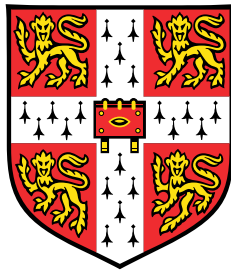


Multi-microscopy Characterisation of III-nitride Devices and Materials



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This dissertation is submitted for the degree of
Doctor of Philosophy

Wolfson College

April 2016

I would like to dedicate this thesis to my loving parents ...

Declaration

I hereby declare that except where specific reference is made to the work of others, the contents of this dissertation are original and have not been submitted in whole or in part for consideration for any other degree or qualification in this, or any other university. This dissertation is my own work and contains nothing which is the outcome of work done in collaboration with others, except as specified in the text and Acknowledgements. This dissertation contains fewer than 65,000 words including appendices, bibliography, footnotes, tables and equations and has fewer than 150 figures.

Christopher Xiang Ren
April 2016

Acknowledgements

And I would like to acknowledge ...

Abstract

This is where you write your abstract ...

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Nomenclature

Acronyms / Abbreviations

1-D	One-Dimensional
AlN	Aluminium Nitride
BSF	Basal-plane Stacking Fault
EBL	Electron Blocking Layer
EL	Electroluminescence
ET	Electron Tomography
FIBT	Focussed Ion Beam Tomography
FSR	Free Spectral Range
GaAs	Gallium Arsenide
GaN	Gallium Nitride
InN	Indium Nitride
LED	Light-Emitting Diode
PD	Partial Dislocation
PSF	Prismatic Stacking Fault
SPS	Single Photon Source
WGM	Whispering Gallery Mode

Chapter 1

Introduction

Gallium nitride (GaN) has been termed the 'most important semiconductor material since silicon' [1], and indeed the influence of this incredible material and its associated alloys (termed III-nitrides) is pervasive in modern society. The impact of III-nitride materials is perhaps best evidenced by the global transition from traditional lighting sources to semiconductor lighting solutions based on III-nitride materials. Since the first demonstration of a high-brightness blue light emitting diode (LED) in 1991 by Shuji Nakamura [2], the widespread use of LEDs for general lighting purposes has blossomed into a multi-billion pound industry. The extraordinary optical properties of III-nitride materials have enabled their application outside of the lighting industry: the development of III-nitride based lasers has found applications in telecommunications [3], medicine [4] and data storage. Furthermore, III-nitride optical emitters have been used as single photon sources (SPSs) which have applications in cryptography for secure communications [5].

The optoelectronic properties of III-nitride materials are somewhat astonishing: GaN suffers from a defect density several orders of magnitude higher than other optically active semiconductor materials such as gallium arsenide (GaAs) [6] yet is still optically active. Despite this, the effects of defects originating from the heteropitaxial growth of GaN are clearly deleterious when considering III-nitride device operation. This work aims to explore the manner in which the microstructural properties of photonic III-nitride devices affect their performance by combining multiple microscopy techniques, an approach we term 'multi-microscopy', thus allowing us to link specific structural features with emissive properties at the device level. The experimental research in this thesis is separated into four main sections.

The first section details the investigation of inhomogeneous electroluminescence (EL) of indium gallium nitride (InGaN) quantum well (QW) LEDs. By employing the use of scanning probe techniques, electron microscopy and spectroscopy the underpinning cause of LED behaviour was elucidated and reported.

The second section involves microscopy-based investigation into the mechanisms behind incomplete etching in the fabrication of III-nitride based microdisk cavities and the effect of this issue on the overall optical performance of these cavities

The third section describes the microscopy of one dimensional (1-D) photonic crystal cavity (PCC) 'nanobeam' cavities. The intrinsic resistance of III-nitride based materials can often result in improperly etched features, which can results in high optical losses in cavities. This section concerns the use of tomographic techniques such as electron tomography (ET) and focussed ion beam tomography (FIB-T) to investigate the effect of these issues on the emission of III-nitride nanobeam cavities.

1.1 III-Nitride Material Properties

1.1.1 Crystal Structure

GaN can crystallise into two distinct crystal structures: hexagonal (wurtzite) and cubic (zinc blende and rock salt). Under ambient conditions, wurtzite GaN is the most commonly studied form as it is the most structurally stable. Thus, the work discussed in this thesis concerns wurtzite III-nitrides. A schematic of a wurtzite III-nitride crystal structure is shown in Fig.?? and consists of stacked hexagonal close-packed planes following an ABABAB stacking sequence. Atoms of the respective elements are tetrahedrally bonded to one another. However, in the case of III-nitrides this structure deviates from ideal tetrahedral bonding and results in a non-zero dipole moment for each unit cell which will be discussed in the following sections.

A 4-index Miller-Bravais notation (*hkil*) is used to denote the crystal planes where the index *i* is defined by the relation:

$$i = -(h + k) \quad (1.1)$$

The crystallographic planes (0001), (1-100) and (11-20) shown in Fig.?? are often termed the *c*, *m* and *a*-planes in the literature. The fundamental unit cell of the wurtzite GaN crystal structure and its associated lattice parameters **a** and **c** is shown in Fig.1.1

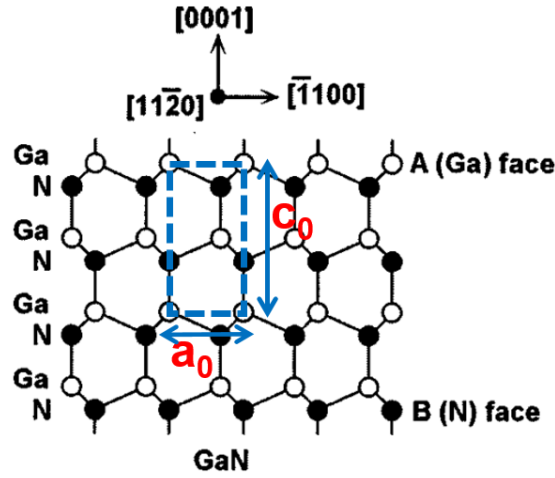


Fig. 1.1 Unit cell (dashed line) for GaN crystal structure and lattice parameters a_0 , c_0 . Adapted from [7]

Other members of the III-nitride materials such as indium nitride (InN) or aluminium nitride (AlN) have different lattice parameters due to the differing atomic radii of aluminium and indium relative to gallium.

Alloy	a (Å) at $T = 300\text{K}$	c (Å) at $T = 300\text{K}$
GaN	3.189	5.185
InN	3.545	5.703
AlN	3.112	4.982

Table 1.1 Room temperature lattice parameters for GaN, InN and AlN [8].

III-nitride photonic devices are often heterostructures consisting of ternary alloys the materials shown in Table ???. Lattice parameters of a relaxed ternary alloy $A_xB_{1-x}N$ can be estimated using Vegard's law [9]:

$$a = xa_{AN} + (1 - x)a_{BN} \quad (1.2)$$

$$c = xc_{AN} + (1 - x)c_{BN} \quad (1.3)$$

Typical indium compositions for blue LEDs range between 15-20 %, which leads to a considerable lattice mismatch of approximately 2 %, resulting in considerable amounts of strain in these GaN/InGaN heterostructures.

1.1.2 Band Structure

One of the principal driving factors behind the interest in III-nitrides for photonic devices is their direct bandgap which collectively spans the visible spectrum and beyond. The bandgap of III-nitride binary alloys is given below in Table.1.2.

Alloy	Bandgap (eV)
GaN	3.51
InN	0.78
AlN	6.25

Table 1.2 Direct bandgaps of GaN, InN and AlN [8].

Ternary alloying modifies the bandgap as shown in Fig.1.2. In theory the entire range of 0.78-6.25 eV is accessible through alloying, though material limitations reduce the full effective range for III-nitride devices [10].

