Thermal conductivity of $\text{Al}_x\text{Ga}_{1-x}\text{N}$ and $\beta\text{-Ga}_2\text{O}_3$ semiconductors

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ABSTRACT

For the high-power (HP) electronic applications the existing Si-based devices have reached the performance limits governed by the material properties. Hence the device innovation itself is unable to enhance the overall performance. GaN, a semiconductor with wide bandgap, high critical breakdown field, and high electronic saturation velocity is regarded as an alternative of Si. The material properties of GaN make it very suitable for fast-switching HP electronic devices and contribute to the fast growing of GaN technology. The state-of-the-art GaN devices operating up to 650 V have recently become commercially available. Further goal is to reach higher breakdown voltage which can be done via device engineering and material growth optimization.

$\text{Al}_x\text{Ga}_{1-x}\text{N}$ is an ultrawide-bandgap (UWBG) semiconductor which is considered as a natural choice for next generation in the development of GaN-based HP electronic devices. This material attracts particular interest due to the possibility for bandgap tuning from 3.4 eV to 6 eV which allows nonlinear increase of avalanche breakdown field. Furthermore, both *n*- and *p*-type conductivity can be achieved on this material permitting variety of device design with reduced energy losses during operation. $\beta\text{-Ga}_2\text{O}_3$ is also a promising material for HP electronics because of its ultra-wide bandgap (4.8 eV) and a huge value of Baliga's figure of merit (FOM) exceeding by far that of GaN. More interesting feature making this material attractive is the availability of low-cost natural substrates, and then the possibility to obtain high crystal quality of device structures.

For the HP electronic devices thermal conductivity is one of the key parameters determining the device's performance. The initial studies have shown that the thermal conductivity of $\text{Al}_x\text{Ga}_{1-x}\text{N}$ and $\beta\text{-Ga}_2\text{O}_3$ is quite low comparing with that of GaN. This is one of the biggest challenges slowing the development of these materials for HP device applications. Nevertheless, $\text{Al}_x\text{Ga}_{1-x}\text{N}$ - and $\beta\text{-Ga}_2\text{O}_3$ -based field-effect transistors and Schottky-barrier diodes have been demonstrated showing performances superior to that of GaN. To optimize and maintain good performance and reliability, heat generated in the device active regions has to be effectively dissipated. Therefore the thermal conductivity of the materials in the device structures needs to be systematically studied and accurately determined. This information is critically important for the thermal management of the devices.

Transient thermoreflectance (TTR) is a contactless nondestructive method for measuring of the thermal conductivity of materials. TTR, which is based on a pump-probe technique, has shown its potential in evaluation of the thermal conductivity in bulk crystals as well as in thin layers in hetero-epitaxial structures. The method requires an analysis of experimental data based on the fit of thermoreflectance transients with the solution of the

one-dimensional heat transport equations by a least-square minimization of the fitting parameters. Such a procedure allows to extract not only the thermal conductivity of the constituent materials in the structures, but also the thermal boundary resistance at different hetero-interfaces.

The main research results of the graduate studies presented in this licentiate thesis are summarized in three scientific papers.

Paper I. In this paper thermal conductivity of $\beta\text{-Ga}_2\text{O}_3$ and high Al-content $\text{Al}_x\text{Ga}_{1-x}\text{N}$ thin layers was studied. For $\beta\text{-Ga}_2\text{O}_3$ the effects of Sn doping and phonon-boundary scattering on the reduction of thermal conductivity were discussed. For the $\text{Al}_x\text{Ga}_{1-x}\text{N}$ we studied the effect of Al-Ga alloying which gives rise to phonon-alloy scattering. It was found that this scattering process accounts for low thermal conductivity of this material. Finally, a comparison for the thermal conductivity of the two materials was made.

Paper II. In this paper the effect of layer thickness on the thermal conductivity of $\text{Al}_x\text{Ga}_{1-x}\text{N}$ layers grown by HVPE were investigated. Due to Al alloying the thermal conductivity of this material is degraded and reduced by more than one order of magnitude. On top of that we also observed further reduction of thermal conductivity when the layer thickness goes thinner. The mechanism of this phenomenon has been revealed by studying the phonon transport properties in bulk crystal and thin layer.

Paper III. This study emphasizes the role of defects in GaN and $\text{Al}_x\text{Ga}_{1-x}\text{N}$ to the thermal conductivity of these materials. The dislocations, impurities, free carries, and random alloying have been separately studied and discussed. Thermal conductivity of samples containing these defects with various concentrations was measured and the results were interpreted by a theoretical model based on relaxation time approximation (RTA).

POPULÄRVETENSKAPLIG SAMMANFATTNING

Elektronik för högeffektstillämpningar är ett fält som ständigt växer, såväl kommersiellt inom trådlös teknologi som militärt inom högfrekvensområden. Existerande teknik är kisel-baserad (Si) och har nu nått sin kapacitetsgräns med tanke på hur höga effekter som kan genereras. Galliumnitrid (GaN) är en halvledare likt Si, men med ett brett bandgap, ett högt kritiskt överslagsfält samt en hög elektronisk satureringshastighet, som potentiellt är en ersättare till Si-tekniken. GaN besitter materialegenskaper som är väldigt användbara inom snabbväxande högeffektselektronik och bidrar till nu till en mängd teknologiska framsteg. GaN används redan kommersiellt i komponenter upp till 650 V, tex strömbrytare och kraftöverföring. Ett mål är att pressa tekniken till högre effekter och materialet tillåter det. Aluminiumgalliumnitrid ($\text{Al}_x\text{Ga}_{1-x}\text{N}$) är en halvledare med ett ultrabrett bandgap (UWBG) som tillåter justerbarhet av bandgapet mellan 3.4 och 6 eV genom kontroll av legeringens komposition (x). Dessutom kan både n- och p-ledningsförmåga genereras i $\text{Al}_x\text{Ga}_{1-x}\text{N}$ vilket kan leda till bättre designmöjligheter och därav en minskning av energiförluster. $\beta\text{-Ga}_2\text{O}_3$ är ytterligare ett material som ansetts kunna konkurrera inom högeffektstillämpningar på grund av dess höga bandgap (4.8 eV) samt det extremt höga värdet på Baliga's skalan ($\text{BV}^2/\text{R}(\text{on})$), som beskriver hur komponenter kan graderas genom att skatta överslagsspänningen (BV) relativt på-lägesresistansen ($\text{R}(\text{on})$). Dessutom kan $\beta\text{-Ga}_2\text{O}_3$ växas kostnadseffektivt på lättillgängliga substrat vilket ökar möjligheterna för tillväxtoptimering. Termisk ledningsförmåga är den primära materialegenskapen som sätter prestandan på högeffektselektronik. Tidigare studier har visat att termisk ledningsförmåga i $\text{Al}_x\text{Ga}_{1-x}\text{N}$ och $\beta\text{-Ga}_2\text{O}_3$ är relativt låg jämfört med GaN. Därför är det en av de största utmaningarna som forskningen ställs inför för att kunna utveckla tekniken. Trots detta har $\text{Al}_x\text{Ga}_{1-x}\text{N}$ och $\beta\text{-Ga}_2\text{O}_3$ fälteffektstransistorer och Schottky-barriärdioder realiserats med en prestanda som överglänser GaN. För att optimera och stabilisera komponentprestanda krävs god kontroll av termisk avledning. Det är därför termisk ledningsförmåga är så viktigt att studeras systematiskt och kunna bestämmas med hög signifikans. Sådan information är kritisk för termisk kontroll inom komponenter. Termoreflektivita transienter (TTR) är en kontaktlös icke-destruktiv karaktäriseringsmetod för att mäta termisk ledningsförmåga. TTR har visats effektiv både för mätning av bulkmaterial och för tunna filmer och heterostrukturer. Metoden kräver en modell-baserad analys av transienter och resulterar i en endimensionell värmetransportsekvation genom minstakvadratmetodminimering. Proceduren tillåter både elementbaserad termisk ledningsförmåga att studeras samt en utvärdering av den termiska gränsresistansen.

I denna licentiatavhandling har den termiska ledningsförmågan i tunna lager av

$\text{Al}_x\text{Ga}_{1-x}\text{N}$ och $\beta\text{-Ga}_2\text{O}_3$ studerats. Effekten av legering av Al-Ga i $\text{Al}_x\text{Ga}_{1-x}\text{N}$ utvärderades samt effekten av Sn-dopning undersöktes för $\beta\text{-Ga}_2\text{O}_3$. Även lagertjocklek och defekter och dess effekt på termisk ledningsförmåga har undersökts. Sammantaget visar vi med hjälp av teoretiska modeller hur fononspridning, fonontransport påverkas av legering, tjockleksberoendet, och hur olika orenheter influerar den termiska ledningsförmågan.

PREFACE

The licentiate thesis is written based on knowledge and research results accumulated during the graduate studies of Tran Quoc Dat at the Center for III-Nitride Technology (C₃NiT-Janzen), the Department of Physics, Chemistry and Biology in the Linköping University from September 2018 to December 2020. The licentiate thesis is based on scientific papers and contains two main parts: the first part provides a brief introduction to the research field and the second part presents the main research results summarized in three scientific papers.

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Part I

1.1 Introduction

The worldwide quest for a stable energy supply and a reduced greenhouse gas emission has increased demands for new energy sources to replace fossil fuels as well as for innovative technologies for efficient energy usage. One of these technologies is the power electronics - a technology for conversion and control of electrical power for energy generation, transmission, and storage. Key elements of any power electronic systems are the semiconductor switching devices. Over the last several decades, Si-based devices are building blocks in the majority of power electronic converters and their performance has reached maximum limit imposed by Si material properties.

