ELECTRICAL CHARACTERIZATION OF ALD AL₂O₃/ALN/GAN HEMTS AND INTEGRATION WITH MICRO/NANO MAGNETS FOR NOVEL DEVICE APPLICATION

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Abstract

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The III-Nitrides are direct band gap semiconductors that span a large range of band gaps from ~0.6 eV (InN) to 6.2 eV (AlN). In recent years High Electron Mobility Transistors (HEMTs) based on III-Nitride semiconductors have proved capable of high power, high voltage and high frequency operation. Integration of ALD oxides with GaN will enable lower gate leakage currents, high breakdown voltages, and surface passivation of HEMTs. In this work we present a comprehensive characterization of AlN/GaN MOS-HEMT (grown by PAMBE) gate stacks with ALD Al$_2$O$_3$ of various thicknesses. Through capacitance-voltage and Hall-effect measurements, we find the presence and propose an origin of benign donor-type interface charge $Q_{it} \sim 6 \times 10^{13}$ cm$^{-2}$ at the AlN/Al$_2$O$_3$ junction, and relate its presence to the polarization charges in AlN. The role of ALD oxygen layers as possible modulation dopants can offer opportunities for various innovative designs in III-Nitride electronic devices. Most importantly by controlling the effective $Q_{it}$ (for example by compensation doping, or by varying the
polarization through the composition), pinch-off voltage $V_p$ can be made to increase, remain independent, or decrease with dielectric thickness $t_{ox}$.

There is a strong recent interest in studying the effect of magnetic interaction of various ferromagnetic structures leading the pathway for nanomagnet logic operation (NML). However, there is a very little effort so far to identify the potential effect of high current density of III-Nitride heterostructure on micro/nanomagnetic structures High-current drive nitride devices can potentially eliminate the need for auxiliary arrangements of switching and may enable the integration of logic and memory in the same device.

The successful integration of micro/nano magnets (supermalloy made) with III-Nitride heterostructure and the ability of imaging the single/multi-domain featured nano/micromagnets on III-Nitrides presented in this work opens the possibility of active device driven switching of nanomagnets in MQCA logic operation. Although simulation result has shown that the field requirement in nanomagnet switching is higher compared to the H field obtained (~1mT) using the drain current (1.4Amps/mm) of a normal HEMT, it is possible to use other nitride structures (n-GaN) to deliver current density more than 10amps/mm as shown in this work. Such a high current density produces a magnetic field in excess of 8mT. This field can be used affectively to move the domains of the micromagnetic structures, and possibly nanomagnets switching too.
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CHAPTER 1

INTRODUCTION

1.1 Motivation

The III-Nitrides are direct band gap semiconductors that span a large range of band gaps from ~0.6 eV (InN) to 6.2 eV (AlN). In recent years High Electron Mobility Transistors (HEMTs) based on III-Nitride semiconductors have proved capable of high power, high voltage and high frequency operation. However, gate leakage current limits the performance of these devices. To overcome this problem in this work we present a comprehensive study of AlN/GaN MOS-HEMT gate stacks with atomic layer deposited (ALD) Al$_2$O$_3$ as the gate dielectric. Moreover the high current carrying capability makes III-Nitride semiconductors a natural choice to be used as active device for switching nano/micro magnets in Magnetic quantum cellular automata (MQCA) logic operation. This work also explores that avenue.

1.2 ALD Al$_2$O$_3$ as the gate dielectric in HEMTs

III-Nitride semiconductors with wurtzite crystal structure show large built-in polarization fields with spontaneous and strain-induced piezoelectric components.
Figure 1.1: Wurzite crystal structure and the resulting polarization charges giving rise to polarization field in GaN [1].

The wurtzite crystal structure that gives rise to spontaneous polarization in III-Nitrides is shown in Figure 1.1. Large band offsets coupled with the highest possible spontaneous and strain induced polarization fields in AlN/GaN heterostructures have led to the formation of two-dimensional electron gas (2DEGs) with densities in excess of $1 \times 10^{13}$ cm$^{-2}$. 
However, the gate leakage current which limits the drain break down voltage limits the performances of these devices. Therefore, gate insulator dielectrics like $\text{Si}_3\text{N}_4$ have been studied [3, 4] in detail for the performance improvement of HEMTs. Only in recent past, atomic layer deposited (ALD) high band gap (6.5eV) [5], high $k$ (9.1) $\text{Al}_2\text{O}_3$ has drawn the attention of the community due to its excellent dielectric property coupled with thermal and chemical stability. In this work we present a comprehensive electrical characterization study of $\text{Al}_2\text{O}_3/\text{AlN/GaN MOS-HEMTs}$ and find the presence and propose an origin of benign donor-type interface charge ($Q_{it}$) at the AlN/$\text{Al}_2\text{O}_3$ junction, and relate its presence to the polarization charges in AlN. If cleverly exploited, this finding can enable various novel device applications.

1.3 Integration of Micro/nano magnet with III-Nitride Heterostructure

There had been increasing effort in nanomagnet logic operation, and consequently Magnetic quantum cellular automata (MQCA) [6,7] has drawn attention recently for novel logic operation. Passive external magnetic fields [7] and current carrying copper lines [8] have been used for the excitation of nanomagnets. It has not yet been investigated whether the current drive of active devices can influence the switching of micro/nano magnets. The choice of III-Nitrides comes naturally due to the high current drive [9] reported in HEMTs. However, the magnetic field originating from the current flowing in the 2DEGs of the HEMT devices is a few mili Tesla (mT), and inadequate for nanomagnet (75nm x 100nm x 30nm) switching (which requires ~100mT field), hence Object Oriented Micro Magnetic Framework (OOMMF) simulations have been
performed in this work to explore suitable micro-magnetic structures capable of showing irreversible changes at these low magnetic fields. Moreover other III-Nitride structures are investigated (apart from conventional HEMTs) in the quest of enhancing the magnetic field resulting from the flow of current in the channel, which might prove capable for switching the nanomagnets

1.4 This Work

In this work we have systematically analyzed the electrical characterization of ALD $\text{Al}_2\text{O}_3$/AlN/GaN MOS-HEMTs. The work also investigates the integration of nano/micro magnets made of supermalloy with ALD deposited III-Nitride structures and their possible applications. All the structures have been grown here using plasma assisted molecular beam epitaxy (PAMBE) in Veeco Gen 930 system. Chapter 2 outlines the MBE growth mechanism and the growth process of III-Nitride heterostructures for fabricating HEMT and other devices (for integration with nano/micro magnets) used in this work. The subsequent characterization of grown crystals using Reflection High Energy Electron Diffraction (RHEED), Atomic Force Microscopy (AFM) has also been described in Chapter 2. The working principle of Magnetic Force Microscope (MFM) for magnetic imaging and the Hall Effect system for electrical characterization have also been touched upon in chapter 2.

In chapter 3 we investigate the scaling properties of gate-stacks consisting ALD $\text{Al}_2\text{O}_3$/III-Nitride heterojunctions, and find interface charges that appear closely linked to the polarization charges of the underlying nitride substrate. Through capacitance-voltage
measurement on a series of samples of varying dielectric thicknesses, we find the presence, and propose an origin of benign donor-type interface charges ($Q_{it} \sim 6 \times 10^{13}$ cm$^{-2}$) at the AlN/Al$_2$O$_3$ junction. The polarization-related dielectric/AlN interface charge and the role of oxygen in the dielectric as a possible modulation dopant potentially offer opportunities for various device applications.

In chapter 4 we investigate the effect of high current drive in HEMTs on various micro/nanomagnetic structures. In this chapter first we briefly describe the NML logic operation followed by the description of the difference between single and multi-domain ferromagnetic structures. Next we explore the possibility of integration and the effect on micro/nanomagnetic structures by the magnetic field created from the HEMT device current. Object Oriented Micromagnetic Framework (OOMMF) simulation results are then presented, followed by the experimental efforts thus far in this path of III-Nitride device-current driven micro/nanomagnetic transitions. This work is novel in the sense it can also open up the possibility of hybrid logic (HEMT + MQCA) technology in the same device architectural frame work.

In chapter 5 we conclude and outline the future direction of this work.

1.5 References


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CHAPTER 2

MBE GROWTH AND CHARACTERIZATION OF GROWN III-NITRIDE HETEROSTRUCTURES BY MBE

This chapter describes the historical development and growth mechanism of epitaxial thin films by Molecular Beam Epitaxy (MBE). The MBE growth process of III-Nitride heterostructures for fabricating HEMTs and other devices (for integration with Nano/Micro magnets) used in this work is outlined in this chapter with a brief overview of the Gen 930 MBE system used for this purpose. Finally a detailed description of the tools and methods used to characterize the MBE grown structures during and after growth is provided.