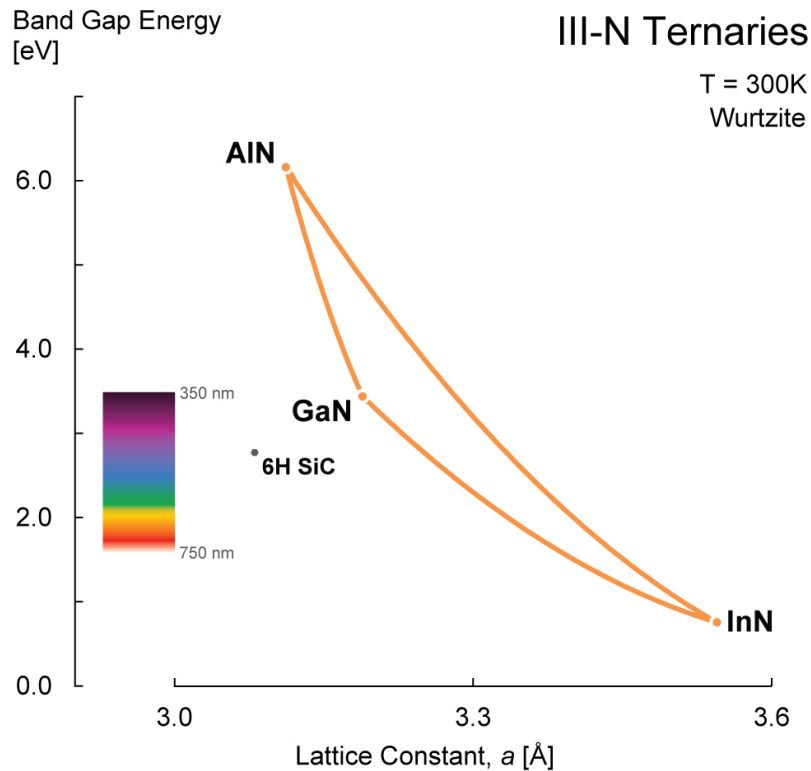


Fig. 1.2 Bandgap at room temperature for III-nitride materials with the visible spectrum shown on the left. Courtesy of K. Montgomery.

The bandgap of a ternary alloy $A_xB_{1-x}N$ is given by a modified Vegard's Law:

$$E_g = xE_g^{AN} + (1-x)E_g^{BN} - x(1-x)C \quad (1.4)$$

Where C is a bowing parameter which accounts for deviation from a linear relation between ternary alloy composition and bandgap energy. The value of the InGaN bowing parameter has been widely debated in the literature due to the lack of a reliable value for the bandgap energy for InN [8]. Although the current value of 1.4 eV is reported, there are also suggestions the bowing parameter may be composition dependent [11–13].

In considering the optical properties of III-nitride materials it is also important to consider the effects of impurities and defects. Crystal disorder introduces further energy states which would be 'forbidden' in an ideal crystal lattice leading to an effective 'smearing' of the bandgap. Sub-bandgap absorption can occur due to the introduction of these defective states. The smearing out of the absorption edge of the material is known as the 'Urbach tail', and can be a highly deleterious source of loss in III-nitride cavity structures [14].

Quantum Confinement Effects

The first prototype high-brightness blue III-nitride LED consisted of a GaN p - n junction, or a 'homojunction' [2], however modern LED structures consist of heterostructures known as quantum wells. QWs consist of a thin layer of low bandgap material between two quantum barriers with a higher bandgap. Carriers in the low bandgap material are effectively confined in one direction, hence the term 'quantum well'. This confinement leads to the discretisation of the carrier wavefunctions within the well, as shown schematically in Fig.1.3.

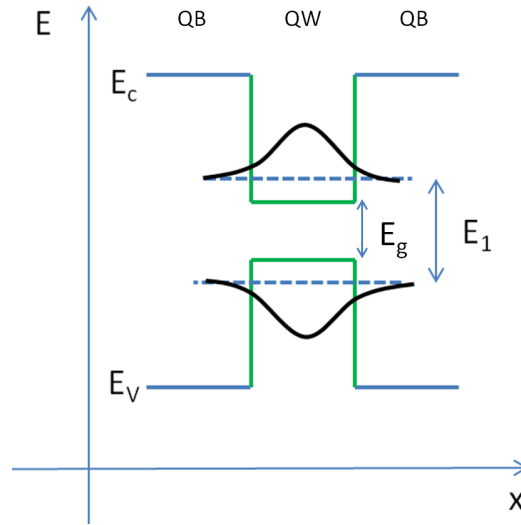


Fig. 1.3 Band diagram of a quantum well. The bandgap of the well material is denoted E_g , the energy of the ground state transition is denoted E_1 and the conduction and valence bands are denoted E_c, E_v respectively [15].

Thus the energy of the transition in the QW is given by the following relation:

$$h\nu = E_1 - E_{ex} \quad (1.5)$$

where E_1 is the energy of the ground state transition and E_{ex} is the exciton binding energy. For an infinite potential well of thickness L , the ground state E_1 is given by:

$$E_1 = \frac{\hbar^2 \pi^2}{2m^* L^2} \quad (1.6)$$

where \hbar is the reduced Plank constant, m^* is the carrier effective mass. As such, the energy of the optical transition can be related to the thickness of the well.

1.1.3 Built-in Fields

III-nitride materials in wurtzite structure are termed 'polar' materials, due to the fact they exhibit a spontaneous polarisation field [16]. This occurs due to III-nitride bonding structure deviating from an ideal tetrahedral structure along the (0001) axis along the crystal, combined with the ionicity of the bond [15]. This deviation causes each unit cell to possess a non-zero dipole moment along the principal axis of the tetrahedral bonding structure, resulting in an overall spontaneous polarization in the crystal. As the III-nitride wurtzite structure is non-centrosymmetric, the direction of the polarization depends on whether the crystal exhibits (+c) or (-c) polarity, as shown in Fig.1.4

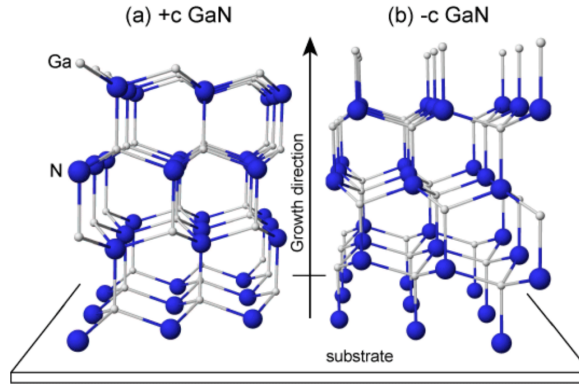


Fig. 1.4 Illustration of Ga-face (+ c) and N-face (- c) GaN wurtzite crystal exhibiting polarity along the c -axis [17].

This non-zero dipole moment is particularly strong for III-nitrides relative to other III-V semiconductors due to the strong electronegativity and small size of nitrogen compared to other group V elements, resulting in a metal-nitrogen bond with greater ionicity than other III-V bonds [18]. Fig.1.5 shows a GaN unit cell with lattice parameters c and a denoted.

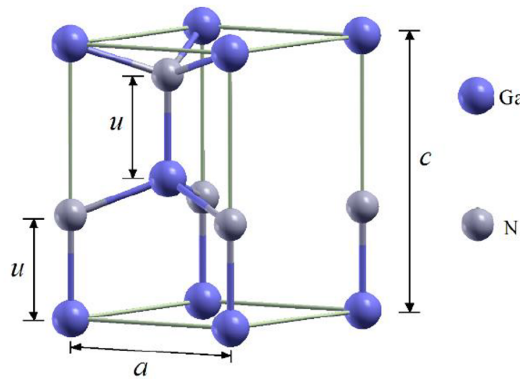


Figure 1. Unit cell 1x1-MN/GaN ($M = V, Cr$ and Mn) multilayers.

Fig. 1.5 GaN unit cell with lattice parameters c and a [19]

If all nearest neighbour bond lengths are equal, an ideal hexagonal closed packed crystal exhibiting zero spontaneous polarisation would have a ratio of lattice parameters denoted by:

$$\frac{c}{a} = \left(\frac{8}{3}\right)^{0.5} = 1.63299 \quad (1.7)$$

The degree of spontaneous polarisation observed in III-nitride materials is thus determined by the amount their lattice parameter ratio deviates from this ideal value. The values for bulk III-nitride materials are given in Table.1.3.