Recently, power electronic devices based on wide-bandgap semiconductors such as SiC and GaN have penetrated into the market, which are just in the beginning stage of their development. Compared with the traditional Si, the wide-bandgap semiconductors show superior material parameters including the high critical breakdown field, high electron mobility, and high electron saturation velocity. Therefore, SiC- and GaN-based power devices have excellent electrical properties including high voltage capacity, fast switching speed, and small on-state resistance [1].

The study of wide-bandgap materials for power electronic devices has attracted enormous attention from broad audiences and it is now reaching a mature state. Thus, the research community has started to look to the next generation of semiconductor materials, the so-called ultra wide-bandgap (UWBG) semiconductors. $\text{Al}_x\text{Ga}_{1-x}\text{N}$ alloy with tunable bandgap between GaN (3.4 eV) and AlN (6.0 eV) has received a lot of attention. The main factor driving this interest is the dependence of the critical electric field for the avalanche breakdown (E_c) on the bandgap energy (E_g), which is predicted to be $E_c \propto E_g^n$ with $n = 2 \div 2.5$ [2]. Furthermore, E_c is the key parameter in the determination of the figure-of-merit (FOM) of the power device performance. This feature together with the

well-established GaN technology makes $\text{Al}_x\text{Ga}_{1-x}\text{N}$ a natural material choice for the next generation power electronic devices.

$\beta\text{-Ga}_2\text{O}_3$ has also gained great attention as it could offer power device performance beyond the capabilities of the current SiC and GaN technology. In addition, to the matured crystal growth and n -type doping technique which enable high crystal quality and low-cost features, $\beta\text{-Ga}_2\text{O}_3$ possesses more superior material properties than GaN and SiC such as larger bandgap of 4.8 eV [3] leading to E_c of 8 MVcm^{-1} . The Baliga's figure of merit (BFOM) of $\beta\text{-Ga}_2\text{O}_3$ is about 3444, several times greater than those of SiC and GaN [4]. These material properties provide strong motivation for the development of $\beta\text{-Ga}_2\text{O}_3$ -based power electronic devices.

Since the power electronic devices target an operation at high currents and high voltages, the Joule heat generation in the device active region become a problem for the device performance and the reliability. The self-heating leads to increased junction temperature which not only reduces the electron mobility and the electron saturation velocity, but also causes a thermal degradation of the devices. Self-heating has been extensively studied in GaN high electron mobility transistors (HEMTs) and shown to cause a current degradation as well as thermal-induced reduction of the breakdown voltage [5–7]. The self-heating behavior of $\text{Al}_x\text{Ga}_{1-x}\text{N}$ - and $\beta\text{-Ga}_2\text{O}_3$ -based devices have been not well explored. Recently, it has been shown that the self-heating problem in $\text{Al}_x\text{Ga}_{1-x}\text{N}$ -channel HEMTs is more severe than that in GaN-channel HEMTs because of the anticipated low thermal conductivity in $\text{Al}_x\text{Ga}_{1-x}\text{N}$ alloys [8, 9]. Similarly, severe self-heating in $\beta\text{-Ga}_2\text{O}_3$ field-effect transistors (FETs) has been also reported [10] and related to the quite low thermal conductivity of this material. Therefore, for a proper thermal management it is necessary to know the accurate thermal conductivity of the thin and thick layers used in the device structures, and more importantly to figure out the reason for the reduced thermal conductivity of UWBS materials.

Over the years significant processes have been made in development of thermal conductivity characterization techniques [11]. Among the commonly used steady-state heat flow and 3ω methods, transient thermoreflectance have shown its potential for characterization of hetero-epitaxial layers in device structures. With the use of a pump-probe technique, this method offers advantages of contactless and nondestructive measurements. There are several variants of the transient techniques including transient thermoreflectance (TTR) [12], time-domain thermoreflectance (TDTR) [13], and frequency-domain thermoreflectance (FDTR) [14]. All of them require a subsequent data analysis by doing fitting of the thermoreflectance transients based on the solution of the thermal transport equation and least-square minimization of the fitting parameters. The data analysis allows to obtain not only the thermal conductivity but also other thermal properties such as thermal boundary

resistance at the hetero-interfaces and the specific heat.

1.2 Fundamentals of heat transport in solids

1.2.1 Heat transport equation – Fourier law

Conduction, radiation, and convection are three commonly known forms of heat transport. In solid state materials, electrons and phonons play as energy carriers allowing the heat to propagate predominantly by conduction mechanism. An empirical law stated by Joseph Fourier says that *the heat flux resulting from thermal conduction is proportional to the magnitude of temperature gradient and opposite to it in sign*. If we call the proportionality constant, k , the Fourier's law can be expressed by [15]

$$\vec{Q} = -k\nabla T \quad (1.1)$$

where $\vec{Q}$ is heat flux vector, k is the thermal conductivity and T is the temperature. The minus sign of the right term reflects the direction of heat flux, which is always flows from the high temperature region to the low temperature region. On the basis of the first law of thermodynamics for a closed system and the Fourier's law of heat transfer, the heat diffusion equation in three dimensions can be expressed by [15]

$$\nabla k \nabla T + \dot{Q} = \rho C \frac{\delta T}{\delta t} \quad (1.2)$$

with $\dot{Q}$ the volume metric heat, ρ the density, and C the specific heat.

1.2.2 Phonon transport – Boltzmann transport equation (BTE)

In semiconductors, the heat transport is predominantly carried by phonons. According to quantum mechanics the energy of a vibration mode with a frequency ω is quantized as of $(n + 1/2)\hbar\omega$, where n is a positive integer, ω is angular frequency and $\hbar$ is reduced Planck's constant. Due to the energy exchange among different modes, the average value, N^0 , of n at thermal equilibrium follows Bose-Einstein distribution

$$N^0 = \frac{1}{\exp(\hbar\omega/k_B T) - 1} \quad (1.3)$$

The N^0 can be understood as the average number of thermally excited quanta of energy $\hbar\omega$. The quanta of vibration energy are called phonons by analogy with the photons of electromagnetic theory.

The phonons of normal vibration modes are spread uniformly throughout the crystal. If there exists a temperature gradient in a crystal which has finite conductivity, there must be interactions which attenuate the vibration energy leading to an inexact correspondence between atomic motions and the normal modes. The thermal energy is considered as propagating of wave packets composed of almost normal modes, which are well localized and travel with the phonon group velocity $v_g = d\omega/dq$. The attenuation is taken into account by letting the number of phonons to vary with position. Then, in a particular region, the number of quanta in the mode q forms a part of wave packet now becomes $N(q)$.

In non-equilibrium thermodynamic system, e. g. in the presence of a temperature gradient within a defined space, the evolution of the distribution function can be described by Boltzmann transport equation (BTE) which has been introduced by Ludwig Boltzmann in 1892 and is expressed as

$$\left(\frac{\partial N}{\partial t}\right)_{drift} + \left(\frac{\partial N}{\partial t}\right)_{scatt} = 0 \quad (1.4)$$

The first term in the equation describes the rate of the phonon number change in the temperature gradient, the second term captures phonon scattering and is called the collision term. The drift term can be expressed in the form [16]

$$\left(\frac{\partial N}{\partial t}\right)_{drift} = -(v_g \cdot \nabla T) \frac{\partial N}{\partial T} \quad (1.5)$$

Following relaxation time approach, the scattering processes are assumed to restore the phonon distribution to the thermal equilibrium one at a rate proportional to the departure of the distribution from equilibrium. Then, the scattering rate can be described as

$$\left(\frac{\partial N}{\partial t}\right)_{scatt} = \frac{N^0 - N}{\tau} \quad (1.6)$$

with τ being the relaxation time. τ , N^0 , and N all depend on wave vector q and phonon polarization.

1.2.3 Solution of the BTE – first principles calculations

By solving linearized BTE equation, the thermal conductivity in the direction of the temperature gradient α can be calculated by

$$k = \sum_{\lambda} C_{\lambda} v_{\lambda\alpha}^2 \tau_{\lambda\alpha} \quad (1.7)$$

Here C_λ is the specific heat capacity per phonon mode and the sum is over all phonon modes $\lambda = (\mathbf{q}, j)$ with $\mathbf{q}$ being the phonon wave vector and j representing the phonon branches. $\tau_{\lambda\alpha}$ is the phonon life time, which involves both the intrinsic anharmonic scattering processes and the extrinsic scattering processes such as isotope impurities, defects, and crystal boundary. $\tau_{\lambda\alpha}$ can be extracted from the solution of the linearized BTE and is expressed by [17–19]

$$\tau_{\lambda\alpha} = \tau_\lambda^0 + \tau_\lambda^0 \left[\sum_+ \Delta^+ (\Gamma_{\lambda\lambda'\lambda''}^{(+)} \tau_{\lambda'\alpha'} \tau_{\lambda''\alpha}) + \sum_- \Delta^- (\Gamma_{\lambda\lambda'\lambda''}^{(-)} \tau_{\lambda'\alpha'} \tau_{\lambda''\alpha}) \right]. \quad (1.8)$$

