3<br>Secondary peak: 2.0                 |
| Carrier density obtained by integration of ECV carrier concentration ( $\times 10^{13} \text{ cm}^{-2}$ ) | 0.23     | Primary peak: 1.35<br>Secondary peak: 0.28<br>Total: 1.63 | Primary peak: 1.0<br>Secondary peak: 0.45<br>Total: 1.45 |
| Sheet resistance, $R_{\text{SH}}$ ( $\Omega/\text{sq.}$ )                                                 | 73121    | 6271                                                      | 5857                                                     |
| Hall Mobility, $\mu_{\text{H}}$ ( $\text{cm}^2/\text{V}\cdot\text{s}$ )                                   | 62       | 90                                                        | 95                                                       |
| Hall Sheet carrier density, $N_{\text{S}}$ ( $\times 10^{13} \text{ cm}^{-2}$ )                           | 0.137    | 1.11                                                      | 1.12                                                     |
| $R_{\text{SH}}$ after $\text{Al}_2\text{O}_3$ ( $\Omega/\text{sq.}$ )                                     | 32206    | 5595                                                      | N/A                                                      |
| $\mu_{\text{H}}$ after $\text{Al}_2\text{O}_3$ ( $\text{cm}^2/\text{V}\cdot\text{s}$ )                    | 64       | 94                                                        | N/A                                                      |
| $N_{\text{S}}$ after $\text{Al}_2\text{O}_3$ ( $\times 10^{13} \text{ cm}^{-2}$ )                         | 0.303    | 1.19                                                      | N/A                                                      |

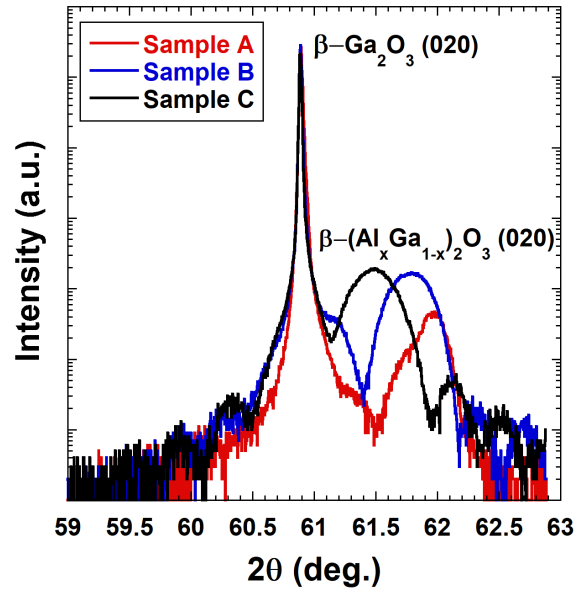


FIG. 1. X-Ray diffraction spectra of three  $\beta-(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3/\text{Ga}_2\text{O}_3$  heterostructures grown by  $\text{O}_3$ -MBE on (010) Fe-doped  $\text{Ga}_2\text{O}_3$  substrates.

The aluminum composition ratio,  $x$ , in the three samples was estimated via X-ray diffraction using the shift in the (020)  $\beta-(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3$  reflection using the relationship  $x \cong 0.4727 \cdot \Delta\theta_{020}$ , as illustrated in Fig. 1 and Fig. S1 in the supplementary material.<sup>24</sup> It was observed that the Al composition increased with a higher temperature of the Al cell, as shown in Table I. Sample A exhibited an Al concentration of about 23.1%; sample B had a slightly lower Al concentration of about 21.2% in the  $(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3$  layers. The 890  $^{\circ}\text{C}$  Al cell temperature for the growth of sample C resulted in about 14.2% Al in the  $(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3$  layers. In addition, the

thickness  $t$  of the  $(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3$  layer was estimated based on the XRD positional difference in the satellite reflections using the relationship  $t = (L_1 - L_2)\lambda / 2(\sin\theta_1 - \sin\theta_2)$ , where  $\lambda$  is the X-ray wavelength (0.1540598 nm),  $L$  is the satellite peak order, and  $\theta$  is the satellite angle.<sup>25</sup> From this analysis,  $t$  for samples B and C was estimated to be 28.1 nm and 26.0 nm, respectively. Due to the lack of satellite peaks in sample A, indicating a lower quality of that  $(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3$  film, the thickness was estimated to be the same as that of sample B based on growth time.

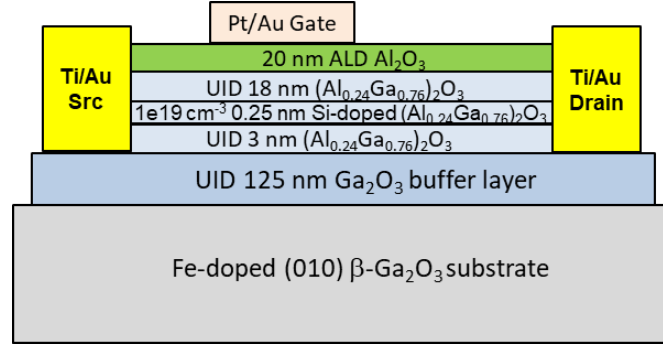


FIG. 2. Schematic of the  $\beta$ -( $\text{Al}_x\text{Ga}_{1-x}$ ) $_2\text{O}_3$ / $\text{Ga}_2\text{O}_3$  heterostructure grown by  $\text{O}_3$ -MBE and fabricated transistor devices.

### 3.3.3 Device Fabrication Details

Heterojunction field-effect transistor (HFET) fabrication was performed beginning with a 3-minute inductively-coupled plasma (ICP) dry etch in  $\text{BCl}_3$  plasma (800 W ICP, 80 W RIE, 40 sccm  $\text{BCl}_3$ , 5 mT) to form mesa isolation regions. Device feature patterning was performed via maskless photolithography performed using a Heidelberg MLA150 maskless aligner. Following Ohmic contact lithography, an 80 W reactive-ion  $\text{BCl}_3$  plasma etch for 20 seconds in the patterned Ohmic regions was performed with the intention to improve contact resistance, similar to Higashiwaki et al.<sup>22</sup>, immediately followed by e-beam evaporation of the Ti/Au Ohmic contacts. Following lift-off and photoresist removal, a 470 °C, 1-min. rapid-thermal anneal was carried out in  $\text{N}_2$  ambient. A contact resistance of 155.23  $\Omega$ -mm was measured on sample B using the linear transfer-length (LTLM) method with a  $\pm 1$  V range due to nonlinearity in the raw I-V data (see Fig. S2), indicating that  $n^+$   $\text{Ga}_2\text{O}_3$  regrowth or ion implantation in the Ohmic regions will be necessary in subsequent process iterations to obtain low contact resistance. Room temperature Hall measurement results before and after  $\text{Al}_2\text{O}_3$  gate dielectric deposition are summarized in Table I. Sample C (no  $\text{Al}_2\text{O}_3$  deposited) indicated a sheet carrier density ( $N_s$ ) of  $1.12 \times 10^{13} \text{ cm}^{-2}$ , Hall mobility  $\mu_H = 95 \text{ cm}^2/\text{V}\cdot\text{s}$ , and a sheet resistance of 5857  $\Omega/\text{sq}$ . Processing of the lateral HFETs continued with a 20 nm thick ALD  $\text{Al}_2\text{O}_3$  gate dielectric layer on samples A and B. Device fabrication finished with the lift-off of a 20/200 nm thick Pt/Au gate metal stack. No additional passivation dielectric was utilized in this experiment. The resulting device structure is depicted in the schematic shown in Fig. 2. Fabricated transistors had a gate lengths of 2.5-5  $\mu\text{m}$ , gate-source spacing of 2.5  $\mu\text{m}$ , and gate-drain spacing varying from 2.5 to 15  $\mu\text{m}$ , and width of 104  $\mu\text{m}$ . Direct-current (DC) current-voltage and 1 MHz capacitance-voltage characterization was performed using a Keithley 4200SCS semiconductor parameter analyzer. Pulsed current-voltage (I-V) measurements were performed using a Accent DiVA D265EP analyzer (200 ns on-state pulse width, 1 ms pulse separation, 0.02% duty cycle). Electrochemical capacitance-voltage (ECV) measurements were performed using a Nanometrics ECV Pro system at Novel Crystal Technology. Temperature-dependent Hall measurements were performed on a closed-He loop Lakeshore cryogenic Hall prober. Breakdown voltage measurements were performed at room

temperature in Fluorinert<sup>TM</sup> using a Keithley 237 source-measurement unit. Band diagram simulation was performed using self-consistent coupled Schrödinger-Poisson calculation. For the band diagram simulation, the delta-doping level ranged between  $1 \times 10^{18}$  and  $1 \times 10^{20} \text{ cm}^{-3}$ . The background doping level was chosen in the  $6 \times 10^{16}$  to  $2 \times 10^{17} \text{ cm}^{-3}$  range based on the ECV data in Fig. 3b. A detailed simulation study on the effect of heterostructure parameters (delta doping concentration and layer thickness, Al concentration, dopant redistribution, etc.) has been performed and published separately.<sup>26</sup>

### 3.3.4 Experimental Results

Figure 3a shows the ECV data and Fig. 3b presents the corresponding depth profiles of the measured free carrier concentration in the three samples. The measurements were performed on the as-grown samples where no  $\text{Al}_2\text{O}_3$  was present yet; thus, the 0 nm data point corresponds to the  $\text{Al}_2\text{O}_3/(\text{AlGa})_2\text{O}_3$  interface of the fabricated devices. Peak measured free carrier concentrations  $N_D - N_A$  were  $7.9 \times 10^{18} \text{ cm}^{-3}$  for sample A,  $3.3 \times 10^{19} \text{ cm}^{-3}$  for sample B, and  $4.0 \times 10^{19} \text{ cm}^{-3}$  for sample C (Table I). Inset in Fig. 3a are the ECV depth profiles showing  $N_D - N_A$  within approximately 30 nm from the surface, where a second peak in free carrier concentration was detected.

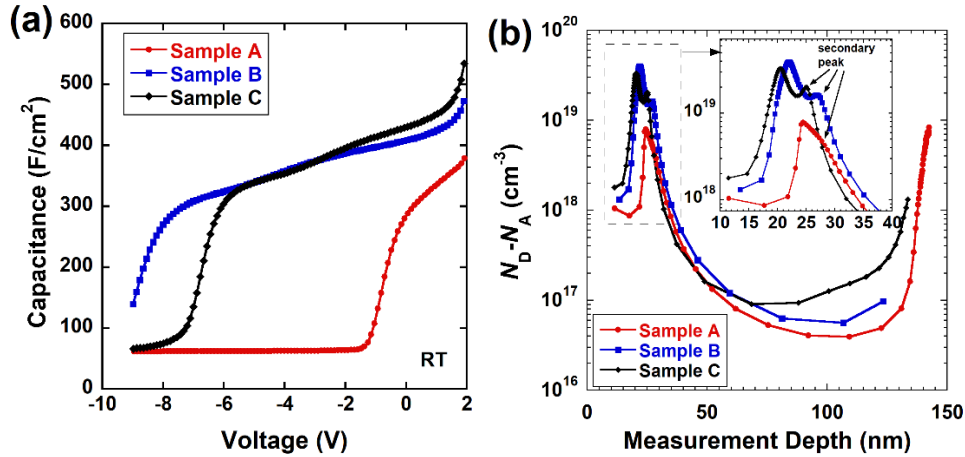


FIG. 3. a) Electro-chemical  $C$ - $V$  measurements and b) free carrier concentration depth profile of samples A (red curve), B (blue curve) and C (black curve). Fig. 3b inset: Depth profile within  $\sim 30$  nm from the  $\beta$ -( $\text{Al}_x\text{Ga}_{1-x}$ ) $_2\text{O}_3$  surface.

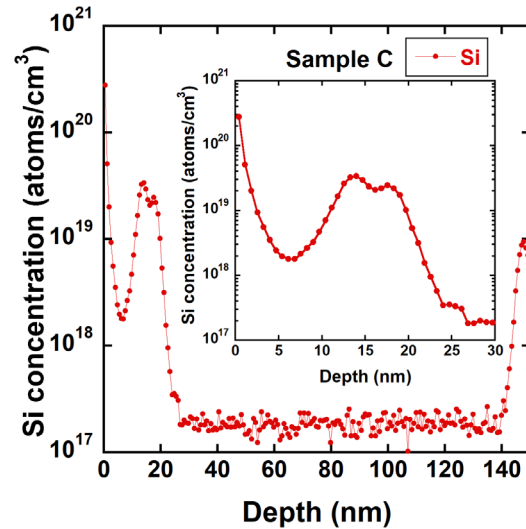


FIG. 4. Secondary ion mass spectroscopy (SIMS) trace of the Si dopant in sample C.

Secondary ion mass spectroscopy (SIMS), shown in Fig. 4, was performed on sample C after Ohmic contact formation, showing good agreement with the ECV data and thus confirming the presence of a secondary carrier channel originating from unintentional incorporation of Si near the  $(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3/\text{Ga}_2\text{O}_3$  interface. The presence of this unintentional Si peak was due to an interruption of the growth process in these first growth experiments in order to ramp up dopant flux for the subsequent delta-doped layer growth. A possible cause for unintentional incorporation of Si at the  $(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3/\text{Ga}_2\text{O}_3$  interface was evaporation of Si from the cell shutter or source during the interrupted growth window. We note that the oxygen flow in the reactor was stopped during heating of the dopant cell. This Si charge accumulation is undesirable as likely leading to reduced mobility due to additional carrier scattering effects. For samples B and C, the ECV-measured values for the secondary peak maxima were  $1.6 \times 10^{19} \text{ cm}^{-3}$  and  $2.0 \times 10^{19} \text{ cm}^{-3}$ , respectively. Integration of the total free charge measured by ECV yielded  $1.63 \times 10^{13} \text{ cm}^{-2}$  and  $1.45 \times 10^{13} \text{ cm}^{-2}$  for samples B and C, respectively. We further separated the contribution of each peak to this carrier density (Table I), obtaining carrier concentrations within the same range as the Hall-measured  $N_s$  but allowing us to quantify the relative contribution of each peak to the cumulative  $N_s$ . Dopant profile optimization in the  $(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3$  barrier layers has been performed in our subsequent growths with the goal of removing unintentional channel effects and improving carrier profile.