Alloy	$\frac{c}{a}$
GaN	1.6259
InN	1.6116
AlN	1.6010

Table 1.3 Bulk $\frac{c}{a}$ ratios for GaN, InN and AlN [15].

A lower $\frac{c}{a}$ ratio indicates a higher angle between the three bonds at the base of the tetrahedral bonding structure, resulting in a lower compensation polarisation along the (0001) axis and a higher spontaneous polarisation. Thus according to Table.1.2 the strongest spontaneous polarisation is observed in AlN and the weakest in GaN.

It is important to note that materials which exhibit spontaneous polarisation also exhibit piezoelectric polarisation [16]. Strain experienced by the material results in the distortion in of the crystal lattice, which can either alleviate or exacerbate the deviation from the ideal tetrahedral structure resulting in an additional polarisation. This piezoelectric polarization is a crucial consideration in III-nitride devices which often consist of QW heterostructures: lattice mismatches with underlying layers result in the expansion or contraction of III-nitride films. Interestingly two different polarisation configurations are obtained for AlGaN and InGaN coherently strained to GaN. In the case of InGaN the piezoelectric field acts against the spontaneous field, whilst the opposite is true for AlGaN strained to GaN. Within the context of visible light LEDs, InGaN containing QWs are dominated by the piezoelectric contribution to the polarization fields [20] due to the sizeable lattice mismatch between GaN and InN (11%) [21].

The Quantum Confined Stark Effect

As previously discussed, III-nitride photonic devices often make use of heterostructures known as quantum wells, which enhance radiative efficiency by confining carrier wavefunctions over a range of several nanometres. Given the presence of built-in fields in III-nitride materials, it is important to consider the effect polarisation fields will have on the band structure and thus optical properties of quantum wells as shown in Fig.1.6

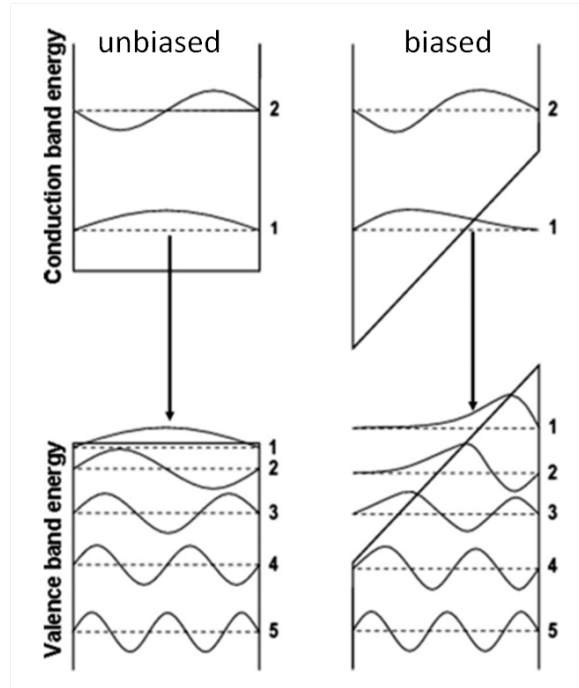


Fig. 1.6 Unbiased and biased quantum well energy levels with associated carrier wavefunctions. Under an applied field the overlap between the electron and hole carrier wavefunctions is reduced [22].

The transition from a rectangular to a 'sawtooth'-shaped potential well results in the reduction in energy of the optical transition, meaning the photons emitted from the QW are red-shifted. However, as the carrier density within the QW is increased, by either optical or electrical injection, the polarization fields are effectively screened resulting in a carrier density-dependent optical transition energy.

A further effect of the polarization fields is to spatially separate the carrier wave functions, thus reducing their overlap as shown in Fig.1.6. This results in a reduced probability for the radiative recombination carriers thus reducing the efficiency of III-nitride QW emitters.

1.1.4 Defects in III-nitrides

Many issues with III-nitride based optoelectronic devices arise from the high defect densities present. Dislocation densities tend to be several orders of magnitude higher for nitride devices relative to other III-V materials due to the lack of a low cost, widely-available lattice matched substrate [6]. Lattice mismatch and alloy-dependent growth temperatures results in the presence of imperfections in the crystal structure of the epitaxial film known as defects. These defects can result in perturbations to the electrical and optical properties of an 'ideal' crystal, and are often classified based on their spatial dimensions. 0-D defects are often

referred to as point defects, 1-D defects are commonly termed linear defects or dislocations, 2-D defects are known as planar defects or stacking faults, and there are a variety of 3-D defects known as volume defects.

0-D Defects

Point defects exist in four main forms, shown in Fig.1.7. Vacancies, where an atom is missing from the lattice, and self-interstitial point defects are termed 'native defects': there is no inclusion of foreign atoms. These two types of intrinsic point defects are shown in Fig.1.7 a) and b) respectively.

In the case of GaN, three types of vacancies can exist: gallium vacancies, nitrogen vacancies and divacancies. The gallium vacancy (V_{Ga}) which has a low formation energy in *n*-type GaN and is acceptor-like, this vacancy has a low migration barrier. Due to this low migration energy, it is expected that gallium vacancies form complexes with more stable defects. Gallium vacancies and associated complexes are thought to be the cause of yellow luminescence observed in *n*-type GaN. Nitrogen vacancies initially attracted a large amount of interest due to the common belief that their energy levels were close to or within the conduction band. Due to this, the *n*-type conductivity of undoped GaN was attributed to nitrogen vacancies. However, calculations have shown the thermal equilibrium of nitrogen vacancies to be too low to account for the observed conductivity. Nitrogen vacancies are also expected to have relatively low migration barriers, indicating complexes involving more stable defects may occur during high-temperature growth or annealing, especially in *p*-type GaN [23]. Divacancies have high formation energy in GaN and are not expected to form in large concentrations [23].

The inclusion of foreign atoms can result in a foreign interstitial point defect, or a substitutional impurity, both are shown in Fig.1.7 c) and d) respectively. The formation of self-interstitial or antisite (swapping of Ga and N lattice positions in the lattice) have a low occurrence due to the small lattice constant of GaN and large size mismatch between Ga and N atoms [23].

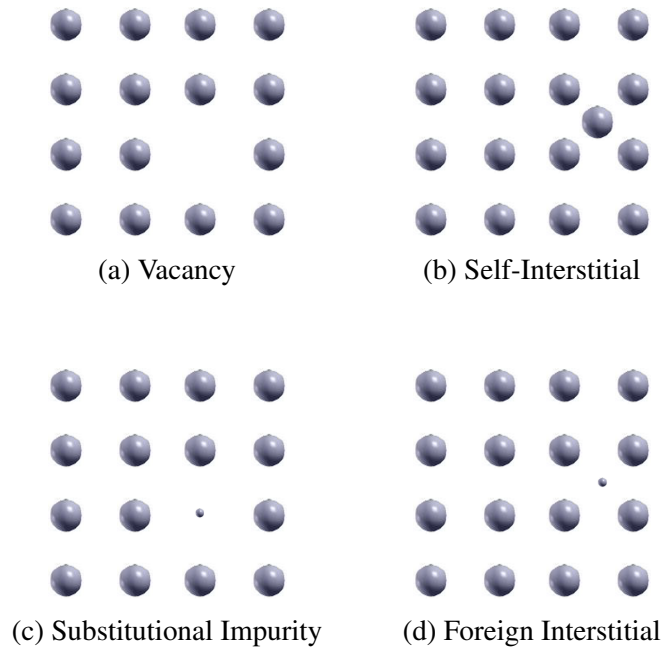


Fig. 1.7 Point Defects: vacancy, self-interstitial, substitutional impurity and foreign interstitial.

Point defects are responsible for a plethora of deleterious effects at the device level in III-nitrides: they can reduce radiative efficiency, produce undesired luminescence and act as parasitic current paths [23].

