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III-Nitride Based Light Emitting Diodes and Applications

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Preface

Over the last two decades, significant progress has been made in the growth, doping and processing technologies of III-nitride based semiconductors which paved the way for high brightness light emitting diodes (LEDs). LEDs have already penetrated traffic signals, signage lighting, and automotive applications. However, its ultimate goal is to replace traditional lamps, such as incandescent and fluorescent varieties fuelled by the fact that LED lighting substantially reduces energy consumption and thus the carbon footprint. Despite breathtaking advances in LED technologies (e.g., growth, doping and processing technologies), there remain critical issues for further improvement for the realization of LED lighting. Considering that there are already several books dealing with known and basic issues in III-nitride-based optoelectronic devices including LEDs and laser diodes, this book aims to provide the readers with contemporary LED issues which have not been comprehensively discussed and on which the performance of LEDs is critically dependent. For example, most importantly, there must be a breakthrough in the growth of high-quality nitride semiconductor epitaxial layers with low dislocation density, in particular, for Al-rich and In-rich varieties. The materials quality is directly dependent on the substrates used, such as sapphire, Si, etc. In addition, the loss in efficiency at high injection levels, growth on different orientations (giving rise to polar, semi-polar and non-polar material), and polarization are also important. Chip processing and packaging technologies are key issues as well.

Chapter 1 chronicles the low-temperature-deposited buffer layers and the impact they had on nitride research and the subsequent development of devices. Chapter 2 deals with a perspective on the future of LED efficiency. Chapter 3 reviews the development of GaN epitaxial growth on Si using different methods and nitride LED performance on Si. Chapter 4 presents the properties and the fabrication processes associated with patterned sapphire substrate (PSS) and applications of this technology to growth of non-polar and semi-polar GaN layers. Chapter 5 provides an overview of epitaxial growth and optical properties of IIInitride LEDs on non- and semi-polar orientations. Chapter 6 presents different techniques utilized to assess the internal quantum efficiency (IQE) in LEDs. Chapter 7 reviews the IQE loss mechanisms. Chapter 8 describes electrical properties, reliability, and electro-static

discharge robustness of InGaN-based LEDs. Chapter 9 treats simulations of light extraction efficiency (LEE) as a function of the major materials parameters and geometries in the mainstream LED structures. Chapter 10 deals with fabrication methodologies for high efficiency LEDs for relatively high LEE. Chapter 11 provides an overview of phosphors and LED packaging configurations for while light emission. Chapter 12 discusses the optoelectrical characteristics, various designs, and developments of high voltage and alternating current (AC) LEDs. Chapter 13 gives an overview of the fundamentals of chromaticity and color rendering, the two important aspects of color quality for general lighting. Chapter 14 describes possible future system level applications of LEDs from the viewpoint of emerging trends in lighting related to human health, communication, and display technologies.

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Contents

Chapter 1
Introduction Part A. Progress and Prospect of Growth of Wide-Band-Gap III-Nitrides

Hiroshi Amano

GaAs, InP, and their related compound semiconductors can be grown on native substrates, whereas such growth was not possible for group III nitride semiconductors in the 1970s and 1980s. Despite this drawback, researchers were able to use novel functions of group III nitride semiconductors by growing their thin films on nonnative substrates such as sapphire and Si. Today, white light emitting diodes (LEDs) based on nitrides on sapphire substrates are some of the most important key devices for energy savings given their use in general lighting. Also, LEDs and HFETs on Si are considered to be the next-generation high-performance and low-cost energy-saving devices. In this chapter, the author reviews the history of the growth of GaN by metalorganic vapor phase epitaxy using a buffer layer on nonnative substrates. The history of the low-temperature-deposited buffer layer and its impact on nitride research and the subsequent development of devices are explained.

1.1 History of III–V Research (1950s to 1970s)

Group III nitride semiconductors (nitrides) belong to group III–V compound semiconductors (III–Vs) such as GaAs and InP. One of the most specific differences between nitrides and other III–Vs is their stable crystal structure. The stable structure of nitrides at room temperature at atmospheric pressure is hexagonal wurtzite, while that of other III–Vs is cubic zinc-blende. The history of nitride research started earlier than those of As- and P-based III–V compounds. AlN and InN were first reported in 1907 [1] and 1910 [2], respectively. GaN has a shorter history and was first reported in 1932 [3]. The author would like to focus on GaN in this chapter.

As- and P-based III–Vs were first reported by Welker in 1952, which was 20 years after the first report on GaN [4]. However, the progress of the research

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on III–Vs was very fast. Ten years later, in 1962, Holonyak and Bevacqua realized a visible red GaAsP laser diode (LD) [5], which triggered the remarkable progress in the development of technologies ranging from crystal growth to device fabrication, and the term "Alloy Road" was coined [6]. Thanks to this progress, various devices have been practically realized and are enriching our lives today, including red and green light-emitting diodes (LEDs), infrared LDs for high-speed optical communications, infrared and visible red LDs for optical recording, photovoltaic cells, high-electron-mobility transistors, and heterojunction bipolar transistors. In the 1960s, many scientists attempted or considered attempting to grow nitride crystals, which belong to the III–Vs, as mentioned above, with the aim of realizing the blue LED, because at that time only the blue LED among the three primary colors remained to be achieved. In June 1968, RCA announced flat-panel liquid crystal displays. At that time, the radio system was changed from vacuum-tube-based circuits to transistor-based electronic circuits. Thus, many researchers felt that even the TV system should be changed from the Braun tube to the solid-state display system. Therefore, flat panel TVs were considered a very important piece of equipment for the development of future electronics systems. Accordingly, the remaining solid-state blue light source was highly desired. It was not until 1969, seven years after the first red LD was developed, that Maruska and Tietjen first reported the successful growth of GaN single crystals, which are currently used for blue LEDs [7]. The first blue LED using GaN was reported in 1971, two years after the growth of its crystals was reported [8]. Nevertheless, the development of the GaN-based blue LED was not straightforward. The serious obstacles that prevented the development of GaN-based blue LEDs were the difficulty in growing high quality and low-defect-density crystals and the difficulty in controlling conductivity, particularly realization of p-type GaN.

The largest difference between III–Vs and nitrides is whether bulk single crystals with a low defect density can be practically grown. For example, the melting point of GaAs is 1238 °C at which the dissociation pressure of As is approximately 1 atm [9]. In contrast, the theoretical melting point of GaN is 2500 °C [10] at which the dissociation pressure of N is estimated to be as high as 45000 atm [11]. This means that bulk GaN single crystals cannot be easily realized. The first reported GaN single crystal was also grown on a sapphire substrate by the halogen transport vapor phase epitaxy (HVPE) method, which uses the chemical reaction between GaCl and NH_3 and is also called the hydride vapor phase epitaxy method [7]. Naturally, many research teams shifted their research target from group III–V compound semiconductors to other materials, such as SiC and ZnSe-based group II–VI compound semiconductors, or adopted wavelength conversion or second-harmonic generation with the aim of realizing blue LEDs. Some research teams, however, continued to study group III nitride semiconductors and methods of growing their thin films on non-native substrates.

Eventually, blue LEDs were practically applied using group III nitride semiconductors grown on non-native substrates, i.e., sapphire substrates. The blue-violet LDs used for Blu-ray Discs were also realized on thick-film GaN substrates grown from seed sapphire or GaAs crystals. High-power high-frequency high-electron-mobility transistors (HEMTs) on SiC substrates are already in widespread use in

base stations for cell phones. AlGaN/GaN heterojunction field-effect transistors (HFETs) on Si substrates will shortly become available for use as mid-voltage switching devices. These devices can markedly reduce energy consumption. For example, LED lamps may be used in 55 % of households worldwide by 2025, and the worldwide reduction in power consumption is estimated to be 1,100 TWh per year, assuming a luminous efficacy of 150 lm/W [12]. This value sufficiently exceeds the total power consumption in Japan in fiscal 2006 of 960 TWh. High expectations are also being placed on scanning laser displays including head-mounted displays in terms of reducing power consumption, because they require neither polarizers nor In-containing transparent electrodes, both of which are currently used in liquid crystal displays. Moreover, the power consumption of motors accounts for approximately 60 % of the total power consumption. It is highly expected that inverters with GaN-based switching devices will reduce the power consumption of motors by half or even two-thirds compared with motors without inverters. The rate of inverter employment is high in Japan, in contrast to not only developing countries but also Europe and the US. In the future, switching devices using group III nitride semiconductors are expected to be a key strategic tool for reducing power consumption along with Si-based insulated gate bipolar transistors and SiC switching devices. In addition, InGaN alloys, which cover almost the entire range of the solar light spectrum, are expected to be used to realize photovoltaic cells with an efficiency of over 60 % by utilizing multijunction structures, light collection, and multiphoton absorption.

It is astonishing to know that such tremendous impacts to human society have been and will be brought about by the nonperfect crystals grown on non-native substrates. This was mainly due to the physical properties of group III nitride semiconductors; that is, the diffusion length of excited (injected) carriers for group III nitride semiconductors is shorter than that for As- and P-based compound semiconductors, and group III nitride semiconductors are less affected by crystal defects [13].

Nevertheless, it may be meaningful to explain how researchers tackled the growth of GaN on nonnative substrates. In the next section, the history of the development of the low-temperature-deposited buffer layer for the growth of GaN on a sapphire substrate.

1.2 Dawn of GaN Research (1970s to Mid 1980s)

In Japan, Osamura et al. reported in 1972 the absorption edge of polycrystalline GaN, InN and their alloys [14]. Isamu Akasaki, Matsushita Research Institute, Tokyo, started ammonia MBE of GaN from 1974 and succeeded in the thin-film growth of GaN that emitted blue luminescence by cathodoluminescence. Then, the former Ministry of International Trade and Industry, MITI, decided to embark on three-year national projects on the solid-state blue light emitter. One was the second harmonic generator-based technology proposed by Dr. Izuo Hayashi, NEC, who was famous for the room-temperature CW operation of the AlGaAs-based laser diode,

and the other one was the GaN-based technology proposed by Dr. Isamu Akasaki.
At that time, the quality of the ammonia MBE system was not so good compared
with today's system. According to claim, the GaN film grown by ammonia MBE
always showed highly conductive *n*-type behavior. In addition, the growth rate was
very low. Therefore, Akasaki's group also started HVPE in addition to ammonia
MBE.

In the 1970s, several research groups around the world were pursuing the GaN
growth by HVPE, and some groups such as RCA [8], Stanford University [15],
Oki Electric [16] and Philips [17] demonstrated the mis-type LEDs based on GaN.
However, no group had succeeded in the commercialization of the blue LEDs. The
research group of Matsushita Research Institute, Tokyo, concentrated on the com-
mercialization of GaN-based LEDs and developed several important technologies
such as selective area growth using a SiO_2 mask and flip-chip bonding for high
light extraction [18]. Finally, they succeeded in the sample shipping of the mis-
type GaN-based LEDs in 1982. However, the company stopped the development
of GaN-based LEDs. Thus, Dr. Isamu Akasaki moved to Nagoya University in
1982.

At that time, many researchers had abandoned the "III–V alloy road" for blue
light emitters and moved to other materials such as SiC [19] or II–VIs [20]. Only a
few groups carried out GaN research [21–23].

1.3 Low-Temperature-Deposited Buffer Layer, *p*-Type GaN and Highly Luminescent InGaN (Late 1980s)

The author joined Professor Akasaki's group in 1983. In 1985, Yasuo Koide and
I developed the vertical-type MOVPE reactor as shown in Fig. 1.1.