Temperature-dependent Hall measurements were performed on sample C on ungated Van der Pauw test structures. No ALD  $\text{Al}_2\text{O}_3$  gate dielectric was deposited on that sample. The results are summarized in Fig. 5. A peak Hall mobility of  $244 \text{ cm}^2/\text{V}\cdot\text{s}$  was measured at 150 K. The reduction in mobility near room temperature was limited by polar optical phonon scattering, whereas ionized impurity scattering lead to reduced mobility at cryogenic temperature.<sup>27,28</sup>  $N_s$  decreased to  $6.9 \times 10^{12} \text{ cm}^{-2}$  at 150 K, indicating that the channel likely was not degenerately doped, unlike previous reports by Zhang showing constant sheet carrier density at cryogenic temperature as well as corresponding Shubnikov de Haas oscillations.<sup>29</sup>

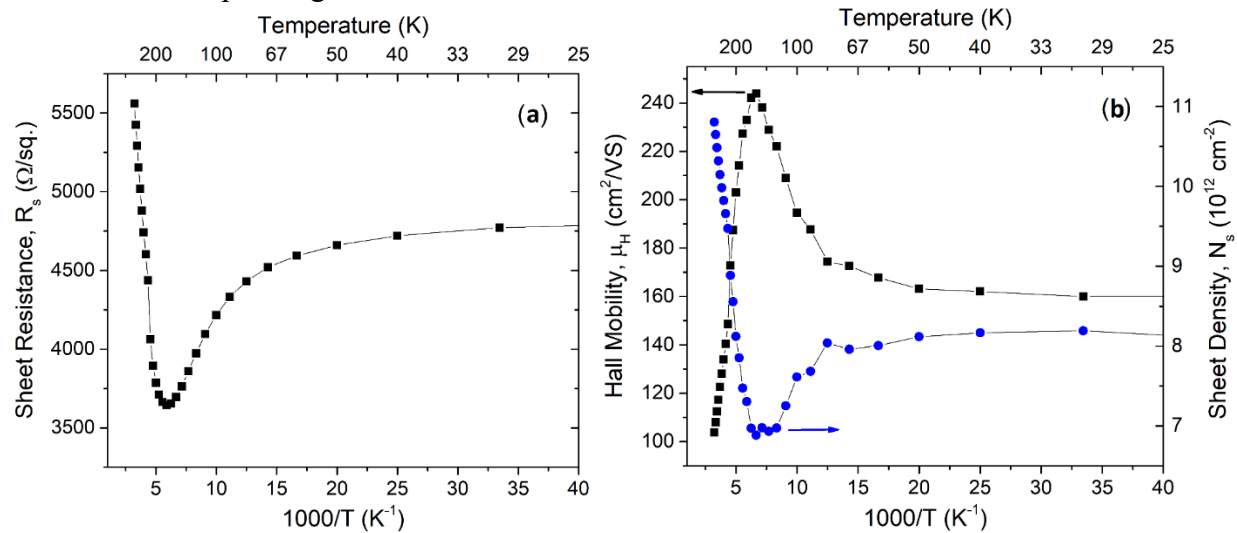


FIG. 5. Temperature-dependent Hall measurements on sample C showing a) sheet resistance, b) Hall mobility and sheet carrier concentration as a function of temperature.

Figures 6 and 7 present the capacitance-voltage and current-voltage characteristics measured from the fabricated transistors on samples A and B, respectively, at room temperature. The capacitance-voltage measurements were performed on transistor devices from depletion towards accumulation. Compared to the ECV data in Fig. 3a, a lower capacitance density was measured due to the presence of the ALD  $\text{Al}_2\text{O}_3$  dielectric resulting in a lower total accumulation capacitance. Sample A (Fig. 6a) showed comparable flatband voltage ( $V_{\text{FB}}$ ) relative to the ECV data; however, sample B (Fig. 7a) exhibited significant negative shift in  $V_{\text{FB}}$  likely due to higher interface trap density at the ALD  $\text{Al}_2\text{O}_3/(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3$  interface. For the case of  $\text{Ga}_2\text{O}_3$  transistors, ALD  $\text{Al}_2\text{O}_3$  has been shown to exhibit higher density of trapped charge compared to, for example,  $\text{HfO}_2$ .<sup>30</sup> We speculate that the higher bandgap of  $(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3$  and the resulting reduction in conduction band offset with  $\text{Al}_2\text{O}_3$  would lead to even more pronounced interfacial charge effects in ternary heterostructure devices. Detailed studies of dielectric/ $\beta$ -( $\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3$  interfacial defects are presently lacking and are the subject of our subsequent experiments.

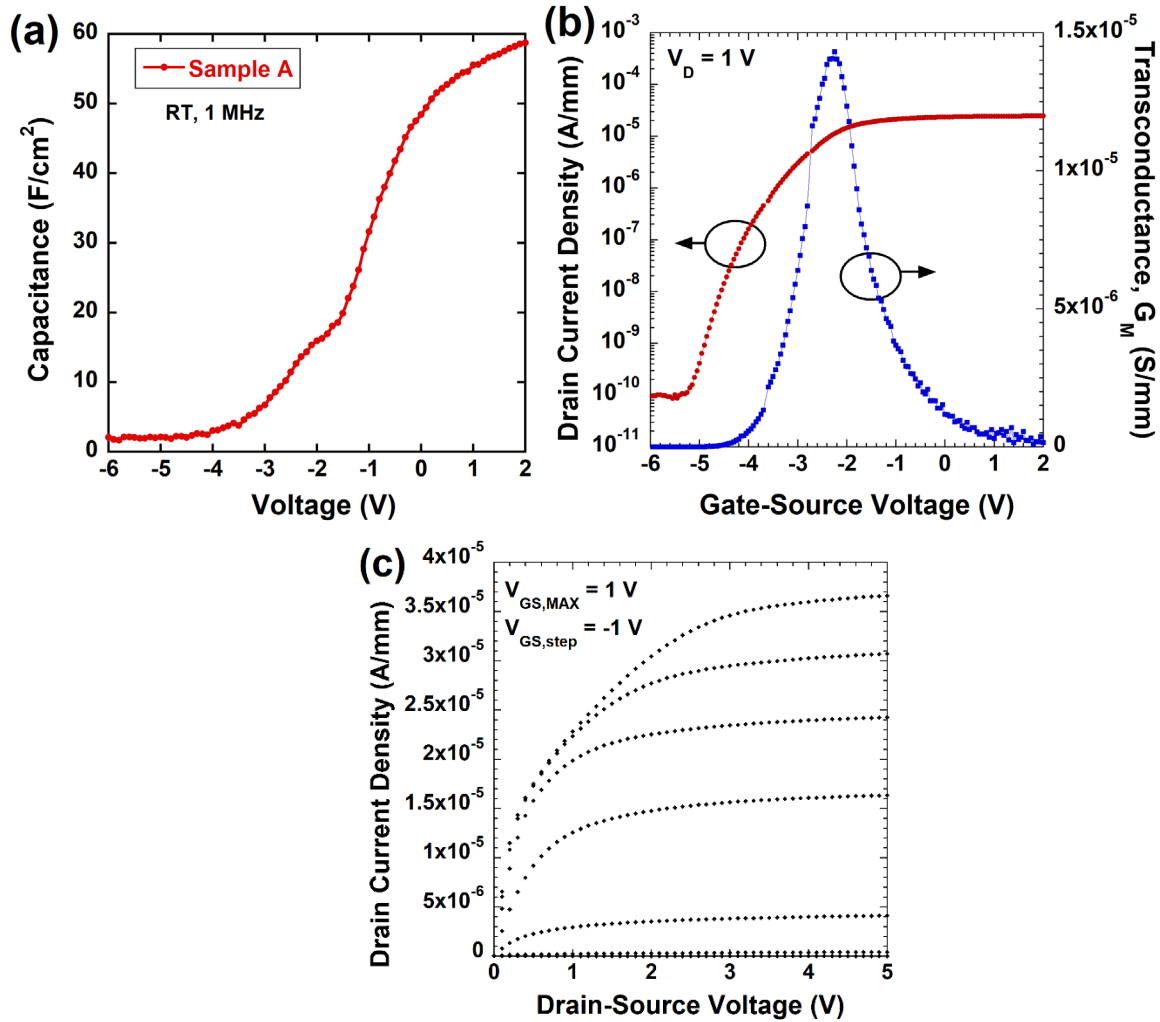


FIG. 6. a) Capacitance-voltage measurement at 1 MHz at room temperature, b) static (DC) input ( $J_{\text{DS}}-V_{\text{GS}}$ ) and c) output ( $J_{\text{DS}}-V_{\text{DS}}$ ) characteristics of sample A.

Depletion mode operation was observed in both samples A and B with threshold voltage values of -5.2 V (Fig. 6b) and -26 V (Fig. 7b), respectively, in good agreement with the capacitance voltage data in Fig. 6a and Fig. 7a. The walkout of the drain current linear region in Fig. 7c-d

indicated that a low contact resistance process is still required to fully take advantage of the high carrier density in these heterostructures. The higher electron concentration in sample B, compared to sample A, caused both a negative threshold voltage shift as well as a significant increase in total saturation drain current ( $I_{DS,SAT}$ ) in sample B ( $I_{DS,SAT} = 22$  mA/mm, Fig. 7c). An additional transconductance peak was measured on sample B (Fig. 7b) around  $V_{GS} = -17$  V. This is consistent with a parasitic channel caused by the unintentional Si at the  $(Al_xGa_{1-x})_2O_3/Ga_2O_3$  interface observed in the *ECV* data for samples B and C and confirmed by SIMS for sample C. As a more negative gate voltage is applied on the gate contact, electrons closer to the surface would deplete first, followed by electrons at the  $(Al_xGa_{1-x})_2O_3/Ga_2O_3$  interface, to comprise the total measured drain-source current.

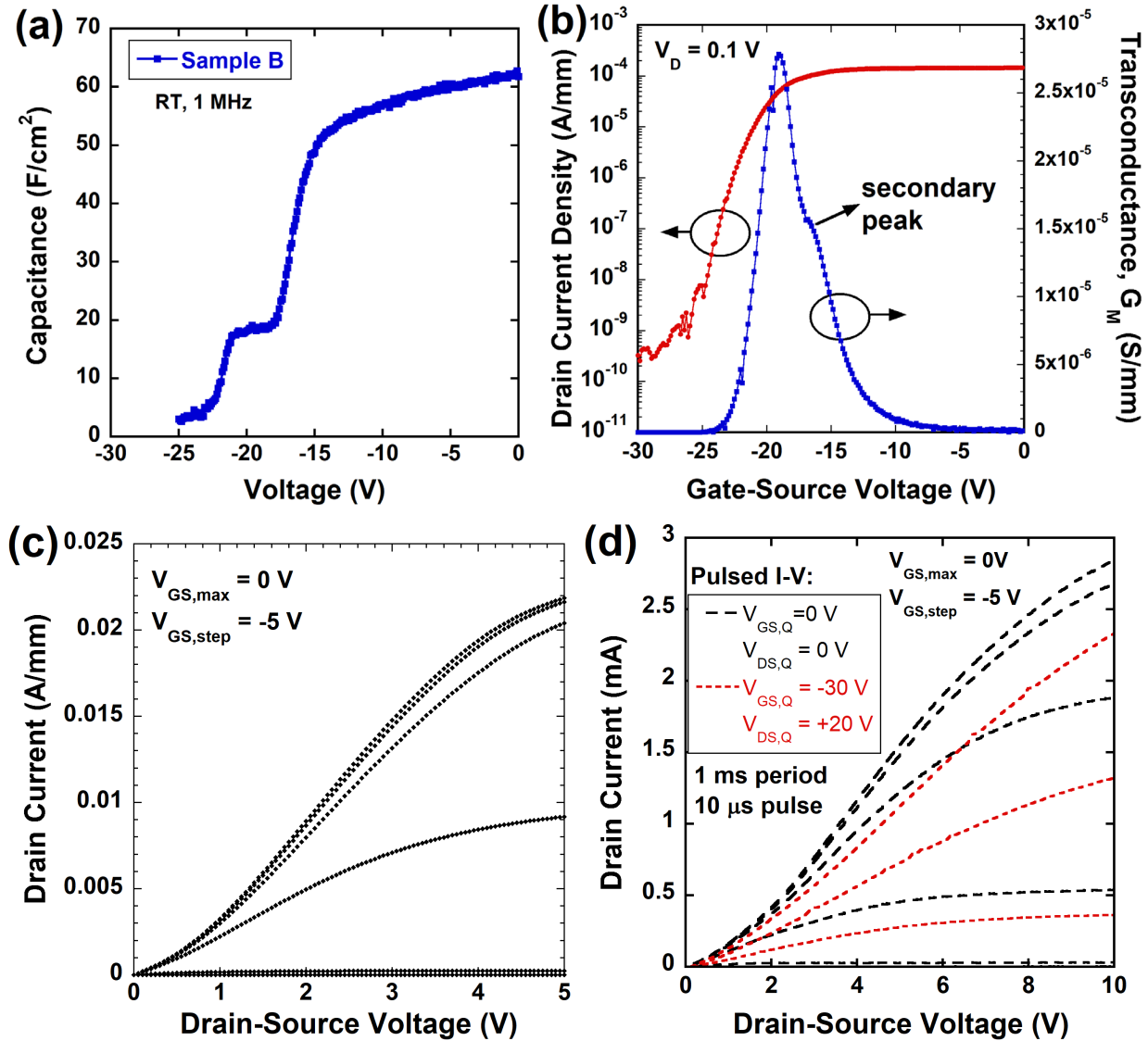


FIG. 7. a) Capacitance-voltage measurement at 1 MHz at room temperature, b) static (DC) input ( $J_{DS}-V_{GS}$ ) and c) output ( $J_{DS}-V_{DS}$ ) characteristics of sample B. d) pulsed I-V characteristics of sample B measured with (red) and without (black) quiescent bias stress on gate and drain contacts.