1-D Defects

Dislocations in GaN epilayers are categorised in two main forms: misfit dislocations (MDs) and threading dislocations (TDs). The origins of misfit dislocations are quite well understood: they occur through the release of misfit strain at interfaces between two crystals of differing lattice constants. The process is shown in Fig.1.9: a film with a lattice parameter greater than the substrate is grown as is typical for III-nitride epilayers (GaN on sapphire or InGaN on GaN) and as a result the grown layer experiences compressive stress and forms pseudomorphic layer. The top layer is strained and matched to the lower layer due to its smaller lattice parameter. Strain relaxation occurs as the pseudomorphic relationship is broken when the top film reaches a critical thickness and results in the formation of misfit dislocations.

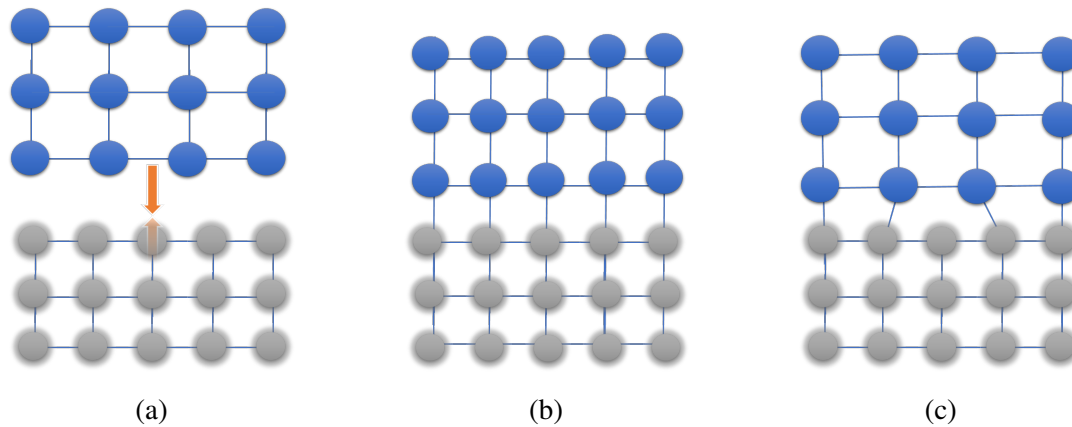


Fig. 1.8 Misfit dislocation formation through strain relaxation for heteroepitaxial growth: a) the film is grown on a substrate of smaller lattice size b) film maintains a pseudomorphic relationship with the substrate c) the film relaxes through the formation of a dislocation.

The origins of TDs are far less well understood. TDs are not believed to relieve mismatch stress, and typically propagate perpendicular to the planar surface. TDs are classified into three categories based on their Burgers vector, as shown in Table.1.4.

Dislocation type	Burgers vector
a type	$\frac{1}{3} \langle 11\bar{2}0 \rangle$
c type	$\langle 0001 \rangle$
a + c type	$\frac{1}{3} \langle 11\bar{2}3 \rangle$

Table 1.4 Burgers vectors for pure edge (**a**), pure screw (**c**) and mixed (**a + c**) TDs

It was initially reported that GaN islands on sapphire during the initial stages of growth may be misorientated with respect to one another and that during the coalescence of these misorientated islands [24]. This was seemingly disproven by a transmission electron microscopy data from a study on partially coalesced GaN on sapphire layers at various growth stages, which indicated the large majority of TDs seemed to initiate from within the nucleation layers at the GaN/sapphire interface rather than at coalescence boundaries. Oliver *et al.* used silane treatment to enlarge dislocation pits and observe them using atomic force microscopy, finding no significant relationship between boundary regions and the locations of dislocations [25]. It was however suggested that dislocations may arise from the overgrowth of smaller islands by larger ones. Thus, while there is convincing evidence that TDs do not originate due to island coalescence, the actual mechanism behind their generation remains poorly understood.

2-D Defects

Stacking faults are defects which disrupt the regular stacking sequence of the crystal structure, in non-polar heterostructures they can intersect the QW layers. As a result of this, stacking faults are a more pressing concern than dislocations in epitaxial films grown along alternative directions to the c -plane, as in polar materials stacking faults tend to remain in the nucleation layers [10]. Fig.1.9 shows different forms of stacking faults in $(11\bar{2}0)$ GaN (a -plane) on r -plane sapphire. Basal-plane stacking faults (BSFs) are atomic layers with a modified stacking sequence in the wurtzite crystal matrix. These BSFs can transfer to another stacking plane through prismatic stacking faults (PSFs). BSFs can also be bound by partial dislocations (PDs).

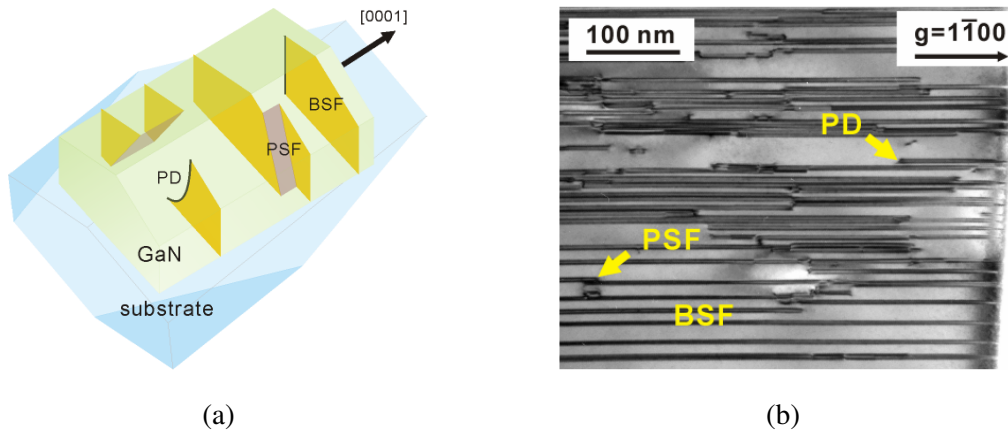


Fig. 1.9 a) a -plane GaN showing basal plane stacking faults (BSFs), prismatic stacking faults (PSFs) and stacking faults bounded partial dislocations (PDs) shown schematically b) and in TEM. Adapted from [26].

Three types of BSFs exist in wurtzite crystals, they are classified based on their displacement vector \vec{R} as shown in Table.1.5.

BSF type	Displacement vector \vec{R}
I_1	$\frac{1}{6} \langle 20\bar{2}3 \rangle$
I_2	$\frac{1}{3} \langle 10\bar{1}0 \rangle$
E	$\frac{1}{2} \langle 0001 \rangle$

Table 1.5 BSF types in wurtzite materials.

Whilst TDs are considered completely undesirable due to the adverse effects they may have on radiative processes and carrier transport, it has been suggested the presence of BSFs on the optical properties may be beneficial in some cases. Indeed, BSFs have been shown to

enable radiative recombination through confinement and thus act as QWs with an exciton binding energy of 45 meV [27].

3-D Defects

There are many forms of 3-D defects in III-nitrides such as voids, nanopipes and cracks. In this particular section we will focus on those relevant to the work featured in later chapters of this work.

Inverted hexagonal pyramid defects, also known as V-pits, are defects commonly found at the surface of InGaN/GaN QW structures. They form as a result of a TD intersecting the QW layers. It is believed that the low temperatures required for the growth of the InGaN layers allow even minute perturbations of the surface to persist into inclined facets with low growth rates, such as the $(1\bar{1}01)$ facets. The apex of TDs thus provide optimal conditions for the formation of V-pits during the growth of InGaN layers [28]. Interestingly, TEM studies have shown that the QW layers disrupted by the defect grow along the semi-polar facets at a lower thickness [28–30]. Fig.1.10 shows a TEM image of a V-pit in an InGaN/GaN multiple QW structure with a schematic describing this defect and how it affects the growth of the QWs.