First, we coped with the flow dynamic issue. At that time, the maximum hy-
drogen flow rate of our purifier was very low for growing GaN at a reasonable
rate. Therefore, the author used a very tiny and narrow quartz tube and gathered

Fig. 1.2 High speed gas
injection type MOVPE
reactor [24]
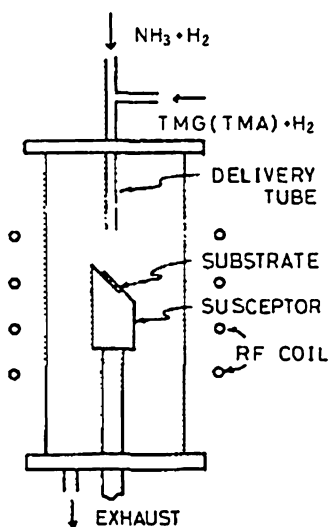
all the gases including hydrogen, metalorganics and ammonia to realize a high flow
rate of about 5 m/s. Figure 1.2 schematically shows the concept of gas injection
tube.

Next, the author tackled the problem of large lattice mismatch of about 16 % be-
tween GaN and a sapphire substrate. In general, it is quite difficult to proceed with
epitaxial growth if the lattice mismatch is larger than 1 %. Therefore, the grown
crystal is mainly composed of tiny grains aligned to the c-axis but the in-plane crys-
tal direction is largely misaligned with each other. The author could not grow GaN
with an atomically flat surface for a long time. A c-plane sapphire with an offset
angle larger than 1° was found to be effective for growing GaN having a relatively
flat surface as shown in Fig. 1.3. However, the surface morphology was far from
perfect, even worse than that of GaN grown by HVPE. The surface flatness of AlN
and AlGaN grown by a colleague was better than that of GaN grown by the au-
thor. Thus, the author tried to use thin AlN as a buffer layer between GaN and the
sapphire substrate in the autumn of 1985. He knew that single-crystal AlN should
be grown at a much higher temperature than that of GaN. However, unfortunately,
when he started to grow GaN, he found that the power oscillator did not work prop-
erly. Then, he remembered the comment of the associate professor who sometimes
said that the substrate surface should be as clean as possible for epitaxial growth.
However, if it is impossible to realize an atomically clean surface, it may be ef-
fective to deposit a not so high-quality layer that acts as a nucleation layer. Thus,
he deposited AlN at a low temperature of approximately 450 °C. In the meantime,
the power oscillator recovered to work again. Then, he continued the growth of
GaN at the normal temperature of 1,000 °C. When he removed the sample from
the reactor after the growth, he first thought that he had forgotten to supply met-
alorganics because the surface was almost the same as that of the bare sapphire.
However, when he observed the surface using a Nomarski microscope, he found

Fig. 1.3 Optical microscopic
images of GaN directly
grown on sapphire. (From top
to bottom, 3° off toward
⟨11–20⟩ of sapphire, just cut,
3° off toward ⟨10–10⟩ of
sapphire)
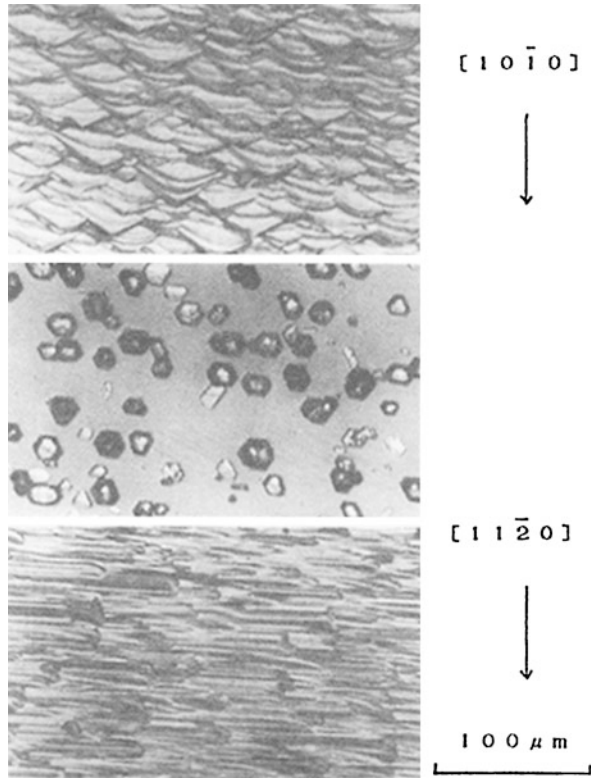
that he had overcome the problem of surface roughness. Now, this process is called
the low-temperature-deposited buffer layer and becomes the de-fact-standard tech-
nology of the growth of GaN by MOVPE on a sapphire substrate. Figure 1.4 shows
the effect of the low-temperature buffer layer on the growth of GaN on a sapphire
substrate [24].

Before publishing this remarkable result to the academic journal, the author
thought he should submit a patent. At that time, he knew similar results had been
already published. For example, M. Akiyama et al. reported that low-temperature-
deposited GaAs buffer layer was effective for the improvement of the quality of
GaAs on Si by MOVPE [25]. S. Yoshida et al. reported an increase of electron mo-
bility and improvement of the cathodoluminescence property of GaN on a sapphire
substrate grown by ammonia MBE by coating with single crystalline AlN [22]. The
author knew that low-temperature-deposited GaN buffer layer was also effective to
improve the surface morphology of GaN. He also knew that high temperature de-
posited thin AlN also showed some effect to improve the surface morphology of
GaN. However, he was afraid that people might think that it was only the imitation
of GaAs on Si. Therefore, he wrote a draft of the patent in which the buffer layer
was limited to only low-temperature deposited AlN buffer layer [26]. Later on, he
knew that his intension was superficial [27].
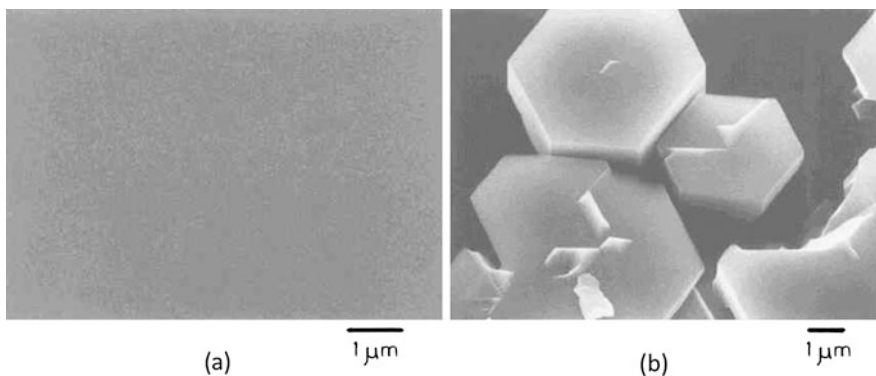
Fig. 1.4 SEM images of GaN grown on a sapphire (0001) substrate (**a**) with and (**b**) without low-temperature deposited AlN buffer layer

Thirdly, the author coped with the realization of *p*-type GaN. At that time, Zn was commonly used as an acceptor dopant because Zn-doped GaN shows a relatively bright blue luminescence. In 1988, the author found that the blue luminescence intensity monotonically increases during low-energy electron beam irradiation [28]. This phenomenon was already published and reported by Saparin et al. in 1983 [29] and 1984 [30]. When he read the textbook written by Phillips and Lucovsky [31], he found that Mg or Be should be better than Zn as an acceptor dopant because of the lower activation energy. Unfortunately, he could not use Be because of safety issues. Thus, he changed the acceptor dopant from Zn to Mg and tried to grow Mg-doped GaN. The as-grown GaN:Mg did not show *p*-type behavior. However, when it was treated with low-energy electron beam irradiation, he found that the *p*-type GaN had been successfully realized for the first time. This process was called LEEBI treatment [32]. At that time, he thought he should clarify the mechanism. Firstly, he thought that the mechanism was caused by the heating during electron beam irradiation, which was unexciting to him. Therefore, he investigated another possible mechanism. But, he could not find a good explanation of LEEBI effect. Then, in 1991, he also noticed that he had a shallow idea [33].

Anyway, the author presented this result to the International Symposium on GaAs and Related Compounds in 1989, which was held at Karuizawa, Japan. The author attended the program committee meeting on behalf of Professor Isamu Akasaki and found that his paper was just below the borderline, that is, close to being accepted as a poster presentation, it was rejected during the first screening. However, when the sessions were finally arranged, one slot became vacant. Then, he recommended himself of his paper and it was very fortunate that his abstract was allowed to be presented as a poster [34]. At the same symposium, Dr. Takashi Matsuoka of NTT at that time presented the growth of InGaN by MOVPE for the first time [35]. But, a very few people were interested in his presentation.

1.4 Summary

The history of the initial stage of the development of GaN, especially MOVPE growth of GaN on a nonnative sapphire substrate using a low-temperature-deposited buffer layer, and the realization of *p*-type GaN in the 1980s and 1990s were reviewed. In 2011, more than 3,400 academic papers on GaN and related compounds academic journal papers were published. Also, the number of attendees of the International Workshop on Nitride Semiconductors 2012, which was held in Sapporo, Japan was more than 1000. In addition, the market size of LED displays and lighting will exceed one billion US dollars in 2012. No one probably expected such a vast expansion of the research field in the 1980s. The potential of nitrides should not be limited to LEDs, but should also expand to other fields such as high-frequency and high-power electron devices, laser diodes and photovoltaic devices. Another breakthrough will surely lead to the second expansion of the research field and market of these material systems in the very near future.

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Chapter 2
Introduction Part B. Ultra-efficient Solid-State Lighting: Likely Characteristics, Economic Benefits, Technological Approaches

Jeff Y. Tsao, Jonathan J. Wierer Jr., Lauren E.S. Rohwer, Michael E. Coltrin, Mary H. Crawford, Jerry A. Simmons, Po-Chieh Hung, Harry Saunders, Dmitry S. Sizov, Raj Bhat, and Chung-En Zah

Abstract Technologies for artificial lighting, as illustrated on the left side of Fig. 2.1, have made tremendous progress over the centuries: from fire, with an efficiency of about a tenth of a percent; to incandescent lamps, with an efficiency of about 4 %; to gas discharge lamps, with an efficiency of about 20 %; and soon to solid-state lighting (SSL), with efficiencies that in principle could approach 100 %.

At this point in time, there is virtually no question that SSL will eventually displace its predecessor technologies. A remaining question, however, is what the final efficiency of SSL will be. Will it be, as illustrated on the right side of Fig. 2.1, 50 %, which is what the community (Haitz and Tsao in Phys. Status Solidi A 208:17–29, 2011) has long targeted as its "efficient" lighting goal? Will it be 70 % or higher, which is what some (Phillips et al. in Laser Photon. Rev. 1:307–333, 2007) have called the "ultra-efficient" lighting goal? Or will it be even beyond an effective efficiency of 100 %, something that might be enabled by smart lighting (Kim and Schubert in Science 308:1274–1278, 2005), in which one doesn't just engineer the efficiency with which light is *produced*, but the efficiency with which light is *used*?

In this chapter, we give a perspective on the future of SSL, with a focus on ultra-high efficiencies. We ask, and sketch answers to, three questions. First, what are some of the likely characteristics of ultra-efficient SSL? Second, what are some of the economic benefits of ultra-efficient SSL? And, third, what are some of the challenges associated with the various technological approaches that could be explored for ultra-efficient SSL?