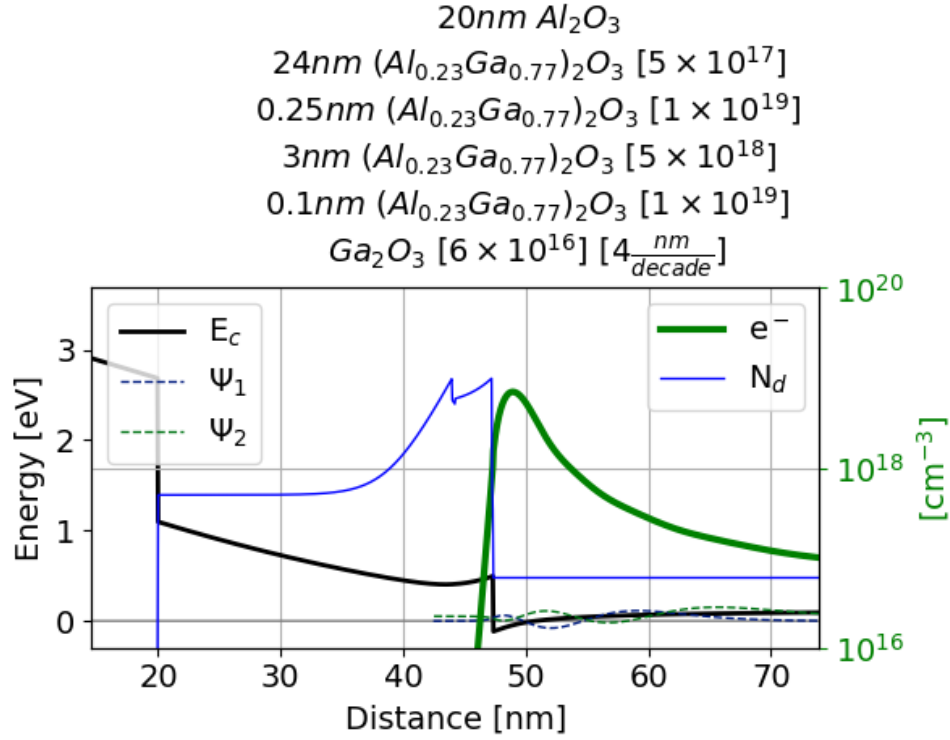


FIG. 8. Simulated band diagrams of  $\beta\text{-(Al}_x\text{Ga}_{1-x})_2\text{O}_3/\text{Ga}_2\text{O}_3$  heterostructures with a 20 nm thick  $\text{Al}_2\text{O}_3$  dielectric deposited on the surface for a graded Si dopant level with a maximum of  $10^{19} \text{ cm}^{-3}$  (sample A).

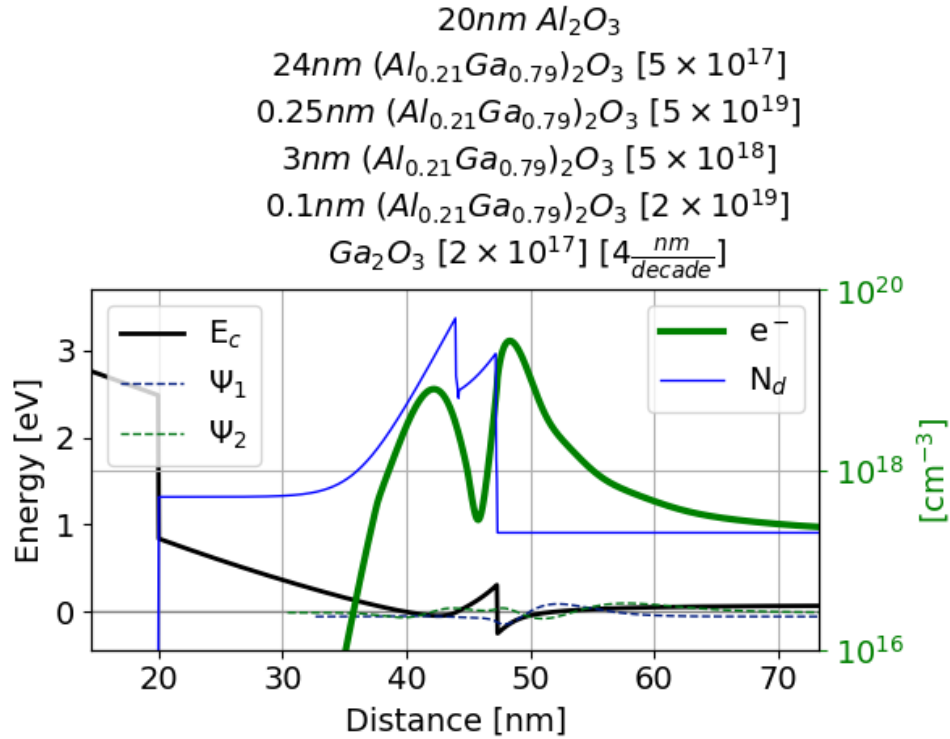


FIG. 9. Simulated band diagram of  $\beta\text{-(Al}_x\text{Ga}_{1-x})_2\text{O}_3/\text{Ga}_2\text{O}_3$  heterostructures with a 20 nm thick  $\text{Al}_2\text{O}_3$  dielectric deposited on the surface and a graded Si dopant profile with a maximum level of  $5 \times 10^{19} \text{ cm}^{-3}$  (sample B).

Pulsed-mode  $I$ - $V$  measurements were performed in addition to the DC measurements in sample B using a 1  $\mu$ s pulse width and a 1 ms pulse period (0.1% duty cycle). Quiescent bias on the drain and gate contacts was initially kept to 0 V ( $V_{GS,Q} = 0$  V,  $V_{DS,Q} = 0$  V). Quiescent stress with bias values of  $V_{GS,Q} = -30$  V and  $V_{DS,Q} = +20$  V showed about 25 percent reduction in output drain current (Fig. 7d). While this could be indicative of possible current collapse in these samples in the absence of a robust passivation dielectric, the shortest pulse width available for this measurement was not sufficient to completely suppress the onset of the self-heating effect, which would lead to additional thermal-induced drain current lowering. Combined with the extremely low thermal conductivity of  $\text{Ga}_2\text{O}_3$ , it is likely that self-heating was present regardless of biasing conditions. Detailed electrothermal characterization of these samples is needed in order to separate thermal and electron trapping effects in these devices once a robust surface dielectric passivation process is established in our fabrication process.

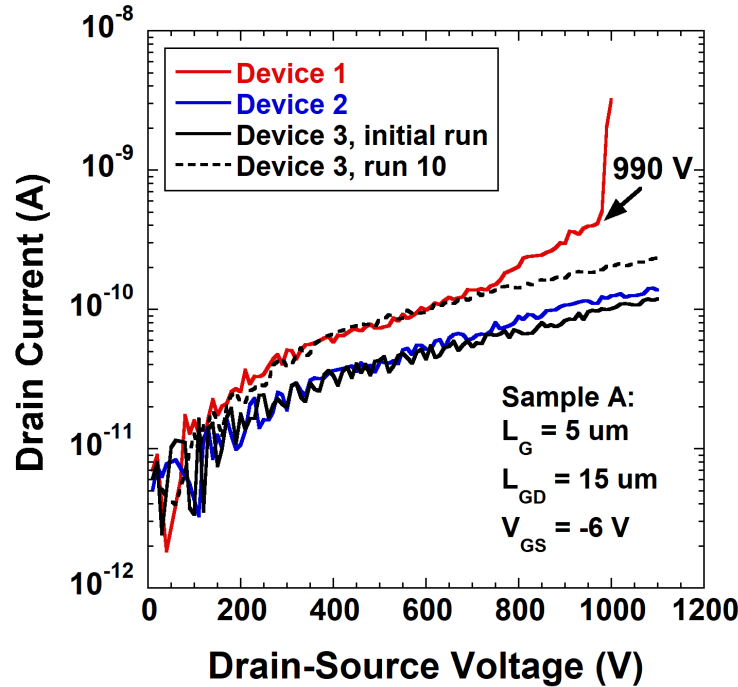


FIG. 10. Breakdown voltage measurements in forward blocking mode ( $V_{GS} = -6$  V,  $V_{DS}$  up to 1100 V) on three devices on sample A.

The presence of Si donors at the delta-doped  $(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3$  layer and near the  $(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3/\text{Ga}_2\text{O}_3$  interface, detected via SIMS/ECV and presumably the reason for the dual electron channel behavior measured in the  $I$ - $V$  data from sample B, was further simulated to confirm the effect of the SIMS/ECV-measured dopant profile on the electron concentration in the heterostructure. Figures 8 and 9 show the simulated band diagrams for samples A and B, respectively, in the presence of 20 nm  $\text{Al}_2\text{O}_3$  deposited on the surface. In addition, band diagram simulations with and without  $\text{Al}_2\text{O}_3$  for all three samples are provided in the supplementary section, and a more detailed simulation study was carried out separately in Ref. 25. We note that the effect of traps at the  $\text{Al}_2\text{O}_3/(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3$  interface was neglected in the simulations and will be accounted for once experimental data becomes available. Unlike the delta-doped Si profile in that study, here the simulated dopant distribution profile in the  $(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3$  layers were obtained from the experimental ECV data. Thus, a slightly graded dopant profile of 4 nm/decade of dopant

density (Fig. S3-S5) was used as an input with Al composition ratio varied from 23 percent (sample A, Fig. 8 and S3) to 14 percent (sample C, Fig. S5). Increasing the concentration of Si in the 3 nm thick  $(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3$  layer increased the concentration of electrons in the potential well, as expected. Furthermore, the simulated band structure profiles showed it was possible to form a secondary channel if Si was present at the  $(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3/\text{Ga}_2\text{O}_3$  interface and Si dopant redistribution took place in the  $(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3$  barrier. This was the case for sample B (Fig. 9) but not for sample A (Fig. 8), where the Si concentration in the barrier was not sufficiently high. The deposition of  $\text{Al}_2\text{O}_3$  gate dielectric further increased the conduction band energy towards the surface, resulting in a slightly higher electron concentration in the channel; however, the simulated band diagrams showed that the effect of the additional  $\text{Al}_2\text{O}_3$  was minor and thus unlikely to have caused secondary channel formation detected by CV and transconductance measurements in Fig. 7. For the case of sample C (Fig. S5b), it appeared that the same conditions were met as in sample B, with the additional reduction in Al mole fraction causing even lower ratio of primary/secondary channel electron concentration.

Forward blocking mode ( $V_{\text{GS}} = -6$  V)  $I$ - $V$  measurements were performed on-die in inert liquid (Fluorinert<sup>TM</sup>) on representative devices with 15  $\mu\text{m}$  gate-drain spacing and 5  $\mu\text{m}$  gate length from sample A up to the instrument limit of 1100 V. To preserve samples for future studies, samples B and C were not tested for breakdown voltage. Destructive breakdown voltage of 990 V was measured on one device, as shown in Fig. 10. The other two devices tested for breakdown on sample A did not exhibit breakdown within the measurement range after ten successive measurements. Figure 10 shows the initial and final current-voltage measurement for device 3, where only a slight increase in off-state leakage current was observed.

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### 3.4 Characterization of $\beta$ -Ga<sub>2</sub>O<sub>3</sub> Homoepitaxial Films and MOSFETs Grown by MOCVD at High Growth Rates

#### 3.4.1 Introduction

The ultra-wide bandgap semiconductor gallium oxide, especially the beta phase ( $\beta$ -Ga<sub>2</sub>O<sub>3</sub>), has been the subject of intense research efforts over the past several years[1,2]. The immense interest in the material lies in its fundamental properties (large bandgap, critical electric field and electron saturation velocity) and the existence of a manufacturable native substrate which make it highly suitable for application in and for the advancement of high power electronics.

One of the key challenges to realizing the full potential of  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> for such applications is the ability to grow electronic-grade material at high growth rates. Although the material's large critical electric field strength reduces the required thickness of low-doped layers needed to withstand high voltages, the growth rate of traditional growth methods makes it practically cost-prohibitive to produce them. For example, a 20 kV device would require a 40  $\mu\text{m}$  thick, low-doped drift layer. Molecular beam epitaxy (MBE) growth rates for  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> are typically in the 0.1 to 0.8  $\mu\text{m}/\text{h}$  range [3, 4], which would require a growth longer than 50 hours. Similarly, conventional metalorganic chemical vapor deposition (MOCVD) growth rates are typically 0.4  $\mu\text{m}/\text{h}$  [5,6], which would require a growth longer than 100 hours.

Realizing this critical technology challenge, we have designed and implemented a novel MOCVD growth reactor to achieve growth rates in the 1.0 – 4.0  $\mu\text{m}/\text{h}$  range [7] while simultaneously achieving the low background doping ( $< 2 \times 10^{16} \text{ cm}^{-3}$ ) needed for high performance voltage blocking layers.

Here, we report on the characterization of such films through a set of optical, structural and electrical methods to better understand the quality of such films and to assess their suitability for high-voltage power switch application. We also report on the fabrication of a lateral transistor device using a thin,  $n^+$  Si-doped channel.

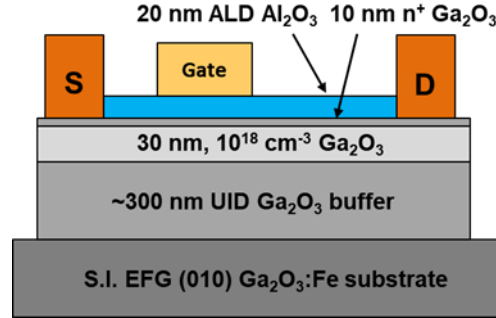
#### 3.4.2 Experimental Details

##### 3.4.2.1 Homoepitaxial Growth

The homoepitaxial  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> films were grown using Agnitron Technology's Agilis 100 R&D MOCVD reactor [7-10] on either unintentionally-doped (UID) or semi-insulating, Fe-doped (010)  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> substrates (Novel Crystal Technology, NCT) as described in Table I. Triethylgallium (TEGa) and pure oxygen (99.999%) were used as a precursor for Ga and oxidation, respectively. Ar (99.9999%) gas was used as a carrier to deliver the TEGa vapor into the reactor. The Ar and O<sub>2</sub> gases were passed through point-of-use purifiers to further reduce the impurity level to  $< 1$  ppm. The FET structure (264A-B) consisted of an  $\sim 300$  nm thick UID  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> buffer layer, followed by a 30 nm thick,  $10^{18} \text{ cm}^{-3}$  channel layer and a 10 nm thick  $n^+$  ( $\sim 10^{19} \text{ cm}^{-3}$ ) contact layer (Fig. 1). Silane (SiH<sub>4</sub>) diluted in helium precursor was used for doping. The quality of the buffer layer was assessed by growing three  $\sim 2$   $\mu\text{m}$  thick UID  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> films (one (261A) on a UID substrate and two (261B-C) on semi-insulating, Fe-doped substrates) as control samples and which were a large part of the characterization effort reported here. The layers were grown at  $\sim 1.0$   $\mu\text{m}/\text{hr}$ .

**Table I.** Summary of sample structures used in study.

| Sample ID    | Substrate      | Epilayer              |
|--------------|----------------|-----------------------|
| FET (264A&B) | Fe-doped (010) | FET structure         |
| 261A         | UID (010)      | 2 $\mu\text{m}$ , UID |
| 261B & 261C  | Fe-doped (010) | 2 $\mu\text{m}$ , UID |

Figure 1. Schematic of fabricated  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> Si-doped channel MOSFET transistor devices grown by the MOCVD method.