2.1 History behind the Development of MBE Technique

There are various methods for the epitaxial growth of thin films, such as Liquid Phase Epitaxy (LPE), Solid Phase Epitaxy, Metalorganic Vapor Phase Epitaxy (MOVPE), Halide Vapor Phase Epitaxy (HVPE), and Molecular Beam Epitaxy (MBE). Among them MBE is one of the most advanced thin film deposition techniques, pioneered in the late 1960s by A. Y. Cho and J. R. Arthur [1] at Bell Labs. Before the advent of MBE, thin films were grown in vacuum by thermally evaporating single
elements. But thin film growth of compound semiconductors was difficult to achieve due to the inherent problem of different vapor pressures of the constituent materials at same temperature. Günther [2] was the first to establish the basic ideas for one of the key requirements of MBE, i.e. the deposition of stoichiometric films of III-V semiconductors from molecular beams of the constituent elements impinging onto a heated substrate. In his experiment Günther was able to grow stoichiometric films of InAs on polycrystalline glass substrates. Later, Arthur and Cho developed Günther’s method into the MBE technology to grow epitaxial III-V thin GaAs films on heated crystalline substrates in UHV conditions. Since then many materials have been grown in the thin-film form epitaxially by MBE. Semiconductors of the III-V group have received the most attention because of their superior high frequency and optical properties as compared to Si and the need for precise compositional control in heterostructures.

The use of ultra high purity sources, low growth temperature and ultra high vacuum (UHV) environment results in the following advantages of MBE over other material growth techniques:

- Superior control over layer thicknesses, due to low growth rate (<1ML/s) and low inter-diffusion effects, and
- In-situ and real-time characterization by Reflection High-Energy Electron Diffraction (RHEED) and Auger Electron Spectroscopy (AES).

2.2 Epitaxial Growth Mechanism by MBE

Epitaxy is defined as the deposition of a monocrystalline film on a monocrystalline substrate while maintaining a certain lattice alignment. The epitaxial growth process
does not only depend on the thermodynamic properties of the surface, but also on the kinetics of adatom. The epitaxial growth in MBE on the substrate surface can be divided into a few stages as given below:

- **Adsorption**: The impinging atoms on the substrate surface get adsorbed and at this point they adhere to the surface by weak van der Waals force. No chemical bond has formed at this stage.

- **Diffusion**: After the atoms are adsorbed they diffuse on the surface at high substrate temperatures.

- **Incorporation**: During diffusion atoms finally get incorporated into the crystal by forming a chemical bond with the surface atoms.

- **Desorption**: During diffusion atoms can also desorb from the surface due to high substrate temperature. Although diffusion is aided by high substrate temperatures, too high a substrate temperature results in higher energy for atoms and might also lead to desorption and dissociation of the atoms from the surface.

![Figure 2.1: Schematic showing different atomistic processes that occur on surface during growth [3].](image-url)
In general crystal growth can occur in three basic modes namely: Volmer-Weber (VW), Frank-van der Merwe (FM) and Stranski-Krastanov (SK) growth mode. In the VW growth mode the interaction between adatoms is stronger than that between adatom and surface and as a result small clusters get nucleated directly on the substrate surface and then grow into small islands, leading to 3D island growth. This mode of growth is displayed by many metals growing on insulators. FM mode on the other hand displays the opposite characteristic where adatoms attach preferentially to surface sites leading to smooth thin-film formation. A new layer doesn’t grow until the previous layer has completely formed resulting in a 2D layer by layer growth. This form of growth mode is generally common in lattice matched systems. SK growth mode is a combination of VW and FM modes where both two dimensional (2D) layer by layer growths as well as three dimensional (3D) island growth takes place. Schematics of these three growth modes are shown in Figure 2.2.
2.3 PAMBE Growth of III-Nitrides

Since the growth of MBE GaN crystals by Akasaki [4] back in 70’s there have
been lots of efforts to grow high quality III-Nitrides on various substrates such as
sapphire, SiC, Si, bulk GaN etc. However, it was only in early 90s when GaN growth
improved drastically. Today MBE is used extensively in optical industry (for LEDs and
lasers), as well as for growing III-Nitride HEMTs. In our research group, room
temperature mobility of more than 1500 cm$^2$/V.s has been reported for MBE grown
AlN/GaN HEMT structures. Figure 2.3 shows the schematic of a typical VEECO Gen
930 MBE system used in this work for the growth of III-Nitride HEMT structures. The
growth process involves the following. After Acetone +Methanol+ Isopropanol cleaning
the substrate is first loaded into the load-lock chamber where it is baked overnight at
200°C to remove water vapor and gases adsorbed on the surface. Then it is transferred
into a buffer chamber where it is baked for a few hours at a higher temperature of ~450°C
for further outgassing. Thereafter, it is transferred to the growth chamber. The growth
chamber is maintained under Ultra High Vacuum (UHV) ($10^{-9} - 10^{-10}$ Torr) by the use of
a cryogenic pump. Effusion cells containing ultra high purity constituent elements: Ga,
In, Al (Group III elements), Si, Mg (dopants) are heated using resistive heaters and the
material flux is controlled accurately by keeping the cell temperature stable.
RF plasma is used to produce active N atoms from inert Nitrogen (N\textsubscript{2}) gas molecules used as group V element. By measuring the pressure difference (using ion gauge) in the growth chamber while opening and closing the cell shutter, the Beam Flux Monitor (BFM) determines the flux of the source material being evaporated. In Figure 2.3 the Continuous Azimuthal Rotation Stage (C.A.R.) is shown which contains the substrate holder, heater and the filament. Before growth, the C.A.R. is rotated such that the BFM faces the molecular beam. By controlling the cell temperature the desired flux of the constituent growing material is determined. Then by controlling the substrate heater, the desired substrate temperature is set.
Figure 2.4: Phase diagram of Effective Ga flux vs Substrate temperature at RF=275W (developed in Notre Dame). AFM scan (2µm)² of Ga and N-rich grown samples have also been shown. (Courtesy John Simon)

During growth, a Reflection High Energy Electron Diffraction (RHEED) electron gun along with a fluorescent screen and a RHEED camera which are installed in the growth chamber are used for in-situ monitoring of the growth rate and surface morphology (details provided later). Throughout the growth liquid nitrogen flows through a cryogenic panel to cool down (77k) the growth chamber walls. This prevents impurities such as oxygen from entering the growing crystals by effectively ‘condensing’ on the walls.

Of the two commonly used substrates (SiC, sapphire) for GaN growth, sapphire (with a higher lattice mismatch with GaN compared to SiC) is used most frequently due
to lower cost. For this work semi-insulating (S.I.) GaN templates grown by MOCVD on top of sapphire were used. There are three growth regimes for III-Nitrides: metal rich, intermediate and nitrogen rich. Figure 2.4 shown above captures the phase diagram [5] of GaN growth in three different regimes at a fixed plasma power (275W). In the nitrogen rich regime, the nitrogen flux reaching the substrate surface is more than that of the metal flux. As shown in Figure 2.4 in this case a rough surface results which is confirmed during growth by spotty RHEED pattern (shown later). When the III/V ratio is made ~1, the surface morphology improves & RHEED becomes streakier. Under this condition a thin layer of Ga wetting layer covers the whole surface and helps the diffusion of N atoms to the surface and enhances the uniform incorporation. However, as shown in the Figure 2.4, the intermediate regime is very narrow and it is difficult to control the metal flux and substrate temperature precisely to stay in this regime. The metal-rich growth regime (III/V ratio is made > ~1), results in metal droplets on the surface after growth as observed through optical microscope in Figure 2.5(a). The metal droplets are subsequently removed post-growth by etching in HCl. It is easy to remove Ga but comparatively difficult to remove Al droplets. As confirmed in Figure 2.5, since metal rich growth conditions lead to smoother surface morphologies with less defect density it is the regime of choice for all the growths performed in this work.
Figure 2.5(a): Ga droplets after growth, before removal (HCL); (b): AFM scan (2µm)$^2$ showing smooth GaN surface of the sample after metal rich growth; (c): streaky RHEED image after growth.
Next we discuss the various characterization tools used during and after growth.

2.4 Reflection High Energy Electron Diffraction (RHEED)

For in-situ analysis of the surface morphology during crystal growth the Reflection high-energy electron diffraction (RHEED) technique [6] is very useful. A RHEED system consists of an electron gun, a detector screen and the sample to be analyzed. The beam of electrons generated from the gun hits the sample at a grazing angle. The resulting diffracted electrons from the sample surface, which interfere constructively, form a pattern on the phosphor detector screen. Figure 2.6 presented below shows the basic set up of a RHEED system. The formation of RHEED diffraction spots require the reciprocal lattice vector \( G \) to satisfy the condition: \( k_i - k_f = G \) (Bragg diffraction condition) where \( k_f \) and \( k_i \) are the wave vectors of the diffracted and incident beams.

![Figure 2.6: Schematic diagram of RHEED geometry showing the incident beam hitting the sample surface and creating REED pattern on the screen [7].](image-url)
The conditions for diffraction in elastic scattering requires $|k_i| = |k_f|$, which is provided by the Ewald sphere construction, depicted in Figure 2.6. The reciprocal lattice of a two-dimensional plane is a lattice of infinitely thin rods, perpendicular to the surface. For the diffraction pattern to be produced by elastic scattering the reciprocal lattice rods need to intersect the Ewald sphere (radius $1/\lambda$) and the intersection points will produce RHEED spot patterns on the screen. The wavelength $\lambda$ of an electron corresponding to an accelerating voltage $V$ is given to a good approximation by $\lambda = 12.247/\sqrt{V(1+10^{-6}V)}$.