Here, "+" and "-" represent two possible mechanisms for three-phonon interaction including "creation" and "annihilation" of modes that satisfy the energy and momentum conversation. The functions $\Delta^\pm$ depend on phonon lifetimes in mode λ' and λ'' and on the anharmonic scattering rate for processes $\Gamma_{\lambda\lambda'\lambda''}^{(+)}$ and $\Gamma_{\lambda\lambda'\lambda''}^{(-)}$. τ_λ^0 is given by

$$1/\tau_\lambda^0 = \sum_{\lambda'\lambda''} \Gamma_{\lambda\lambda'\lambda''}^{(+)} + 1/2 \sum_{\lambda'\lambda''} \Gamma_{\lambda\lambda'\lambda''}^{(-)} + 1/\tau^{ext}. \quad (1.9)$$

and the intrinsic three-phonon probabilities are

$$\Gamma_{\lambda\lambda'\lambda''}^{(\pm)} = \frac{\pi\hbar}{4N_0\omega_\lambda\omega_{\lambda'}\omega_{\lambda''}} \left(\begin{matrix} N_{\lambda'}^0 - N_{\lambda''}^0 \\ N_{\lambda'}^0 + N_{\lambda''}^0 + 1 \end{matrix} \right) |\Phi_{\lambda,\pm\lambda',-\lambda''}^{(\pm)}|^2 \delta(\omega_\lambda \pm \omega_{\lambda'} - \omega_{\lambda''}). \quad (1.10)$$

In the above equation, N_0 is the number of unit cells in the crystal and $\Phi_{\lambda,\pm\lambda',-\lambda''}$ is the three-phonon matrix elements. The later are given by

$$\Phi_{\lambda,\pm\lambda',-\lambda''} = \sum_{\kappa} \sum_{l'\kappa'} \sum_{l''\kappa''} \sum_{\alpha\beta\gamma} \Phi_{\alpha\beta\gamma}(\kappa, l'\kappa', l''\kappa'') \frac{e_{\alpha\kappa}^{\lambda'} e_{\beta\kappa'}^{\lambda'} e_{\gamma\kappa''}^{\lambda''}}{\sqrt{M_\kappa M_{\kappa'} M_{\kappa''}}} e^{iq' R_{l'}} e^{iq'' R_{l''}}. \quad (1.11)$$

where $\Phi_{\alpha\beta\gamma}(\kappa, l'\kappa', l''\kappa'')$ are third-order anharmonic interatomic force constant (IFCs), $e_{\alpha\kappa}^{\lambda}$ denotes the α^{th} component of the phonon eigenvector for the κ^{th} atom in the unit cell with mass M_κ , and R_l is the lattice vector in real space locating in the l^{th} unit cell.

1.2.4 Solution of the BTE – Relaxation Time Approximation (RTA).

Callaway's and modified Callaway's models

The relaxation time approximation (RTA) for solution of BTE assumes that the phonon distribution is restored to the equilibrium state by phonon scattering via intrinsic and extrinsic processes. Furthermore, the phonon distribution under a temperature gradient is

not change greatly from the equilibrium one, so $\partial N / \partial T = \partial N^0 / \partial T$. From Eqns (1.4), (1.5), and (1.6) the one-dimensional transport equation can be written as

$$v_g^z \frac{\partial N^0}{\partial T} \frac{\partial T}{\partial z} = \frac{N^0 - N}{\tau} \quad (1.12)$$

The total heat current $\mathbf{h}$ carried by all phonon modes is

$$\mathbf{h} = \sum N(q) \hbar \omega v_g(q) = - \sum \hbar \omega (v_g^z)^2 \tau \frac{\partial N^0}{\partial T} \frac{\partial T}{\partial z} \quad (1.13)$$

and the thermal conductivity is given by

$$\kappa = - \frac{h}{\partial T / \partial z} = \sum \hbar \omega (v_g^z)^2 \tau \frac{\partial N^0}{\partial T} \quad (1.14)$$

By integrating over wave vector q the thermal conductivity becomes

$$\kappa = \int \hbar \omega (v_g^z)^2 \tau \frac{\partial N^0}{\partial T} f(q) dq \quad (1.15)$$

where $f(q) dq$ is number of phonon modes between q and dq per unit volume. v_g^z represents group velocity of a single mode along z direction. Because of three degree of freedom we can express $(v_g^z)^2$ as $\frac{1}{3} v_g^2$. By applying the fundamental relation $f(q) dq = q^2 / 2\pi^2 dq$ for single mode we obtain

$$\kappa = \frac{\hbar^2}{6\pi^2 k_B T^2} \int \tau \frac{\exp(\hbar \omega / k_B T)}{[\exp(\hbar \omega / k_B T) - 1]^2} v_g^2 \omega^2 q^2 dq \quad (1.16)$$

For simplicity the Debye approximation in which the acoustic phonons are assumed to have no dispersion, i. e. $\omega(q) = vq$, where v is the phase velocity, is applied. By using ω as integration variable, the thermal conductivity of a single mode becomes

$$\kappa = \frac{\hbar^2}{6\pi^2 k_B v T^2} \int_0^{\omega_{max}} \tau \frac{\exp(\hbar \omega / k_B T)}{[\exp(\hbar \omega / k_B T) - 1]^2} \omega^4 d\omega \quad (1.17)$$

Instead of variable ω , a new variable $y = \hbar \omega / k_B T$ is commonly introduced. The maximum frequency can be expressed in terms of Debye temperature with $\omega_{max} = \theta_D (k_B / \hbar)$. Then, Eqn (1.17) is written as

$$\kappa = \frac{k_B^4}{6\pi^2 \hbar^3 v} T^3 \int_0^{\theta_D / T} \tau \frac{y^4 e^y}{(e^y - 1)^2} dy \quad (1.18)$$

In the relaxation time τ , all resistive phonon scattering processes, i. e. those which do not conserve wave vector q , are included via the Matthiessen rule, so $\tau = \tau_R$. However, there is a three-phonon scattering process which conserves both the energy and the wave

vector, so-called normal phonon scattering. This process does not contribute to the heat resistance but affect the phonon distribution. If the normal scattering time τ_N is added into τ_R , the thermal conductivity calculated by Eq. (1.18) will be underestimated. In such a case, a correction term needs to be added. In 1959, Callaway has derived an expression for the thermal conductivity in solids within the frame of RTA and introduced a correction term to compensate the lost of thermal conductivity due to τ_N [20]. In his derivation, the three acoustic mode are assumed to have the same sound velocity and to equally contribute to thermal conduction. Lately, the model was improved to separately account for each phonon mode [21]. This is commonly known as a modified Callaway model. In modified Callaway model, the lattice thermal conductivity is given by [22]

$$\kappa = \sum_s \frac{k_B^4}{6\pi^2\hbar^3 v_s} T^3 \int_0^{\theta_D^s/T} \tau_C^s \frac{y^4 e^y}{(e^y - 1)^2} dy + \sum_s \frac{k_B^4}{2\pi^2\hbar^3 v_s} T^3 \frac{\left[\int_0^{\theta_D^s/T} \frac{\tau_C^s}{\tau_N^s} \frac{y^4 e^y}{(e^y - 1)^2} dy \right]^2}{\left[\int_0^{\theta_D^s/T} \frac{\tau_C^s}{\tau_N^s \tau_R^s} \frac{y^4 e^y}{(e^y - 1)^2} dy \right]} \quad (1.19)$$

where $(\tau_C^s)^{-1} = (\tau_N^s)^{-1} + (\tau_R^s)^{-1}$ is total scattering time and the summation is over the all phonon modes (s).

Another improvement of the Callaway's model is to go behind the Debye approximation for the phonon dispersion. In such a case, the real dispersion of the acoustic phonons is taken in to account and the wave-vector dependence of the phonon group velocity is included. Then, the thermal conductivity is expressed by

$$\kappa = \frac{\hbar^2}{6\pi^2 k_B T^2} \int n_0(1 + n_0) \tau_C v_g^2 \omega^2 q^2 dq + \frac{\hbar^2}{6\pi^2 k_B T^2} \int n_0(1 + n_0) \frac{\tau_C \tau_R}{\tau_N} v_g^2 \omega^2 q^2 dq \quad (1.20)$$

with n_0 is the Bose-Einstein distribution which is expressed by $n_0 = 1/(\exp(\hbar\omega/k_B T) - 1)$.

1.2.5 Physics of the phonon scattering

1.2.5.1 Phonon-phonon scattering

Many thermal properties of materials such as the specific heat, the thermal expansion, and the thermal conductivity at low temperatures can be well described by the harmonic approximation for the phonon interactions since the atoms have a small oscillation around their equilibrium positions. At high temperatures, however, this is not the case because of the interaction constant becomes a nonlinear function of the atom displacement. Regarding the thermal conductivity by applying the harmonic approximation, phonons would never

be scattered leading to an infinite thermal conductivity. For a correct description at all temperatures, the anharmonicity in the interaction between the normal modes need to invoke. According to Ziman [23] the Hamiltonian of the cubic term in the phonon-phonon interaction can be written as

$$H_{pp} = \frac{1}{3!} \sum_{l,b;l',b';l'',b''} \eta_{l,b} \eta_{l',b'} \eta_{l'',b''} : \mathbf{A}_{l,b;l',b';l'',b''} (1.21)$$

where $\mathbf{A}_{l,b;l',b';l'',b''}$ is a tensor of third rank in Cartesian coordinates, $\eta_{l,b}$ denotes displacement vector, and $:$ represents the triple inner product.