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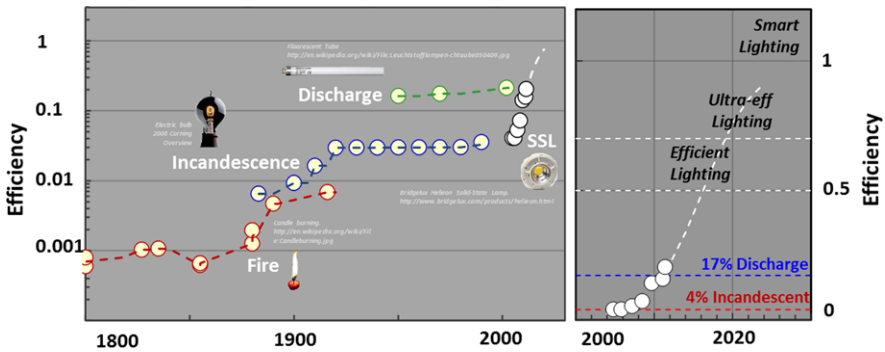
Fig. 2.1 200-year evolution of the efficiency of various lighting technologies. Adapted and updated from: Tsao [29, 30], Bergh [2], Haitz [10, 11], and Nordhaus [21]

2.1 Some Likely Characteristics of Ultra (>70 %) Efficient SSL

We first ask the question: what are some of the likely characteristics of ultra-efficient SSL? We believe there are at least two likely characteristics, both mutually compatible and complementary.

The first likely characteristic of ultra (>70 %) efficient SSL is that, unlike in the current dominant paradigm, it not make use of the wavelength downconversion that is acceptable for (>50 %) efficient SSL. Because about 80 % of white light power is green or red, and because the Stokes deficit on converting from blue to green and red is about 25 %, SSL in which wavelength downconversion is used to produce green and red is automatically at most $1 - 0.8 \times 0.25 = 80$ % efficient [30], and likely even less because of other loss factors. So the first likely characteristic of ultra-efficient SSL is that it make use of efficient electroluminescence, not wavelength downconversion, across the visible spectrum. Of course, as illustrated in Fig. 2.2, this will be challenging due to the well-known green-yellow gap in the electroluminescence efficiency of compound semiconductors [4]. However, it will be necessary for ultra-efficient SSL.

The second likely characteristic of ultra-efficient SSL is that, also unlike the current dominant paradigm in SSL technology, its white light spectrum not be broadband and continuous, but rather narrowband and "spiky." To see this, Fig. 2.2 shows the spectral power densities of the output optical power from various white light sources.

Incandescence   The least efficient source is the incandescent lamp, with the continuous spectral power density of black-body radiation shown in red, extending extremely far out into the infrared where the human eye does not see at all. All the light produced in the infrared is wasted, and the spectral efficiency of the light is very poor.

PC-LED   The second least efficient source is the phosphor-converted white LED shown in black, which has very broad green and red phosphor emission extending
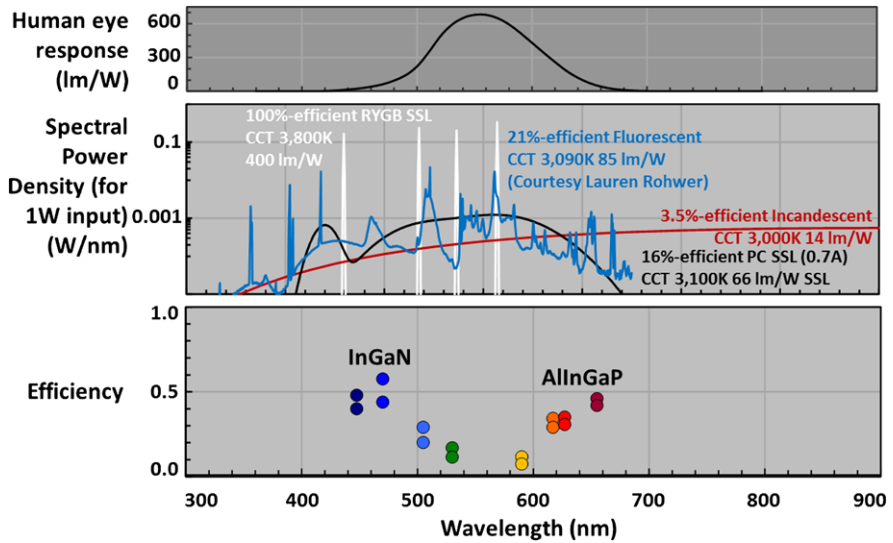
Fig. 2.2 *Top*: Photopic (daytime) human eye response. *Middle*: Spectral power densities of the output power of various sources for 1 W input electrical power to the source. For the fluorescent lamp, the red phosphor is Y_2O_3:Eu^{3+}, the blue phosphor is $BaMgAl_{10}O_{17}$:Eu^{2+} (BAM), while the green phosphor is either $LaPO_4$:Ce^{3+},Tb^{3+} or $CeMgAl_{11}O_{19}$:Tb^{3+}. *Bottom*: Power conversion efficiencies of state-of-the-art commercial monochromatic light-emitting diodes (LEDs) driven at 350–750 mA/mm^2 [26]

into the deep red which the human eye can see but not very efficiently. Just as with the incandescent lamp, though not as severely, that wasted power in the deep red means that the spectral efficiency of the white light is poor.

Discharge   Compared to the two lamps just discussed, the discharge-based fluorescent lamp is a much more efficient source. It does not have much wasted light in the deep red, with a spike in its spectral power density near the ideal shallow-red wavelength of 615 nm. But its efficiency is still limited, for the following reasons: the gas discharge is itself only 65 % efficient; the Stokes loss from the UV mercury lines at 254 nm into the visible around 555 nm adds an additional 50 % inefficiency; and the quantum yield of the phosphor blend is only about 85 % efficient. So, despite its spiky and efficient spectral power density [23, 33], its ultimate maximum efficiency is only about 25 %.

100 %-Efficiency   In fact, the best lamp of all would be one that produces the four spikes of light shown in white. There are just enough spikes, spaced widely enough apart, to fill the visible spectrum and render quite well the colors of objects typical in the environment around us [30]. Additionally, the spikes are concentrated within the visible spectrum, without spillover of light into wavelengths at which the human eye is insensitive, resulting in the highest spectral efficiency possible.

2.2 The Ultimate SSL Source Is Spiky

One might now ask: how is it that the spiky source illustrated in white in the middle panel of Fig. 2.2 could be a good white-light illuminant? First, how could such a spiky four-color spectrum really render the colors of objects in the environment around us well, and be a visually pleasing white light illumination source, even while it is missing so much spectra in between the spikes? And, second, does such a spiky four-color spectrum also maximize luminous efficacy of radiation? In this section, we discuss these two questions in turn.

2.2.1 Spiky Spectra Give Good CRI

Let us start with how it is that spiky spectra can give good color rendering quality.

In fact, spiky sources are not only predicted to be a visually pleasing white-light illuminant by color science [22], but have been shown to be so in a recent experiment [20]. In that experiment, an *extremely* spiky illumination source was created using four visible lasers: blue, green, yellow and red. The relative powers of the lasers were tuned to maximize color rendering quality and to match color temperature and overall lumen output of a number of reference sources, including the output of an incandescent lamp. Then, two side-by-side scenes were illuminated, one with the four-color laser illuminant and the other with the various reference sources.

One of the results is illustrated in Fig. 2.3. Although the similarity of the side-by-side photos does not automatically imply similarity in human visual experience (because the image sensor in the camera is not the human eye), the human visual experiences of the two scenes were in fact quite similar. As might be expected, when any one of the four colors is removed, the scene looks visually displeasing. But, when that color is put back in, the scene looks visually quite pleasing. In fact, the four-color laser illuminant was, in double-blind tests, on average preferred over the phosphor-converted neutral-white and cool-white LED lamps, and on average neither preferred nor dispreferred over phosphor-converted warm-white LED and incandescent lamps.

Why is this? The reason is that most objects in the world have relatively broad reflectance spectra. It is a rare object, like a diffraction grating, that has a narrow reflectance spectrum that would enable the human eye to distinguish between spiky and non-spiky illuminants.[1]

So the striking conclusion of theory and experiment is that spiky sources do indeed give excellent color rendering quality.

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[1]Rare, but not non-existent. Any iridescent object which supports optical interference phenomena (e.g., opals, soap bubbles, some butterfly wings) would distinguish between spiky and non-spiky illuminants.
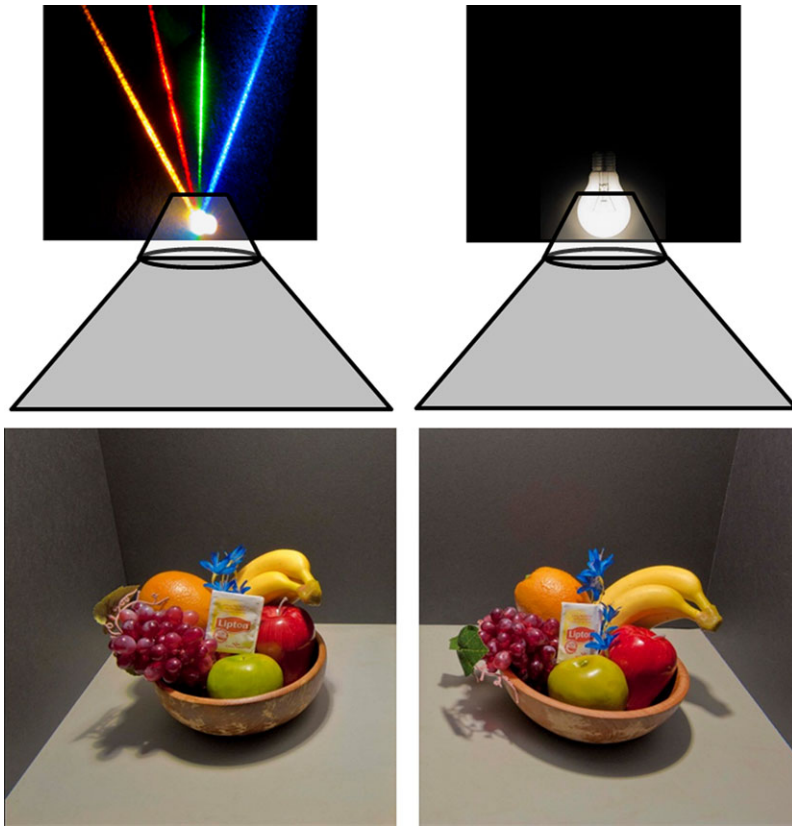
Fig. 2.3 Schematic of side-by-side test of the color-rendering quality of a four-color "spiky" laser illuminant and a number of reference sources (including the incandescent lamp depicted). To give a sense for the color-rendering quality of the laser illuminant, the photos of the *left* and *right* bowls of fruit were illuminated, respectively, by the four-color laser illuminant and by an incandescent lamp. Although the image sensor in the camera is not the human eye, so the similarity of the photos does not automatically imply similarity in human visual experience, in fact the human visual experience *was* quite similar

2.2.2 Spiky Spectra Give the Highest MWLERs

Let us now consider whether spiky spectra also give good luminous efficacies of radiation.