### 3.4.2.2 Characterization Methods

Homoepitaxial layers were characterized for their optical, structural and electrical properties. Optical reflectance of control samples (261A-C) was measured for wavelengths between 250 and 850nm at normal incidence using a halogen lamp as a source, and the reflected beam was dispersed through an Ocean Optics S2000 spectrometer with a 50  $\mu\text{m}$  slit. The large area film morphology was evaluated using Nomarski microscopy and surface roughness by a Bruker Dimensions FastScan atomic force microscopy (AFM) in tapping mode. Note the samples were cleaned prior to microscopy by ultrasonic solvent clean for 5 min followed by a 5 min HCl soak and 5 min H<sub>2</sub>O<sub>2</sub> soak. To understand homoepitaxial quality of Ga<sub>2</sub>O<sub>3</sub> films high-resolution x-ray diffraction (XRD) measurements were made using a Rigaku SmartLab system with a 9 kW, (45 kV, 200 mA) copper (Cu) rotating anode source. Cu-K $\alpha$  radiation was used in conjunction with a D/tex 1-D detector. The crystalline structure and orientation of the films were also investigated by in-plane grazing incidence diffraction (IP-GID) using a Rigaku SmartLab with Cu rotating anode x-ray source, upstream 0.5 degree parallel slit collimator, and downstream 0.114 degree parallel slit analyzer. Raman spectra were taken using a confocal Thermo DXRxi system with 532 nm laser excitation and 4 mW incident power. Cathodoluminescence (CL) emission was excited with an electron beam current of 3  $\mu\text{A}$  and a beam accelerating voltage of 3 kV in ultra-high vacuum chamber at cryogenic (5 K) and at room temperatures ( $\sim$ 300 K). Under this accelerating voltage condition the estimated maximum carrier penetration is 80 nm. The CL emission, collected with a combination of f-number matching lens and mirrors, was analyzed by a compact fiber optic spectrometer. The electron spin resonance (ESR) measurements were performed at 300 K with a Bruker 9.5 GHz spectrometer. Samples with dimensions of 1 mm x 3 mm were cleaved from the parent Ga<sub>2</sub>O<sub>3</sub> epilayer/(010) Ga<sub>2</sub>O<sub>3</sub> substrates. Microwave powers of 0.4 – 1.0 mW and 0.05 – 1 Gauss modulation amplitudes were employed for these experiments.

An average free carrier concentration in the MOCVD epilayer of about  $5.3 \times 10^{15} \text{ cm}^{-3}$  was measured on control samples using the capacitance-voltage technique on a vertical Schottky diode processed with Pt/Au Schottky contacts and Ti/Au blanket backside Ohmic contact (Fig. 2).

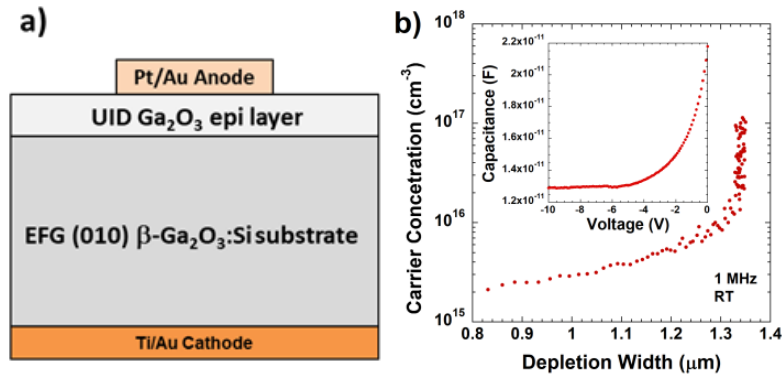


Figure 2. a) Schematic of Schottky diode control sample grown on unintentionally Si-doped (010) EFG substrate and b) free carrier concentration measured in the undoped buffer layer grown by MOCVD. Inset: Capacitance-voltage measurement data used for the carrier concentration analysis.

### 3.4.2.3 Fabrication Details

Lateral device structures were fabricated for electrical testing. Mesa isolation regions were etched in BCl<sub>3</sub> plasma (800 W ICP, 40 sccm BCl<sub>3</sub>, 5 mTorr, 3 min.) and Ti/Au Ohmic contacts were lifted off using optical non-contact photolithography (Heidelberg MLA150). A 1-minute, 470 °C rapid-thermal anneal in N<sub>2</sub> atmosphere was performed. Linear transfer length method (LTLM) measurements in the  $\pm 0.1$  V range yielded a contact resistance of  $5.26 \times 10^{-4} \Omega \cdot \text{cm}^2$ . A 20 nm thick Al<sub>2</sub>O<sub>3</sub> gate dielectric was deposited via atomic layer deposition (ALD), followed by a Pt/Au gate metal stack. All metallization was performed using electron-beam evaporation. No passivation dielectrics were deposited on these samples.

### 3.4.3 Results & Discussion

#### 3.4.3.1 Optical Characterization

Transmittance of the three control samples are presented in Figure 3. The optical bandgap was calculated from a linear extrapolation of  $(\alpha h\nu)^2$  as a function of  $h\nu$ . An optical bandgap 4.50 eV was extracted for the UID Ga<sub>2</sub>O<sub>3</sub> on undoped Ga<sub>2</sub>O<sub>3</sub> substrate sample (261A). Referring to Ricci et al. [11], it is expected that unpolarized transmittance of Ga<sub>2</sub>O<sub>3</sub> through [010] should display an optical bandgap near 4.50 eV, owing to anisotropy of the optical bandgap. An optical bandgap of 4.38 eV was extracted for the UID Ga<sub>2</sub>O<sub>3</sub> on Fe-doped Ga<sub>2</sub>O<sub>3</sub> substrate samples (261B and 261C). In Figure 3, there is a clear decrease in transmittance at 275 nm to 350 nm in the structure with Fe doping in the substrate as compared to the structure with a UID substrate.

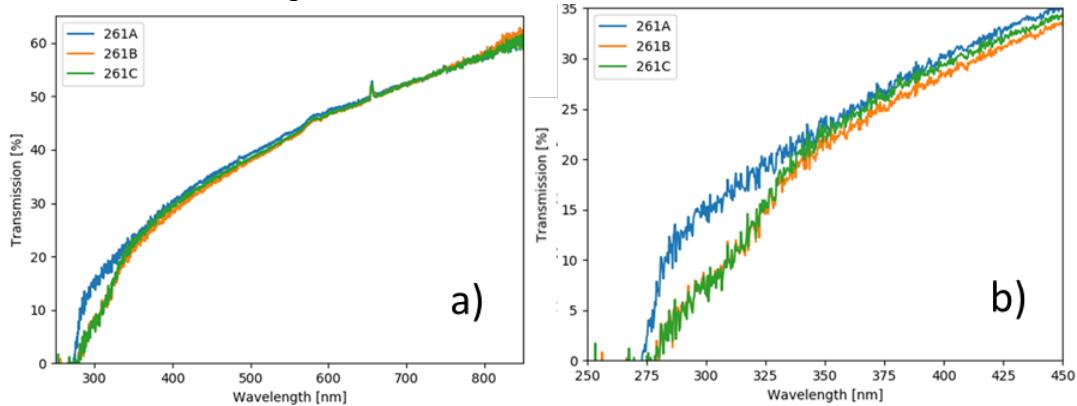


Figure 3. (a) UV and visible and (b) UV and near-UV transmittance of Ga<sub>2</sub>O<sub>3</sub> structures display a cutoff at 275 nm.

### 3.4.3.2 Structural Characterization

#### 3.4.3.2.1 Surface Morphology

Samples with the thick UID epi are decorated with oblong features for all three control samples as determined by Nomarski microscopy. Sample 261B, shown in Fig. 4, exhibited the lowest density of features at the center (Fig. 4 (a)) and highest density at the sample edge (Fig. 4 (b)), while sample 261A exhibited lower density at the edge and higher density at the center. The features in all samples are along one direction, likely parallel to the growth offcut direction.

To evaluate the surface roughness and features observed in Nomarski micrographs, the samples were also studied using AFM. The oblong features appear as pits in the surface, with depths on the order of 100's of nm as seen in Fig. 5 (a). The surface morphology for all three samples looks relatively similar and an example for sample 261B is shown in Fig. 5 (b), where steps along the sample are clearly seen in the image. The surface roughness for samples 261A and 261B are 1.4 and 1.0 nm RMS, respectively. Sample 261C is not measured as there were particles across the surface, see Fig 5 (c).

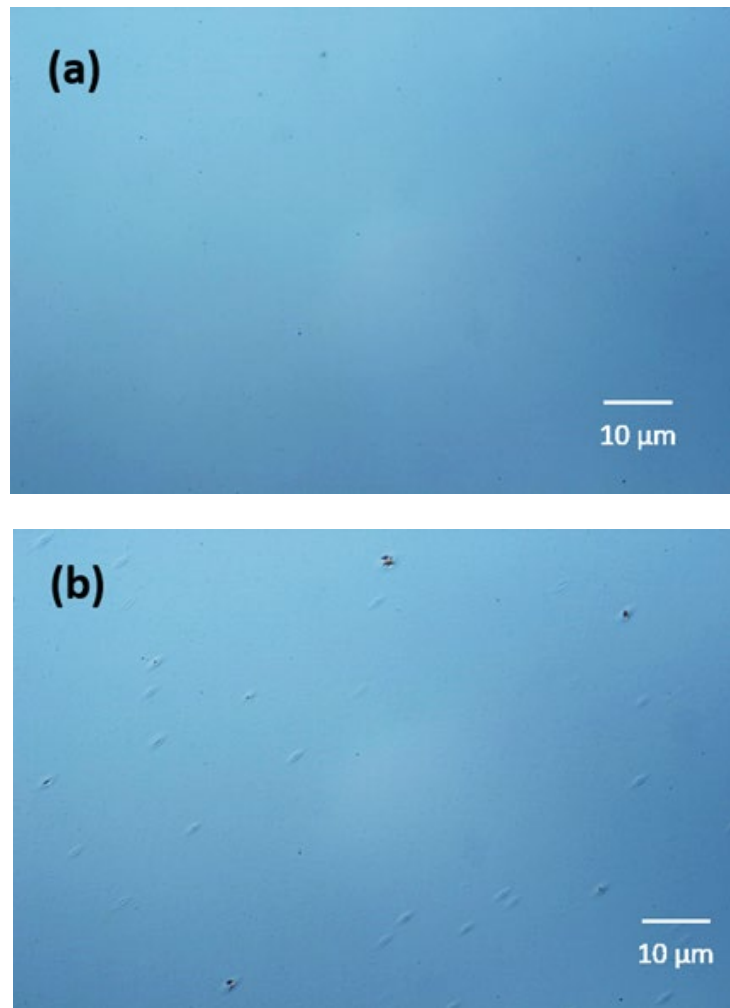


Figure 4. Nomarski micrographs of Ga<sub>2</sub>O<sub>3</sub> epi sample 261B in the (a) center and (b) edge of the sample.

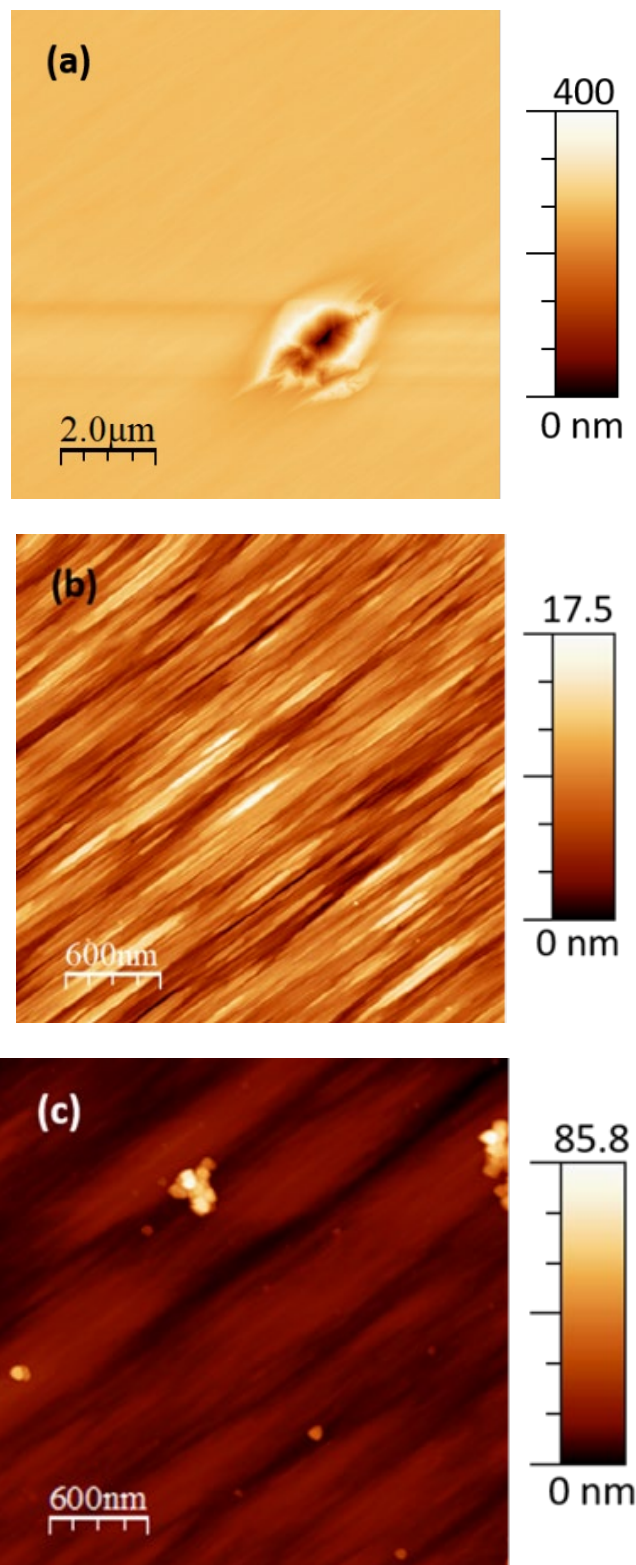


Figure 5. AFM micrographs of (a) pit and (b) surface morphology for sample 261B, and (c) particles on sample 261C.

### 3.4.3.2.2 Microstructure

Figure 6 shows symmetric and asymmetric XRD measurements of homoepitaxial  $\text{Ga}_2\text{O}_3$  films sample 261A (blue) and sample 261B (red) grown on (010) bulk substrate from NCT. Based on the XRD peak positions, film and substrate both have (020) and (22-1) peaks for both samples, with the same (020) peak, however the (22-1) peak for sample 261A is larger indicating smaller  $d$  spacing along this direction.