Now the radius of the Ewald sphere for electron with energy 5keV is calculated using the above expression as $\sim 37\text{Å}^{-1}$. For GaN surface with lattice constant (a) 3.19 Å the distance between adjacent rods in reciprocal space will be 1.9 Å$^{-1}$. As the radius of the Ewald sphere is very much larger than the separation of the rods, hence the rods intersect the sphere tangentially (for smooth 2-D surface), resulting in a streaky, rather than a spotty appearance. For growth of III-Nitride by MBE, metal rich growth conditions result in smooth surfaces. Hence following the argument above; it should result in a streaky RHEED pattern. Figure 2.7(a) shows the RHEED pattern of a GaN surface grown under such conditions. In case of rough surfaces, the rods intersect the sphere obliquely and result in spots rather than streaks. Hence, the N-rich growth condition which typically results in rough surfaces should display spotty RHEED, and indeed it is observed as shown in the Figure 2.7(b).
Figure 2.7(a): Streaky RHEED pattern characteristic of Ga-rich GaN surface; (b): Spotty RHEED pattern characteristic of N-rich GaN surface.

The growth dynamics of MBE is well understood by monitoring the temporal variations in the intensity of a specular RHEED spot [9]. In addition to providing calibration of beam fluxes, determination of alloy composition and surface morphology, RHEED oscillations can be used for the determination of the growth rate in real time. The number of oscillation peaks in the intensity curve relates to the number of monolayer (For GaN one monolayer = 2.6nm) grown. Figure 2.8 shows that one monolayer is grown during the time between two successive peaks of RHEED oscillation. Hence growth rate can be obtained by studying Rheed intensity vs time plot as shown in Figure 2.8.
2.5 Atomic/Magnetic Force Microscopy (AFM/MFM)

Gerd Binning, Quate and Gerber developed the first atomic force microscope (AFM) back in 1986 [10]. The AFM uses a cantilever with a sharp tip that physically touches the sample surface.

Figure 2.8: RHEED diffraction Intensity vs time curve as the layer grows.

Figure 2.9: Schematic illustration of Atomic Force Microscopy (AFM) setup [11].
The cantilever scans the sample while its oscillation amplitude is measured via a segmented photodetector which tracks the reflection of a laser from the back of the cantilever. This is shown schematically in Figure 2.9 [11]. The change in the cantilever oscillation amplitude is correlated to the sample surface topography, and thus the topographic image is captured by recording these changes and closing the z feedback loop. In this work tapping mode AFM has been used to scan the surface of grown III-Nitride crystals and gives the root mean square roughness which is a metric of crystal growth quality.

When the AFM technique is used to capture magnetic information of a sample it is called MFM (magnetic force microscopy). In ‘Lift Mode’ first an AFM scan is performed to extract the surface topography of the sample. Then the tip is lifted and maintained at a fixed height above the sample. The magnetic force between the area under the scan and the magnetic tip influences the resonance frequency of the tip. The change in resonance frequency is detected by the phase change. Figure 2.10 shows the schematic and the working principle of a MFM setup. In this work standard MESP MFM probe tips (from VEECO) containing magnetic coating of Co/Cr alloy with cantilever length of 225µm and spring constant \((k)\) of 1N/m were used for MFM imaging.

Initially, the cantilever vibrates at its resonance frequency \(\omega_0 (=\sqrt{\left(k_0/m\right)})\), where \(m\) is the effective mass and \(k_0\) is the natural spring constant of the cantilever. When there is an interaction between the tip and the sample, the spring constant of the cantilever gets modified from its natural spring constant \(k_0\) as:
\[ k = k_0 - \frac{\partial F}{\partial z} \] (2.1)

Here \( \frac{\partial F}{\partial z} \) is the derivative of the interaction force with respect to the perpendicular coordinate \( z \). Thus, depending on the interaction of the sample and the probe tip the resonance frequency of the cantilever changes. This in turn modulates the phase of the cantilever and thus maps the magnetic information of the sample.

2.6 Electrical Characterization by Hall Effect Measurement

For extracting the carrier mobility, carrier type and carrier concentration, Hall Effect measurement in van der Pauw geometry is performed on the samples grown in this work. After MBE growth and HCl cleaning, four Indium dots are mounted on the four corners of the sample (as shown in the Figure 2.11). By applying a magnetic field \( (B) \) perpendicular to the direction of injected electrical current \( (I) \), the transverse Hall voltage

Figure 2.10: MFM Lift mode scanning principles [12]. (1&2) Cantilever traces surface topography on first trace and retrace. (3) Cantilever ascends to the lift scan height. (4&5) Lifted cantilever profiles topography while responding to magnetic influences on second trace and retrace.
($V_H$) is measured. To measure the Hall voltage $V_H$, a current $I_{13}$ is forced through the opposing pair of contacts 1 and 3 and the Hall voltage $V_H$ ($= V_{24}$) is measured across the remaining pair of contacts 2 and 4. Once the Hall voltage $V_H$ is acquired, the sheet carrier density $n_s$ can be calculated using $n_s = IB/q|V_H|$ from the known values of $I$, $B$, and $q$. The Hall mobility ($\mu_H$) can then be extracted from the expressions given below:

$$\mu_H = \frac{V_H}{R_{sh}IB}$$

Where $R_{sh}$ is the sheet resistance of the sample.

Figure 2.11: Hall measurement technique in van der Pauw geometry.

The characterization tools namely RHEED, AFM, Hall system described in this chapter have been used for electrical and surface morphology characterization of MBE grown samples discussed in the next chapter. The MFM imaging technique outlined here has been utilized for acquiring magnetic information of the samples described in chapter 4. The next chapter highlights the electrical characterization and interface (ALD/AlN) analysis of the AlN/GaN HEMT structures grown by PAMBE.
References


CHAPTER 3

BARRIER HEIGHT, INTERFACE CHARGE & TUNNELING EFFECTIVE MASS IN ALD Al₂O₃/AlN/GaN HEMTs

Atomic layer deposited (ALD) high band gap (~6.5eV), high $k$ (~9.1) Al₂O₃ has emerged as an attractive candidate to support vertical scaling of AlN/GaN HEMTs and its variants owing to its outstanding dielectric, thermal, and chemical properties. Integration of ALD oxides with GaN will enable lower gate leakage currents, high breakdown voltages, and surface passivation of HEMTs. In this work we present a comprehensive characterization of AlN/GaN MOS-HEMT gate stacks with ALD Al₂O₃ of various thicknesses. Through capacitance-voltage and Hall-effect measurements, we find the presence and propose an origin of benign donor-type interface charge ($Q_{it}$) at the AlN/Al₂O₃ junction, and relate its presence to the polarization charges in AlN. By studying tunneling transport in corresponding (Ni/Al₂O₃/Ni) M-I-M diodes, we extract the Ni/Al₂O₃ surface barrier height ($\Phi_B$), the electron tunneling effective mass in Al₂O₃.
3.1 Background Research

Since late 80’s aggressive scaling of transistors have paved the pathway for meeting several technological requirements of CMOS IC industry, namely high speed and low static power dissipation. However, persistent miniaturization of the MOS transistor creates the demand of ever thinner gate oxides. Although thinner gate oxides can provide higher gate capacitance \( C_{ox} \sim k / t_{ox} \) which in turn helps boost transistor speed, it presents an ineffective barrier between gate and channel which causes higher gate leakage current due to direct tunneling. Hence as an alternative to SiO\(_2\), high-k dielectrics like HfO\(_2\) [1] ZrO\(_2\) [2] and Al\(_2\)O\(_3\) [1] have been studied in detail for future generation Si CMOS ICs. Gate insulators like SiO\(_2\) [3], Si\(_3\)N\(_4\) [4,5] HfO\(_2\) [6] and Al\(_2\)O\(_3\) [7] have been investigated for wide gap III-V Nitrides for the reduction of gate leakage current which limits the drain break down voltage and hence the performance of HEMT devices. The passivation of surface traps in GaN based devices is also crucial for high power, high frequency operation. Hence insulators such as PECVD Si\(_3\)N\(_4\) have been used to suppress current collapse [8] in these devices. However, only recently atomic layer deposited (ALD) Al\(_2\)O\(_3\) has drawn the attention of the community due to its outstanding dielectric [9] and passivation [10] property. The thickness control of the ALD film for vertical scaling makes it a superior choice over PECVD Si\(_3\)N\(_4\) and SiO\(_2\). The superior quality (in terms of uniformity) of ALD over sputtering and electron-beam deposition, coupled with high band gap (~6.5eV) [11], high dielectric constant (~9.1), high break down field (~10MV/cm), high thermal (amorphous ~1000°C) and chemical stability of Al\(_2\)O\(_3\) makes ALD Al\(_2\)O\(_3\) a natural choice as a gate insulator for AlN/GaN
HEMTs [12] and its variants. The study of ALD Al\(_2\)O\(_3\) and III-V Nitride interface is of prime importance for understanding and designing the device characteristics of AlN/GaN HEMTs. Till date there are few reports [7, 13] of thickness-dependent electrical characterization study of sputter coated Al\(_2\)O\(_3\) on AlGaN/GaN HEMTs. To fit the experimentally measured 2DEG density, in [13], negative oxide interface charge and in [7] positive oxide interface charge has been assumed. However, no justification has been given for these assumptions. In this chapter we present a comprehensive characterization of AlN/GaN MOS-HEMT gate stacks with ALD Al\(_2\)O\(_3\) of various thicknesses. Through capacitance-voltage and Hall-effect measurements, we find the presence and propose an origin of benign donor-type positive interface charge \((Q_{it})\) at the AlN/Al\(_2\)O\(_3\) junction, and relate its presence to the polarization charges in AlN. The presence of \(Q_{it}\) explains the trend of pinch-off voltage and 2DEG density with ALD thicknesses both qualitatively and quantitatively.