To answer this question, the right side of Fig. 2.4 shows recent simulations of spectral power densities, unconstrained by whether the spectra are continuous or spiky, which maximize white luminous efficacy of radiation (MWLER) given particular values of standard color rendering index[2] (CRI) and correlated color tem-

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[2] In our treatment here, we mean the standard color rendering index R_a [3].
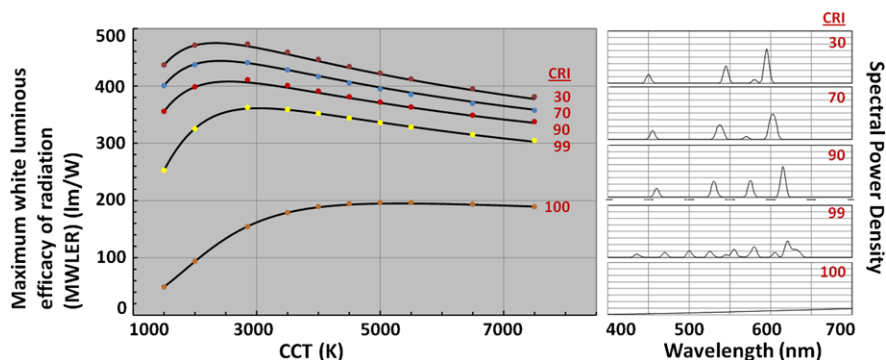
Fig. 2.4 *Left*: Maximum white luminous efficacy of radiation versus correlated color temperature (CCT) for various standard color rendering indices (CRIs). *Right*: Relative spectral power densities for various CRIs for a CCT of 2856 K. After Hung [13]

perature (CCT) [13]. In fact, the spectral power densities that maximize the white luminous efficacies of radiation are, except for the very highest CRI (of 100), all spiky. Of course, for a very high CRI, as in the panel with CRI of 99, there may be more spikes, but nevertheless the spectral power densities are all spiky. We can conclude that spikiness is in fact a general feature of spectral power densities that maximize luminous efficacy or radiation.

Now, *given* a spiky spectral power density, the maximum white luminous efficacy of radiation does of course depend on CCT and CRI. One can see from the family of curves illustrated in the left half of Fig. 2.4, that, for all CRI values, the MWLER peaks at a particular CCT and drops off at higher or lower color temperatures on either side. The physical reason for this is that, starting from a very high CCT, as CCT decreases, spectra shifts first from the blue to the green and then from the green to the red. Because luminous efficacy of radiation is highest in the green, the MWLER first increases then decreases.

One can also see from the family of curves illustrated in the left half of Fig. 2.4 that, as CRI increases, all luminous efficacies of radiation decrease. The physical reason for this is that the higher the CRI becomes, the more "spread out" the white light spectrum (even if it is spiky) becomes. The more spread out the white light spectrum becomes, the further the spectrum penetrates into the wings of the human eye response (hence further from the peak of the human eye response at 555 nm), and the lower the luminous efficacy of radiation.[3]

We can conclude that spiky spectra are preferred over non-spiky (continuous) spectra: they not only give good color rendering quality, but also maximize white luminous efficacy of radiation.

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[3]Note that the higher the CRI, the higher the CCT at which the MWLER maximizes, as discussed in Hung [13].

2.3 Economic Benefits of Ultra-efficient SSL

Up to this point, we have discussed the characteristics of ultra-efficient SSL: it must not make use of wavelength downconversion, and its spectrum should be spiky. But what are the economic consequences of ultra-efficient SSL? The benefits (i.e., profits) of efficient lighting to the SSL industry will have already manifested themselves by the time SSL has achieved 50 % efficiency. At that point, SSL will have beaten both incandescent and gas discharge lamps and will have begun to replace them. Are there additional economy-wide benefits (beyond profit to the SSL industry) that would make it worthwhile to pursue ultra-high efficiencies of 70 % and beyond?

To understand the possible answers to this question, in Fig. 2.5 we sketch two possible scenarios for the energy economics of light. Using world gross domestic product (GDP) and world consumption of light and energy in 2011 as a baseline, we ask what the consequences of changes in the efficiency of lighting would be, *all other factors in the economy held constant*. We assume throughout that the impacts of a change in efficiency are twofold: first, less energy is consumed for a given amount of light consumed; and second, the effective cost of light (typically dominated by the energy used to produce the light) decreases.

2.3.1 Scenario 1: Light Is Not a Factor of Production

The first scenario is sketched in the left panels of Fig. 2.5. This is a scenario in which light is *not* a factor of production in the global economy. Thus, if the efficiency with which light is produced were to increase, world GDP wouldn't change, but would stay constant at its value in 2011 (about 60.5 trillion US dollars). This is also a scenario in which, if light were to become more efficient and hence cheaper, we would have no tendency to consume more (i.e., consumption of light is saturated).

Hence, in this scenario, the main benefit to increased efficiency is that the consumption of energy associated with the production of light decreases. Since the consumption of energy associated with the production of light is about 1/15 of all consumption of energy, as efficiency of lighting increases, consumption of energy could in principle decrease by nearly 1/15—from about 44.5 PWh/yr to about 42 PWh/yr.

But one can also see that the biggest decreases in energy consumption occur when efficiency is low. As efficiency increases there is a diminishing return or diminishing benefit to continuing to increase efficiency. For example, when efficiency has reached 0.9, increasing efficiency further from 0.9 to 1.0 only decreases energy consumption by 35 TWh/yr, which represents only about $4.3 B/yr.[4] This is not a small benefit, but it is *relatively* small, and it is not obvious that an efficiency of 1.0 would be worth the required R&D effort.

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[4]Assuming a light-usage-weighted average world electricity price of $120/MWh [31].
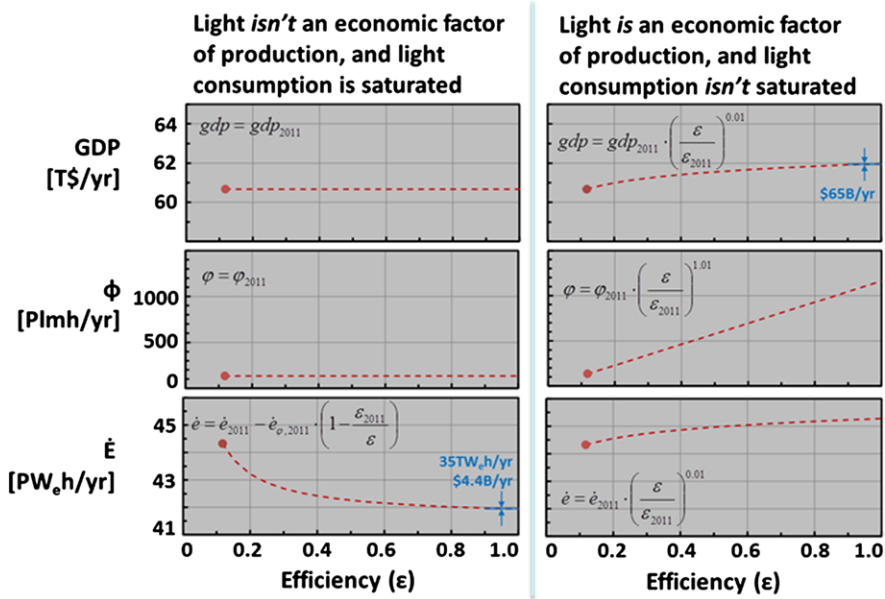
Fig. 2.5 Economic scenarios for world gross domestic product (GDP), world consumption of light (Φ) and world consumption of energy ($\dot{E}$). The scenario on the left is one in which light *is* considered to be a factor of production in the economy; the scenario on the right is one in which light is *not* considered to be a factor of production in the economy

2.3.2 Scenario 2: Light Is a Factor of Production

The second scenario is sketched in the right panels of Fig. 2.5. This is a scenario in which light *is* a factor of production in the global economy. So, if the efficiency with which light is produced were to increase, the cost of light would decrease, more light would be consumed, and world GDP *would* increase (above where it otherwise would be, again all other things in the world of 2011 held constant). This scenario has been analyzed in some detail using a simple Cobb-Douglas model which treats light as a factor of production in the global economy [32].

The result of that analysis is that there is no direct energy (or environmental) benefit to increased efficiency. Since the consumption of light increases as efficiency increases, energy consumption doesn't decrease. Indeed, because GDP increases as light consumption increases, there is even a slight increase in energy consumption. So, in this scenario, the main benefit to increased efficiency is an increase in global GDP. In fact, this benefit is huge. Although there is a diminishing return to the increase in GDP, if efficiency were even as high as 0.9, and were increased to 1.0, global GDP would increase by $65 B/yr. This economic benefit is huge, nearly 15× larger than in the first scenario, despite the diminishing returns.

Moreover, although in this scenario there is no direct energy (and environmental) benefit to increased efficiency, some of the huge increase in GDP could presum-
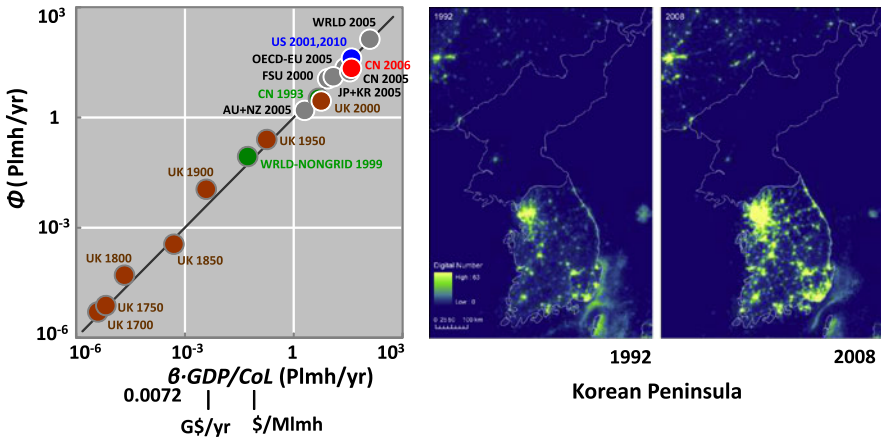
Fig. 2.6 *Left*: Consumption of light versus ratio between gross domestic product (GDP) and cost of light (CoL) for various countries at various times in history, after Tsao [31]. *Right*: Satellite images of the Korean peninsula taken in 1992 and 2008, showing the dramatic increase in consumption of light, correlated with the increase in national GDP during that same time period. After Henderson [12]

ably be harnessed to improve technologies *with* direct energy (and environmental) benefit, including those for renewable energy or for adaptation to the deleterious environmental effects of carbon-emitting energy technologies.

Thus, if this is indeed the scenario that best describes reality, there is a large direct economic benefit and hence large potential indirect energy and environmental benefits to continuing to improve efficiency.

2.3.3 A Qualified Nod to Scenario 2: More Light = More Productivity

Which of the above two scenarios best describes reality? We cannot know for sure, but if the past is any indication, then it is the second scenario.

The reason is illustrated in the plot on the left of Fig. 2.6. The vertical axis of this plot is consumption of light, in Plmh/year. The horizontal axis of the plot is a fixed constant, β, times the ratio between gross domestic product (GDP) and cost of light (CoL). If we use as our units for GDP billions of dollars per year, and for cost of light $/Mlmh, we see that this ratio has the same units, Plmh/year, as the units of the vertical axis. And if we choose the fixed constant, β, to be 0.0072, we see that the empirical data fall very closely along a line of slope unity and zero offset [31].

The ability of this simple formula to predict consumption of light across several centuries and a wide range of countries and groups of countries has two implications. The first implication is that, as GDP has increased, consumption of light has

increased. That is, the wealthier we are, the more light we consume.[5] One can see this graphically for the case of South Korea, where the consumption of light, as seen from outer space, has increased markedly from 1992 to 2008. The second implication is that, as cost of light has decreased, consumption of light has increased as well. That is, the cheaper light is, the more we have consumed.

In fact, these results are exactly what one would predict from the second scenario, the one in which light *is* a factor of production in the economy, so that the more efficiently light is produced, the more we consume and the more productive we are. Now, we do not know whether the future will be like the past, but if it is, then this analysis shows that it *is* worth the effort to continue to improve SSL efficiency, even to ultra-high efficiencies and beyond.

2.4 Two Competing Approaches: Low and High Power Densities

We have just discussed the potential economic benefit of achieving the highest possible efficiencies. In fact, there are two approaches, discussed below, that could be pursued to achieve this: one in which the emitters are run at low power densities and the emission is spontaneous; and one in which the emitters are run at high power densities and the emission is stimulated. Both have potential, but both also face challenges.