The three-phonon scattering can be classified into two different mechanisms 1) a new phonon is generated at the expense of destroying two other phonons, and 2) the splitting of a phonon into two other phonons. In both processes, the energy conservation law must be satisfied, i. e.

$$\hbar\omega + \hbar\omega' = \hbar\omega'' (1.22)$$

where $\hbar\omega$, $\hbar\omega'$ and $\hbar\omega''$ are the energies of the three phonon involved. The relation between the phonon wave vectors is given by

$$q + q' = q'' + g (1.23)$$

where q , q' and q'' are the phonon wave vectors, and g is a zero or non-zero wave vector. If g equals zero, i. e. the wave vector is conserved, the three-phonon scattering is called normal (N) phonon-phonon scattering. The N -scattering is a non-resistive process and does not contribute to the thermal resistance. This process, however, modifies the phonon distribution and thus affects all other phonon scattering processes. If g is non-zero, i. e. the wave vector is not conserved, the three-phonon scattering is called Umklapp (U) phonon-phonon scattering. The U -scattering is a thermally resistive process and usually dominates at high temperatures.

The possible three-phonon scattering processes between the transverse (T) and longitudinal (L) acoustic phonons are constrained by some selection rules [23]: (i) there is no process in which all three phonon belong to the same polarization branch of the spectrum, (ii) the created phonon must from a branch lying higher in energy than at least one of the destroyed phonons, and (iii) low-frequency longitudinal phonons can only take part in processes if the total phonon number is proportional to the square of the frequency. Therefore, the allowed processes are

$$T + T \longrightarrow L (1.24)$$

$$T + L \longrightarrow L (1.25)$$

$$L \longrightarrow T + T \quad (1.26)$$

$$L \longrightarrow T + L \quad (1.27)$$

Regarding the optical phonons, their contribution to the thermal conductivity is usually quite small, especially in semiconductors with a large energy gap between the optical (O) and acoustic (A) branches. This is the case for GaN and AlN as described in Section 1.4.1.2. The possible three-phonon scattering processes involving optical phonons are rather restricted by the energy conservation law. If the phonon energy gap is not so large the only processes that can give a contribution are of type $A + A \rightleftharpoons O$ and $A + O \rightleftharpoons O$ [18, 24].

In the modified Callaway model, the N -scattering rate is expressed by [22, 25]

$$(\tau_N^s)^{-1} = \left(\frac{k_B}{\hbar}\right)^{a+b} \frac{\hbar(\gamma_s)^2 (V_o)^{(a+b-2)/3}}{M(v_s)^{a+b}} y^a T^{a+b} \quad (1.28)$$

where M is the atomic average mass, V_o is the atomic volume, v_s is the acoustic phonon velocity, and γ_s is the mode Gruneisen parameter which describe the anharmonicity of the of the crystal lattice. For the case of hexagonal crystals (like GaN and AlN), $(a, b) = (2, 3)$ for LA phonons and $(a, b) = (1, 4)$ for TA phonons at low temperatures, while at high temperatures $(a, b) = (2, 1)$ for LA phonons and $(a, b) = (1, 1)$ for TA phonons.

The U -scattering gives the main contribution to the thermal conductivity in pure and defect-free crystals, and leads to a characteristic $\kappa \propto T^{-1}$ dependence at high temperatures. The form of U -scattering rate is given by [22]

$$(\tau_U^s)^{-1} = \frac{k_B^2(\gamma_s)^2}{\hbar M(v_s)^2 \theta_D} \exp\left(-\frac{\theta_D^s}{3T}\right) y^2 T^3 \quad (1.29)$$

1.2.5.2 Phonon-defect scattering

In principle, a truly pure ("ideal") crystal is hard to be synthesized. Lattice imperfections, referred as defects, always present in the real crystals. The phonons can be scattered by these defects leading to a reduction of the thermal conductivity. There are three groups of lattice imperfections that affect the thermal properties - (i) point defects (impurities, vacancies, atom isotopes, alloying), (ii) structural defects (dislocations, stacking faults, grain boundaries), and (iii) crystal boundaries. The first introduction of the phonon-defect interaction was given by Peierls in 1929 [26]. A detailed analysis of the different phonon-defect scattering processes was given by Klemens [27] and later refined and expanded by Ziman [23].

Below we describe the scattering rate for the phonon-isotope (I), the phonon-point-defect (PD), the phonon-dislocation (D), the phonon-alloy (A) and the phonon-free-electron

(*FE*), and the phonon-boundary (*B*) scattering. In the discussion, we emphasize on the case of $\text{Al}_x\text{Ga}_{1-x}\text{N}$ semiconductor.

I-scattering is a result of the atomic mass fluctuation when various isotopes of the constituent chemical elements are present in the material. In semiconductors, this scattering process is shown to be significant at low temperatures and to reduce the maximum thermal conductivity (typically occurring at $T \approx 0.05\theta_D^s$) by 30 – 50% [22, 28]. In high temperature range ($T > 180$ K for GaN and $T > 250$ K for AlN), the *I*-scattering process still affects the thermal conductivity in undoped or low-doped semiconductors. The *I*-scattering rate is expressed by [22, 27]

$$(\tau_I^s)^{-1} = \left(\frac{k_B}{\hbar}\right)^4 \frac{V_o}{4\pi(v_s)^3} \Gamma_I y^4 T^4 \quad (1.30)$$

In the case of ternary compounds (as $\text{Al}_x\text{Ga}_{1-x}\text{N}$) the strength of the phonon-isotope scattering Γ_I can be written as

$$\Gamma_I = 2 \sum_i \left(c_i \frac{M_i}{M_{tot}}\right)^2 \Gamma_i \quad (1.31)$$

where Γ_i is the scattering parameter for each single chemical elements ($i = \text{Ga}, \text{Al}, \text{N}$), M_i is the atomic mass, c_i is the corresponding fractional composition (i. e. c_i is x for Al, $(1 - x)$ for Ga, and 1 for N), and $M_{tot} = xM_{Al} + (1 - x)M_{Ga} + M_N$, where x is the Al composition. Γ_i scattering parameters are given by [22]

$$\Gamma_i = \sum_j c_j^i \left(\frac{m_j^i - M_i}{M_i}\right)^2 \quad (1.32)$$

with

$$M_i = \sum_j c_j^i m_j^i \quad (1.33)$$

where m_j^i is the atomic mass of j^{th} isotope and c_j^i is the fractional atomic natural abundance.

PD-scattering originates from the lattice disorder caused by point defects (substitutional impurities or vacancies). The expression for the *PD*-scattering rate is written as [27, 29]

$$(\tau_{PD}^s)^{-1} = \left(\frac{k_B}{\hbar}\right)^4 \frac{V_o}{4\pi(v_s)^3} \Gamma_{PD} y^4 T^4 \quad (1.34)$$

where Γ_{PD} is the the strength of the *PD*-scattering. In this scattering process, the phonons are scattered by (i) mass fluctuation between the host atom and the impurity (vacancy), (ii) the change in the force constant of the bonds neighboring to the defect, and (iii) the strain field caused by the size difference between the host atom and the impurity (vacancy). Due

to the lack of information about the force constant in $\text{Al}_x\text{Ga}_{1-x}\text{N}$ the contribution of the force-constant term is assumed to be equivalent to that of the strain-field term. Then, Γ_{PD} can be written as [27, 29]

$$\Gamma_{PD} = \sum_p f_p \left[\left(\frac{M_p - M}{M} \right)^2 + 8(\gamma_s)^2 (1 + Q_b)^2 \left(\frac{\Delta R_p}{R} \right)^2 \right] \quad (1.35)$$

where f_p and M_p are the fractional concentration and the atomic mass of the p^{th} impurity, ΔR_p is the average local displacement in the host lattice due to the p^{th} impurity, and R is the average nearest-neighbor atomic radius in the host crystal. The displacement ΔR_p is calculated from the difference between the average bond length in the host crystal and the average bond length in the presence of an impurity in a relaxed lattice. The constant Q_b equals 4.2 assuming that the nearest-neighbor bonds of the impurity have the same anharmonicity as the bonds in host crystal.[27]

D -scattering arises from the phonon scattering on both the core of dislocation lines and the elastic strain field on the dislocation lines. Taking into account the contributions of edge and screw dislocations, the total D -scattering rate is given by [27, 30]

$$(\tau_D^s)^{-1} = (\tau_{\text{core}}^s)^{-1} + (\tau_{\text{strain}}^s)^{-1} \quad (1.36)$$

where

$$(\tau_{\text{core}}^s)^{-1} = \eta N_D \left(\frac{k_B}{\hbar} \right)^3 \frac{(V_o)^{4/3}}{(v_s)^2} y^3 T^3 \quad (1.37)$$

$$(\tau_{\text{strain}}^s)^{-1} = 0.06\eta \left(\frac{k_B}{\hbar} \right) (\gamma_s)^2 \left[N_S b_S^2 + N_E b_E^2 P_E^2 \right] y T \quad (1.38)$$

and

$$P_E^2 = \frac{1}{2} + \frac{1}{24} \left(\frac{1 - 2\sigma}{1 - \sigma} \right)^2 \left[1 + \sqrt{2} \left(\frac{v_L}{v_T} \right)^2 \right]^2 \quad (1.39)$$

$(\tau_{\text{core}}^s)^{-1}$ and $(\tau_{\text{strain}}^s)^{-1}$ are the scattering rates from the dislocation core and the elastic strain field, respectively. In Eqs. (1.37-1.39), N_D is the total dislocation density, N_S and N_E are the densities of screw and edge dislocations, respectively, η is the weight factor accounting for the mutual orientation of the dislocation line and the direction of the heat transport (assumed to be 0.55 for a random distribution of all dislocation [27, 30], b_S and b_E are the magnitudes of Burgers vectors for the screw and edge dislocations, and σ is the Poisson ratio. Numerical calculations using Eqs. (1.36-1.39) give a threshold dislocation density (i. e. the density at which the thermal conductivity starts to decrease) in the range of $10^{10} - 10^{11} \text{ cm}^{-2}$ [30]. However, the experimental data for GaN [31, 32] indicate that the threshold dislocation density is much lower. Then, we add a dimensionless scaling factor

in the right side of Eq. (1.36) in order to match the experimentally determined threshold density. Paskov *et. al.* [29] have proposed a scaling factor of 1×10^3 . Based on our data for the dislocation density dependence of the thermal conductivity of GaN we have obtained a scaling factor of 4×10^2 , which value is used in the analysis of the experimental data included in this thesis.