3.2 Experimental Procedure

For this study the AlN/GaN HEMT structures were grown in a Veeco Gen 930 MBE system on semi insulating 0001 GaN templates (2\(\mu\)m) on sapphire under metal rich conditions. Ga flux of ~2.1 \(\times\) 10\(^{-7}\) Torr, Al flux of ~1.6 \(\times\) 10\(^{-7}\) Torr and N\(_2\) supplied from a Veeco rf source at a plasma power of 240W (Pressure : 2 \(\times\) 10\(^{-5}\) T) were used. The MBE layer structure consisted of a thin 1.5nm AlN nucleation layer (to eliminate buffer leakage [14]), followed by 240nm UID GaN, and \(t_{AlN} = 4\)nm AlN barrier layer grown at a substrate thermocouple temperature of 660\(^{\circ}\)C. Mesa isolation was achieved with
BCl₃/Cl₂ plasma reactive ion etching, followed by source/drain ohmic metallization using Ti/Al/Ni/Au (15/120/40/60 nm) stack deposition followed by rapid thermal annealing in N₂/Ar atmosphere for 30sec at 600°C. A saturation current of 1.3 Amps/mm was measured. The same sample was then cleaved into four parts, and four different Al₂O₃ thicknesses (t_ox = 2nm, 4nm, 6nm, 8nm) were deposited on the AlN surface by ALD with Tri Methyl Aluminum (TMA) and H₂O as the precursors at 200°C. Finally, Ni/Au (50/150 nm) gate metal stacks were deposited simultaneously on the 4 samples. The layer structure of the sample is shown in Figure 3.1(a). Figure 3.1(b) & (c) shows the cross-section TEM image of gate stack of the (t_ox = 4nm) sample, confirming the thicknesses, and shows thickness variations within an acceptable window.

Figure 3.1(a): Schematic layer structure of the sample; (b): High Resolution Transmission Electron Microscope image along zone axis [100] showing the layer structure of the sample; High-resolution lattice image of the gate stack (inset) showing the crystalline AlN barrier layer, the amorphous ALD Al₂O₃ and Ni as gate metal.
The C-V characteristics measured at 1 MHz on circular diode patterns of area $A=\pi \times (10\mu\text{m})^2$ on the four samples with different $t_{\text{ox}}$ shows negligible hysteresis. As shown in Figure 3.2(a) the pinch-off voltage $V_p$ increased with the thickness of the ALD oxide layer, from $V_p = -3.6$ V for $t_{\text{ox}} = 2\text{nm}$ to $V_p = -8.8$ V for $t_{\text{ox}} = 8\text{nm}$. The carrier profile $n(z)$ is extracted [15] from the C-V measurement and using $n(z) = (C^3 / q\varepsilon_s)(dC / dV)^{-1}$.

Figure 3.2(b) indicates the varying depth of the 2DEG channel from the gate metal.

Figure 3.3 shows the decreasing gate leakage current with increasing $t_{\text{ox}}$, the property that will enable high breakdown voltages in AlN/GaN HEMTs.

Figure 3.2 (a): C-V plots (@1MHz) of the four samples with different $t_{\text{ox}}$, showing $V_p$ increases with increasing $t_{\text{ox}}$; (b) Plots showing the charge profile and varying depth of 2DEG channel from the gate metal for samples with different $t_{\text{ox}}$. 

![C-V plots](image1.png)
Figure 3.3: Gate leakage current density vs. gate voltage plots indicating the reduction of $J_G$ with increasing $t_{ox}$.

3.3 ALD/AlN Interface

To explain the increase in $V_p$ with $t_{ox}$ for the structures we investigate all possible sources of fixed/mobile charges and associated energy band diagrams. Apart from fixed polarization charges a fixed positive sheet charge ($Q_o$) at the ALD/AlN interface (justification given later) and interface dipole charges [16] ($+\sigma_o/-\sigma_o$) arising due to the electronegativity difference between Al and O atoms at the Al$_2$O$_3$/AlN interface have been considered.
Figure 3.4: Charge distribution and simulated band diagram of Ni/Al₂O₃/AlN/GaN at $V_G=0$ (w and w/o $Q_{it}$)

Figure 3.4 shows the charge and energy-band diagram simulated using a self-consistent 1-D Poisson-Schrödinger solver [17] for the structure Ni/Al₂O₃/AlN/GaN (50/2/4/240 nm) at $V_G=0$V with and without Al₂O₃/AlN interface oxide charges.

At pinch-off condition ($V_G=V_p$) the charge and energy band diagram have been shown in Figure 3.4. From this figure, the generalized expression for $V_p$ can be written as:
\[ qV_p = -\phi_B + (qV_{AlN}^p + qV_{ox}^p + qV_0) + (\Delta E_{c}^{AlN/GaN} - \Delta E_{c}^{ox/AlN}) \]  

(3.1)

Where \( V_{AlN}^p, V_{ox}^p, V_0 \) are expressed as:

\[ V_{AlN}^p = qQ_{P(AlN/GaN)}(t_{AlN} / \varepsilon_{AlN}) \]  

(3.2)

\[ V_{ox}^p = q(Q_{it} - Q_{P(GaN)})(t_{ox} / \varepsilon_{ox}) \]  

(3.3)

\[ V_0 = q(\sigma_0)(t_0 / \varepsilon_{ox}) \]  

(3.4)

Here \( Q_{P(AlN/GaN)} \) (~6.1x10\(^{13}\) cm\(^{-2}\)) is the polarization charge density at the AlN/GaN interface, \( Q_{P(GaN)} \) (~1.8x10\(^{13}\) cm\(^{-2}\)) is the total polarization charge density of GaN, \( t_{AlN} \) is the thickness of AlN, \( t_{ox} \) is the thickness of the ALD oxide layer, \( \varepsilon_{AlN} \) (~9.1 \( \varepsilon_0 \)) is the dielectric constant of AlN, \( \varepsilon_{ox} \) (~9.1 \( \varepsilon_0 \)) is the dielectric constant of ALD oxide, \( \Delta E_{c}^{AlN/GaN} \) (~1.9eV) is the conduction band offset between AlN/GaN and \( \Delta E_{c}^{ox/AlN} \) (~1.3eV) [18, 19] is the conduction band offset between Al\(_2\)O\(_3\)/AlN. \( \sigma_0 \) is the dipolar sheet charge density (~6x10\(^{14}\) cm\(^{-2}\)) [16], \( t_0 \) is the separation between the two dipole sheet charges which is less than the sum of atomic radii of the constituent atoms (~1Å). The Ni/Al\(_2\)O\(_3\) surface barrier height used is the extracted value \( \Phi_B = 2.9\) eV (described in the next section). From expression (3.1) and the energy band diagram presented in Figure 3.5, it is evident that when \( Q_{it} \) and \( \sigma_0 \) are assumed to be zero then with the increase of oxide thickness, \( V_p \) decreases. This contradicts the experimental observation presented in Figure 3.2(a). However, when \( Q_{it} \sim 6x10^{13}\) cm\(^{-2}\) is assumed at the (Al\(_2\)O\(_3\)/AlN) interface then \( V_p \) increases in accordance with the experimental findings as
shown in Figure 3.4. On the other hand if interface charge equals the GaN polarization charges i.e. if \( Q_{it} = Q_0 = Q_{P(GaN)} \) is assumed, then from the expression of \( V_{ox}^p \)

\[
V_{ox}^p = q(Q_{it} - Q_{P(GaN)})(t_{ox} / \varepsilon_{ox})
\]

it is seen that \( V_p \) should not change with \( t_{ox} \), the simulation result in Figure 3.5 shows this expected behavior.