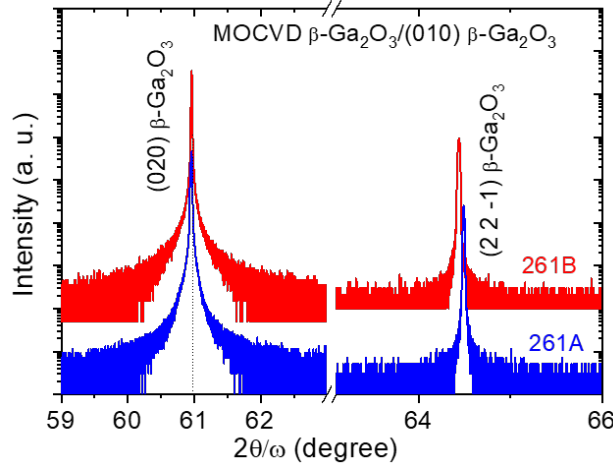


Figure 6. Symmetric and asymmetric XRD peaks for homoepitaxial  $\text{Ga}_2\text{O}_3$  films 261A and 261B.

Rocking curves (RC) of samples 261A and 261B along (020) and (22-1) directions are shown in Fig. 7 (a) and (b), respectively. The rocking curve (RC) full-width at the half maximum (FWHM) of 36 and 26 arcsec were measured for sample 261A along symmetric and asymmetric directions. Similarly RC FWHM values are 23 and 36 arcsec for sample 261B. These are narrow RC FWHM and can have contributions from the substrate. However, if MOCVD grown films have lower quality, it should broaden either symmetric, asymmetric or both directions in these RC measurements. Assuming both substrates are identical, it is clear that along out-of-plane direction, sample 261B has better crystalline quality. However, along asymmetric direction sample 261A has better crystallinity indicating a lower density of in-plane defects.

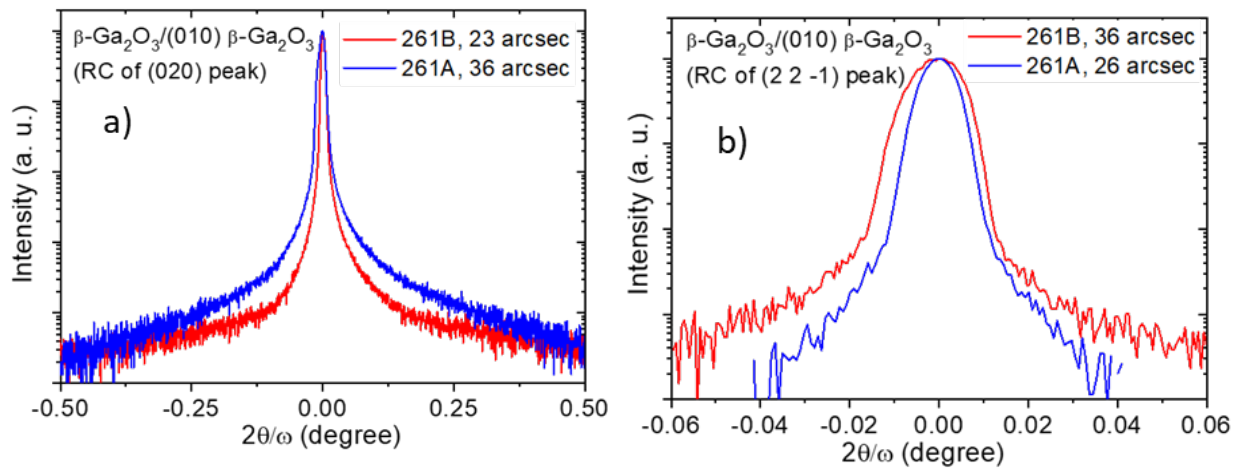


Figure 7. Rocking curves of samples 261A and 261B along (020) and (22-1) directions shown in (a) and (b), respectively.

In a further effort to understand crystalline quality of the epilayers,  $\phi$  rocking scans were performed at the (400) reflection using a surface sensitive IP-GID geometry, as shown in Figure 8(a) for sample 261A. All samples demonstrated the expected two-fold rotational symmetry for the (400) reflection and also exhibited a pair of weak diffraction peaks that was offset from the primary peaks by approximately 30.3 degrees. The weak diffraction peaks can be attributed to the (40-1) reflection, as the corresponding lattice spacing is very close to that of (400). Gaussian fits of the (400)  $\phi$  scan peaks for samples 261A, 261B, and 261C are shown in Figures 8(b), 8(c), and 8(d), respectively. The peak FWHM increases slightly from sample 261A to sample 261B, and from sample 261B to sample 261C, indicating a broadening of the mosaic twist distribution.

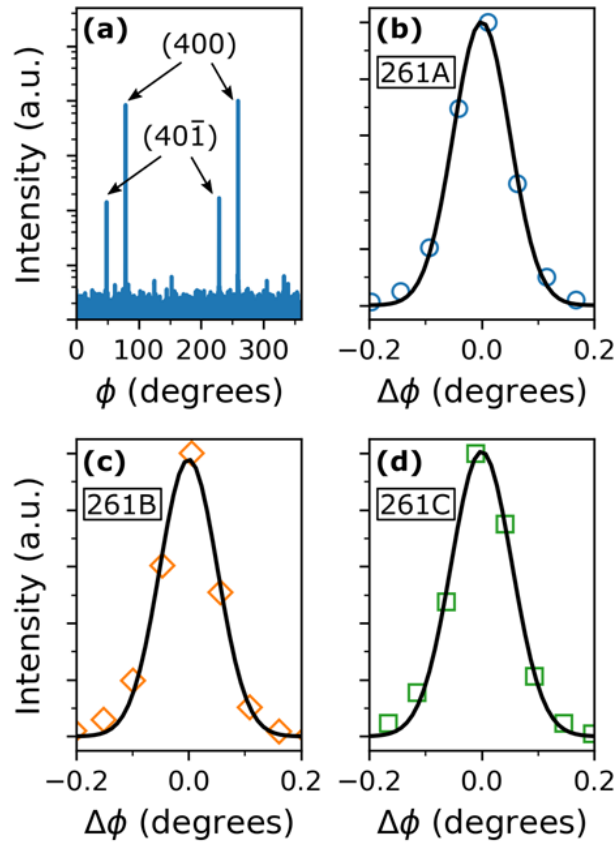


Figure 8. IP-GID  $\phi$  scans taken at the (400) reflection. (a) The complete scan for sample 261A, and Gaussian fit of the first (400) peak for (b) sample 261A in which the FWHM = 426 arcsec, (c) sample 261B in which the FWHM = 434 arcsec, (d) sample 261C in which the FWHM = 457 arcsec.

Coupled  $2\theta\chi/\phi$  scans were performed in both the [400] and [40-1] directions, as shown in Figure 9(a) for sample 261C. The weak peaks at  $2\theta\chi \approx 27$  degrees are the result of Cu-K $\beta$  radiation, which was present due to the lack of an upstream monochromator, and can be ignored. Gaussian fits of the (400) peak for samples 261A, 261B, and 261C are shown in Figures 9(b), 9(c), and 9(d), respectively. Starting from the assumption that angle  $\beta = 103.83$  degrees, the corresponding  $a$  lattice constants are calculated from the peak positions as 12.2163 Å, 12.2215 Å and 12.2061 Å. Gaussian fits of the (40-1) peak for samples 261A, 261B, and 261C are shown in Figures 9(e), 9(f), and 9(g), respectively. Using previously determined  $a$  lattice constants, the

corresponding  $c$  lattice constants are calculated from the peak positions as 5.7888 Å, 5.7539 Å and 5.7688 Å. These values are in close agreement with those reported by Åhman et al., [12]:  $a = 12.214$  Å,  $c = 5.7981$  Å.

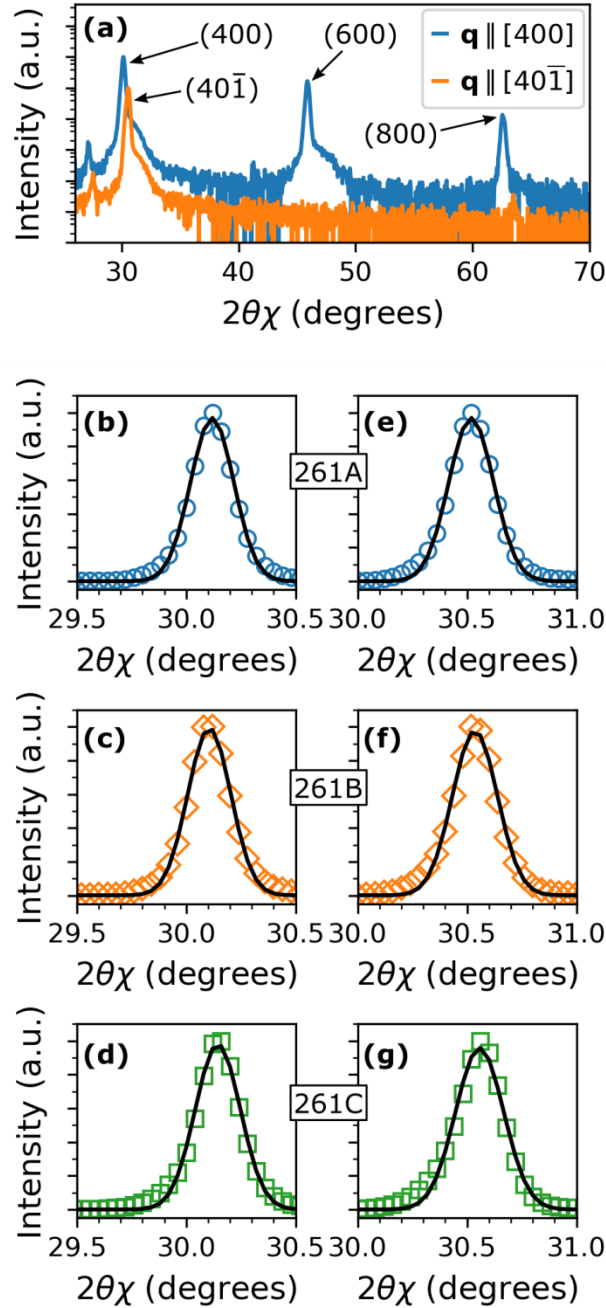


Figure 9. IP-GID coupled  $2\theta\chi/\phi$  scans. (a) The complete scans for sample 261C performed with momentum transfer  $q$  aligned in the  $[400]$  and  $[40\bar{1}]$  directions, in which the former has been offset vertically for clarity. Gaussian fit of the (400) peak for (b) sample 261A, which is centered at 30.1186 degrees with FWHM = 840 arcsec, (c) sample 261B, which is centered at 30.1055 degrees with FWHM = 835 arcsec, and (d) sample 261C, which is centered at 30.1444 degrees with FWHM = 874 arcsec. Gaussian fit of the (40 $\bar{1}$ ) peak for (e) sample 261A, which is centered at 30.5188 degrees with FWHM = 870 arcsec, (f) sample 261B, which is centered at 30.5358 degrees with FWHM = 871 arcsec, and (g) sample 261C, which is centered at 30.5572 degrees with FWHM = 931 arcsec.

### 3.4.3.2.3 Point Defects

To assess local order in the films, Raman spectra were taken in three locations for each sample and a representative spectrum (center of sample) is shown in Fig 10. There are 10 Raman active modes for samples 261A and 261B, while only 9 for sample 261C. The peak positions are in good agreement with Yao et al.[11]. Comparing the peak positions of the two epilayers grown on doped substrates with the undoped substrates, all peaks have the same position with the exception of a few which had small peak shifts of  $\sim 1 \text{ cm}^{-1}$ . For sample 261B, the following have blue shifted from sample 261A; 320.87, 347.87 and  $476.11 \text{ cm}^{-1}$ . The peak at  $630.4 \text{ cm}^{-1}$  blue shifted and  $766.35 \text{ cm}^{-1}$  red shifted for sample 261C. Comparing the peak intensity of the UID to the doped substrate samples, both epilayers on doped substrates showed variations in intensity for the majority of the peaks, however, sample 261C had the most variation as this was the highest doped sample. This is likely due to phonon-plasmon coupling as it is expected to show significant peak shifts and changes in intensity from excess free carriers [13].

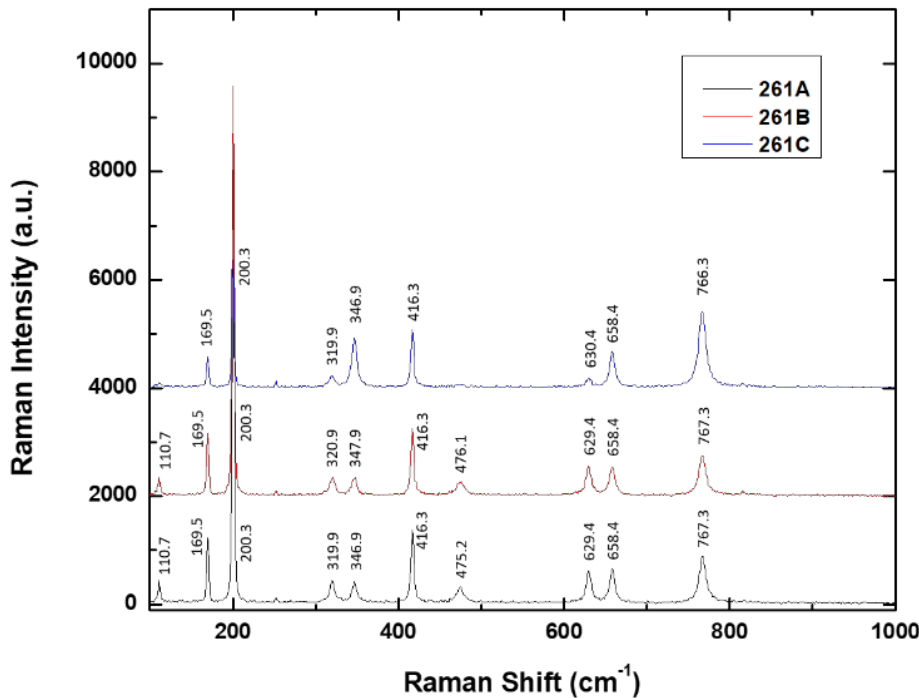


Figure 10. Raman spectra of  $\text{Ga}_2\text{O}_3$  epitaxy, samples 261A-C, showing 10 Raman active modes.

Low temperature CL spectra of a UID  $\text{Ga}_2\text{O}_3$  substrate used for sample 261A and an undoped (n-type)  $\text{Ga}_2\text{O}_3$  epilayer deposited on a Fe-doped  $\text{Ga}_2\text{O}_3$  bulk substrate (261B) are represented in Figures 11(a) and 11(b), respectively. The spectra acquired from both samples are quite similar. There were very small variations of the broad emission band intensities on sequential spectra acquired with 10 sec integration time. Peak intensity normalization indicates that no line shape change is observed under present experimental conditions. The broad emission band seems to

result from the overlap of dominant emission bands at 2.95, 3.25, and 3.50 eV. Emission bands near 3.0 and 3.5 eV have been tentatively assigned to Ga and O vacancies, respectively [14].