The A -scattering is described in the framework of the virtual crystal approximation, in which the disordered alloy is replaced by an ordered virtual crystal with virtual atomic mass, atomic volume and lattice constants [33–35]. Then, the phonons are scattered by the perturbations of the virtual crystal. The expression for A -scattering rate is similar to that for the PD -scattering rate [27], but modified to take into account the coexistence of the substitutional Al into GaN lattice and substitutional Ga into AlN lattice.

$$(\tau_A^s)^{-1} = \left(\frac{k_B}{\hbar} \right)^4 \frac{V_o}{4\pi(v_s)^3} \Gamma_A y^4 T^4 \quad (1.40)$$

where

$$\Gamma_A(x) = \sum_l c_l \left[\left[\frac{\Delta M_l}{M(x)} \right]^2 + 8\gamma_s(x)^2 (1 + Q_b)^2 \left[\frac{\Delta R_l}{R(x)} \right]^2 \right] \quad (1.41)$$

Here, the index l denotes group-III atomic constituents (Ga and Al), c_l is the atomic fractional compositions, $\Delta M_l = M_l - M(x)$, $\Delta R_l = R_l - R(x)$, $M(x) = xM_{Al} + (1 - x)M_{Ga}$, and $R(x) = xR_{Al} + (1 - x)R_{Ga}$. $M(x)$ and $R(x)$ are the averaged atomic mass and atomic radius, respectively, and $\gamma_s(x)$ are the mode Grüneisen parameters of $\text{Al}_x\text{Ga}_{1-x}\text{N}$ alloy. The second term in Eq. (1.41) combines the difference in the atomic radius and the difference in the stiffness constant of the nearest-neighbor bonds.[29] Assuming that the incorporation of Al(Ga) into GaN(AlN) does not change significantly the anharmonicity of nearest-neighbor bonds, the parameter Q_b has a value of 4.2 [27].

FE -scattering is believed to give a negligible contribution to the thermal conductivity in undoped and low-doped semiconductors. In the case of GaN, the FE -scattering was shown to start to play a role at electron concentrations above 10^{17} cm^{-3} [29]. Within the effective mass approximation, the FE -scattering rate is formulated by [36, 37]

$$(\tau_E^s)^{-1} = \frac{(E_1)^2 (m_e)^2 k_B T}{2\pi \hbar^4 \rho v_s} \left[y - \ln \left(\frac{1 + \exp(Z + y/2)}{1 + \exp(Z - x/2)} \right) \right] \quad (1.42)$$

and

$$Z = \frac{y^2 k_B T}{8m_e (v_s)^2} + \frac{m_e (v_s)^2}{2k_B T} - \frac{E_F}{k_B T} \quad (1.43)$$

where E_1 the deformation potential for the electron-acoustic-phonon interaction, m_e the electron effective mass, and E_F the Fermi-level energy measured from the conduction band

edge.

In bulk materials and thick layers, B -scattering is important only at very low temperatures, where the phonon mean free path (MFP) becomes comparable to the dimensions of the sample under investigation [38, 39]. In thin layers, however, the contribution of this scattering process to the thermal conductivity is significant even at room temperatures. The B -scattering rate is assumed to be independent of temperature and phonon frequency and can be expressed as [32, 40]

$$(\tau_B^s)^{-1} = \frac{v_s}{aL} \quad (1.44)$$

where L is the layer thickness and a is a dimensionless parameter. aL can be regarded as an effective MFP along the direction of the heat transport and is essentially determined by the size effect. A value of 2.38 for the a parameter has been found by fitting Eq. (1.44) with the solution of the Boltzmann transport equation for a thin layers [32] and used to model the thickness dependence of the thermal conductivity of GaN [32, 40]. In our analysis, we use $a = 1.5$, the value extracted from the experimental thickness dependence of the thermal conductivity of $\text{Al}_x\text{Ga}_{1-x}\text{N}$ with various compositions [41].

1.3 Experimental techniques for thermal conductivity measurements in solid-state materials

1.3.1 Steady-state method

Steady-state method is a fundamental technique for measuring of thermal conductivity which relies on the Fourier's law of heat conduction. In this method, the temperature difference ΔT between two points along the heat flow through the sample is measured [42]. The thermal conductivity κ is determined by

$$\kappa = \frac{QL}{A\Delta T} \quad (1.45)$$

where Q is the heat flux flowing through the sample, A is the cross-section area, and L is the distance between two sensors with which the temperature difference ΔT is measured. The accuracy of measurements depends strongly on the parasitic heat losses possibly caused by convection and radiation. If the measurement system is optimized, the heat losses can contribute less than 2% of the total heat flow through the sample.

The steady-state methods requires a uniform sample with a large size. Only bulk materials can be measured, not the epitaxial layers. Some advanced variants of this technique such as comparative method and radial heat flow method have been developed.

The former utilizes a standard material with a well-known thermal conductivity, thus avoiding the difficulties in the determination of heat flux and eliminating the uncertainty caused by the parasitic heat losses. The later is commonly used at very high temperatures (above 1000 K).

1.3.2 Dynamic techniques

1.3.2.1 Laser flash method

The flash method for measuring of the thermal diffusivity, the heat capacity, and the thermal conductivity was introduced by Parker *et. al.* in 1961 [43]. In the earlier applications of this method, the sample is heated on the front side by a flash lamp and at the back side the temperature is monitored by a thermocouple. Since the heating is uniform, the heat transport along the sample can be considered as one-dimensional and the temperature changes recorded at the back side is described by the following equation

$$T(z, t) = \frac{Q_A}{C_p} \left[1 + 2 \sum_{n=1}^{\infty} (-1)^n \exp \left[-(\pi n)^2 \frac{t}{\tau_0} \right] \right] \quad (1.46)$$

where Q_A is the total heat absorbed by the sample, C_p is the heat capacity and $\tau_0 = d^2/\alpha$ is the characteristic time for heat diffusion across the sample, where d and α are the sample thickness and the thermal diffusivity, respectively.

The technique was further developed by introducing the use of a laser beam directed by a optical fiber aiming to reduce the error due to nonuniform heating. Temperature at the back side is measured by a fast infrared radiation thermometer allowing a decrease the error due to nonlinear temperature detection [44]. This technique requires several conditions to be held - (i) the laser pulse should be shorter compared to the heat diffusion time, (ii) the sample needs to be uniformly heated, (iii) the sample should be adiabatic during the measurement, and (iv) the sample is opaque to the pulse of light and thermal radiation.

1.3.2.2 3ω method

In 3ω method, a metal strip on the sample surface which plays a role of not only the heat source but also of the thermometer detector, is deposited.

The heat is provided by applying an AC current $I(t) = I_0 \cos(\omega t)$ through the metal strip leading to Joule heating with a power $P(t) = P_0 \cos(2\omega t)$, with $P_0 = I_0^2 R_0/2$, where R_0 is the time-independent electric resistance of the metal. Considering an infinitely narrow heater line the temperature rise ΔT as a function of coordinate r is given by [45]

$$\Delta T(r) = (P/l\pi\kappa)K_0(qr) \quad (1.47)$$

where l is the length of the stripe and K_0 is the zero-order modified Bessel function. The complex quantity $1/q = (D/i2\omega)^{1/2}$ is the thermal penetration depth, where $D = \kappa/\rho C_p$ is the thermal diffusivity. In the limit $|qr| \ll 1$ the above equation can be approximated by [45]

$$\Delta T = \frac{P}{l\pi\kappa} \left[\frac{1}{2} \ln \frac{D}{r^2} + \ln 2 - 0.5772 - \frac{1}{2} \ln(2\omega) - \frac{i\pi}{4} \right] \quad (1.48)$$

In this equation, the real and imaginary terms, which correspond to in-phase ΔT_{in} and out-of-phase ΔT_{out} , respectively, are separated. The thermal conductivity can be linked to the real part of temperature oscillation by

$$\frac{\Delta T_{in}}{P} = \frac{1}{2\pi l\kappa} \ln(\omega) + C \quad (1.49)$$

Here, the constant C is a ω -independent parameter. The in-phase and out-of-phase temperature changes are obtained from the output signal $V_{3\omega}$ by

$$V_{3\omega} = V_{1\omega} \frac{\alpha(\Delta T_{in} + i\Delta T_{out})}{2} \quad (1.50)$$

Finally, the thermal conductivity can be deduced by taking derivative of ΔT_{in} with respect to $\ln(\omega)$. Note that the out-of-phase temperature change ΔT_{out} can also give the thermal conductivity, however, the uncertainty in this case is quite high.

The 3ω method as described above is suitable for measuring the thermal conductivity of bulk crystals or bare substrates. For determination of the thermal conductivity in thick and thin layers an advanced analysis of the raw experimental data has to be applied [46, 47].