Figure 3.5: Charge distribution and simulated band diagram of Ni/Al₂O₃/AlN/GaN at \( V_G = V_p \) for different \( Q_{it} \).
The increase in the experimental $V_p$ with $t_{ox}$ is quantitatively justified (with $Q_{it} \approx 6 \times 10^{13}$ cm$^{-2}$) by using Self-consistent 1-D Poisson-Schrödinger simulation as shown in Figure 3.6. The interface density $Q_{it} \approx 6 \times 10^{13}$ cm$^{-2}$ is remarkably close to the surface polarization charge of a strained AlN layer. We argue that since the AlN surface is metal (Al)-face, in the first steps of ALD growth, oxygen atoms attach to Al and can be viewed as substituting the nitrogen site [20]. Thus oxygen atoms act as donor dopants by electron counting rules. The picture is essentially identical to modulation doping: the positive sheet charge at the Al$_2$O$_3$/AlN interface (inset, Figure 3.6) neutralizes negative polarization charges of the AlN surface, increasing the 2DEG density at the AlN/GaN heterojunction.

Figure 3.6: Experimental and simulated (w and w/o $Q_{it}$) $V_p$ for different $t_{ox}$. The atomic arrangement at the Al$_2$O$_3$/AlN interface and the positive donor dopants giving rise to $Q_{it}$ are shown in the inset.
Figure 3.7 shows that the increase in the experimental 2DEG density $n_s$ (from Capacitance Voltage measurement) with $t_{ox}$ can be explained if $Q_{it} \sim 6 \times 10^{13} \text{cm}^{-2}$ is assumed. The addition of dipolar sheet charges in this model will cause an increase of $V_0$ (~1V) irrespective of the $t_{ox}$ values, in the calculation of $V_p$ as shown in the inset of Figure 3.5. However, it should be noted that dipolar sheet charges are crystallographic orientation dependent [21] and hence would not effect $V_p$ if III-Nitrides are grown other than [001] direction.

![Figure 3.7: Experimental and simulated (w and w/o $Q_{it}$) $n_s$ for different $t_{ox}$.](image)
The finding of interface (ALD/AlN) charge ($Q_{it} \sim 6 \times 10^{13} \text{cm}^{-2}$) is verified with another AlN/GaN structure with a thinner AlN barrier. Figure 3.8 shows the Hall-effect measured charge before and after ALD deposition ($t_{ox}=5\text{nm}$). Charge density loci predicted by Poisson-Schrödinger simulations for various $Q_{it}$ (after ALD) and $\Phi_B$ are shown in Figure 3.8, confirming that $Q_{it} = 6 \times 10^{13} \text{ cm}^{-2}$ and $\Phi_B = 2.9 \text{eV}$ match the experimental data.

![Figure 3.8: Simulation data plot showing the variation of $n_s$ with $\Phi_B$ before and after 5nm ALD (for various $Q_{it}$). The Hall measured $n_s$ before and after ALD ($t_{ox}=5\text{nm}$) are compared with the simulation result.](image)
3.4 Ni/ALD Al₂O₃ Interface

For this study, four semi-insulating-GaN samples were used. 150nm Ni was deposited on them, then four different Al₂O₃ thicknesses (8nm, 9nm, 10nm, 12nm) were deposited on the Ni surfaces by ALD (with TMA and H₂O as the precursors at 200°C). Finally 150nm Ni was deposited on top. The layer structure of the sample is shown in Figure 3.9 (inset). Figure 3.9 shows the measured tunneling current density (J) vs. the oxide electric field (Eox) for the four samples. The current density decreases with increasing Al₂O₃ thickness as a result of decrease in the tunneling current.

Figure 3.9: J versus Eox for the M-I-M structures with different tox. Inset: Schematic layer structure of the M-I-M diode.
For the M-I-M structure, when the applied voltage is high compared to the surface barrier height \((qV > \Phi_B)\), the Fowler-Nordheim (FN) tunneling takes place and the tunneling current density through a triangular barrier can be expressed as \([22]\)

\[ J = K_1 E_{ox}^2 \exp(-\frac{K_2}{E_{ox}}) \tag{3.5} \]

Where \( E_{ox} = V / t_{ox} \), \( K_1 = 1.54 \times 10^{-6} \left(1 / m^* \phi_B \right) \) and \( K_2 = 6.92 \times 10^7 (m^*)^{1/2} (\phi_B)^{3/2} \)

The expression can be written as: \( \ln(J / E_{ox}^2) = \ln K_1 - k_2 / E_{ox} \). From the plot of \( \ln(J / E_{ox}^2) \) vs \( 1/E_{ox} \), the slope can be extracted which is \( -K_2 \). From the expression of \( K_2 \) the tunneling electron effective mass \( (m^*) \) and \( (\Phi_B) \) are extracted. Figure 3.10 shows the Fowler-Nordheim plot of \( \ln(J / E_{ox}^2) \) vs \( 1/E_{ox} \) from the experimental data shown in Figure 3.9. The schematic band-diagram at high bias voltage \((>\Phi_B)\) is shown in the inset of Figure 3.10. The Table 3.1 presents the extracted effective mass and surface barrier heights for various oxide thicknesses. The extracted value of tunneling electron effective mass \( (~0.16m_0) \) is quite reasonable when compared to the earlier reports \([19]\). The extracted surface barrier height \( (\Phi_B) \approx 2.9eV \) is also consistent with the reported electron affinity difference between Ni \( (\Phi_{Ni} \approx 5.4eV) \) \([23]\) and ALD Al\(_2\)O\(_3\) \( (\chi_{Al2O3} \approx 2.5eV) \) \([19]\).
Figure 3.10: FN plots for the samples with $t_{ox}=10$nm, 12nm. The expression in the inset is used to extract $\Phi_B$ & $m_T^*$ from the slope.
### TABLE 3.1
SHOWING THE EXTRACTED $\Phi_B$ AND $m^*$ FROM THE FN PLOT FOR VARIOUS ALD Al$_2$O$_3$ THICKNESSES

<table>
<thead>
<tr>
<th>Oxide thickness (nm)</th>
<th>Extracted $\Phi_B$ (eV)</th>
<th>Fitted effective mass ($m^*$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>10</td>
<td>2.7</td>
<td>0.16</td>
</tr>
<tr>
<td>12</td>
<td>2.9</td>
<td>0.16</td>
</tr>
</tbody>
</table>

3.5 Conclusion

The extraction of the polarization-related ALD/AlN interface charge, the Ni/ALD barrier height, and the tunneling effective mass of electrons through ALD Al$_2$O$_3$ into AlN/GaN heterojunctions reported here is expected to accelerate the choice of optimal gate stacks for nitride HEMTs. Our result has indicated that the negative polarization charge at the AlN surface is effectively compensated by a fixed positive charge at the ALD/AlN interface. The electrostatic model presented here predicts the influence of the ALD Al$_2$O$_3$ thickness on the device characteristics like 2DEG density and pinch-off voltage. The extraction of the polarization-related ALD/AlN interface charge and the role of ALD oxygen layers as possible modulation dopants offer opportunities for novel designs in III-Nitride electronic devices. Recently it has been demonstrated [24] that the deposition of ALD Al$_2$O$_3$ /Si$_3$N$_4$ on subcritical AlN(<2nm) /GaN structure can induce 2DEG ~ 1x10$^{13}$ cm$^{-2}$ and can be used for E/D mode HEMTs. As the interface roughness
scattering is inherently low [14] in thinner barrier GaN HEMTs, the ALD/AlN interface properties presented here can indeed accelerate the subcritical AlN or AlGaN barrier GaN HEMTs research. In addition, the analysis presented here is based on Al face AlN/ALD Al$_2$O$_3$ interface. The interface properties of N-face GaN HEMTs with ALD Al$_2$O$_3$ should find similarities.

The high current carrying capability of III-Nitride HEMTs reported here, makes them a natural choice to be used as active device for switching nano/micro magnets in MQCA logic operation. Next chapter investigates the possibility of integration of various nano/micro magnetic structures with III-Nitride HEMTs and its variants.

3.6 References


CHAPTER 4

INTEGRATION OF MICRO/NANOMAGNET WITH III-NITRIDE HETEROSTRUCTURE

The high 2DEG density as discussed in the last chapter in III-Nitride heterostructure is useful for various device applications. In this chapter we investigate the effect of high current drive in HEMTs on various micro/nanomagnetic structures. There is a strong recent interest in studying the effect of magnetic interaction of various ferromagnetic structures leading the pathway for nanomagnet logic operation (NML).

However, there is a very little effort so far to identify the potential effect of high current density of III-Nitride heterostructure on micro/nanomagnetic structures. High-current drive nitride devices can potentially eliminate the need for auxiliary arrangements of switching and may enable the integration of logic and memory in the same device. In this chapter first we briefly describe the NML logic operation and motivation of the work followed by the description of the difference between single and multi-domain ferromagnetic structures. Then we explore the possibility of integration and the effect on micro/nanomagnetic structures by the magnetic field created from the HEMT device current. Object Oriented Micromagnetic Framework (OOMMF) simulation results are
then described, followed by the experimental efforts thus far in this path of semiconductor device driven micro/nanomagnetic transitions.