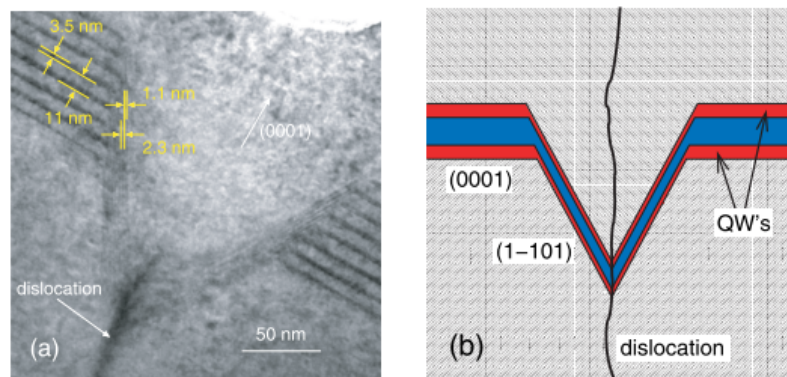


Fig. 1.10 a) TEM image of a hexagonal inverted pyramid defect b) Schematic of a V-pit with its associated TD [28].

The existence of TDs as non-radiative recombination centres has been well documented [6], however it is believed that V-pits suppress this non-radiative recombination and provide an increase in light emission efficiency in III-nitride devices by providing an energy barrier surround TDs [28]. Hangleiter *et al.* suggested the thinner wells grown along the semi-polar facets of the V-pit provided an energy barrier of several hundred meV relative to the normal *c*-plane QWs [28], thus providing a potential landscape shielding carriers from TDs, as shown in Fig.1.11.

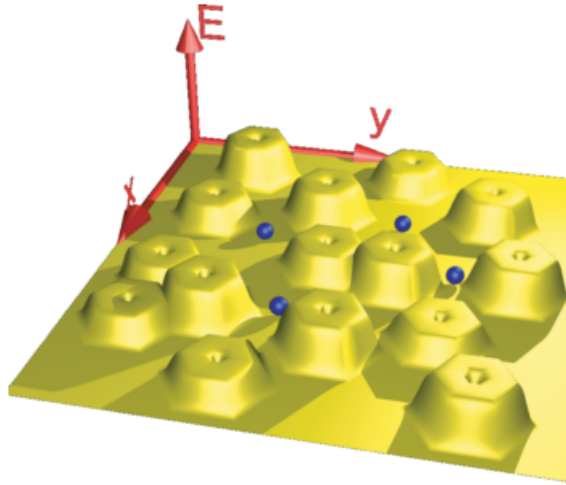


Fig. 1.11 Potential landscape due to V-pits decorating the apex of TDs: carriers (blue) need to overcome the energy barriers to recombine non-radiatively at the TDs [28].

1.2 III-nitride Devices

1.2.1 Light Emitting Diodes

Light emitting diodes (LEDs) are the most common application of III-nitride materials. These devices typically consist of a p-n junction. This consists of material containing excess acceptor (*p*-doped) and another containing excess donor impurities (*n*-doped) which are brought into contact. This allows holes from the *p*-type material and electrons from the *n*-type material to diffuse across the junction until an equilibrium state is reached, a region where the electric field from the charged dopants on either side prevents diffusion is formed known as the depletion region. The application of forward bias reduces the built-in potential across the depletion region and allows for the flow of electrons and holes across the junction, as shown schematically in Fig.1.12.

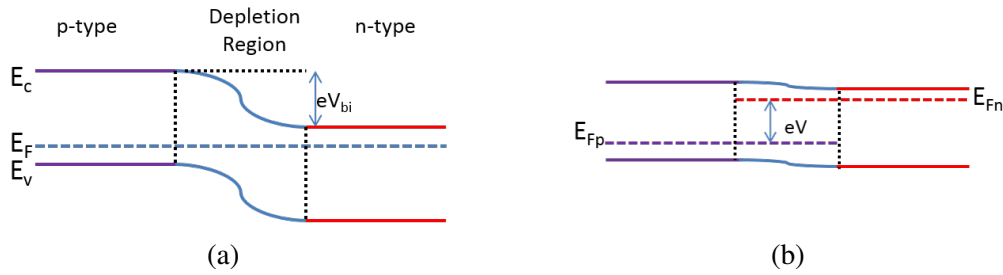


Fig. 1.12 a) p - n junction at equilibrium, with the conduction band, Fermi level and valence band denoted E_c , E_F and E_v respectively, the built in potential across the junction is denoted as V_{bi} b) under forward bias of V .

A schematic of a general LED structure is shown in Fig.1.13. Visible light LED structures typically contain a magnesium doped p -region and a silicon doped n -region. An electron blocking layer (EBL) consisting of a material with a higher bandgap than GaN (in this case AlGaIn) is used to prevent the leakage of electrons into the p -doped region and confine them in the InGaIn QW active region.

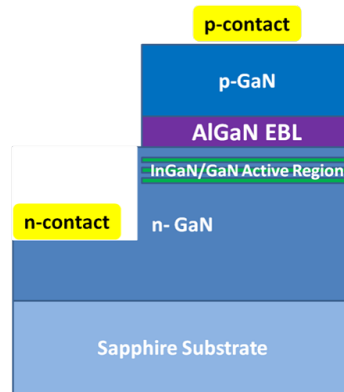


Fig. 1.13 Typical visible light LED structure grown on a sapphire substrate [15].

1.2.2 Microcavities

Microcavity emitters possess rather singular optical properties due to their dimensions. By matching one or more dimensions of the cavity to the order of the wavelength of confined light a plethora of effects can be produced such as low-threshold lasing, directional luminescence and enhanced nonlinear conversion [31]. By confining a dipole within a microcavity, one can modify its emissive properties by altering the photon density of states. The interaction rate between the confined dipole and a cavity photon relative to the average rate of dissipation of a cavity determines whether the microcavity operates in the weak-coupling or strong-coupling

regime.

Weak coupling occurs when dissipation overwhelms the dipole-cavity photon interaction: in essence, the effect of the microcavity in this case is to alter the vacuum description of the dipole lifetime, resulting in an increase in spontaneous emission for on-resonance cavity modes, known as the Purcell effect [32]. Weakly-coupled microcavity systems have applications across a wide range of optoelectronic devices due to this effect: from enhancing the recombination rate and extraction efficiency of embedded single photon emitters [33] to the development of high efficiency, low threshold lasers [34].

An interaction occurring on shorter timescales than the average dissipation rate of the cavity photon is defined as being in the strong coupling regime, and results in the formation of admixed eigenstates populated by quasiparticles known as polaritons, which are hybrid particles combining a photon and an electric dipole. The bosonic nature of these quasiparticles has led to the observation of spontaneously emitted coherent light from condensates of exciton-polaritons, a phenomenon also known as polariton lasing [35]. The expected threshold energy for coherent emission from a polariton laser is expected to be much smaller than that of a conventional laser due to the lack of the requirement of population inversion, thus rendering polariton lasers extremely attractive as low-threshold lasing applications [31]. Beyond polariton lasing, strong coupling in microcavities is also required for key quantum information processing tasks such as the entanglement of distinguishable quantum systems and controlled coherent coupling [36, 37].

Cavity Parameters and Design

The ability of a microcavity to confine light is thus a crucial parameter in producing the required effects and is known as the cavity 'quality factor' which is described by Eq. 1.8.

$$Q = \frac{\nu_0}{\delta \nu_0} \quad (1.8)$$

Where ν_0 is the resonant frequency of the cavity mode and $\delta \nu_0$ is the mode bandwidth. Cavity quality factor can be understood as a parameter describing the rate of energy decay the resonant mode undergoes within the cavity and thus may be alternatively described using an exponential characteristic decay constant τ_{cav} as shown in Eq.1.9, where Q^{-1} is the proportion of energy lost during a single cavity round-trip.

$$Q = \pi \tau_{cav} \nu_0 \quad (1.9)$$

The manner in which the resonant mode fields interact with the cavity geometry is also determined by the effective modal volume of the cavity, which is described by Eq.1.10.

$$V_{eff} = \int_V \frac{\epsilon_0(\mathbf{r})|\mathbf{E}(\mathbf{r})|^2}{\max[\epsilon_0(\mathbf{r})|\mathbf{E}(\mathbf{r})|^2]} dV \quad (1.10)$$

where $|\mathbf{E}(\mathbf{r})|^2$ is the normalised electric field amplitude, ϵ_0 is the dielectric constant and V is the quantization volume. V_{eff} describes the manner in which cavity supports the distribution of the resonant mode, thus in some cases a large evanescent field component must be included in the calculation of the modal volume.