1.3.2.3 Transient thermoreflectance (TTR)

Transient thermoreflectance method is based on pump-probe technique and was first proposed by Paddock and Easley in 1986 [48]. The method employs a pulsed pump laser for heating and a continuous-wave probe laser for monitoring the temperature change from the sample surface, hence, it is a contactless and nondestructive. After every pulses of heating laser, the sample experiences a transient decay of temperature on the surface. The change of temperature is recorded by measuring the temperature-dependent reflectance using probing laser. With the use of a ultra-fast photodetector the temperature transient can be fully mapped. Since the temperature decay contains thermal transport information of the material, several thermal properties of the material can be deduced by analyzing

the transients. The TTR method is suitable for various types of samples including opaque or transparent samples. In the case of transparent sample, a metal transducer layer is deposited on top of the sample. The power of the pump laser will heat up the surface of the transducer to a temperature higher than the ambient temperature. Then, the temperature decays due to the heat dissipation through the samples. TTR method was employed for all thermal conductivity measurements in this thesis. The system we used (Transometer D8 (TMX Scientific)) is schematically shown in Fig. 1.1.

Figure 1.1: Schematic setup of the TTR system used in our measurements [49]

By adjusting the laser aperture and the optics, the size of the pump laser beam (with a Gaussian space shape) on sample surface can be defined. By setting the spot size of the pump laser much wider than the size of the probing laser, one-dimensional thermal transport through the sample can be assure. Assuming constant and uniform thermal properties of the sample, the one-dimensional thermal diffusion equation can be written as

$$\frac{\kappa}{\rho C_p} \frac{\partial^2 T(z, t)}{\partial x^2} - \frac{\partial T(z, t)}{\partial t} = \frac{q'''}{\rho C_p} \quad (1.51)$$

where q''' is the heat power per unit volume. The initial boundary conditions applied to

the above equation are $T(z, t) = T_0$ for $t \leq 0$ and $T \rightarrow T_0$ when $z \rightarrow \infty$, where T_0 is the sample temperature before the heating, i. e. ambient temperature. The thermal diffusion equation can be solved by taking Laplace transformation [49]. Since in the most of TTR measurements a thin metal transducer is used, a zero volumetric heat source $\dot{q}'''$ can be assumed. Then, the solution of Eq. (1.51) for the surface temperature becomes

$$T(t) = T_0 + \frac{1}{(\kappa\rho C_p)^{1/2}} \frac{\partial^{-1/2} \dot{q}''(t)}{\partial t^{-1/2}} \quad (1.52)$$

with $\dot{q}''$ being the surface heat flux given by

$$\dot{q}''(t) = (1 - R)I_0 \exp \left[- \left(\frac{t - t_0}{\tau} \right)^2 \right] \quad (1.53)$$

where R is the reflectivity of the metal transducer, and I_0 and τ are the intensity and the pulse width of the pump laser.

The above equations describe the surface temperature decay which is governed by the thermal conductivity and other thermal properties of the material underneath. In the case of multi-layers structures (Fig. 1.2), a more complex analysis of the heat transport across the sample is required. The thickness L , heat capacity C_p , and thermal conductivity κ of each layer, as well as the interface thermal resistance R need to be specified. Only two of those parameters can be extracted by least-square fitting of the reflectance transients. Therefore, series of samples with different number of layers in the stack have to be measured in order to characterize the entire structure.

Figure 1.2: Schematic illustration of a multi-layer structure with an indication of material parameters needed for the reflectance transients analysis.

As seen from Eqs. (1.52 and 1.53), the analysis of the reflectance transients by the

Figure 1.3: Refractive index and extinction coefficient of Au transducer layer.

Figure 1.4: Measured reflectance transients of $\text{Al}_x\text{Ga}_{1-x}\text{N}$ layers with different compositions.

thermal diffusion equation solution requires some input parameters. The surface heat flux and the absorbed power depends on the optical property of the transducer. In our measurements, an Au metal layer is employed as a transducer. The refractive index n and extinction coefficient k of Au layer measured by spectroscopic ellipsometry are shown in Fig. 1.3. Also needed for the analysis is the specific heat C_p for each layers. We have used the available literature data for Au [50], GaN [51], AlN [52], $\beta\text{-Ga}_2\text{O}_3$ [53], Al_2O_3 [54],

and 4H-SiC [55]. The C_p of $\text{Al}_x\text{Ga}_{1-x}\text{N}$ is assumed to depend linearly on the composition between the values for binary compounds.

Figure 1.4 shows the reflectance transients measured for $\text{Al}_x\text{Ga}_{1-x}\text{N}$ layers with different Al compositions. In measurements, the frequency of the pump laser is fixed at 50 Hz, this corresponds to a duration time of 0.02 s between two subsequent laser pulses. Thus, even for a low thermal conductivity material such as $\text{Al}_x\text{Ga}_{1-x}\text{N}$ with $x = 0.5$, the heat is completely dissipated before the coming of next laser pulse. Therefore, the accumulation effects are negligible in our TTR measurements. Note that the TTR method is suitable for measurements of materials with a wide range of thermal conductivity. The maximum and minimum thermal conductivity limits of our system are 400 and 0.2 W/mK, respectively.

1.3.2.4 Time-domain thermoreflectance (TDTR)

TDTR is also a pump-probe technique working on the base principle of the TTR method. The change of the surface temperature after the pump laser heating is measured from the probe beam reflectance transients of a metal transducer. The only difference is that a picosecond pump laser is used. Commonly the spots of heating and probing lasers are of the same order, and then in the analysis, the heat transport in lateral and vertical directions is convoluted. As a consequence, both the in-plane and the out-of-plane thermal conductivities can be extracted. The general equation for temperature variation can be written by [13].

$$C_p \frac{\partial T}{\partial t} = \frac{\kappa_r}{r} \frac{\partial}{\partial r} \left(r \frac{\partial T}{\partial r} \right) + \kappa_z \frac{\partial^2 T}{\partial z^2} \quad (1.54)$$

where κ_r and κ_z are the in-plane and the out-of-plane thermal conductivities, respectively.

A schematic illustration of TDTR setup is given in Fig. 1.5 (a). The pump laser has a frequency of 80 MHz corresponding to the interval time between two pulses of 12.5 ns. This interval is small compared to the diffusion time in the most of studied semiconductor, especially those having low thermal conductivity. Therefore, the heat accumulation from the subsequent laser pulses have to be taken into account. In TDTR measurements, the pump laser is commonly modulated by an electro-optical modulator with a sine function with a frequency in the range of 0.2-2 MHz (Fig. 1.5 (b)). The use of modulation frequency is necessary because it serves as a reference for the lock-in detection of the reflectance signal. By controlling the delay time of the probe laser in respect to the pump laser, the surface temperature response can be extracted from the output signal (Fig. 4 (c, d)). The TDTR has been proved to be a powerful method for thermal characterization of bulk materials and epitaxial layers with low [53] and high [56, 57] thermal conductivity, as well as of multi-layer structures [13].

Figure 1.5: (a) A typical TDTR setup, (b) the shape of modulated pump pulsed laser, (c) the surface temperature response and the probed signals at a fixed delay time, and (d) output signals as functions of delay time. [13]

1.3.2.5 Frequency-domain thermoreflectance (FDTR)

FDTR is another pump-probe method for thermal conductivity measurements. The FDTR setup is basically similar to that is used in the TDTR. However, instead of detection of the amplitude of the output signal, in the FDTR method the change of output signal phase is measured. By analysing the phase changes the thermal properties of the studied samples such as the thermal conductivity, the thermal boundary resistance and the heat capacity can be attained. Using a lock-in detection technique, the amplitude A and the phase ϕ of the output signal, which is the reflectance from the transducer, are monitored in FDTR measurements. Mathematically, the detected signal can be expressed as a complex number $Z(\omega_0)$ such that the output of the lock-in amplifier for a reference wave $e^{i\omega_0 t}$ is given by [14]

$$Ae^{i\omega_0 t + \phi} = Z(\omega_0)e^{i\omega_0 t} \quad (1.55)$$

where ω_0 is the modulation frequency of pulsed pump beam and $Z(\omega_0)$ is

$$Z(\omega_0) = \beta \sum_{k=-\infty}^{\infty} H(\omega_0 + k\omega_s) e^{ik\omega_s \tau_d} \quad (1.56)$$

Here, β is a factor containing the information of sample's thermoreflectance coefficients and the power of lasers, τ_d is delay time between pump and probe pulses, $H(\omega_0)$ is the thermal frequency response of the sample weighted by the material properties and the

pump laser intensity.

The main advantage of FDTR method is the possibility to attain the thermal properties (the thermal conductivity and the interface thermal boundary resistance) in stacked epitaxial structures with very thin layers. However, the data analysis is much more complicated than that in the case of TTR and TDTR [14, 58, 59].

1.4 UWBG semiconductors: properties and growth

1.4.1 GaN, AlN, and AlGaN

1.4.1.1 Crystal structure and electronic properties

The group-III nitride compounds exist in two forms of crystal polytypes, the zinc-blende (ZB) structure of space group $F\bar{4}3M$ and the wurtzite (WZ) structure of space group $P6_3MC$. In both structures each atom (metal or N) is tetrahedrally surrounded by four atoms of other type and the inter-ionic distances between the close packed plane are approximately the same. These two structures can be distinguished from each other by lattice stacking sequence, which is ABAB for (0001) plane of the WZ and ABCABC for (111) plane of the ZB as shown in Fig. 1.6. Since the formation energy per atom in WZ crystal is smaller than that in ZB crystal [60], the WZ structure is the stable one and ZB structure is metastable.