4.1 MQCA Logic: Motivation

Moore predicted that the packing density of IC chips would double every two years [1], which has more or less remained true for the last fifty years. However, it seems that any further reduction in device size will generate quantum mechanical effects that hinder device operation. With an eye toward the demise of the transistor scaling paradigm, many molecular sized devices are being investigated for their feasibility for electronic logic. Among them quantum-dot cellular automaton (QCA) [2, 3] is promising. The first demonstrated working QCA logic device was based on metal-dot [4] system and required low temperature for operation. Room temperature operation requires more advanced fabrication technology and which is still under investigation. On the other hand, realization of the QCA technology using the coupling between nanomagnets and magnetic fields seems more promising. Magnetic quantum cellular automata (MQCA) logic has been used to successfully implement majority logic gate [5] and other primitive gates like NAND and NOR at RT.
A single domain nanomagnet must show bistability so that it can be used to represent a binary state (either ‘0’ or ‘1’). Bistability for a nanomagnet means that in the absence of an external field, the magnetization will point in either of the two directions along an axis called the easy axis (Figure 4.1).

The stable state of any nanomagnet or micromagnet strongly depends on the principle of total energy minimization of the magnet. Hence we need to know the different energy components that govern the net magnetization of any ferromagnetic material in the nm or µm length scale. This basic knowledge will also guide the choice of the suitable magnetic structures.
4.2. Single/Multi Domain Structures: Different Energy Components

There are five energy components that can affect the magnetization of a ferromagnetic material. These energy contributions in turn determine the formation and structure of domains. The formation of magnetic domains is the result of energy minimization. The total magnetic energy of any ferromagnetic material can be written as:

\[ E_T = E_{ex} + E_{ms} + E_a + E_{mt} + E_z \]  (4.1)

Where \( E_{ex} \) is the exchange energy, \( E_{ms} \) is the magnetostatic energy, \( E_a \) is the magnetocrystalline energy, \( E_{mt} \) is the magnetostriction energy and \( E_z \) is the zeeman energy.

The exchange energy \( E_{ex} \) is the energy associated with the interaction and coupling between spins. The exchange energy for two atoms \( i \) and \( j \), that have one electron each is given by [7]

\[ E_{ex} = -2J m_i m_j \]  (4.2)

Where \( J \) is the exchange integral and \( m \) is the spin moment of electron. For ferromagnetic materials, the exchange integral is positive and hence tends to keep adjacent spins (magnetic moment) parallel to each other. If a ferromagnetic material is single domain, i.e. all the magnetic moments are aligned along the same direction then the exchange energy is minimized.

When a piece of magnetic material is magnetized by an external field in a given direction, the magnetic field lines go from the north to the south pole of the magnet outside the magnet. However, the direction of the field inside the material (Figure 4.2) is in the opposite direction to the applied field. This is called the demagnetizing field. The
demagnetizing field results in a magnetostatic energy, which can be reduced by reducing the demagnetizing field. The formation of domains reduces the magnetostatic energy as shown in Figure 4.2. ‘Ferromagnetic domains are small regions in ferromagnetic materials within which all the magnetic dipoles are aligned parallel to each other’ [8]. The reduction in magnetostatic energy of course increases the exchange energy since $m_i m_j < 0 \Rightarrow E_{ex}$. Hence, there is always a balancing act between these two energy components (provided other energy components remain unaffected).

Shape anisotropy is the tendency for a magnetic material to have magnetization parallel to its long axis. The shape anisotropy is a direct consequence of the effort of the structure to reduce the magnetostatic energy. The magnetostatic energy is given by

$$E_{ms} = \mu_0 N_d M^2 / 2$$

(4.3)

Where $\mu_0$ is the free space permeability, $N_d$ is the demagnetizing factor and $M$ is the magnetization of the sample along a specific direction. The demagnetizing factor is highest along the short axis and lowest along the long axis. So it is easy to magnetize a material along its long axis (also referred to as the easy axis).

![Figure 4.2: Magnetic domain structures](image)

Figure 4.2: Magnetic domain structures [8].
The magnetization in ferromagnetic crystals prefers to align along certain preferred crystallographic directions. The preferred directions are termed as easy axes, because to magnetize the sample along the preferred crystallographic direction the applied field required is the least as shown in Figure 4.3(a). For bcc iron with cubic anisotropy, <100> are the easy axes and <111> directions are the hardest. To minimize the magnetocrystalline energy, domains will form so that their magnetizations point along easy crystallographic directions. The magnetocrystalline anisotropy energy for the cubic crystals (Fig 4.3(b): bcc iron) is given by

\[ E_a = K_0 + K_1 (\alpha_1^2 \alpha_2^2 + \alpha_2^2 \alpha_3^2 + \alpha_3^2 \alpha_1^2) + K_2 (\alpha_1^2 \alpha_2^2 \alpha_3^2) \]  

(4.4)

Where \( \alpha_i \) is the direction cosines of the magnetization along the three easy axes and \( K_i \) is the anisotropic constant.

Figure 4.3(a): Schematic magnetization curves of a ferromagnet where field pointing toward hard and easy axis; (b): easy and hard directions in a unit cell of bcc iron crystal [8].
A piece of ferromagnetic material undergoes a change in length called magnetostriction when it gets magnetized. Magnetostriction causes the triangular domains of closure to extend horizontally whereas vertical domains try to elongate vertically (shown in Figure 4.4) causing an elastic strain energy term to be added to the total energy. Magnetocrystalline energy supports the formation of smaller domains by the introduction of additional domain walls.

Zeeman energy arises due to the application of external magnetic field. This energy is minimized when the magnetization aligns along the direction of an applied external field and is given by

\[ E_z = -HM\cos\theta \]  

Where \( H \) is the applied field, \( M \) is the magnetization and \( \theta \) is the angle between \( H \) and \( M \).

Whether a magnetic structure will remain single domain or multi-domain depends on the minimization of the total energy components described above. The total energy depends on the size and geometry of the structure. Below a critical size single domain is energetically favorable. For larger structures the formation of multiple domains is
favored as shown in Figure 4.5. It should be noted that the field required to switch a single domain nanomagnet from one stable state to other is higher compared to the field required to change the wall orientation of a multi-domain micromagnet. More quantitative data is presented in the following sections. The next section will describe the switching of a single domain nanomagnet followed by comparison of single and multi-domain structures. These studies will clarify the motivation of this work.

Figure 4.5: Schematic energy landscape of single and multi-domain particles.
4.3. Nanomagnet Switching

With the application of an external magnetic field the magnetic moments of a series of nanomagnets can be aligned into a neutral logic state, called ‘null’ state, against the easy axis of magnetic anisotropy. As this state is not stable, when the field is removed, the chain of nano-magnets relaxes into the antiferromagnetically ordered ground state [8]. With the help of an input magnet (whose long axis is along the direction of the applied field) the first element of the chain can be influenced during relaxation which in turn switches the neighboring nanomagnet and so on setting the switching of the whole line along the desired direction as shown in Figure 4.6.

While the use of external magnetic field to excite nanomagnets is useful to show the basic operating principles of NML wires and logic gates, for on chip implementation it is problematic. It has been demonstrated [9] very recently that on-chip NML system implementation is possible by using Cu line with permalloy cladding/yoke to generate on-chip magnetic fields by passing electric current through the Cu lines. Figure 4.7 captures the implemented design architecture.

![Figure 4.6: External field assisted switching of a line of nanomagnets.](image)
Figure 4.7: Current through the Cu line creates magnetic field that excites nanomagnets to the null state [courtesy of Tanvir Alam]

Although the use of external magnetic field or current carrying copper lines for excitation of nanomagnets paves the pathway for alternative logic operation other than CMOS; there is little effort so far to use active device currents as a means of magnetic excitation. The high current carrying capacity of the III-Nitrides makes them a natural choice as a substrate platform for such integration with magnets. The HEMTs which find applications in high power high voltage and high frequency operation, if coupled with NML, can indeed prove attractive for such device applications. Moreover as nano/micromagnets can retain their states without an external power source, they can also be used in non-volatile memory applications. Hence, the integration of nano/micro-magnets with HEMTs can also be advantageous for coupling logic and memory in the same device. The next sections explore the possibility of integration of nano/micro-magnets with III-Nitride HEMTs.
4.4. Integration with III-V Nitrides: Single Domain case

From the previous section, let us first evaluate the magnetic field available from the drain current flowing in the III-Nitride HEMT devices fabricated in our group. Figure 4.8 shows a schematic of the HEMT structure and nanomagnets to be integrated with it.

The gate has not been shown intentionally in the figure, because the primary focus of this section is to explore the switching of nanomagnet by drain current. Although the highest reported drain current density for short gate lengths is \(~2.3\) Amps/mm \([10]\), in general for AlN/GaN HEMTs we obtain drain current in the range of 1.4-1.6 Amps/mm.

![HEMT structure with nanomagnets](image)

Figure 4.8: HEMT structure with nanomagnets to be put between metal contacts. The direction of the current and produced magnetic field direction has also been shown.
for longer gate lengths. Using ‘Maxwell’ simulation, the magnetic field obtained for that current density (1.4 Amps/mm) is slightly less than 1mT. Using Ampere’s law ($\int H \cdot dl = I$) the analytical expression for the resultant magnetic field is obtained as:

$$H = \frac{J}{2} \hat{x} \quad \text{(for } z > 0)$$

Where $J$ is the current density along + (ve) y direction as shown in Figure 4.8. Using the above expression the $H$ field is calculated to be ~0.9mT for $J=1.4$Amps/mm.