To illustrate these challenges, let us focus on just the blue light emitter. Eventually, as discussed above, it will be important to find green, yellow and shallow-red light emitters for ultra-efficient SSL. For our purpose here, however, let us assume the paradigm of white light based on blue light emitters and wavelength downconversion. Then, in Fig. 2.7, we show the implications of input power density into the blue light emitter on the efficiency and cost constraints of the blue light emitter, and also on the resulting heat-sink-limited area and white light flux producible from that area.

The key panel in Fig. 2.7 is the bottom one: the power-conversion efficiency of the InGaN-based blue light emitter versus the input power density into the emitter. On the one hand, spontaneous-emission-based blue LEDs (red line) cannot be driven at an input power density more than a few W/cm^2 before efficiency decreases—the so-called "efficiency droop" problem [4, 15, 28]. On the other hand, stimulated-emission-based blue lasers (blue line), because of the finite lasing thresholds, must be driven at an input power density more than about a few thousand W/cm^2 before they "turn on" and begin to achieve appreciable efficiency. There is a "valley of death" in input power density within which InGaN-based blue light emitters are not very efficient, and, thus, there is currently a tremendous amount of research aimed at circumventing this valley of death by improving the efficiency droop. If it cannot be circumvented, the valley of death motivates two very different approaches [29] to ultra-efficient SSL.

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[5]Indeed, this correlation between GDP and light consumption is being explored as a means to "measure" GDP. See, e.g., Henderson [12].

Fig. 2.7 Economics of low-power-density and high-power-density approaches to SSL
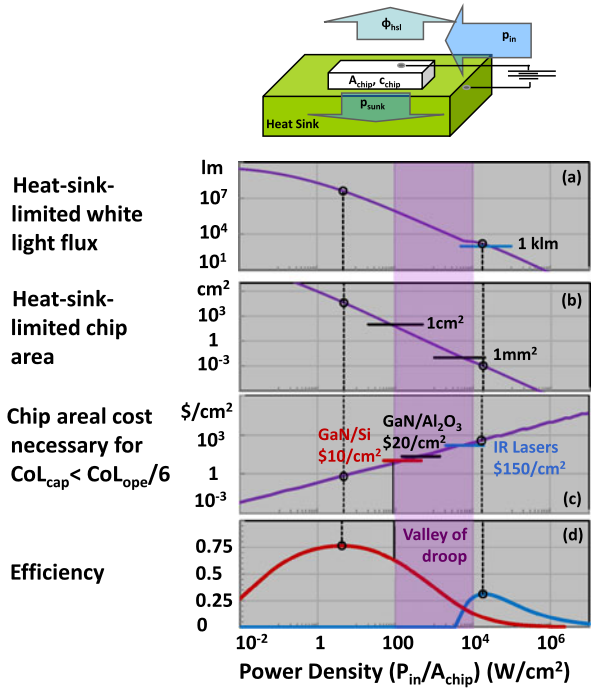
2.4.1 Low Power Density Approach (LEDs)

The first approach is to drive a blue LED at a low input power density, near peak efficiency. The red curve in Fig. 2.7(d) shows efficiency calculated using "ABC" coefficients [5] based on a state-of-the-art Philips Lumileds LED [26]. Peak efficiency is at a power density of roughly 5 W/cm^2, or at about a 15–20 mA drive current into a 1 mm^2 chip [35].

The problem with this approach is that, if the chip isn't driven very hard, not much light is output from the chip per unit area. Large integrated chip areas would be required to produce an appreciable light output, so the chip cost per unit area must be low. To see how low, consider the two costs associated with light [30].

First there is the capital cost of light, which can be written as

$$CoL_{cap} = \frac{\alpha \cdot c_{chip} \cdot A_{chip}}{MWLER \cdot \varepsilon_B \cdot \varepsilon_{PP} \cdot P_{in} \cdot L}, \qquad (2.1)$$

where α is the ratio between retail lamp price to the consumer and chip cost, c_{chip} is the areal cost of the chip (in $/cm^2), A_{chip} is the area of the chip (in cm^2), $MWLER$ is the maximum white luminous efficacy of radiation (in lm/W), ε_B and ε_{PP} are the blue emitter and phosphor + package efficiencies, respectively, P_{in} is the electrical power input into the chip (in W), and L is the lifetime of the lamp (in Mh).

Second, there is the operating cost of light, which can be written as

$$CoL_{ope} = \frac{CoE}{MWLER \cdot \varepsilon_B \cdot \varepsilon_{PP}},$$
(2.2)

where *CoE* is the cost of electricity (in \$/Mlmh).

For traditional (non-SSL) lighting, the capital cost of light is small relative to (and about 1/6 of) the operating cost of light, and one can anticipate a similarly low ratio between the two costs for SSL in its steady-state future. If so, we would have $CoL_{cap} = CoL_{ope}/6$, and could then deduce that the areal cost of the chip must be roughly

$$c_{chip} = \frac{L \cdot CoE}{6 \cdot \alpha} \cdot (P_{in}/A_{chip}).$$
(2.3)

In other words, the higher the power density (P_{in}/A_{chip}) input into the chip, the higher the areal chip cost can be while still satisfying a low capital-to-operating cost ratio of 1/6. Note that the absolute efficiencies of the blue emitter or of the phosphor + package do not enter in, as these affect both the capital and operating costs in the same manner and cancel out.

If we now assume an input-power-density-dependent lifetime that decreases smoothly[6] from 0.03 Mh (at a typical LED input power density of 225 W/cm^2) to 0.01 Mh (at a typical laser input power density of 240 kW/cm^2), a retail cost of electricity of roughly 120 \$/MWh, and a retail-lamp-to-chip cost ratio[7] of roughly $\alpha \sim 10$, then we get the diagonal purple line plotted in Fig. 2.7(c). If one would like to drive the chip at 5 W/cm^2 power density, where efficiency is maximum for current state-of-the-art blue LEDs, one finds that chip cost per unit area must be lower than \$1/cm^2.[8]

There is thus a strong motivation for developing extremely inexpensive chip technologies, like GaN/Si or GaN nanowires. However, future scenarios [34] for GaN/Si chip technologies predict roughly 2× reductions from GaN/sapphire, not the 20× reductions that appear to be necessary. Thus, areal chip cost will be a key but difficult challenge for the low-power-density approach to ultra-efficient SSL.

A related challenge is to increase the power density at which ultra-high-efficiencies may be obtained, so that more expensive chips can be used.

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[6]We use a simple logarithmic dependence of lifetime on input power density: $\log(L/0.03 \text{ Mh}) = 0.23 \cdot \log[(P_{in}/A_{chip})/(225 \text{ W/cm}^2)]$, where L is lifetime in Mh and P_{in}/A_{chip} is input power density in W/cm^2.

[7]This multiplier between the cost of a retail lamp to the cost of the chip within the lamp includes various sub-multipliers that connect the chip to the package, the package to the wholesale lamp, and finally the wholesale lamp to the retail lamp [7, 8]. Note that this multiplier is surprisingly similar to those for higher-power-density chips such as high-power IR lasers [16, 19] inserted into retail laser modules and for low-power-density chips such as solar cells inserted into residential retail panels [1, 6, 18]. Hence, we use the same multiplier across the range of input power densities considered here.

[8]To put this in perspective, current chip cost per unit area for state-of-the-art GaN/sapphire chips is much higher (about 20 \$/cm^2), while for single-crystal Si solar cells is much lower (about \$0.02/cm^2).

2.4.2 High Power-Density Approach

The second approach is to drive a blue laser at a high input power density where, due to stimulated emission, efficiency can again be higher. The blue curve in Fig. 2.7(d) shows a threshold-and-slope-efficiency calculation of efficiency based on a state-of-the-art Osram blue laser, showing that efficiency peaks at a power density of roughly 20 kW/cm^2, or at about a 70 A drive current into a 1 mm^2 chip [35].

In this approach, because the chip is being driven very hard, with a lot of light out per unit area, the chip cost per unit area can be high. In fact, if we apply the same economic constraints we applied before, that the cost to buy the lamp should be about 1/6 the lifetime cost to run the lamp, we find that, at the power density at peak efficiency for this blue laser, chip cost per unit area must be less than $400/cm^2. In fact, high-efficiency [24] high-power IR laser chips, at $150/cm^2 [16], are already much cheaper than this, with the possibility that they could become much cheaper still [16]. This provides an existence proof that such costs might someday also be achievable with laser chips in the visible.

What is less obvious is whether one can get a reasonable amount of light out of a high-input-power-density light emitter. A limit to the amount of light will be imposed by heat sinking. To determine roughly what that limit might be, we use the usual scaling relationship [9, 17, 29] for the thermal resistance of a conduction-cooled (and hence relatively inexpensive) chip:

$$R_T = \frac{1}{2\kappa_T \cdot \sqrt{4A_{chip}/\pi}},\tag{2.4}$$

where κ_T is the thermal conductivity of the heat sink, and A_{chip} is the area of the chip.[9] If the power sunk by the heat sink is $P_{sunk} = P_{in} - P_{out} = P_{in} \cdot (1 - \varepsilon_B \varepsilon_{PP})$, then the thermal resistance that enables the temperature increase of the chip due to this sunk power to be less than a maximum allowable ΔT_{max} is:

$$R_T = \frac{\Delta T_{max}}{P_{in} \cdot (1 - \varepsilon_B \varepsilon_{PP})}.\tag{2.5}$$

Equating the two expressions for the thermal resistance allows us to solve for the "heat-sink-limited" maximal area[10] of the chip for a given input power density:

$$A_{hsl} = \left[\frac{2\kappa_T \cdot \sqrt{4/\pi} \cdot \Delta T_{max}}{(1 - \varepsilon_B \varepsilon_{PP}) \cdot (P_{in}/A_{chip})} \right]^2.\tag{2.6}$$

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[9]Note that this is an underestimate of the thermal resistance for a laser chip, as such chips may be non-square with a large aspect ratio.

[10]In other words, for a given input power density, efficiency and heat-sink properties, there is a maximum chip size that enables the temperature rise of the chip to remain manageable. This maximum chip size depends strongly on (inversely as the square of) input power density because the chip size that gives a particular thermal resistance depends strongly on (inversely as the square of) that thermal resistance.

The "heat-sink-limited" maximal light output of the chip, also for a given input power density, is proportional to this heat-sink-limited maximal area, and can thus be written as:

$$\Phi_{hsl} = MWLER \cdot \varepsilon_B \varepsilon_{PP} \cdot \left(\frac{P_{in}}{A_{chip}} \right) \cdot A_{hsl}$$

$$= \frac{MWLER \cdot \varepsilon_B \varepsilon_{PP}}{(P_{in}/A_{chip})} \left[\frac{2\kappa_T \cdot \sqrt{4/\pi} \cdot \Delta T_{max}}{(1 - \varepsilon_B \varepsilon_{PP})} \right]^2. \tag{2.7}$$

Both of these heat-sink-limited quantities, chip area and chip white light output are plotted in Figs. 2.7(b) and 2.7(a) for a *MWLER* of 400 lm/W, blue light emitter efficiencies given by those in Fig. 2.7(d), a phosphor + package efficiency of $\varepsilon_{PP} \sim 0.7$, a maximum allowable temperature rise of $\Delta T_{max} \sim 100$ K, and a thermal conductivity of $\kappa_T \sim 2$ W cm^{-1} K^{-1} (between those of copper and silicon).[11]

One can see that, at the input power density of 20 kW/cm^2, at which the efficiency of a state-of-the-art blue laser maximizes, the chip must be less than a tenth of a mm^2 in area, but the heat-sink-limited light flux out of such a chip would still be over 1,000 lm (the rough equivalent of the light flux from a 75 W incandescent lamp). In other words, single high-power-density heat-sink-limited chips should have no difficulty generating useful amounts of light.