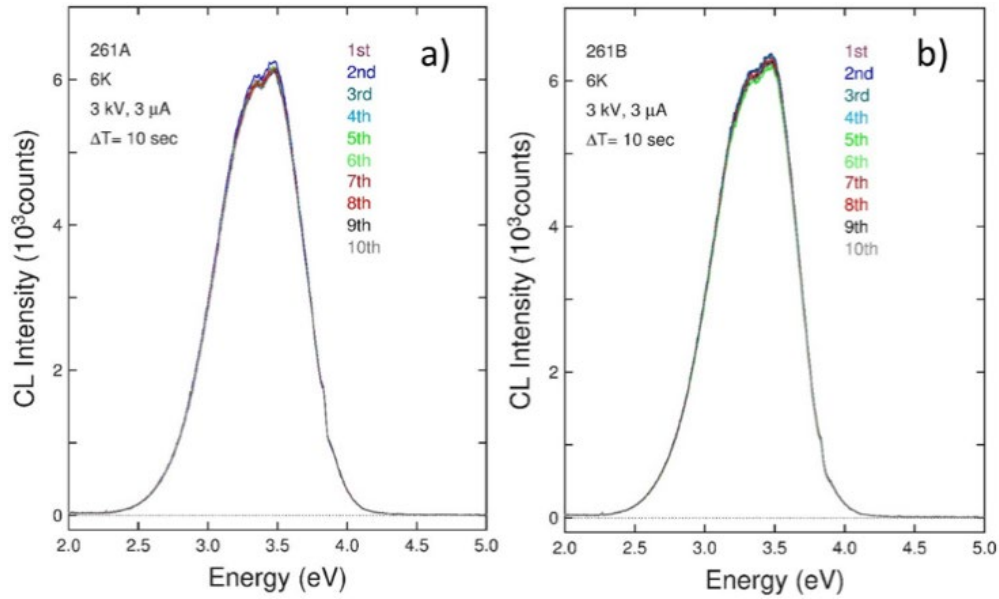


Figure 11. Low temperature CL spectra of (a) undoped (n-type)  $\text{Ga}_2\text{O}_3$  substrate used for sample 261A growth, and (b) sample UID epilayer 261B deposited on a Fe-doped  $\text{Ga}_2\text{O}_3$  bulk substrate.

ESR spectra are shown in Fig. 12 for undoped (n-type)  $\text{Ga}_2\text{O}_3$  epilayer sample 261B deposited on a semi-insulating, Fe-doped  $\text{Ga}_2\text{O}_3$  bulk substrate with the applied magnetic field oriented perpendicular and parallel to the [010] crystalline direction. The dominant signals observed for these very wide magnetic field scans (0 – 7000 G) have been previously reported [15] and are associated with  $\text{Fe}^{3+}$  defect centers on the Ga octahedral sites in the parent  $\text{Ga}_2\text{O}_3\text{:Fe}$  substrate. Very narrow scans in the magnetic field range of 3460-3470 G were expected to reveal a signal associated with shallow donors in  $\text{Ga}_2\text{O}_3$  but such a resonance was not observed above the noise level. Given the typical ESR spin detection limit of approx.  $10^{11}$  spins and the epilayer sample dimensions, we estimate that an uncompensated shallow donor concentration (i.e.,  $N_d - N_a$ ) of low- to mid- $10^{16} \text{ cm}^{-3}$  would be required to observe an EPR signal associated with those donors. Most notably, the non-observation of a shallow donor ESR signal is very consistent with the room-temperature C-V measurements (see Fig. 2) presented above for a sister epilayer sample (261A) deposited on a UID (n-type)  $\text{Ga}_2\text{O}_3$  bulk substrate during the same growth run that revealed a carrier concentration of  $\sim \text{mid-}10^{15} \text{ cm}^{-3}$ .

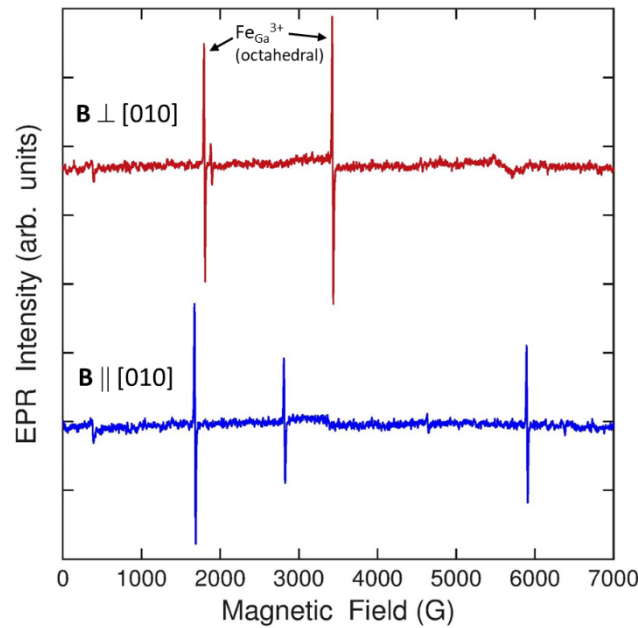


Figure 12. ESR spectra obtained at 300K for UID (n-type)  $\text{Ga}_2\text{O}_3$  epilayer/(010)  $\text{Ga}_2\text{O}_3\text{:Fe}$  sample 261B with  $\mathbf{B}$  oriented perpendicular and parallel to the [010] direction. The various strong signals are associated with  $\text{Fe}^{3+}$  centers on the Ga octahedral sites in the parent substrate.

#### 3.4.3.2.4 Electrical Characterization

Room temperature Hall measurements indicated a sheet carrier density ( $N_{\text{SH}}$ ) of  $1.74 \times 10^{12} \text{ cm}^{-2}$ , a record high Hall mobility ( $\mu_{\text{H}}$ ) for a  $\text{Ga}_2\text{O}_3$  transistor of  $170 \text{ cm}^2/\text{V}\cdot\text{s}$ , and a sheet resistance of  $21.02 \text{ k}\Omega/\text{sq}$  [16-18]. Temperature-dependent Hall characteristics, presented in Fig. 13, show a peak mobility of  $1003 \text{ cm}^2/\text{V}\cdot\text{s}$  measured at 220 K. Much higher peak mobility has been reported near 50 K for low-doped MOCVD  $\beta\text{-Ga}_2\text{O}_3$  [8, 19]. Higher room temperature mobility has also been reported from delta-doped  $(\text{Al}_x\text{Ga}_{1-x})_2\text{O}_3/\text{Ga}_2\text{O}_3$  ternary heterostructures grown by MBE [20]. In this case, the presence of the highly-doped channel layers increased the contribution of the impurity scattering component of the total mobility, resulting in overall lower peak bulk mobility and a higher peak temperature. Figure 14 presents the DC input and output current-voltage characteristics of the MOCVD  $\beta\text{-Ga}_2\text{O}_3$  MOSFET. Depletion-mode operation with a threshold voltage of -15 V was obtained with  $I_{\text{ON}}/I_{\text{OFF}}$  ratio of almost  $10^8$ . Field effect mobility was calculated using the relationship  $\mu_{\text{FE}} = g_{\text{m}} \cdot (L/W) / (C_{\text{OX}} \cdot V_{\text{DS}})$  using a FATFET transistor geometry at drain voltage values of 0.1 V and 1 V (Fig. 15). Peak field-effect mobility values obtained at these bias levels were  $5 \text{ cm}^2/\text{V}\cdot\text{s}$  and  $56 \text{ cm}^2/\text{V}\cdot\text{s}$ , respectively. Forward blocking mode operation at  $V_{\text{GS}} = V_{\text{T}}$  was characterized up to 1 kV drain bias. Representative  $I_{\text{DS}}\text{-}V_{\text{DS}}$  data measured on four devices is presented in Figure 16. Only a soft breakdown in the 650-800 V range was indicated by the increase in the off-state drain current, without catastrophic breakdown occurring in these devices. A dielectric passivation scheme must be developed for these devices, at which point a number of additional electrical measurements (such as current collapse and breakdown voltage) will be being performed in order to fully characterize electro-thermal performance.

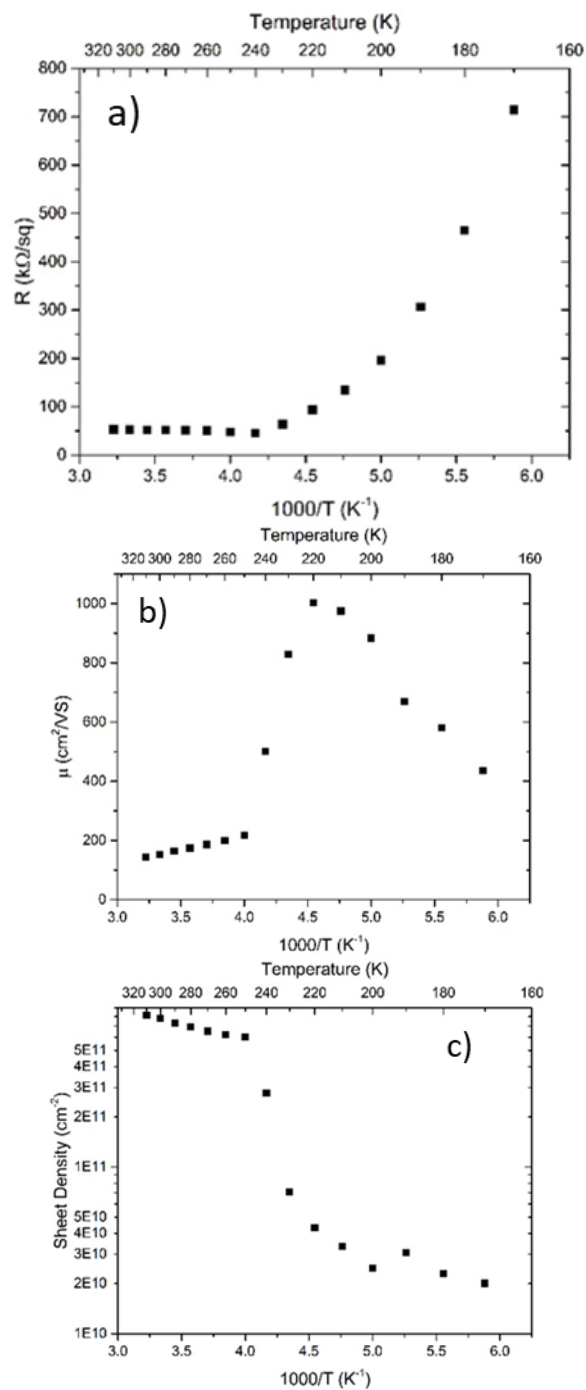


Figure 13. Temperature-dependent Hall measurements measured on a Van der Pauw test structure showing a) sheet resistance, b) Hall mobility, and c) sheet carrier concentration measured as a function of temperature.

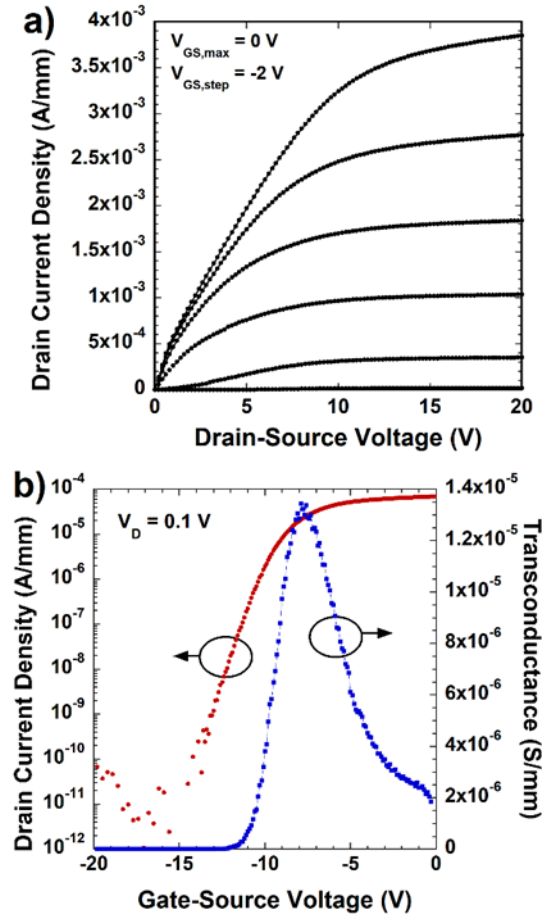


Figure 14. a) Static output ( $J_{DS}$ - $V_{DS}$ ) and b) input ( $J_{DS}$ - $V_{GS}$ ) characteristics measured on the lateral  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> MOSFET schematically shown in Fig. 1.

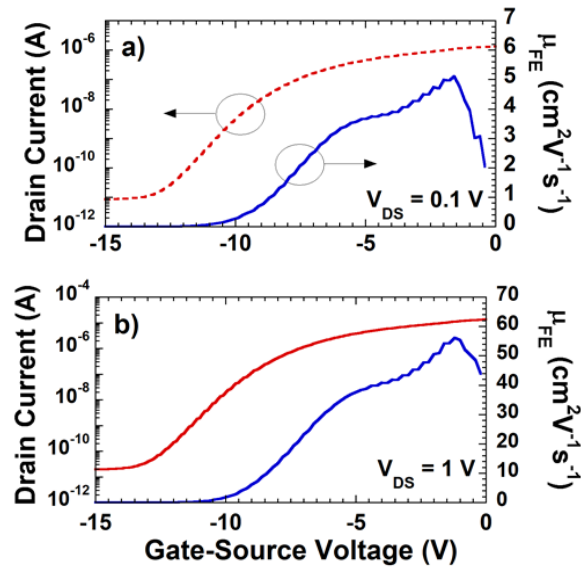


Figure 15. Field-effect mobility as a function of  $V_{GS}$  measured on a FATFET at  $V_{DS}$  values of a) 0.1 V and b) 1 V.

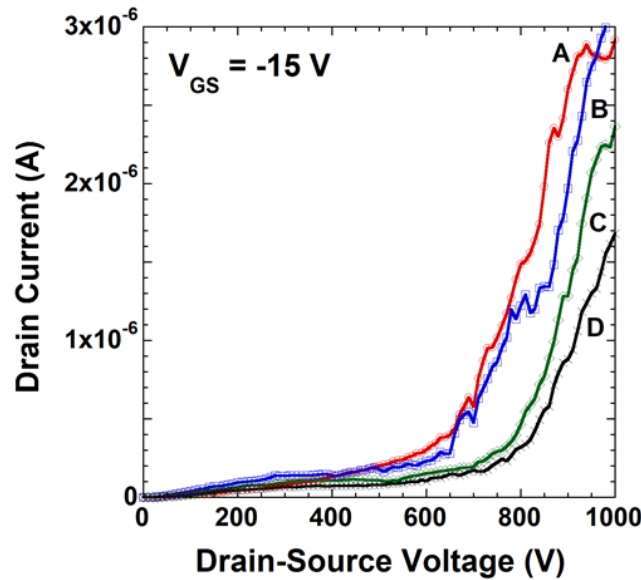


Figure 16. Forward blocking mode operation at  $V_{GS} = -15$  V measured on four representative lateral  $Ga_2O_3$  transistors.