OOMMF micromagnetic simulation has been performed to find the minimum field required to switch a nanomagnet. It is found that the field required to switch a 75nm x 90nm x 1nm supermalloy* (magnetic properties described later) nanomagnet is 9mT—much higher than the field available due to the flow of the current in the 2DEG channel, shown in Figure 4.9. A 1nm thin nanomagnet fabrication is challenging and with the increase of thickness the required switching field increases, as expected by the higher stability due to the large volume.
Figure 4.9: Switching field of a 75nm x 90nm x 1nm supermalloy nanomagnet using OOMMF simulator.

TABLE 4.1

SUPERMALLOY MAGNETIC PARAMETERS

<table>
<thead>
<tr>
<th>Composition</th>
<th>Ni (79%), Fe (16%), Mo (5%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Saturation magnetization</td>
<td>$8 \times 10^7$ Ams/m</td>
</tr>
<tr>
<td>Exchange stiffness</td>
<td>$1.05 \times 10^{-11}$ J/m</td>
</tr>
<tr>
<td>Permeability</td>
<td>200000</td>
</tr>
<tr>
<td>Coercive field</td>
<td>0.02Oe</td>
</tr>
</tbody>
</table>
Since it is difficult to obtain such high H field using the drain current of a normal HEMT; we turn our attention toward multi-domain structures where a lower H field is needed to move domain walls. The next section presents the simulation results performed with OOMMF micromagnetic simulator to explore the effect of the low magnetic field on different micromagnetic structures.

4.5. Multi Domain Structures: Simulation Results

The hysteresis loop of micromagnetic structures made of ferromagnetic materials is very useful in finding various properties of the structures. The dependence of magnetization (M) on the applied field (H) has been shown in Figure 4.10. The value of M at saturation is called the saturation magnetization (\(M_s\)). When the applied field is reduced to zero the material still retains some magnetization called remnant magnetization (\(M_r\)). However, if initial magnetization is interrupted and H is reversed and then re-applied, then the magnetization follows the minor hysteresis loop shown in Figure 4.10. The presence of minor hysteresis loops make micromagnets attractive for this work over nanomagnets because the application of very small H fields can cause a permanent irreversible change in the magnetization.
Coercively of any ferromagnetic material is defined as the magnetic field required to reduce the magnetization of that material to zero after the magnetization of the sample has been driven to saturation. Though not an absolute measure, it provides a rough estimate of the switching field of a micromagnetic structure. Based on this fact, OOMMF simulations have been performed to see if there is a difference in coercive fields between two structures (made of Supermalloy): one with dimension 2µm x 2µm x 30nm and the other with 5µm x 5µm x 30nm dimension. It has been found that the coercive field for the latter is smaller compared to the former (Figure 4.11). This simulation result lays the foundation for the search of dimensionally bigger structures (in the range of micrometer) which might need smaller field (compared to the nanomagnet switching field) to show irreversible changes.

Figure 4.10: Typical hysteresis loop for multi-domain ferromagnetic materials [11].
Figure 4.11(a): M-H curve of 2µm X 2µm X 30nm dimension supermalloy structure along x direction (applied field is along x direction) showing coercivity 2mT; (b): M-H curve of 5µm X 5µm X 30nm dimension supermalloy structure along x direction (applied field is along x direction) showing coercively 1.6mT.

For a 12µm x 12µm x 30nm square supermalloy structure the OOMMF simulation shows that only 2mT field is enough to cause an irreversible change. The colors and arrows shown in Figure 4.12 are indicative of intensity and the direction of the magnetization. Figure 4.12(a) is the initial random magnetization state, Figure 4.12(b) shows the state after the application of 20mT initializing field along – (ve) x direction. Figure 4.12(c) demonstrates the magnetization after the initial field is removed. The effect of applying 2mT field applied opposite to the direction of the initializing field, along + (ve) x direction on the sample is shown in Figure 4.12(d). Figure 4.12(e) shows the final state after the field is removed.
Figure 4.12(a)-(e): Showing different stages of 2-D magnetization of 12µm x 12µm x 30nm dimension supermalloy structure during the application and removal of magnetic field.
The difference in magnetization in Figure 4.13 features a different type of multi-domain structure [12] which consists of a submicron thin (200nm) magnetic wire with a circular pad on one side and a sharpened end on the other side. This type of structure is used to controllably nucleate domain walls on the wire at a relatively low magnetic field \( H \). OOMMF simulation shows that only 5mT field can cause irreversible changes.

Comparing Figure 4.13(b) and Figure 4.13(d), it is seen that the magnetization of the long tapered line extending from the circular round pad can be completely flipped by the application of 5mT field.

Figure 4.13(a)-(d): Showing different stages of 2-D magnetization of supermalloy structure during the application and removal of magnetic field.
OOMMF simulation thus identifies that supermalloy structures of different shapes and sizes indeed just need few mT of magnetic field to undergo irreversible changes. We further investigate multi-domain features and whether aspect ratio plays a role in determining the required minimum magnetic field to affect irreversible changes. Figure 4.14 shows that two structures of the same volume but different aspect ratios respond differently when a magnetic field of same magnitude is applied to them. The structure with higher aspect ratio (18µm/4µm=2.5) shows irreversible changes at H=2mT, whereas the structure with aspect ratio ~1 (9µm/8µm) does not show any change at this level.

Figure 4.14(a): Showing different stages of 2-D magnetization of 18µm x 4µm x 30nm supermalloy structure during the application and removal of magnetic field. (b): Showing different stages of 2-D magnetization of 9µm x 8µm x 30nm supermalloy structure during the application and removal of magnetic field.
As there would be MFM imaging challenges (discussed later) involved in detecting the minute changes in the domain wall orientation of micromagnetic structure, we look for more distinguishable changes. The simulation result shows that only $H \sim 4\text{mT}$ field is enough to cause complete flipping of magnetization from one direction to the other for the structure shown in Figure 4.15. Figure 4.15(c) shows the reference state (after initializing field is removed) and Figure 4.15(e) shows the final state after the 4mT field is removed. The difference in magnetization is evident from the simulation result.
Figure 4.15(a)-(e): Showing different stages of 2-D magnetization of 15µm x 4µm x 30nm dimension supermalloy structure during the application and removal of magnetic field.
OOMMF simulation thus far indicate that the magnetic field required to affect a detectable change in multi-domain micromagnetic structures is in the range of 2-5mT. However, the field generated from a conventional HEMT structure is in the range of 1mT (discussed in section 4.4). Hence, to enhance the field we have searched for III-Nitride structures and have found that highly n-doped (Si) GaN can deliver higher current density, and as a result higher magnitude of magnetic field can be obtained from such structures. Figure 4.16(a) shows the schematic of the structure. It is found experimentally that with ~100nm doped n⁺-GaN layer, a current density of 4Amps/mm is obtained. Since the current density (current per unit width) increases linearly with the thickness of the doped layer, the Figure 4.16(b) shows the magnitude of magnetic field resulting from various doping thicknesses. The ‘Maxwell’ simulator has been used in this case to calculate the H field. This result is also encouraging since a high magnetic field results from these structures. This field can be used affectively to move the domains of the micromagnetic structures, and possibly nanomagnets too (Figure 4.16(b)).
The simulation and experimental results described in this section lays the foundation for future experimental investigation of integrating nano and micromagnets with GaN. However, the first challenge is the process integration and imaging of the nanomagnets (on GaN wafer) and detection of the domain patterns of the multi-domain features. Using MFM this experimental process development is described next.

4.6. Experimental Endeavor

The processing step involved in integrating supermalloy magnetic structures with III-Nitrides is now described. The steps are: MBE growth of AlN(4nm)-GaN(235nm) heterostructure on semi-insulating GaN on sapphire wafers. Cleaning of the sample (HCL DI water) to remove metal droplets, followed by acetone, isopropanol treatment. Bake 90°C for 1min before spinning the PR. Spin MMA-PMMA double photo-resist
stack (ramp: 2500rpm/s, speed: 4000 rpm). Sample was baked for 3min at 170°C. The dual layer e-beam resist is necessary to ensure good lift-off. Electron beam lithography definition of magnetic patterns in ELS-7700 EBL system at 75 KV. After exposure, the sample was developed in a MIBK-IPA-MEK solution for 40s. It was then rinsed in IPA and DI water for 1 minute. Deposition of 30nm Supermalloy metal (Fe 16%, Ni 79%, Mo 5%) in E-beam evaporator. Lift-off was done using dichloromethane which reacts very fast with e-beam resists.