Instead, for this approach, the main issue will be getting to ultra-high efficiencies. Although peak efficiencies for a laser aren't limited by either the *A* or *C* coefficients in the *ABC* model for efficiency, they are limited by other losses such as injection efficiency, optical absorption and series resistance. So an important challenge will be to understand the origin of those losses, and to design ways of reducing them.

2.5 Summary

First, solid-state lighting is the latest in a series of technologies of ever-increasing efficiencies for producing white light.

Second, getting to the highest possible efficiencies will ultimately require both efficient electroluminescence across the visible as well as spiky spectra. Indeed, the ultimate spiky source, a 4-color laser illuminant, has been shown to provide very good color rendering quality and to maximize white luminous efficacy of radiation, and hence can be thought of as the "gold standard" for white lighting.

Third, getting to the highest possible efficiencies has potentially huge economic benefit, and is likely worth pursuing.

Fourth, and finally, the two different routes for ultra-efficient SSL that could be pursued, low and high power density chips, are both promising but face different

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[11]Note that this thermal conductivity implies, through Eq. (2.4), a thermal resistance of 2.5 K/W for 1 mm^2 chip size, in the range of (but slightly lower than that, 5.5 K/W, for a Philips Lumileds Luxeon K2 package [25].

challenges. The challenge for the low-power-density approach is not so much to improve efficiency, as this is already very high, but to reduce cost while maintaining such high efficiency (or to increase the power density at which such high efficiency can be obtained). The challenge for the high-power-density approach is not so much to reduce cost (as there is an existence proof that this can be low enough), but to increase efficiency.

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Chapter 3
Epitaxy Part A. LEDs Based on Heteroepitaxial GaN on Si Substrates

Takashi Egawa and Osamu Oda

Recently, LEDs based on heteroepitaxial GaN on Si substrates (GaN/Si) are attracting a great attention and are industrially developed by many companies in order to compete with LEDs based on GaN/sapphire and GaN/SiC which are commercially marketed advancing GaN/Si LEDs, though GaN/Si LEDs have been realized early in 2002. The recent attention on GaN/Si is due to the availability of low cost and large wafer size availability (up to 300 mm diameter) of Si substrates. In this article, we review the developments of the GaN epitaxial growth on Si and summarize the developments made in our laboratory including the device structures achieving the GaN/Si LEDs with higher emission efficiency. We describe on GaN/Si using (i) high temperature (HT) AlN/AlGaN intermediate layers, (ii) HT intermediate layers (ILs) and multilayers (MLs), and (iii) strained-layer superlattices (SLS) interlayers and their LED performances respectively. We believe that GaN/Si LEDs with low prices will become important LEDs for general lighting in the near future.

3.1 Introduction

Since there were no low cost and high quality bulk substrates available for the epitaxial growth of group-III nitride materials, InGaN-based light emitting diodes (LEDs) have been fabricated on expensive sapphire [1–13] or SiC substrates [14–21] and these LEDs are commercially fabricated worldwide, not only for blue LEDs but

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also for white LEDs, which are very promising for the reduction of energy consumption replacing incandescent and fluorescent bulbs. This replacement is expected to contribute to the reduction of CO_2 emission with the level of several percent [22]. In order to achieve the replacement of general lighting, we still need the improvement of LEDs from the view point of cost-effectiveness and high output power.

The present widely produced InGaN-based LEDs on sapphire substrates have disadvantages that the output power is limited because of its low thermal conductivity of sapphire substrates and the cost is high because both electrodes has to be deposited only on one side of the LED chip since sapphire is insulating. The tip area is almost doubled compared with the case when the electrode can be made on the back side of the chip. On the other hand, SiC have advantages such as its high thermal and electrical conductivity that make the above mentioned disadvantage solved by using SiC substrates. SiC however is costly because commercial substrates are produced by the physical vapor transport method.[1] Because of the limitation based on the bulk crystal growth method, though 100 mm diameter substrates have been developed and now commercially available [24–28], 150 mm diameter substrates are not yet commercially available although its fabrication was recently reported [29, 30]. The future perspective of larger diameter substrates is thus not yet well established. SiC has therefore a disadvantage from the viewpoint of the substrate cost, though it is improved year by year.

Compared with these disadvantages of sapphire and SiC substrates, GaN/Si is recently taking attention from the viewpoint of its high thermal and electrical conductivity and low substrate cost, for reducing the LED fabrication cost, especially for applying LEDs for general lighting which actually has a huge market counting several tens B\$ [22]. The advantage of GaN/Si based LEDs is its low material cost since Si substrates are obtainable even to the 300 mm diameter with a reasonable price. Moreover, the fabrication process of LEDs on Si substrates becomes easier than that of LEDs on sapphire because one of the Ohmic contacts can be made from the backside through a conductive Si substrate. The other advantage of GaN/Si LEDs is the possibility of the integration with other devices on Si substrate. It is also pointed out that at high output power, GaN/Si is more advantageous than GaN on sapphire since Si is thermally conductive and higher current can be flowed to obtain higher output power. Nevertheless, GaN/Si based LEDs have less developed compared with LEDs based on sapphire and SiC substrates because of the following obstacles.

In order to realize high-quality GaN/Si LEDs, we need to overcome the problems caused by large mismatch in lattice constants (16.9 %) and difference in thermal expansion coefficients (57 %) as shown in Fig. 3.1. These factors attribute to large wafer bending, epitaxial layers with high threading dislocation density ($\sim 10^{10}$ cm^{-2}), deep pits and cracks to relax the large tensile stress induced by the substrate [31]. The heteroepitaxial growth of GaAs on Si has been already extensively studied [32–35] and the heteroepitaxial growth of GaN on Si has been

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[1] See a review in [23].

Fig. 3.1 Thermal and lattice mismatches in the growth of heteroepitaxy
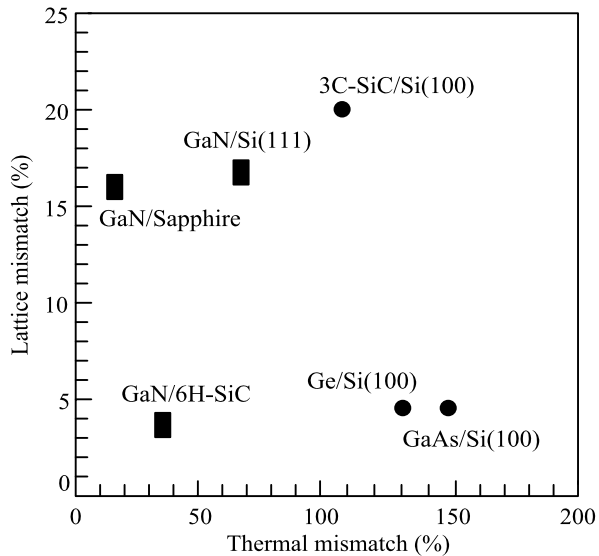
developed based on these precedent GaAs epitaxial growth works. The problem as mentioned above has been solved by applying various intermediate layers (they are also called buffer layers). Since it is known that GaN grown on Si has very low device performances compared with GaN on sapphire, it is important to improve the crystal quality. The growth of GaN on Si has been performed initially by molecular beam epitaxy (MBE) [36–42] and later by metal organic chemical vapor deposition (MOCVD). The various works have been performed using intermediate layers such as 3C–SiC [43], AlN [44–59], GaN [60, 61], AlN/AlGaN [62–79], AlAs [80], oxidized AlAs [81–83], Al_2O_3 [84], Si_xN_y [85–87], ZrB_2 [88–93] and silicon delta-doping [94]. The insertion of AlN/GaN super-lattice structures (SLSs) [95, 96] and AlGaN/GaN SLSs [97–99] were effective to improve the luminosity. The reduction of dislocation densities and the prevention of cracking have been performed by various technologies such as epitaxial lateral over growth (ELO) [100–106], patterned Si [107–114], selective area growth (SAG) [59, 115–137], nanohetero epitaxy (NHE) technologies [138–145] and growth on compliant substrates [146–150]. Using some of these technologies, GaN/Si epitaxial layers up to 100 mm [75] and 150 mm [151, 152] have been achieved. In fact, based on the above technologies, the strain in the GaN/Si was relieved as discussed in various references [153–156].

The other obstacle of GaN/Si is in that the half of emitted light from InGaN-based multi-quantum wells (MQWs) is absorbed by Si substrates [157, 158] which has small band gap of 1.12 eV. In order to solve this light absorption, distributed Bragg reflector (DBR) structures [158, 159], wafer bonding [160], laser lift-off [99, 161], and light reflecting interlayer [160] technologies have been developed. In fact, various LEDs such as ultraviolet LED (UV-LED) and violet LED [46, 47], blue LEDs [49, 67, 71, 72, 93, 96, 111, 113, 114, 117, 119, 127, 132, 134, 160–164], green LEDs [65, 66, 71, 95, 98, 165] using these GaN/Si epitaxial layers have been re-

ported since 1998. Various device structures in order to reduce the series resistance of the vertical LEDs, such as wafer bonding [160], laser lift-off [99] and through-holes [165] have also been studied. Some of these developments have been reviewed in Refs. [35, 151, 166–168].

By solving these above mentioned problems, GaN/Si LEDs have already been commercialized by Sanken Electric Co. early in 2002 that collaborated with us [169] though at that time GaN/Si LEDs were believed to be very inferior to GaN/sapphire or GaN/SiC LEDs and to be difficult to make business. GaN/Si LEDs are now however becoming very promising for general lighting applications. Various companies hereafter are looking for the commercialization based on 100–200 mm diameter Si substrates [170–175], aiming at the cost reduction of 75–80 % by reducing reject rates, minimizing batch time and utilizing automated semiconductor processing equipments, compared with the present LEDs based on sapphire and SiC substrates [176]. Even high bright (HB) LEDs have been achieved with 100–200 lm at 350–1000 mA [177, 178] for applications to domestic, architectural, medical and automobile lighting. GaN/Si LEDs have therefore the possibility to become the main LEDs in the near future for general lighting.

In this article, we review GaN/Si epitaxial growth and LED fabrication developments (as summarized in Table 3.1), mainly focusing on our laboratory's work performed since 1998.

3.2 Epitaxial Growth and Characterization

In order to reduce dislocations due to lattice mismatching and cracks due to the difference of thermal expansion coefficients, various epitaxial structures have been studies.

3.2.1 GaN Growth on Sapphire

Blue LEDs were first developed using ZnSe and SiC substrates.[2] Blue LEDs based on GaN were then reported as metal-insulating-semiconductor (MIS) structure GaN based LEDs [180] but their emission efficiency was too weak for industrialization. The crystal quality of GaN on sapphire substrates was improved by MBE growth [181], hydride vapor phase epitaxy (HVPE) growth [182] and then greatly by MOCVD [2, 6]. *p*-type GaN was first realized by electron beam irradiation [4] followed by thermal annealing [8] and *pn* junction GaN based LEDs have been realized using InGaN as the emission layer grown on sapphire substrates [9–13]. These developments are reviewed in Refs. [157, 183]. These blue LEDs showed a significant emission efficiency and this was attributed to the carrier confinement into

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[2]See reviews in [23, 179].