Figure 1.6: Schematic illustration of the unit cell of (a) WZ and (b) ZB crystal structures [60]

The WZ primitive cell contains 4 atoms and is characterized by a and c lattice parameters. The atomic coordinates are $A_1(1/3, 2/3, 0)$, $A_2(2/3, 1/3, 0)$, $N_1(1/3, 2/3, 0.375+u)$, and $N_2(2/3, 1/3, 0.875+u)$, where A_1 and A_2 denote the group-III atoms, and N_1 and N_2 denote nitrogen atoms [61]. In an ideal WZ structure, $u = a^2/c^2 = 3/8 = 0.375$. For bulk GaN, the experimentally determined lattice constants [62] give almost the ideal u value. For AlN, however, the strong ionicity increases the attractive interaction between unlike atoms and leads to an increase of the u parameter.

[61]. Three of those are the acoustic modes - one longitudinal (LA) and two transverse (TA) modes which have lowest frequencies. The other 9 phonon modes belong to optical modes, two of them, B_1 (low) and E_2 (low), are low-lying branches locating close to the acoustic phonons [68]. *Ab initio* calculated phonon dispersion in AlN and GaN is shown in Fig. 1.8 [69]. The calculated dispersion of the acoustic phonon for both GaN and AlN is found to be in a very good agreement with experimental data from inelastic X-ray scattering measurements [70, 71].

Figure 1.8: Phonon dispersion of WZ GaN and WZ AlN calculated along high symmetry lines [69]

In semiconductors and insulators, the phonons are the main heat carriers, therefore, the thermal properties are governed by the phonon scattering processes. The thermal conductivity is mainly determined by the scattering of acoustic phonons since they have high sound velocity and low frequency. The low-lying optical phonons (B_1 (low) and E_2 (low)), which have a high group velocity, can also contribute to the thermal conductivity but the contribution is relatively small [72]. The effect of the high-frequency optical phonons having a small group velocity, is negligible.

1.4.1.3 Growth methods

Since group-III nitrides have a very high melting temperature, the growth of bulk crystals is exceptionally difficult by the conventional melt solidification crystal growth process. Successful growth of bulk GaN crystal has been done by high-pressure high-temperature (HPHT) growth [73] and by ammonothermal growth [74]. Only the ammonothermal (ATh) method can provide bulk GaN substrates for the epitaxial growth of the device structure. The another way to produce GaN substrates is by Hydride Vapor Phase Epitaxy (HVPE) growth on sapphire. In this case, a thick (> 1 mm) GaN layers are grown and the sapphire substrate and the defective region at the surface interface is removed [75]. So far, all 2-inch GaN substrates available on the market are fabricated by ATh and HVPE. As for AlN, substrates can be fabricated from bulk AlN crystals grown by physical vapour transport [76].

The ATh growth method utilizes ammonia as a solvent for the growth of nitride materials. In general, the growth mechanism can be described as follows: (1) a nutrient such as Ga or GaN is dissolved in ammonia solvent under high pressure (100 – 400 MPa) and high temperatures (400 – 600° C); (2) the dissolved nutrient is transported by convective flow of supercritical ammonia to a seed region; (3) the solvent gets saturated in the seed region at higher temperature compared to the nutrient region; (4) GaN starts to crystallize on seed crystals. Typically all these processes are placed in an autoclave which can sustain high pressure and high temperature. Bulk GaN crystals with dislocation of as low as $5 \times 10^3 \text{ cm}^{-2}$ have been demonstrated [77].

The HVPE method is developed on the basis of the vapor phase epitaxy, which uses chloride in order to realize higher growth rate. In a typical HVPE growth of GaN, volatile species of GaCl is formed by the interaction of HCl and liquid Ga at high temperature of about 850°C and carried by hydrogen gas. On the area of substrate surface GaCl reacts with ammonia to form GaN by the following reactions [78]

For growth of the group-III nitrides epitaxial device structures, Metal Organic Chemical Vapor Deposition (MOCVD) is the most commonly used technique [79]. The MOCVD is based on a common CVD growth technique with the use of metal-organic compounds as precursor for group-III elements. Hydride gas such as NH_3 is commonly used for group-V elements. The MOCVD growth technique has unique features such as high growth rate, high purity, high uniformity, capability for abrupt interfaces, selective growth, and large

scale production.

1.4.2 $\beta\text{-Ga}_2\text{O}_3$

1.4.2.1 Crystal structure and electronic properties

There are five polymorphs of Ga_2O_3 called α , β , γ , ϵ , and κ . Among them, $\beta\text{-Ga}_2\text{O}_3$ is the most thermally and chemically stable phases. $\beta\text{-Ga}_2\text{O}_3$ semiconductor crystallizes in base-centered monoclinic system having a space group of C2/m . The lattice constants are $a = 12.23 \text{ \AA}$, $b = 3.04 \text{ \AA}$, and $c = 5.8 \text{ \AA}$, and the angle between a and c is $\beta = 103.7^\circ$ [80]. There are 20 atoms in the unit cell. The different Ga and O atoms occupy inequivalent crystallographic positions. There two types (two positions) of Ga atoms, Ga(I) and Ga(II), and three types of O atoms O(I), O(II), and O(III). Ga(I) and Ga(II) atoms are tetrahedrally and octahedrally coordinated, respectively, while O(I)/O(II) and O(III) are three- and fourfold coordinated, respectively. O(I) shares two bonds with Ga(II) and one bond with Ga(I), O(II) share three bonds with Ga(II) and one bond with Ga(I), and O(III) share two bonds with Ga(I) and one bond with Ga(II). The atomic arrangement can be seen in the schematic illustration of crystal unit cell shown in Fig. 1.9.

Figure 1.9: Schematic illustration of unit cell of $\beta\text{-Ga}_2\text{O}_3$. The tetrahedrally coordinated Ga(I) atoms are depicted in green and the octahedrally coordinated Ga(II) atoms are in gray. The inequivalent O atoms are indicated in red O(I), pink O(II), and maroon O(III) [81].

The electronic band structure of $\beta\text{-Ga}_2\text{O}_3$ obtained by first-principles calculations is shown in Fig. 1.9 [82]. The calculations were based on density functional theory (DFT) with the hybrid functional, where the exchange-correlation potential is combined with the exact Hartree-Fock exchange. The calculated bandgap of 4.84 eV, which is found to be indirect, is in a good agreement with experiment [3]. The conduction band minimum (CBM) is located at Γ point of the Brillouin zone, while the valence band maximum (VBM) is located on the I-L line which is higher than the valence band at the Γ point by only 40

meV [82]. The strong energy dispersion around the CBM leads to a low electron effective mass [82, 83]. On the contrary, the dispersion at VBM is almost flat indicating a very high hole effective mass and very low mobility, making a possible *p*-type conductivity impractical.

Figure 1.10: Band structure of β -Ga₂O₃ along symmetry lines of the Brillouin zone [82]

Figure 1.11: Phonon dispersion and projected density of state (PDOS) for octahedrally (green) and tetrahedrally (blue) coordinated Ga, and O (red) in β -Ga₂O₃. An reduced primitive cell of 10 atoms was used for the calculation, therefore A₁, A₂, and A₃ are the modified symmetry points to describe the directions of the conventional unit cell basis vectors [84]

1.4.2.2 Phonon properties

With a huge number of atoms in the unit cell, the phonon spectrum in β -Ga₂O₃ is very complicated with dense phonon branches including both acoustic and optical ones (Fig. 1.11). This results in a phonon density of states with no gap separating the acoustic and optical modes. Due the continuous phonon spectrum the three-phonon interaction between the acoustic and optical modes are promoted indicated by exceptionally high phase-space volume for A-A-O and A-O-O processes [84]. This is the major reason causing the low thermal conductivity of β -Ga₂O₃.

1.4.2.3 Growth methods

Bulk β -Ga₂O₃ crystals with controllable size can be grown by several well-developed methods such as Verneuil method, floating zone method, Czochralski method, edge-defined-film-fed growth method, and Bridgman method [85]. In all these methods, a liquid phase of Ga₂O₃ which will crystallize into β -Ga₂O₃ on the seed materials, is needed. For growth of the β -Ga₂O₃ epitaxial layers, MOCVD [86], Molecular Beam Epitaxy (MBE) [87], and Pulse Laser Deposition (PLD) [88], are commonly used. In all these methods, a good control of the layer thickness, crystal quality, and impurity incorporation has been achieved.

Since the β -Ga₂O₃ samples studied in this thesis were grown by PLD, this technique is described in more details. PLD is a physical technique which uses Ga₂O₃ as a source materials (usually in form of powder). With a high power UV pulsed laser the source material is heated up. The laser power is absorbed leading to a heat generation mainly by the lattice vibrations. At a high enough heating, atoms or clusters of atoms are ejected from the source material, form a plasma, and directed to the surface of a substrate, which is heated up to a suitable temperature. In the process, extra oxygen gas has to be supplied in order to avoid the formation of oxygen vacancies [88].

1.5 Material characterization techniques

1.5.1 X-ray diffraction (XRD) technique

Due to the lack of lattice-matched substrates, GaN and other group-III nitride layers are commonly grown on foreign substrates as sapphire, silicon, and silicon carbide. The heteroepitaxial growth results in a high strain in the layers and a high density of extended defects such as dislocations and stacking faults. In order for the devices to work efficiently, the threading dislocation density should be controlled at an acceptable level. XRD is a standard technique for dislocation characterization based on the measurements of the

diffraction peak broadening. It is a non-destructive method which allows a rapid analysis and gives representative results. Also, XRD is the most suitable method for determination of the composition in ternary alloys, as $\text{Al}_x\text{Ga}_{1-x}\text{N}$. In this case, the shift of the diffraction angle compared to the angle in the binary compound is measured.