Figure 4.17 shows the SEM, topographic (AFM) and MFM image of a single nanomagnet (65nm/180nm/30nm). The direction of magnetization of the nanomagnet is indicated by the arrow in Figure 4.17(c). During MFM imaging when the magnetic tip feels the magnetic attraction by the sample then the change of phase angle of the cantilever is translated to red color. Whereas, the repulsive magnetic force is coded with the yellow color. Hence red and yellow color can be viewed as the magnetic north and south poles. This color coded MFM imaging scheme is really useful in determining the magnetization of nano/micromagnets.

Figure 4.17 (a) SEM, (b) Topography and (c) Lift mode phase images (MFM) of a single-domain supermalloy nanomagnet (65nm x 180nm x 30nm).
Sets of nanomagnets with two different aspect ratios (60nm/130nm/30nm and 60nm/160nm/30nm) were fabricated. The SEM image shown in Figure 4.18 captures the image of the row of nanomagnets with aspect ratio (60nm/130nm/30nm).

Initially the magnetization states of the nanomagnets are expected to be aligned along the easy axes, but in a random fashion from a magnet to the next. Hence a magnetic field of 350mT was applied on the sample to align the magnetizations of all the nanomagnets. After the application of the magnetic field, MFM imaging was performed for the two sets of aspect ratio samples. Surprisingly, the direction of the magnetization of two sets is different as shown in Figure 4.19. When the sample was physically rotated 180° and the MFM scan was performed in the direction as before, the magnetization direction of the larger aspect ratio nanomagnets flipped as expected but for smaller aspect ratio nanomagnets, it stayed the same. So it was inferred that although the MFM scan could successfully resolve the magnetization of the larger aspect ratio (60nm/160nm/30nm) nanomagnets, it could not do so for smaller aspect ratio ones. For smaller aspect ratio ones the magnetization direction was always pointing along the scan direction.

Figure 4.18: SEM image of the 60nm x 130nm x 30nm nanomagnets.
Figure 4.19(a): Showing the magnetization (MFM image) of two different sets of aspect ratio nanomagnets (b): Showing the magnetization after the sample was physically rotated 180° (scan direction remains same as before).

As the MFM tip is itself magnetic, to magnetize smaller aspect ratio low-volume nanomagnets the field requirement (H α Volume) is also smaller. Hence it is highly likely that the MFM tip itself magnetizes the smaller volume nanomagnet sample during the scan. To overcome this problem, the scan direction of the tip is rotated 90 degree so that it scans the sample along the hard axis (short axis). It is difficult to flip the magnetization of any sample along the hard axis compared to its easy axis. The improvement is evident and is shown in Figure 4.20. After physically rotating the sample 180° the magnetization direction rotates 180° for both sets of nanomagnets as expected. Thus by changing the scan direction we could successfully resolve the magnetization of almost all the nanomagnets (large and small aspect ratios).
Figure 4.20: The AFM and MFM image has been shown for the hard axis scan of the nanomagnets, before and after rotation (180°) of the sample.

After the successful integration and imaging of nanomagnets on III-Nitrides, we discuss the imaging of domain walls in multi-domain micromagnetic structures. For this purpose various micromagnetic features were fabricated on AlN/GaN heterostructure. The SEM images shown in Figure 4.21 capture some of the structures. During the MFM scan (performed along the hard axis) of these micromagnetic structures, the problem of tip influencing the imaging resurfaced.
Figure 4.21: SEM images of the U-shaped and V-shaped micromagnetic structures made of supermalloy on GaN.

Figure 4.22(1c) shows the MFM image of a rectangular shaped micromagnet (1µm x 2µm x 30nm). When the sample is physically rotated 180° and scanned (same direction as before), the image is as in Figure 4.22(2c). Figure 4.22(3) is the 180° rotated image of the Figure 4.22(1c). If MFM scan could resolve magnetization properly, we anticipate Figure 4.22(2c) should be identical to the Figure 4.22(3). But it is clearly seen that the Figure 4.22(2c) and the Figure 4.22(3) are different. We understand that the magnetic field required for moving a domain wall is much less compared to that needed to flip a nanomagnet. Hence the magnetization of multi-domain features is far-more susceptible to be changed by the MFM tip itself. To avoid such tip-induced flipping a solution was identified. It was found that the influence of the tip arises mainly due to the reason that it comes in direct contact of the structures to be imaged. To avoid direct contact, a thin layer (40nm) of PMMA was coated on top of the structures to be imaged. This prevented the tip from direct contact with the magnets, and reduced the magnetic interaction, solving the problem.
Figure 4.22(1a-1c): Showing the rectangular structure on the sample, its AFM and MFM image, (2a-2c): Showing the rectangular structure on the physically 180° rotated sample, its AFM and MFM image, (3): Showing the 180° rotated image of the Figure 4.22(1c).

Using the PMMA coating even along the easy axis scan, we were able to resolve the magnetization of the multi-domain features properly. Figure 4.23(1c) and Figure 4.23(2c) show that the images are identical as expected by the MFM scan. Clearly resolved domain walls were also observed by the MFM scan as shown in the Figures 4.23(1c) and 4.23(2b).
Figure 4.23(1a-1c): Showing the rectangular structure on the sample, its AFM and MFM image; (2a-2c): Showing the rectangular structure on the physically 180° rotated sample, its MFM image, and the 180° rotated image of the Figure 4.23(2b).

The ability of imaging the multi-domain featured micromagnets on III-Nitrides lays the foundation for exploring the effect of drain current (of the HEMT structure) on the domain wall motion in the structures described in section 4.5. Based on this achievement, an Electron Beam Lithography (EBL) mask has been designed for fabricating various micro and nanomagnetic structures on the III-Nitride substrate. It was identified in section 4.5 that doped n-GaN structures can indeed provide high current and
as a consequence high magnetic field can also be generated for more effective switching. Figure 4.24 shows the schematic of the grown structure and the metal contact that has been formed for driving in current through the channel. A thin layer of oxide has also been deposited on top of the substrate to keep the electrical interaction of the channel and the micro/nanomagnets at a minimum. The current density obtained was more than 10amps/mm (could not be measured beyond this point due to the upper limit of the instrument). Such a high current density produces a magnetic field in excess of 8mT. The future work awaits the promises of integrating the magnetic structures on this sample and to investigate the effect of active current driven domain wall motion and switching of these micro/nanomagnetic structures.
Figure 4.24: Graph showing the current density available from the n-GaN structure. Schematic in the inset shows the grown structure and deposited metal contact (micromagnets to be put in between these contacts). Inset also shows the top view of the processed sample with metal contacts.

4.7. References


CHAPTER 5

CONCLUSION AND FUTURE WORK

In conclusion, we have systematically analyzed the properties of metal-polar AlN/GaN interface with ALD Al$_2$O$_3$ insulator. The increase of pinch-off voltage in the resulting MOS structures from $V_p = -3.6$ V for $t_{ox} = 2$nm to $V_p = -8.8$ V for $t_{ox} = 8$nm is explained by invoking a fixed positive charge ($Q_{it} \sim 6 \times 10^{13}$ cm$^{-2}$) at the ALD/AlN interface. Recent works on ALD Al$_2$O$_3$ as well as other gate oxide dielectrics like Ta$_2$O$_5$ and HfO$_2$ for AlN/GaN HEMTs report similar trends, and interface trap density ($D_{it}$) values comparable to the $Q_{it} \sim 6 \times 10^{13}$ cm$^{-2}$ reported here. The role of ALD oxygen layers as possible modulation dopants can offer opportunities for various innovative designs in III-Nitride electronic devices. Most importantly by controlling the effective $Q_{it}$ (for example by compensation doping, or by varying the polarization through the composition), $V_p$ can be made to increase, remain independent, or decrease with $t_{ox}$. Even without the ALD layer, if the top heterointerface satisfies $Q_{it}=Q_{p(GaN)}$ using either GaN caps or InAlN cap that is polarization matched to GaN will render $V_p$ independent of the cap layer thicknesses. In the future we plan to investigate this pathway. Though the precise origin of the charges at the heterointerface, and the corresponding behavior for the N-polar face remains to be clarified in the future, the strong correlation with the
polarization is expected to further our understanding of ALD/III-Nitride interfaces with useful consequences for high-performance devices. Moreover, the analysis presented here is for samples with as deposited ALD Al$_2$O$_3$. It will also be interesting to investigate how post annealing affects the electrical properties of these MOS structures as well as the resulting HEMTs.

The successful integration of micro/nano magnets (supermalloy made) with III-Nitride heterostructure and the ability of imaging the single/multi-domain featured nano/micromagnets on III-Nitrides presented in this work opens the possibility of active device driven switching of nanomagnets in MQCA logic operation. Although simulation result has shown that the field requirement in nanomagnet switching is higher compared to the H field obtained (~1mT) using the drain current (1.4Amps/mm) of a normal HEMT, it is possible to use other nitride structures (n-GaN) to deliver current density more than 10amps/mm as shown here. Such a high current density produces a magnetic field in excess of 8mT. This field can be used affectively to move the domains of the micromagnetic structures, and possibly nanomagnets switching too. Moreover the future work also promises to explore the possibility of hybrid logic (HEMT + MQCA) technology in the same device architectural framework.