Table 3.1 GaN/Si epitaxial growth and LED fabrication

Growth method	Content	Author	Year	Refs.
LPCVD + MOCVD	GaN on Si(111) through 200 nm 3C–SiC IL	Takeuchi et al.	1991	[43]
MOCVD	GaN on Si(111) through LT GaN IL	Ohkoshi et al.	1991	[60]
MOCVD	100–200 nm AlN IL, effect of growth temperature	Watanabe et al.	1993	[45]
GS-MBE	GaN through AlN/GaN SLS on SiC/SOI substrate	Yang et al.	1995	[146]
CVD + MOCVD	GaN on SOI substrate carbonized by rapid thermal CVD	Steckle et al.	1996	[147]
HVPE	Thick GaN on Si(111) through HT AlN IL	Hashimoto et al.	1996	[80]
MOCVD	AlO_x IL by oxidation of AlAs layer	Kobayashi et al.	1997, 1998	[81, 82]
MOCVD	GaN on SOI substrate, effect of Si overlayer	Cao et al.	1997	[148, 149]
MOCVD	Initial stage of nitridation of Si	Hashimoto et al.	1997	[191]
PA-MBE	First LEDs using GaN/Si	Guha et al.	1998	[46, 47]
PA-MBE	GaN on Si(111) through amorphous like SiN_x IL	Nakada et al.	1998	[36]
LPCVD + MOCVD	GaN through γ-Al_2O_3 IL on Si(111) and orientation relationship	Wang et al.	1998	[84]
MOCVD	Difficulty of LT-GaN growth on Si	Ishikawa et al.	1998	[61]
MOCVD	First InGaN/GaN MQW blue LED using GaN/Si	Tran et al.	1999	[49]
MOCVD	GaN on Si(111) through AlAs/AlN IL	Strittmatter et al.	1999	[83]
MOCVD	SAG on dot- and stripe-patterned Si (111) using AlGaN IL	Kawaguchi et al.	1998–2000	[115–117]
PA-MBE	Two-step growth of GaN on Si through AlN IL	Calleja et al.	1999	[38]
GS-MBE	AlN IL with short period GaN/AlGaN SLS	Nikishin et al.	1999	[37]
MOCVD	AlN IL and the origin of cracks	Follstedt et al.	1999	[51]
MOCVD	Effect of HT AlN to prevent melt-back etching and LED fabrication	Egawa et al.	1999–2002	[62–66]
MOCVD	GaN on N^+ ion-implanted Si(111) through AlN or GaN/AlN IL	Koh et al.	2000	[150]

Table 3.1 (Continued)

Growth method	Content	Author	Year	Refs.
GS-MBE + MOCVD	Blue LED using SAG GaN/Si (300 × 300 mm) through MBE grown AlN IL	Yang et al.	2000	[119]
MOCVD	NEH growth on patterned SOI and theoretical consideration	Zubia et al.	2000	[138, 139]
MOCVD	Appropriate condition for AlN IL growth	Lahreche et al.	2000	[52]
MOCVD	Maskless ELO on AlN IL	Lahreche et al.	2000	[100]
GS-MBE	SAG on patterned Si(111) through AlN IL or AlGaN/GaN SLS	Seon et al.	2000	[118]
MOCVD	AlGaN/GaN structure though AlN IL	Schremer et al.	2000	[54]
MOCVD	Crack-free growth of GaN by LT AlN IL	Dadger et al.	2000	[53]
LPCVD + MOCVD	Interface interaction of Si and GaN	Graupner et al.	2000	[192]
MOCVD	Layer by layer growth by Si dosing	Munkholm et al.	2000	[193]
MOCVD	Residual stress in GaN on Si(111) with the V/III molar flow ratio	Fu et al.	2000	[153]
MOCVD	EL from InGaN/GaN QW on AlAs IL	Dadgar et al.	2000	[162]
MOCVD	HT AlN IL on Si(111) by rapid thermal process MOCVD	Chen et al.	2001	[195]
MOCVD	Crack-free growth of GaN by AlN IL or AlN/GaN ML	Lahreche et al.	2001	[70]
MOCVD	Crack free GaN using AlN/GaN SLS, strain measurement and green LED fabrication	Feltin et al.	2001	[95, 154, 155]
MOCVD	Growth of InGaN QW and LT AlN + GaN/AlGaN ML on grid patterned Si (111) (100 × 100 μm)	Dadgar et al.	2001	[163]
MOCVD	Blue EL emission from InGaN QW on LT AlN IL + GaN/AlGaN ML	Dadgar et al.	2001	[67]
APCVD + MOCVD	Pendeo-epitaixal growth on 3C–SiC/Si CVD template	Davis et al.	2001	[103]
MOCVD	GaN thorough HT AlGaN IL on periodically grooved Si	Detchprohm	2001	[102]
MOCVD	Maskless ELO of GaN on parallel-grooved Si(111)	Strittmatter et al.	2001	[101]
GS-MBE	Crack-free growth of GaN up to 3 μm on Si(111) through AlN IL	Semond et al.	2001	[41, 42]

Table 3.1   (Continued)

Growth method	Content	Author	Year	Refs.
MOCVD	SAG of GaN on dot and stripe patterned Si(111) and (001)	Honda et al.	2001	[120, 121]
MOCVD	HT AlN IL and graded IL from AlN to GaN	Marchand et al.	2001	[68]
UHVCVD	Step-graded AlGaN IL between GaN and AlN IL	Kim et al.	2001	[69]
MOCVD	Si_xN_y IL between AlN BL and GaN	Hageman et al.	2001	[85]
MOCVD	Reduction of cracks by GaN grown on patterned Si	Zamir et al.	2001, 2002	[107–109]
MOCVD	Dislocation density reduction and crack-free epitaxial growth up to 100 mm diameter using HT AlN/AlGaN + AlN/GaN ML, GaN/Si LEDs comparable with LEDs on sapphire	Egawa et al.	2002–2005	[35, 71–78]
MOCVD	Coating of the reactor wall by AlN for HT AlN IL growth	Lee et al.	2002	[196]
MOCVD	Growth of 100 mm diameter GaN/Si	Brown et al.	2002	[56]
MOCVD	Concept of NHE and growth of GaN on Si	Hersee et al.	2002	[140]
MOCVD	Dislocation annihilation by Si δ-doping	Contreas et al.	2002	[94]
MOCVD	Cantilever epitaxy with AlN IL on patterned Si	Katona et al.	2002	[104]
MOCVD	Crack-free 2 inch using two LT AlN ILs with Si_xN_x mask insertion on LT AlN IL	Dadger et al.	2002	[113, 114]
MOCVD	Review of GaN/Si	Krost et al.	2002	[166]
MOCVD	Growth of GaN/AlGaN and QW on (1–101) GaN side facets of truncated triangular GaN grown on Si(111)	Kato et al.	2002	[105]
MOCVD	SAG of (1–101) GaN on off-angled Si(001)	Honda et al.	2002	[122]
MOCVD	SAG of crack-free GaN on grid mask patterned Si(111)	Honda et al.	2002	[123, 124]
MOCVD	Effect of AlN IL on stress reduction	Krost et al.	2003	[167]
MOCVD	Appropriate growth temperatures for LT AlN + HT GaN + LT AlN IL structure	Dimke et al.	2003	[58]
MBE	NHE on nanopore array Si	Liang et al.	2003	[141]
PA-MBE	Growth of AlGaN on Si through ZrB_2 IL	Tolle et al.	2003, 2004	[88, 89]

Table 3.1 (Continued)

Growth method	Content	Author	Year	Refs.
MOCVD	Blue LED on AlN/AlGaN IL and AlGaN/GaN SLS structure	Lee et al.	2004	[96]
MOCVD	Reduction of dislocation density by Si_xN_x layer insertion	Lee et al.	2004	[86]
MBE	Growth of GaN on nonporous array Si surface	Sun et al.	2004	[142]
MOCVD	Improvement of light extraction by DBR structure	Ishikawa et al.	2004	[158, 159]
MOCVD	Crack-free MQW LEDs by metal bonding and selective lift-off	B. Zhang et al.	2005	[79]
MBE	Growth of GaN on Si through ZrB_2 IL	Yamada-Takamura et al.	2005	[91]
MOCVD	NHE overgrowth of GaN on nanopatterned SiO_2 mask deposited on GaN/AlN/Si	Zang et al.	2005	[143]
MOCVD	Lateral epitaxy on patterned Si substrate	Kim et al.	2005	[106]
HVPE	Thick GaN on AlN templated Si (111) substrate	Honda et al.	2005	[59]
MOCVD	InGaN/GaN DH structure on stripe patterned Si(111)	Narita et al.	2005	[125]
MOCVD	Crack-free 5.4 μm thick GaN on 150 mm diameter Si	Dadger et al.	2006	[151]
MOCVD	NHE on nonporous Si surface prepared by anodized Al_2O_3 template as a mask	Zang et al.	2006	[144]
MOCVD	NHE on patterned SiO_2/GaN/Si substrate	Wang et al.	2006	[145]
MOCVD	Mechanical properties of GaN grown on patterned Si	Z. Yang et al.	2006	[110]
PA-MBE	Nitridation of ZrB_2 on Si	Wang et al.	2006	[92]
MOCVD	Precise conditions for SiN IL growth	Rieman et al.	2006	[87]
MOCVD	Blue LEDs on patterned Si through AlN and SiN_x ILs	B. Zhang et al.	2007	[111]
MOCVD	Crack-free Growth of GaN on 150 mm Si wafers	Ubukata et al.	2007	[152]
MOCVD	Blue LEDs with photonic crystals using Si mold substrates	Orita et al.	2008	[201]
MOCVD	Crack-free GaN on patterned Si through HT AlN IL	Lee et al.	2008	[112]

Table 3.1 (Continued)

Growth method	Content	Author	Year	Refs.
MOCVD	Semipolar (1–101) and (11–22) GaN on patterned Si (001) and Si(113) and InGaN/GaN SQW LEDs	Tanikawa et al.	2008, 2009	[127–129]
MOCVD	Nonpolar (11–20) GaN on patterned Si(110)	Tanikawa et al.	2008	[126]
MOCVD	LED structure on Si(110) substrate	Reiher et al.	2009	[164]
MOCVD	Growth of (11–22) GaN on patterned Si(311) with and without SiO_2 mask	Yang et al.	2009	[130, 131]
MOCVD	AlGaN/GaN SLS on AlN IL + AlN/GaN ML, improvement of crystal quality and fabrication of LEDs	Shuhaimi et al.	2009, 2010	[97–99]
MOCVD	Vertical LED with through hole structure, reduction of the series resistance	Wei et al.	2010	[165]
MOCVD	Semi polar (11–22) GaN stripe crystal on Si by dual selective MOCVD	Murase et al.	2011	[135]
AP-MOCVD	Semi polar (1–101) InGaN/GaN MQW LEDs on patterned Si	Chie et al.	2011	[133, 134]
MOCVD	Growth of InGaN/GaN MQW on nonpolar and semipolar GaN on patterned Si	Tanikawa et al.	2011	[136]
MOCVD	High bright vertical LEDs on Si using wafer bonding	Lee et al.	2011	[160]
GS-MBE + MOCVD	InGaN/GaN LEDs on Si through ZrB_2 BL	Blake et al.	2012	[93]
MOCVD + HVPE	In-situ void formation using a AlN shell structure on GaN stripes on Si	Mitsunari et al.	2012	[137]

MOCVD: Metal-Organic Chemical Vapor Deposition, MBE: Molecular Beam Epitaxy, GS-MBE: Gas Source MBE. PA-MBE: Plasma Assisted MBE, HVPE: Halide or Hydride Vapor Phase Epitaxy, CVD: Chemical Vapor Deposition, LPCVD: Low Pressure CVD, UHVCVD: Ultra High Vacuum CVD, APCVD: Atmospheric Pressure CVD, SOI: Silicon-On-Insulator, SAG: Selective Area Growth, ELO: Epitaxial Lateral Overgrowth, NHE: Nano-Hetero-Epitaxy, DBR: Distributed Bragg Reflector, BL: Buffer Layer, IL: Intermediate Layer, ML: Multiple Layer, SLS: Super-Lattice Structure, QW: Quantum Well, MQW: Multiple Quantum Well, HT: High Temperature, LT: Low Temperature, LED: Light Emitting Diode; DH: Double Hetero

In islands formed due to the segregation of In in InGaN [184–186]. Since the In island density is about 10^{20} cm^{-2} which is much higher than the dislocation density 10^9 cm^{-2} and the carrier diffusion length is very short with this In content, it seems

apparent that dislocations do not affect the emission efficiency. It is however known that when the carrier injection rate is very high and carriers overflow the band barrier of In islands, overflowed carriers recombine at dislocations, and then the dislocation density affects the emission efficiency. It is therefore necessary to reduce dislocation densities in order to increase the emission efficiency of high power LEDs. Egawa et al. [187] have shown that by using AlN/sapphire templates, higher current can flow and higher luminosity can be obtained.