Cavities often support more than one optical mode, which gives rise to another parameter, known as the free spectral range (FSR), which is defined as the frequency spacing between successive resonant modes. This parameter is crucial to lasing cavities as the probability of photon into a lasing mode is affected by the number of modes supported by the cavity. The FSR as well as parameters needed to define cavity quality factor are shown schematically in Fig.1.14.

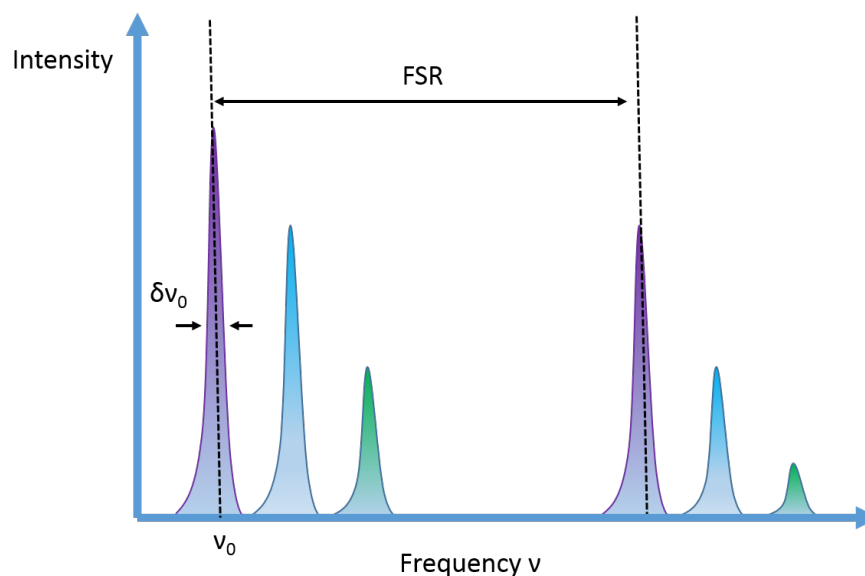


Fig. 1.14 Illustration of the some key parameters such as FSR and resonant cavity mode FWHM.

Cavities modify the optical density of states of an emitter through the generation of standing electromagnetic waves. There are several manners through which to achieve this: at the most basic level an optical cavity is a set of single reflective interfaces, spaced at a specific distance designed to enhance a particular optical mode. However, many cavity designs which employ the use of refractive index mismatches for total internal reflection or

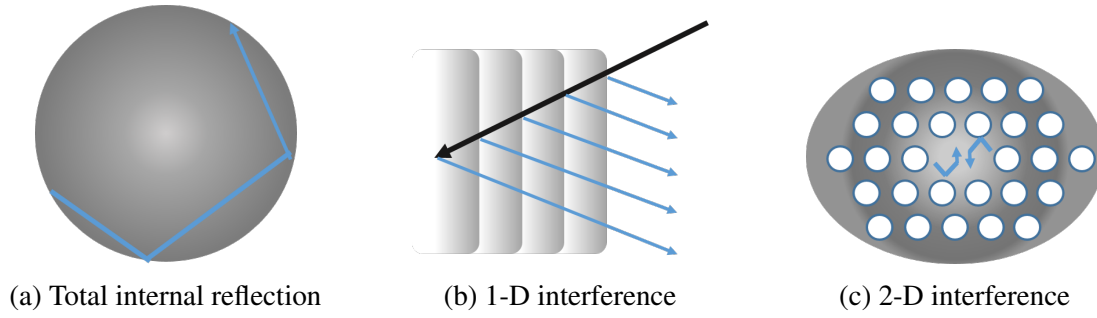


Fig. 1.15 Optical resonators.

an array of boundaries leading to interference enhanced optical modes also exist, as shown in Fig.1.15.

Microdisk Cavities

Total internal reflection can be exploited in circular geometries such as microdisk/ring/sphere devices, in which whispering gallery modes (FSR) (WGMs) propagate around the periphery of the disk.

Microdisks can be described as a cylinder of with a low height:radius ratio supported by a pillar of a small radius (less than half of the cylinder). The position of the WGMs is described by Snell's law: only light satisfying the condition described by Eq.1.11 is contained within the microdisk due to internal reflection.

$$\theta_c = \sin^{-1}\left(\frac{n_2}{n_1}\right) \quad (1.11)$$

Assuming the microdisk thickness is small enough to act as a waveguide in the vertical direction, we can consider the propagation of a ray of light in 2-D as in Fig.1.16. The forbidden position of the WGMs as defined by Eq.1.11 is given by R_{min} .

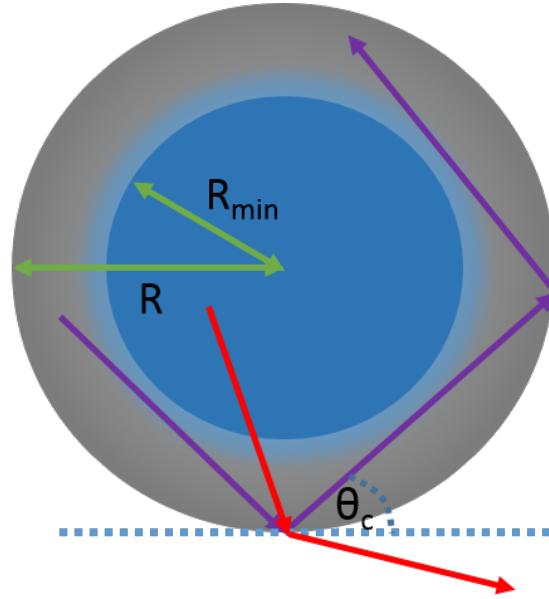


Fig. 1.16 Illustration of the minimum radius R_{min} for WGM propagation. Rays traversing the region $r < R_{min}$ (denoted in red) exceed the critical angle θ_c and thus escape the cavity. Rays travelling outside this region $r > R_{min}$ (denoted in violet) are confined.

Thus the fabrication of microdisks relies heavily on the ability to 'undercut' the microdisk material whilst still leaving a pedestal, as shown in Fig.1.17. This is a particularly difficult problem to address in terms of III-nitride materials due to the excellent thermal and chemical stability of GaN. Early efforts in microdisk nitride fabrication involved dry-etching processes, utilising the refractive index mismatch between the light emitting GaN layers and the sapphire substrate to confine light [38]. Although stimulated emission and lasing was observed by Chang *et al.* in dry-etched GaN microdisk cavities [39], Haberer *et al.* [40] reduced the required excitation power densities for lasing by an order of magnitude by employing photoelectrochemical (PEC) etching to undercut GaN microdisks, thus providing superior optical confinement due to the index contrast of the GaN/air interface relative to the GaN/sapphire interface used by Chang *et al.* [39]. Further improvements in fabrication were achieved by Tamboli *et al.* with room temperature lasing achieved in GaN/InGaN microdisks through enhancements in microdisk circularity and sidewall smoothness [38].

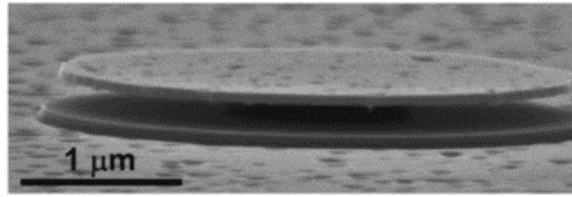


Fig. 1.17 Scanning electron microscope image of a GaN microdisk produced by selective etching of sacrificial AlInN layers [41].

Photonic Crystal Cavities

1.2.3 Nanowire Devices

Chapter 2

Experimental Methods

2.1 Atomic Force Microscopy

Atomic force microscopy is a non-destructive characterisation technique which employs a sharp tip mounted on a cantilever which is rastered across a sample surface. Tip-surface interactions result in changes cantilever position which are measured using the reflection of laser light reflecting off the cantilever and a four-quadrant photodetector as shown in Fig.2.1.