#### 3.4.4 Summary

In summary, homoepitaxial films grown at high growth rates by MOCVD have been extensively characterized for their optical, structural and electrical properties. The films demonstrate relatively smooth surfaces with a high degree of structural order, limited point defectivity ( $N_d - N_a \approx 5 \times 10^{15} \text{ cm}^{-3}$ ) and an optical bandgap of 4.50 eV. Further, the present work reports for the first time a  $Ga_2O_3$  transistor device based on a high growth rate MOCVD technology. We demonstrate record channel mobility, measured by the Hall technique, made possible by the significant improvement in material quality, as discussed above, enabled by the MOCVD growth process. The high growth rate demonstrated via this method paves the road for demonstration of high breakdown voltage ( $V_{BK}$ ) devices on a thick  $Ga_2O_3$  buffer layer.

#### 3.4.5 Acknowledgements

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### 3.5 Towards Controlled Transfer of (001) $\beta$ -Ga<sub>2</sub>O<sub>3</sub> to (0001) 4H-SiC Substrates

#### Introduction

Exfoliation and transfer of films is an important component for the heterogenous integration of semiconductor materials. Part of the ultrawide-bandgap family with materials such as AlN and diamond,  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> has a bandgap of  $\sim 4.85$  eV and is an appealing candidate for next generation high-power ( $> 20$  kV) device applications (1,2). Despite successful reports of  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> device fabrication towards realizing its potential for electronic devices (3,4,5), the most pressing hindrance to  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> power applications is efficient thermal management. The thermal conductivity of  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> is very low ranging from 11 to 27 W/(m·K), which is one to two orders magnitude lower than Si, GaN, 4H-SiC, and diamond (1). The variation in materials properties of  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> along different crystallographic directions is due to its low-symmetry monoclinic crystal structure (6,7,8,9). Poor heat dissipation in either the lateral or vertical  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> device structure degrades device performance (10) and can even cause permanent device failure (11,12). It is evident that efficient heat management during  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> device performance requires integration with other materials with higher thermal conductivities. Thin film  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> growth on various substrates have been demonstrated on sapphire (13), SiC (14), and diamond (15) with high thermal boundary conductance values (thermal conductivity across the interface), but all suffer from reduction in thermal conductivity of the  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> film due to a high density of defects. Growing on SiC or sapphire results in films with a high density of stacking faults or twin boundaries, respectively. For the case of diamond, the resulting film is polycrystalline, and grain boundaries greatly reduce the film's thermal conductivity.

Employing wafer bonding techniques, on the other hand, to integrate single crystal  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> to a high thermal conductivity material would avoid growth-related defects and enable the integration of any  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> orientation. Furthermore, utilizing light atom ion implantation, for example, to induce exfoliation and transfer of  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> films would achieve high-quality  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> thin films of any orientation on a substrate without suffering from defect-induced film thermal conductivity reduction. By engineering the properties of the bonded interface, wafer bonding of  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> could be a pathway to fabricate a heterostructure that exhibits a high interfacial thermal boundary conductance and high thermal conductivity of the film simultaneously. Fundamentally understanding and engineering interfaces via the wafer bonding approach extends beyond thermal management applications and can produce interfaces with emergent properties (e.g., electrical transport) that would not be realizable with only growth techniques.

Light-atom ion implantation of either hydrogen (16) or helium (17) and subsequent annealing was first demonstrated to exfoliate silicon layers to fabricate silicon-on-insulator structures. Annealing an implanted substrate will surface blister due to H<sub>2</sub> or He gas bubble growth near the implanted species projected range beneath the substrate surface. To controllably and reproducibly induce large surface blisters, the anneal involves a low temperature followed by a high temperature step (18). The low temperature anneal serves to initiate the nucleation of either the H<sub>2</sub> or He bubbles while the subsequent high temperature step induces bubble growth. It has been shown that a strong indication of successful exfoliation is the occurrence of surface blistering of an unbonded, implanted substrate. If an implanted substrate were bonded to a handle substrate prior to the high-temperature anneal step, uniform wafer splitting can be achieved after the high-temperature anneal. Large, wafer-scale area thin films are then transferred to a handle substrate after the splitting. The current literature and our previous work have further extended the efficacy and fundamental

understanding of hydrogen and/or helium implantation not only for silicon (19,20,21), but to other materials including germanium (22), various III-V materials (18,23,24,25), and II-VI materials (26) where thin films of each of these materials were successfully exfoliated and transferred. For  $\beta$ -Ga<sub>2</sub>O<sub>3</sub>, implantation of either hydrogen or helium has been successfully demonstrated to exfoliate and transfer the (010) and (201) orientations (27,28). The focus of this current work is to present initial results towards the exfoliation of another technologically relevant orientation – (001)  $\beta$ -Ga<sub>2</sub>O<sub>3</sub>. Currently, 4-inch (001)  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> substrates are commercially available with thick drift layers grown using the hydride vapor phase epitaxy method.

## Experiment

Epi-ready two-inch (001)  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> substrates were implanted with He<sup>+</sup> at an energy of 160 keV with a dose of  $5 \times 10^{16} \text{ cm}^{-2}$  at room temperature, which are the same ion implantation parameters used in our previous work to successfully exfoliate (010)  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> (27). Some of the implanted substrates were then bonded to (0001) 4H-SiC at room temperature under ultrahigh vacuum using a  $\sim 5 \text{ nm}$  thin Ti interlayer to assist with the bond. The primary flats of each substrate were aligned prior to bonding such that the in-plane alignment was  $[100] \beta\text{-Ga}_2\text{O}_3 \parallel [10\bar{1}0] 4\text{H-SiC}$ . High-resolution symmetric  $\omega:2\theta$  X-ray diffraction (XRD) scans were performed using a Bruker-JV D1 diffractometer whose incident X-ray beam is conditioned by a Göbel mirror and a (220) channel-cut silicon crystal to produce a highly collimated monochromatic Cu K $\alpha_1$  beam. The as-implanted XRD was fitted using software that utilizes a genetic algorithm called Differential Evolution (29). An FEI Nova 600 DualBeam focused ion beam system was used to prepare transmission electron microscopy (TEM) samples. An FEI TITAN S/TEM operating at 300 keV was used to measure cross-sectional TEM images of the bonded structure.

## Results and Discussion

XRD scans were measured for the as-implanted, unbonded substrates using the (004)  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> symmetric reflection. Implantation induces strain by causing the lattice to expand due to the intercalation of He and point defects. The underlying substrate material beyond the implanted region imposes in-plane compressive biaxial stress on the implanted region, preventing the lattice within the implanted region to expand in-plane. The typical response to in-plane biaxial compressive stress is out-of-plane expansion – which corresponds to tensile strain. Tensile strain causes peaks to shift towards lower angles in an XRD  $\omega:2\theta$  scan. As shown in Figure 1(a), the as-implanted (004) symmetric  $\omega:2\theta$  scan shows strain fringes left of the main substrate peak towards lower angles. The strain fringes were then fitted using a model to determine the thickness and strain induced by the He implantation. The peak strain value is  $\sim 1.9\%$ , which is nearly a factor of 2 higher than the peak strain value found when implanting (010) substrates under the same conditions (27). This is likely attributed to the higher Poisson's ratio for (001) compared to (010)  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> (30); the theoretical difference in strain due to biaxial stress between (001) and (010) is expected to be  $\sim 2\times$  as well.

The He concentration and displacements-per-atom (DPA) profiles were simulated in SRIM (31) and compared to the simulated strain profile used to model the XRD data as shown in Figure 1(b). The strain profile approximately matches the overlap between the He concentration and DPA profiles, which is what was observed in our previous work when implanting hydrogen in Si using the same dose ( $5 \times 10^{16} \text{ cm}^{-2}$ ) at either 30 keV or 140 keV (21). This suggests that the predominate

cause of strain in these He-implanted substrates are from He-vacancy complexes (i.e., involving both He atoms and the displaced Ga and O atoms in the  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> lattice).

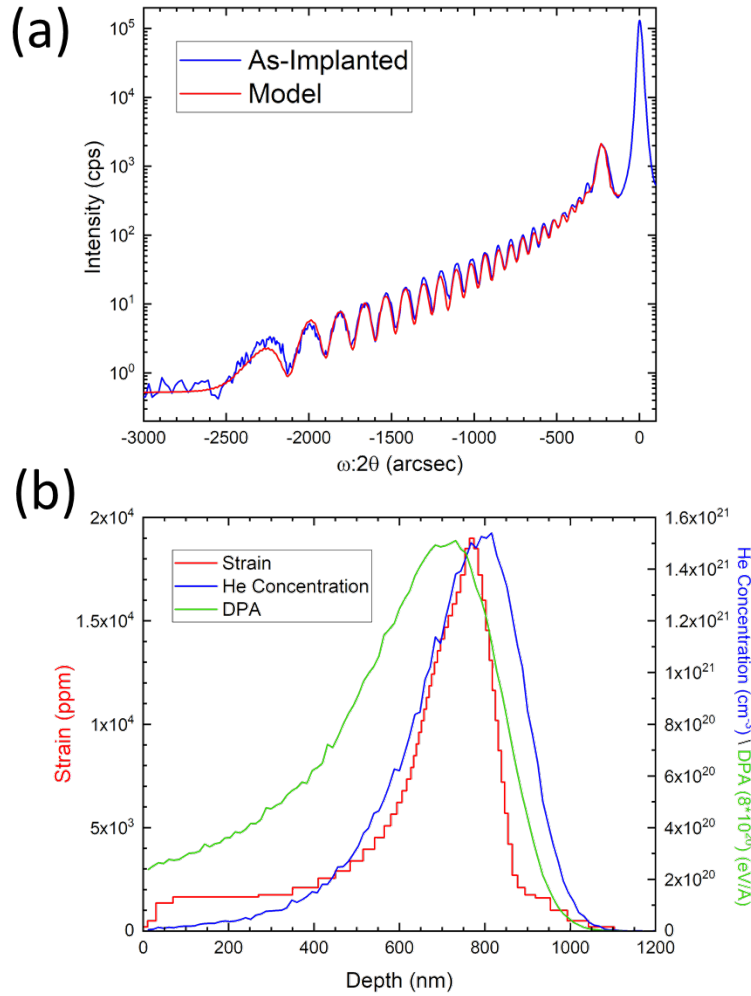


Figure 1. (a) Symmetric (004)  $\omega:2\theta$   $\beta$ -Ga<sub>2</sub>O<sub>3</sub> XRD scan corresponding to the blue curve, where  $\omega:2\theta = 0^\circ$  corresponds to the main substrate away from the implanted region. The Bragg angle for the (004) reflection is  $\sim 33.18^\circ$ . The red curve corresponds to a simulated curve from a model used to fit the strain profile due to the implant. (b) The red curve is the fitted strain profile while the blue and green curves are the He concentration and displacements-per-atom (DPA) profiles simulated with SRIM, respectively.

Following a similar annealing procedure from our previous work on exfoliating He-implanted (010) substrates (27), performing a low-temperature followed by high-temperature anneal on the unbonded implanted substrates reduced the strain induced by the implant. As shown in Figure 2, the strain fringes are removed and only a shoulder remains, which corresponds to leftover strain (on the order of  $\sim 0.2\%$ ) that could be removed upon further annealing. Performing the two-step annealing procedure resulted in surface blisters, as shown in the plan-view optical Nomarski image

in Figure 3. The  $\sim 10\ \mu\text{m}$  light contrast surface features correspond to surface blisters due to He bubble growth that have not broken the surface. Dark contrast features with straight vertical edges are blisters that broke the surface – the vertical edge corresponds to the (100) primary cleavage plane. It has been shown with other materials that demonstrating surface blistering is a strong indication that successful exfoliation and transfer can be achieved.

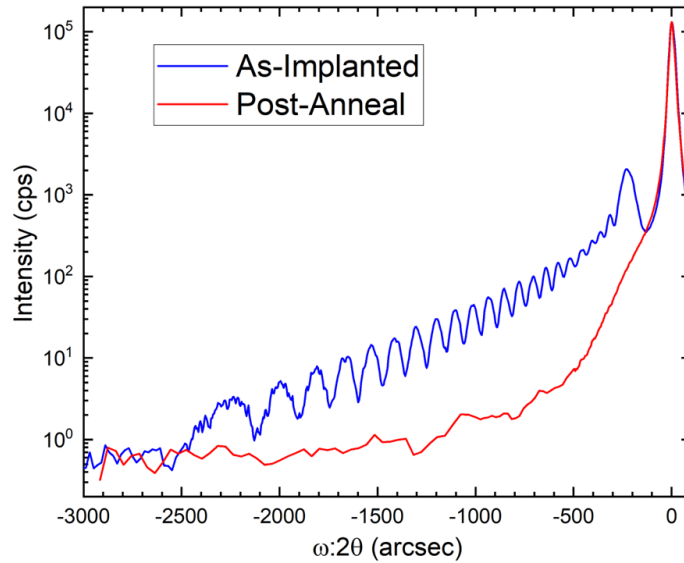


Figure 2. Symmetric (004)  $\beta\text{-Ga}_2\text{O}_3$   $\omega:2\theta$  XRD scan for the as-implanted (blue curve) versus post-anneal (red curve).  $\omega:2\theta = 0^\circ$  corresponds to the main substrate peak. The Bragg angle for the (004) reflection is  $\sim 33.18^\circ$ . After annealing, the strain fringes are removed and only a shoulder is observed, which corresponds to a lower strain of  $\sim 0.2\%$ .

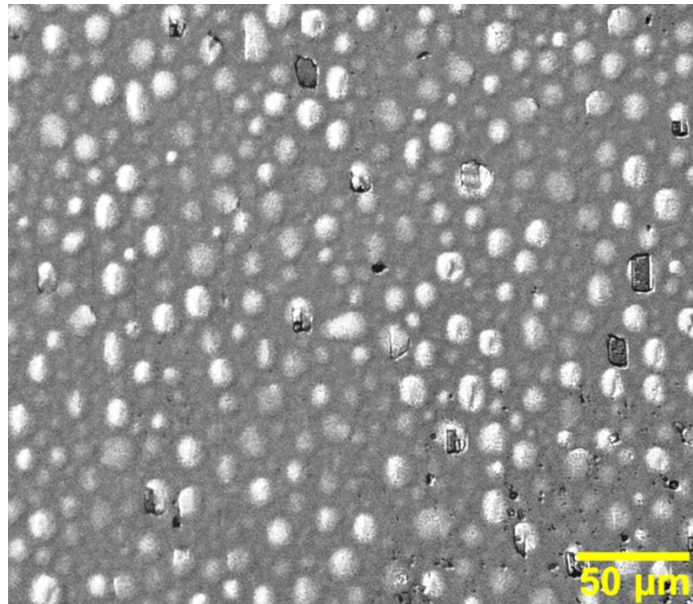


Figure 3. Planview optical Nomarski image of the unbonded (001)  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> surface after the low-temperature followed by high-temperature annealing. The round  $\sim 10\ \mu\text{m}$  light contrast features are He surface blisters. The dark rectangular features are exfoliated regions, with the vertical edges corresponding to the (100) primary cleavage plane.