1.5.1.1 Characterization of the composition

In $\text{Al}_x\text{Ga}_{1-x}\text{N}$, the change of composition is directly related to the change of lattice constants (i. e., a and c) which can be measured by the peak position of $2\theta/\theta$ scans. The lattice constants a and c could be directly determined from the diffraction peaks of planes (0002) and (10-12). In heteroepitaxially grown layers are usually under compressive or tensile strains, leading to a modification of the lattice parameters. The presence of strain makes the composition analysis more complex. By using Vegard's law the lattice parameters at composition x can be estimated by $A(x) = xA_{\text{AlN}} + (1-x)A_{\text{GaN}}$, with A denoting the lattice parameters. In order to exclude the effect of strain one has to take into account the interplay between out-of-plane and in-plane strain which can be described by [89]

$$C_{33}(x) \frac{c_m(x) - c(x)}{c(x)} + 2C_{13}(x) \frac{a_m(x) - a(x)}{a(x)} = 0 \quad (1.59)$$

where $C_{13}(x)$ and $C_{33}(x)$ are the elastic constants, $c_m(x)$ and $a_m(x)$ are the measured lattice parameters, and $c(x)$ and $a(x)$ are the fully relaxed lattice parameters at composition x . Any value of x which satisfies the above equation returns Al composition with excluding the effect of strain. The lattice constants and the elastic constants in GaN and AlN binary compounds used in the analysis are taken from literature data [62, 90–92].

1.5.1.2 Characterization of the dislocation density

Density of the two main types of dislocations, i. e. the screw and edge types, can be determined by an proper analysis of XRD rocking-curve scans.

For calculation of the screw dislocation density, William-Hall (W-H) method has been widely applied so far. The tilt angle α_s is obtained from the slope of $\beta \sin(\theta)/\lambda$ plotted versus $\sin(\theta)/\lambda$ on the basis of three symmetric reflections (0002), (0004), and (0006) (Fig. 1.12(a)). The correlation length parallel to the sample surface can be estimated from intercept y_0 of the fitted line by $L_{||} = 0.9/2y_0$ [93].

For calculation of the edge dislocation density, the twist angle α_E is obtained from the plot of the integral width (β) of the diffraction peaks versus the inclination angle χ from the surface normal. The curve is fitted by the following equation [94].

$$\beta_{hkl}^2 = (\alpha_S \cos \chi)^n + (\alpha_E \sin \chi)^n + \frac{(1/L)^n}{K_{hkl}^2} \quad (1.60)$$

where $n = 1$ for a Lorentzian peak shape, $n = 2$ for a Gaussian peak shape, and $(1 < n < 2)$ for a Voigt peak shape (i. e. a convolution of Lorentzian and Gaussian functions). L is the lateral coherence length, and K_{hkl} is the magnitude of the momentum transfer vector. The last term in Eq. (1.60) is usually much smaller than zero, therefore it can be neglected for high-dislocation density layers. Reflections from the symmetric plane (0002) and five asymmetric planes ($10\bar{1}1$), ($10\bar{1}2$), ($10\bar{1}3$), ($10\bar{1}4$), and ($10\bar{1}5$) are commonly used in the case of wurtzite crystals. In the fitting, the tilt angle obtained from the W-H plot was used as a fixed parameter. An example for β versus χ plot is shown on (Fig. 1.12(b)).

Figure 1.12: XRD analysis for estimation of dislocation density: (a) W-H plot for determination of the tilt angle, and (b) extrapolation fitting for determination of the twist angle. The data are for $1 \mu\text{m}$ thick MOCVD-grown $\text{Al}_{0.57}\text{Ga}_{0.43}\text{N}$ layer.

Then, the dislocation density can be estimated by

$$D_i = \frac{\alpha_i^2}{4.35b_i^2} \quad (1.61)$$

where $i = S, E$, and $b_S = c$ and $b_E = a$ are the Burger vectors for screw and edge dislocations, respectively.[93, 94]

1.5.2 Spectroscopic ellipsometry technique for measurement of the layer thickness

Spectroscopic ellipsometry is a well established techniques for measurement of the material optical constants, e. g. absorption and refractive index [95]. Also, using a single wavelength ellipsometry the thickness of thin layers could be accurately determined.

A common ellipsometer is composed of five components including (1) a lighth source, (2) a polarization state generator (PSG), (3) a sample, (4) a polarization state detector (PSD), and (5) a light detector. The PSG and PSD help to polarize and re-polarize the incident and reflected light beams, respectively. After a reflection from sample surface, the polarization of the light is changed and this change reflects the optical properties of the sample. It has been widely accepted for the polarization of light to be represented by four-element Stocks vector

$$S = \begin{bmatrix} I_0 \\ Q \\ U \\ V \end{bmatrix} = \begin{bmatrix} I_0 \\ I_0 - I_{90} \\ I_{45} - I_{-45} \\ I_{rc} - I_{lc} \end{bmatrix} \quad (1.62)$$

The Stocks vector of incident and reflected beams are connected by a 4×4 Mueller matrices. The intensity of the light beam passing through the ellipsometer can therefore be written as $I = S_{PSD}^T M S_{PSG}$, with M denoting the Mueller matrix which describes the light-matter interaction. For isotropic sample the normalized Mueller matrix has a simple form

$$M = \begin{bmatrix} 1 & -N & 0 & 0 \\ -N & 1 & 0 & 0 \\ 0 & 0 & C & S \\ 0 & 0 & -S & C \end{bmatrix} \quad (1.63)$$

where $N = \cos(2\phi)$, $S = \sin(2\phi)\sin(\delta)$, and $C = \sin(2\phi)\cos(\delta)$. The traditional parameters ϕ and δ are related to the reflection coefficients of different polarizations r_s and r_p by

$$\rho = \frac{r_p}{r_s} = \tan(\phi)e^{i\delta} = \frac{C + iS}{1 + N} \quad (1.64)$$

For thin layers, the refraction coefficients are convoluted into a composite reflection parameters $r_{s,p}$ which contain the information about the thickness. The layer thickness can be obtained by interpreting the measured Mueller matrix by appropriate models. In Fig. 1.13, we show the values of Mueller matrix elements measured for a 350-nm-thick AlN

layer grown on SiC substrates.

Figure 1.13: Mueller matrix elements obtained from UV-VIS spectroscopic ellipsometry measurements on a 350-nm-thick AlN layer grown on SiC substrates. The dashed lines represent the fitting of the experimental data using Cauchy model.

1.5.3 Scanning electron microscopy (SEM) and energy dispersive X-ray spectroscopy (EDX)

The use of SEM enables to observe the surface morphology and the structure of samples. The method utilizes a focused electron beam having a relatively low kinetic energy (up to 50 keV) accelerated in a electrical field [96]. In the electron-matter interaction, the incoming electrons are elastically scattered by both the ionic cores and shell electrons resulting in signals from the secondary electrons (SE), the back-scattered electrons (BSE), and the x-rays. Commonly, both the SE and BSE are detected and used for a structural analysis of the samples. Under a constant energy of the electron beam, the SE intensity is influenced by several factors including the sample morphology, the incoming angle of the electron beam, and the conductance of material. On the basis of this, the scanning lines of electron beam could map the surface morphology or layered structure of samples.

The EDX techniques is able to provide an information of the elemental composition in the material. In this mode of the electron microscopy, the XRD signal that contains information of the atomic structure, is detected [97] It should be noted that this method could not give the accuracy better than 1 % in molar fraction, therefore it is helpful at confirming the existence of host elements in material and additionally of impurity with highly doping concentration.

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Part II

2.1 Publications included in the thesis

Paper I

Dat Q. Tran, Nicholas Blumenschein, Alyssa Mock, Pitsiri Sukkaew, Hengfang Zhang, John F. Muth, Tania Paskova, Plamen P. Paskov, Vanya Darakchieva

"Thermal conductivity of ultra-wide bandgap thin layers - High Al-content AlGa_N and β - Ga₂O₃"
Physica B, **579**, 411 (2020)

I have done thermal conductivity and other measurements and data analysis including theoretical modeling. I have written the paper.

Paper II

Dat Q. Tran, Rosalia Delgado-Carrascon, John F. Muth, Tania Paskova, Muhammad Nawaz, Vanya Darakchieva, Plamen P. Paskov

"Phonon-boundary scattering and thermal transport in Al_xGa_{1-x}N: effect of layer thickness"
Appl. Phys. Lett., **117**, 252102 (2020)

I have done thermal conductivity and XRD measurements and data analysis including theoretical modeling. I have written the paper.

Paper III

Dat Q. Tran, Rosalia Delgado-Carrascon, Motoaki Iwaya, Isamu Akasaki, Bo Monemar, Vanya Darakchieva, Plamen P. Paskov

"Thermal conductivity of Al_xGa_{1-x}N layer grown by MOCVD"
(manuscript prepared for submission)

I have done thermal conductivity and XRD measurements, data analysis, and thermal conductivity modeling. I have wrote the manuscript.

2.2 Publications not included in the thesis

Paper I

Rosalia Delgado Carrascon, Dat Quoc Tran, Pitsiri Sukkaew, Alyssa Mock, Refal Ciechonski, Jonas Ohlsson, Yadan Zhu, Olof Holtin, Bo Monemar, Plamen P. Paskov, Lars Samuelson, and Vanya Darakchieva

"Optimization of GaN nanowires reformation process by metalorganic chemical vapor deposition for device quality GaN templates"

Phys. Status Solidi (b), **257**, 1900581 (2020)

I have measured the thermal conductivity of nanowire-GaN templates and perform data analysis.

Papers

The papers associated with this thesis have been removed for copyright reasons. For more details about these see:

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