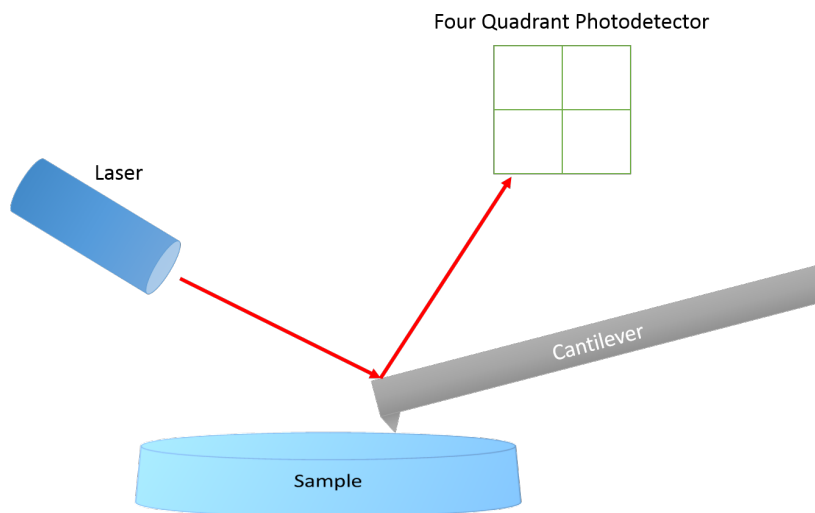


Fig. 2.1 Schematic of an atomic force microscope.

The positioning and movement of the tip is achieved through the use of piezo-electric actuators. In contact mode, a feedback circuit is used to apply a voltage to the piezoelectric crystal in order to maintain a constant tip-sample separation should the tip encounter any features, thus avoiding damage to either the tip or sample. The voltage required to maintain this distance (also known as the setpoint) is registered at each pixel of the scan and is used in

conjunction with calibration data to determine a vertical displacement value, thus generating a topographic image.

An alternative mode of operation known as tapping mode is often referred to contact mode. In this mode of operation the tip is made to oscillate close to its resonant frequency by the piezocrystal. Contact between the tip and the sample is achieved at the lowest point of each oscillation, which damps the oscillation of the tip. The oscillation frequency is maintained by the piezoelectric crystal, thus allowing for the generation of a topographic map. Tapping mode is often preferred to contact mode due to the exclusion of lateral friction which can cause tip wear and sample damage.

The forces experienced by the tip vary depending on the tip-sample separation, as shown in Fig.2.2. Van der Waals forces dominate at large separations attractive the tip to the surface. As the distance is reduced repulsive forces such as hard-sphere repulsion, electron-electron Coulomb interaction and the Pauli-exclusion interaction being to dominate. The sum of these forces result in cantilever deflection, changing the tip-sample interaction.

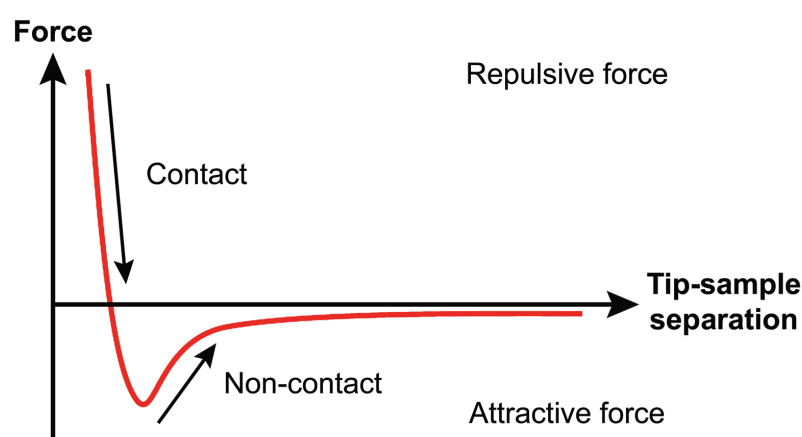


Fig. 2.2 The effect of separation on the tip-sample interaction force [42].

AFM offers excellent vertical resolution limited only by the probes vertical movement and external noise. However, the lateral resolution of this microscopy technique is heavily dependent on the shape and size of the tip employed. This is perhaps best highlighted by Fig.2.3 which depicts a hemispherical tip scanning across a flat-topped island. The apex of the tip is in contact with the surface, but the side of the island also experiences some contact: in this case there is a distinction between the two cases shown in Fig.2.2a) and b) as the error in the measured width of the island varies based on the relative size of the measured object to the tip.

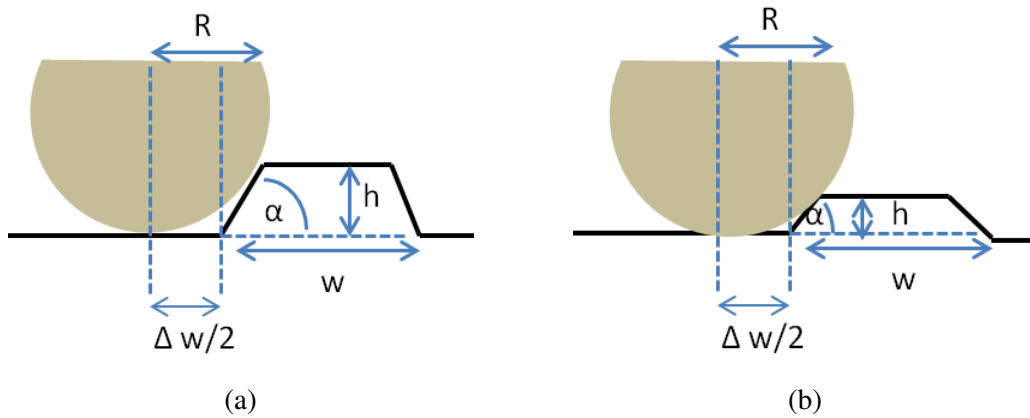


Fig. 2.3 Interaction of a hemisphere with a flat-topped island for the cases a) $h > R(1 - \cos(\alpha))$ and b) $h < R(1 - \cos(\alpha))$ adapted from [43].

Similarly, when measuring depth rather than height the ability of the tip to penetrate into the spaces being measured is also a crucial consideration, as shown in Fig.???. Thus, increasing the gradient of the tip and minimizing the tip apex are desirable to reduce tip-related measurement artefacts when performing atomic force microscopy.

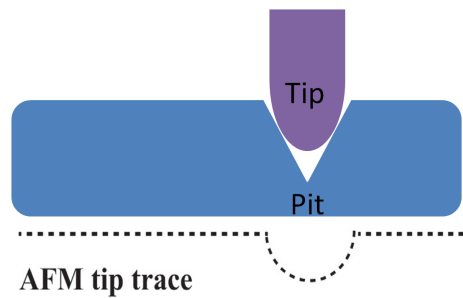


Fig. 2.4 Measurement error in the depth of a pit caused by the finite width of the AFM tip.

Conductive Atomic Force Microscopy

Conductive atomic force microscopy is a technique which combines AFM with local conductivity measurements. In order to perform C-AFM a conductive probe tip is brought close to contact with the sample and a bias is applied between the tip and the sample. The short tip-sample separation causes electron wavefunctions in the tip and sample to overlap, thus allowing a tunneling current to be generated through the application of a bias. C-AFM thus allows for the simultaneous measurement of tip-sample current flow and surface morphology, making it an extremely powerful tool to probe local sample conductivity.