A set of implanted (001)  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> substrates were bonded prior to annealing. The bonding was performed under ultrahigh vacuum with (0001) 4H-SiC using a thin  $\sim 5\ \text{nm}$  Ti interlayer to assist with the bond. Ti has been used in the literature to bond Si|Si (32) and SiC|diamond (33) due to the chemical reactivity between Ti with Si and C and especially because of its high adhesion to SiO<sub>2</sub>. However, utilizing Ti for bonding has not yet been explored in depth for  $\beta$ -Ga<sub>2</sub>O<sub>3</sub>. The cross-sectional high-resolution TEM image of the bonded structure is shown in Figure 4(a). Contrast can be seen within the implanted region, with a darkest contrast band occurring at the projected range of the implant ( $\sim 750\ \text{nm}$ ). The thickness of the modelled strain profile from Figure 1(b) is consistent with this experimentally measured projected range. The dark contrast is due to Z-contrast from the relatively high He content of  $\sim 11\%$ . In addition to the contrast, lattice distortion can be seen on smaller length scales as shown in Figure 4(b) and 4(c). Figure 4(b) shows the pristine lattice planes beyond the implanted region that shows low overall contrast. Figure 4(c), on the other hand, shows fluctuations in contrast that correspond to lattice distortion induced by the implantation. Figure 5 shows a magnified cross-sectional TEM image focused on the bonded interface, which clearly shows the Ti interlayer. Within the TEM sampling region, the Ti is single crystalline. The (10 $\bar{1}$ 0) Ti surface plane is parallel to the (001)  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> surface plane and the [0001] Ti in-plane direction is parallel to the [ $\bar{1}$ 10]  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> in-plane direction. Studies on the impact of interfaces prepared with Ti on thermal boundary conductance are underway as well as the chemical and structural evolution with post-bonding annealing.

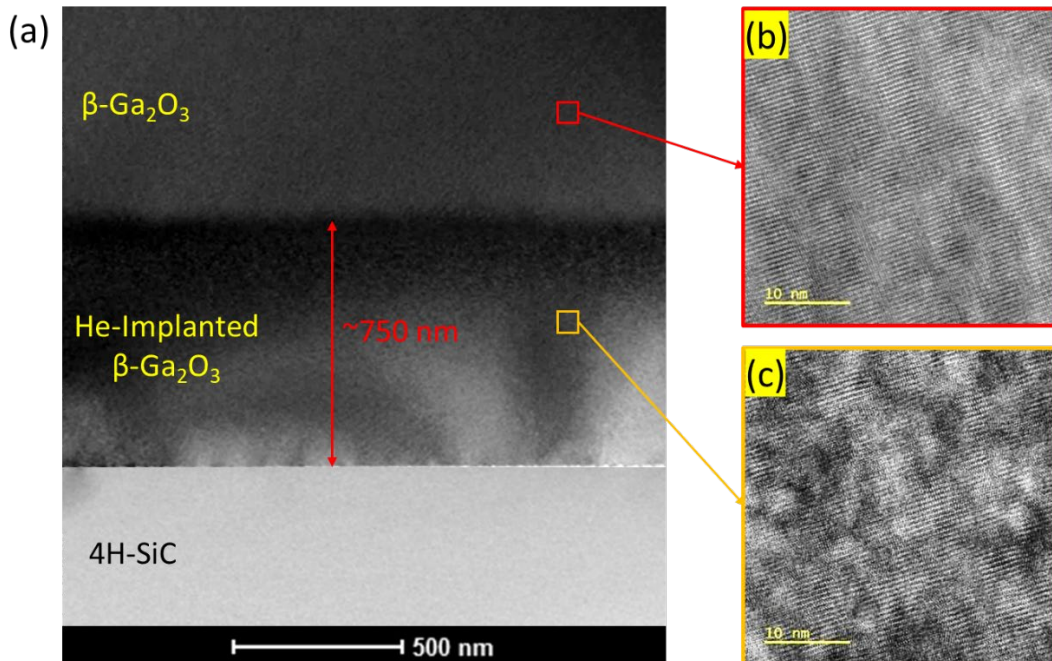


Figure 4. (a) Cross-sectional high resolution TEM image of the bonded structure. The implanted region projected range is  $\sim 750$  nm which can be seen by the dark contrast band. (b) Magnified image within the bulk of the  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> beyond the implant projected range showing pristine, undistorted lattice planes. (c) Magnified image within the implanted  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> region that shows contrast fluctuations due to lattice distortion from the implant.

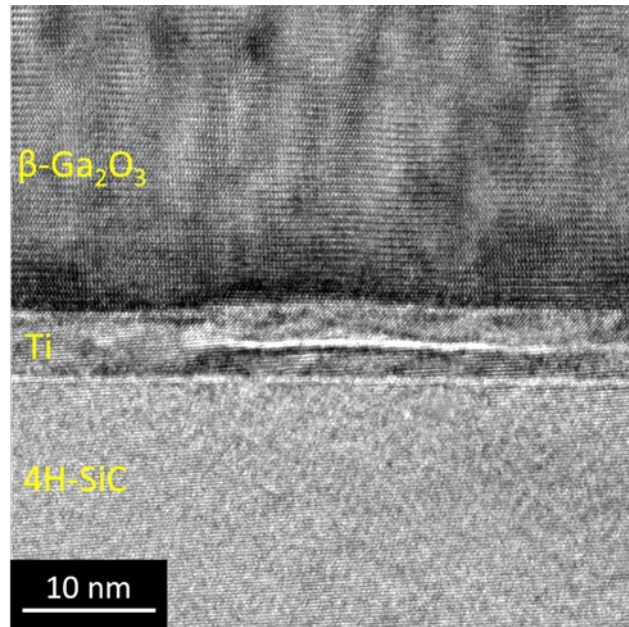


Figure 5. Cross-sectional high resolution TEM image of the bonded structure magnified at the bonded interface.  $\sim 5$  nm of Ti was used to assist with bonding between  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> and 4H-SiC. The zone axis used here is aligned along the  $[110]$   $\beta$ -Ga<sub>2</sub>O<sub>3</sub>. The  $[0001]$  Ti zone axis is found to be parallel to this  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> zone axis.

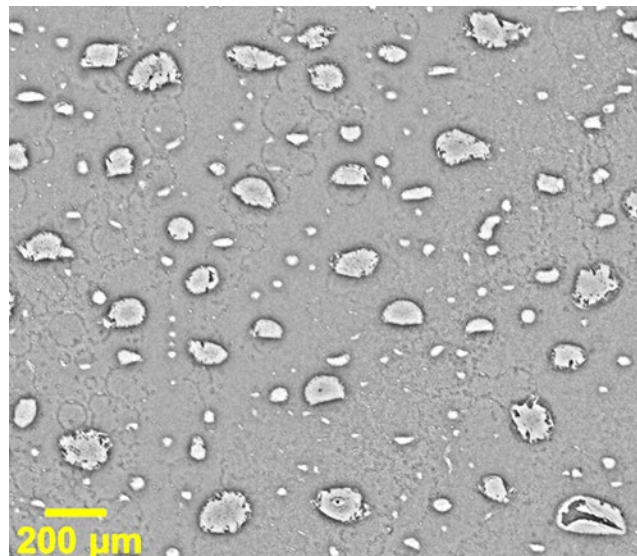


Figure 6. Planview optical Nomarski image of the bonded structure viewed through the  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> wafer after annealing. ~7% of the total bonded area transferred to the (0001) 4H-SiC (round light contrast features). The structure remained bonded and the  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> did not wafer split.

The plan-view optical Nomarski image after annealing the bonded structure using the same anneal conditions that induced surface blistering is shown in Figure 6. While localized regions up to ~200  $\mu$ m wide exfoliated and transferred, the  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> surprisingly did not wafer split. Solutions to achieve wafer splitting and complete  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> film transfer are underway. The reduction in strain shown by the XRD after annealing in Figure 2 and surface blistering in Figure 3 is evidence of He diffusion and bubble growth. The formation of He bubbles at the implant projected range provides mechanical weak points in the  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> at this range beneath its surface. Thus, provided the bond strength with the 4H-SiC is sufficiently strong enough, the bonded structure after annealing as shown in Figure 6 may be able to transfer but subjecting the structure in a pull-test setup. Alternatively, a higher He dose may also be effective in producing a more continuous layer transfer to induce complete wafer splitting.

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#### 4. CONCLUSIONS AND OUTLOOK

The wide bandgap and related high critical breakdown field of  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> enables a high-voltage lateral transistor designed for low on-resistance as well as operation at high-frequency, which reduces the size and weight of the passive circuit elements in the overall system.<sup>21,22</sup> In this work an accurate approximation for mobility as a function of temperature and dopant concentration is presented and employed in an electro-thermal simulation of an (Al<sub>x</sub>Ga<sub>1-x</sub>)<sub>2</sub>O<sub>3</sub>/Ga<sub>2</sub>O<sub>3</sub> MODFET. A decrease in drain current was attributed to a position-dependent mobility reduction caused by a coupled self-heating mechanism and a high electric-field mobility reduction mechanism. Thermal management is a key design issue in all power semiconductor devices. The low thermal conductivity Ga<sub>2</sub>O<sub>3</sub> hampers any design where heat must flow over any appreciable thickness of Ga<sub>2</sub>O<sub>3</sub>. A heatsink via the source contact is discussed and shown to reduce the temperature of the active device channel.

$\beta$ -(Al<sub>x</sub>Ga<sub>1-x</sub>)<sub>2</sub>O<sub>3</sub> films were heteroepitaxially grown on (010)  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> substrates via O<sub>3</sub>-MBE to experimentally determine critical thickness limitations and its dependence on Al content. Films with thicknesses of 50-400 nm and Al content of 6.9-19.5% were studied. Both AFM and XRT confirmed that the most prevalent crack plane was the (100) plane. Additionally, both techniques identified the (001) as a secondary, less frequent crack plane. Despite film cracking, asymmetric RSMs showed that the  $\beta$ -(Al<sub>x</sub>Ga<sub>1-x</sub>)<sub>2</sub>O<sub>3</sub> films were still fully strained. Cross-sectional HR-TEM was used to study crack geometry and propagation; for a ~200 nm thick film with 13.6% Al

content, it was found that the cracks extended from the film surface, through the entire film thickness, and  $\sim 100$ - $200$  nm into the  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> substrate. Finally, experimental data obtained in this study showed good agreement with theoretical calculations for the critical film thickness as a function of Al content as determined by Mu et al.<sup>33</sup>

We presented for the first time growth of  $\beta$ -(Al<sub>x</sub>Ga<sub>1-x</sub>)<sub>2</sub>O<sub>3</sub>/Ga<sub>2</sub>O<sub>3</sub> heterostructures by the O<sub>3</sub>-MBE technique. Epitaxial growth has been demonstrated via this technique in the past for oxide materials such as ZnO, TiO<sub>2</sub>, NiO, and Fe<sub>3</sub>O<sub>4</sub>.<sup>31-34</sup> Here, we have presented the first demonstration of a ternary  $\beta$ -(Al<sub>x</sub>Ga<sub>1-x</sub>)<sub>2</sub>O<sub>3</sub>/Ga<sub>2</sub>O<sub>3</sub> heterostructure growth via O<sub>3</sub>-MBE, and have reported a high sheet carrier concentration in the delta-doped region measured by the Hall technique. Electrochemical  $C$ - $V$  measurements similarly showed a high carrier concentration in samples B and C, and further SIMS data from sample C showed agreement with the ECV data from that sample. Electrical characterization showed depletion-mode HFET transistor performance with maximum drain current density of 22 mA/mm (sample B), which was lower than expected due to the high Ohmic contact resistance in the absence of  $n^+$  Ga<sub>2</sub>O<sub>3</sub> in the Ohmic regions. Simulated band diagrams indicated that carrier redistribution in the barrier layer and the unintentional Si at the interface contributed to the secondary channel observed in transconductance and capacitance-voltage data. A breakdown voltage of 990 V was measured on sample A, and several other measured devices did not exhibit breakdown voltage up to the 1100 V measurement limit. These results demonstrate the potential for ternary (AlGa)<sub>2</sub>O<sub>3</sub>/Ga<sub>2</sub>O<sub>3</sub> HFETs for high voltage power switching applications.

Homoeptitaxial films grown at high growth rates by MOCVD have been extensively characterized for their optical, structural and electrical properties. The films demonstrate relatively smooth surfaces with a high degree of structural order, limited point defectivity ( $N_d - N_a \approx 5 \times 10^{15} \text{ cm}^{-3}$ ) and an optical bandgap of 4.50 eV. Further, the present work reports for the first time a Ga<sub>2</sub>O<sub>3</sub> transistor device based on a high growth rate MOCVD technology. We demonstrate record channel mobility, measured by the Hall technique, made possible by the significant improvement in material quality, as discussed above, enabled by the MOCVD growth process. The high growth rate demonstrated via this method paves the road for demonstration of high breakdown voltage (VBK) devices on a thick Ga<sub>2</sub>O<sub>3</sub> buffer layer.

The first report of surface blistering He implanted (001) oriented  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> is demonstrated. The strain induced by the implant was modelled with a strain profile that matched neither the He concentration profile nor the DPA profile, but rather the overlap between the concentration and DPA profiles. The correlation between this overlap and the strain profile suggests the main contributor to the implant-induced strain is from He-vacancy complexes. This implant-induced distortion is elastic and removed upon annealing. Successful wafer bonding of (001)  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> to (0001) 4H-SiC using a thin  $\sim 5$  nm Ti interlayer is also demonstrated. While Ti has been used to bond Si, SiC, and diamond, bonding  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> using Ti has not yet been utilized. Initial results transferred up to  $\sim 200$   $\mu\text{m}$  wide  $\beta$ -Ga<sub>2</sub>O<sub>3</sub> to 4H-SiC. Further optimization of implant and anneal conditions are expected to increase transfer yield towards full-wafer-scale areas, as has been done with other orientations of  $\beta$ -Ga<sub>2</sub>O<sub>3</sub>.