2.2 Hyperspectral Electroluminescence Mapping

2.3 Scanning Electron Microscopy and Cathodoluminescence

2.4 Photoluminescence

2.5 Dual Beam Scanning Electron microscopy with a Focussed Ion Beam

2.5.1 Fabrication

2.5.2 Tomography

2.6 Transmission Electron Microscopy

2.6.1 Tomography

Chapter 3

Inhomogeneous Electroluminescence in InGaN QW LEDs

3.1 Background

$\text{In}_x\text{Ga}_{1-x}\text{N}/\text{GaN}$ quantum well (QW) structures are key structures in present day light emitting diodes in the visible wavelengths. Despite the growth of III-nitride LEDs into a gigantic market with a projected overall worth of 64 billion EUR by 2020, III-nitride alloys suffer from a plethora of material issues arising from heteroepitaxial growth on foreign substrates with large lattice mismatches [6]. A notorious issue in III-nitride growth is the high density of threading dislocations which are the source of highly undesirable effects in diode structures such as non-radiative recombination [44] and leakage current [6].

Threading dislocations have been shown to result in inverted pyramidal defects at the surface of nitride epilayers, known as 'V defects'. The effect of these defects on LED performance is hotly debated in literature as they are expected by many to hinder LED performance due to their association with TDs. However, it has been shown that narrower QWs along the sidewalls of V-defects serve to screen carriers from the non-radiative centres at TDs

Enumeration

1. The first topic is dull
2. The second topic is duller
 - (a) The first subtopic is silly
 - (b) The second subtopic is stupid

3. The third topic is the dullest

itemize

- The first topic is dull
- The second topic is duller
 - The first subtopic is silly
 - The second subtopic is stupid
- The third topic is the dullest

description

The first topic is dull

The second topic is duller

The first subtopic is silly

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The third topic is the dullest

3.2 Hidden section

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Subplots

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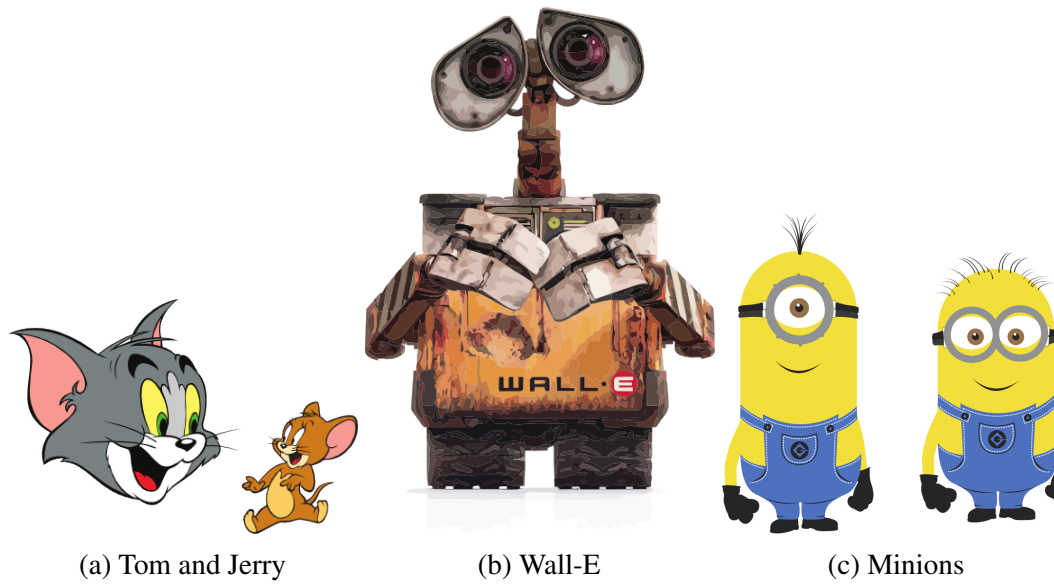


Fig. 3.1 Best Animations

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Appendix A

How to install L^AT_EX

Windows OS

TeXLive package - full version

1. Download the TeXLive ISO (2.2GB) from
<https://www.tug.org/texlive/>
2. Download WinCDEmu (if you don't have a virtual drive) from
<http://wincdemu.sysprogs.org/download/>
3. To install Windows CD Emulator follow the instructions at
<http://wincdemu.sysprogs.org/tutorials/install/>
4. Right click the iso and mount it using the WinCDEmu as shown in
<http://wincdemu.sysprogs.org/tutorials/mount/>
5. Open your virtual drive and run setup.pl

or

Basic MikTeX - T_EX distribution

1. Download Basic-MiK_TE_X(32bit or 64bit) from
<http://miktex.org/download>
2. Run the installer
3. To add a new package go to Start » All Programs » MikTeX » Maintenance (Admin)
and choose Package Manager

4. Select or search for packages to install

TexStudio - T_EX editor

1. Download TexStudio from
<http://texstudio.sourceforge.net/#downloads>
2. Run the installer

Mac OS X

MacTeX - T_EX distribution

1. Download the file from
<https://www.tug.org/mactex/>
2. Extract and double click to run the installer. It does the entire configuration, sit back and relax.

TexStudio - T_EX editor

1. Download TexStudio from
<http://texstudio.sourceforge.net/#downloads>
2. Extract and Start

Unix/Linux

TeXLive - T_EX distribution

Getting the distribution:

1. TeXLive can be downloaded from
<http://www.tug.org/texlive/acquire-netinstall.html>.
2. TeXLive is provided by most operating system you can use (rpm,apt-get or yum) to get TeXLive distributions

Installation

1. Mount the ISO file in the mnt directory

```
mount -t iso9660 -o ro,loop,noauto /your/texlive####.iso /mnt
```

2. Install wget on your OS (use rpm, apt-get or yum install)
3. Run the installer script install-tl.

```
cd /your/download/directory
./install-tl
```

4. Enter command 'i' for installation
5. Post-Installation configuration:
<http://www.tug.org/texlive/doc/texlive-en/texlive-en.html#x1-320003.4.1>
6. Set the path for the directory of TexLive binaries in your .bashrc file

For 32bit OS

For Bourne-compatible shells such as bash, and using Intel x86 GNU/Linux and a default directory setup as an example, the file to edit might be

```
edit ~/.bashrc file and add following lines
PATH=/usr/local/texlive/2011/bin/i386-linux:$PATH;
export PATH
MANPATH=/usr/local/texlive/2011/texmf/doc/man:$MANPATH;
export MANPATH
INFOPATH=/usr/local/texlive/2011/texmf/doc/info:$INFOPATH;
export INFOPATH
```

For 64bit OS

```
edit ~/.bashrc file and add following lines
PATH=/usr/local/texlive/2011/bin/x86_64-linux:$PATH;
export PATH
MANPATH=/usr/local/texlive/2011/texmf/doc/man:$MANPATH;
export MANPATH
```

```
INFOPATH=/usr/local/texlive/2011/texmf/doc/info:$INFOPATH;  
export INFOPATH
```

Fedora/RedHat/CentOS:

```
sudo yum install texlive  
sudo yum install psutils
```

SUSE:

```
sudo zypper install texlive
```

Debian/Ubuntu:

```
sudo apt-get install texlive texlive-latex-extra  
sudo apt-get install psutils
```

Appendix B

Installing the CUED class file

\LaTeX .cls files can be accessed system-wide when they are placed in the $\langle\text{texmf}\rangle/\text{tex}/\text{latex}$ directory, where $\langle\text{texmf}\rangle$ is the root directory of the user's \TeX installation. On systems that have a local texmf tree ($\langle\text{texmflocal}\rangle$), which may be named “ texmf-local ” or “ localtexmf ”, it may be advisable to install packages in $\langle\text{texmflocal}\rangle$, rather than $\langle\text{texmf}\rangle$ as the contents of the former, unlike that of the latter, are preserved after the \LaTeX system is reinstalled and/or upgraded.

It is recommended that the user create a subdirectory $\langle\text{texmf}\rangle/\text{tex}/\text{latex}/\text{CUED}$ for all CUED related \LaTeX class and package files. On some \LaTeX systems, the directory look-up tables will need to be refreshed after making additions or deletions to the system files. For \TeX Live systems this is accomplished via executing “ texhash ” as root. \TeX Live users can run “ initexmf -u ” to accomplish the same thing.

Users not willing or able to install the files system-wide can install them in their personal directories, but will then have to provide the path (full or relative) in addition to the filename when referring to them in \LaTeX .