3.2.2 GaN Growth on SiC

SiC was predicted as a promising substrate for GaN epitaxial growth because of its good lattice matching with GaN [188, 189]. 6H–SiC has a lattice constant which is closer to that of GaN than that of sapphire and the lattice mismatching is only 3.4 % and was expected to be good substrates for GaN-based LED fabrication. The quality of epitaxial growth on C-face and Si-face of SiC substrates [189] and the growth of (GaN/AlN) multi-layers on 6H-SiC are also studied [190].

GaN based LEDs on SiC substrates have been then developed [15] and blue LEDs which compete with GaN based LEDs have been achieved [16], especially applying a structure for better light extraction [17]. These GaN/SiC LEDs have been industrially developed and commercially produced [14, 18–21]. The advantage of GaN/SiC LEDs against GaN/sapphire LEDs is that better emission uniformity could be achieved. Since both electrodes can be formed on top and back surfaces and SiC has a high thermal conductivity, higher current densities can be achieved. The disadvantage of SiC is however the substrate cost as already mentioned.

3.2.3 GaN/Si Using Low Temperature (LT) Intermediate Layers

Previously reported GaN-based LEDs on Si have been grown with a low-temperature thick AlN intermediate layer (~750 °C, 8~30 nm) and AlGaN/GaN active layers [46, 49, 67, 95]. The LT grown intermediate layer technique itself was developed for the growth of epitaxial GaAs and GaP on Si [32–35]. It has been called as the two-step growth technique and widely used in the highly mismatched system such as the growth of GaN on sapphire. It was also proven that the SLS is very effective in improving the crystal quality of GaAs on Si. However, the LEDs on Si with a thick LT-AlN intermediate layer suffer from both high operating voltage and high series resistance, which result from the insulating AlN layer and the large band offset at the AlN/Si interface. In fact, the disadvantage of LT intermediate layers has been shown by Ishikawa et al. [61]. This disadvantage which comes from the meltback of Si due to the interaction of Ga with Si has also been studied precisely by many researchers [31, 191–196].

GaN/LT-AlN/Si   GaN/HT-AlGaN/HT-AlN/Si

200 μm
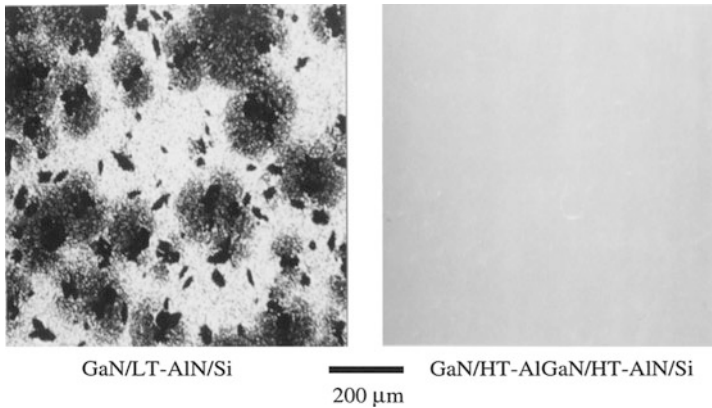
Fig. 3.2 Comparison of surface morphology of GaN on Si using a LT-AlN intermediate layer and a HT-AlN/AlGaN intermediate layer

3.2.4 GaN/Si Using High Temperature (HT) AlN/AlGaN Intermediate Layers

In our laboratory, we have developed several epitaxial growth technologies for GaAs and GaP such as high temperature growth of the intermediate layers and the insertion of SLSs [33] in order to solve the problem of lattice mismatching and cracking. For GaN/Si, we therefore applied these previously developed technologies in order to achieve crack-free large diameter GaN/Si epitaxial growth.

Instead of growing LT intermediate layers, we have grown thin AlN and AlGaN intermediate layers at high temperatures [62–66] on Si substrates with the maximum diameter of 50 mm. This HT growth of intermediate layers is very critical to achieve high quality GaN growth since these HT layers prevent the interaction of Ga with Si (so called meltback phenomenon [31, 191–196]). This is because at high temperatures nitride layers can be grown layer by layer while at low temperatures, the growth takes place with the three dimensional (3D) mode. This 3D mode growth which makes the interaction of Ga with Si and makes it difficult to obtain mirror-like surface and only cloudy surface is obtained.

Figure 3.2 shows the comparison of the surface morphology of the 1-μm thick GaN layers on Si using the conventional LT-AlN intermediate layer (two-step growth technique) and the HT-AlN/AlGaN intermediate layer. The GaN on Si grown by the LT-AlN intermediate layer showed a cloudy surface morphology. However, the surface morphology was improved using the HT-AlN/AlGaN intermediate layer, because the Si substrate was protected from meltback etching by the HT-AlN/AlGaN intermediate layers [63]. The full width at half maximum (FWHM) values of X-ray rocking curve (XRC) symmetric (0004) and asymmetric (20–24) reflections were 670 and 1535 arcsec, respectively.

Figure 3.3 shows the cross-section of the AlGaN/AlN intermediate layer. No hollows were observed at the interface of between the AlN and the Si substrate which

Fig. 3.3 Cross-sectional
SEM image of GaN on Si
with an AlGaN/AlN
intermediate layer (reprinted
from Ref. [63] with
permission, copyright 1999
Japan Society of Applied
Physics)
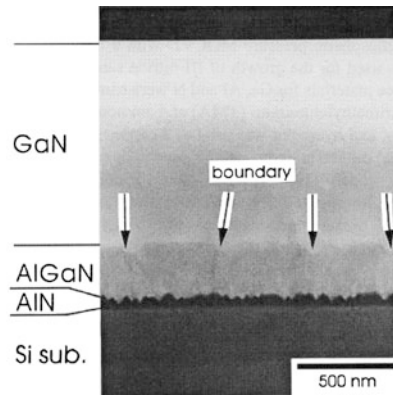
Fig. 3.4 Schematic
cross-sectional structure of
InGaN MQW LED on Si
(reprinted from Ref. [74] with
permission, copyright 2003
The Institute of Electronics,
Information and
Communication Engineers)

p-GaN, 0.2 μm
$p\text{-Al}_{0.15}\text{Ga}_{0.85}\text{N}$, 20 nm
InGaN MQW
n-GaN, 0.2 μm
AlN/GaN (5 nm/20 nm) multilayers (20 pairs)
$n\text{-Al}_{0.27}\text{Ga}_{0.73}\text{N}$, 20 nm
AlN, 3 nm and 120 nm
$n\text{-Si}$ (111) substrate

shows the AlGaN/AlN intermediate layer is effective in preventing the meltback
etching of the Si substrate.

3.2.5 GaN/Si Using HT Intermediate Layers (ILs) and Multilayers (MLs)

In order to improve the luminosity, AlN/GaN MLs have been grown on high temper-
ature grown AlN/AlGaN intermediate layers (HT-AlN/AlGaN ILs) on n^+-Si (111)
substrates with the maximum diameter of 100 mm.

Figure 3.4 shows the typical cross-sectional view of this structure. After ther-
mal cleaning at 1130 °C in H_2 ambient to remove the oxide from the Si surface,
AlN layers have been grown with 3-nm and 120-nm thickness to study the effect
of the thickness of AlN layers. Above these HT intermediate layers, a 20-nm-
thick $n\text{-Al}_{0.27}\text{Ga}_{0.73}\text{N}$ layer, 20-pair of AlN (5 nm)/GaN (20 nm) multilayers at
1080 °C, a 0.2-μm-thick n-GaN layer, an InGaN MQW structure, a 20-nm-thick
$p\text{-Al}_{0.15}\text{Ga}_{0.85}\text{N}$ layer and a 0.2 μm-thick p-GaN layer were deposited subse-
quently. The active layer consists of 3-nm-thick $In_xGa_{1-x}N$ wells and 5-nm-thick

$In_{0.01}Ga_{0.99}N$ barrier layers. The typical In-content of the wells were 17 % and 23 % for the blue and green LEDs, respectively. For comparison, the same structure has been grown on sapphire substrates.

The dark spot densities were measured by using a cathodoluminescence technique and were found to be 5×10^8 and 4×10^9 cm^{-2} for the n-GaN layers on sapphire and Si, respectively. Transmission electron microscopy (TEM) analysis was carried out to examine the microscopic structure. Figure 3.5(a) shows a cross-sectional TEM image for the overall LED structure and Fig. 3.5(b) is an enlarged picture near the active layer. As shown in Fig. 3.5, the V-defects were initiated at the threading dislocations during MQW growth. The V-defect has been observed in the InGaN MQW structure on sapphire [77, 197]. It is noteworthy that the pyramid-shaped structures are formed during the growth of last few QWs and no threading dislocation was observed in these pyramid-shaped structures. The QWs near the top of pyramid-shaped structures become the quantum-dot-like structure. The dots are $\sim$10 nm in diameter and $\sim$3.8 nm in height. Moreover, the valleys between pyramid-shaped structures are filled with p-AlGaN and p-GaN layers during growth at high temperature due to the enhanced surface migration of reactants. Because the resistivities of both p-GaN and p-AlGaN layers are relatively high, the holes could be effectively injected into the QWs near the top of pyramid-shaped structure through the shortest current passage from the p-electrode.

3.2.6 GaN/Si Using SLS Interlayers

The InGaN-based blue LED structure grown with a DBR was reported by Ishikawa et al. to reduce the light absorption in the Si substrate by the Si(111) substrate [158, 159]. We have recently found that the insertion of $Al_{0.06}Ga_{0.94}N$/GaN SLS interlayers to the InGaN-based MQW structures grown on Si(111) substrates with the HT AlN/GaN intermediate layer is very effective to reduce the light absorption by the Si substrate. We adopted the SLS interlayer instead of the DBR layer for its superiority in vertical carrier propagation by tunneling-effect while simultaneously maintaining good reflectivity. Reflectivity can also be improved either by increasing the Al composition or by increasing the pair numbers of the SLS stacks. The SLS structure is also less prone to cracking compared to the DBR structure.

For the SLS sample, a 20-nm-thick AlN layer was grown at 1100 °C on the Si (111) substrates as the intermediate layer, followed by 40 pairs of AlN/GaN ML with respective thickness of 5 and 20 nm. Subsequently, a 400-nm-thick interlayer was grown which consists of 80 pairs of $Al_{0.06}Ga_{0.94}N$/GaN strained-layer superlatives (SLSs) with respective thickness of 2.5 nm for each layer. The $Al_{0.06}Ga_{0.94}N$ portion of the SLS pair was doped with Si ($\sim$1 $\times 10^{18}$ cm^{-3}) to produce a modulation-doped (MD) structure. For comparison, the SLS interlayer was substituted by a Si-doped ($\sim$1 $\times 10^{18}$ cm^{-3}) GaN layer conventionally used for the LED growth. Subsequent layers were kept identical for both samples with a 60 nm n-GaN waveguide layer, 10 pairs of 1.5 nm $In_{0.16}Ga_{0.84}N$ well and 9 nm
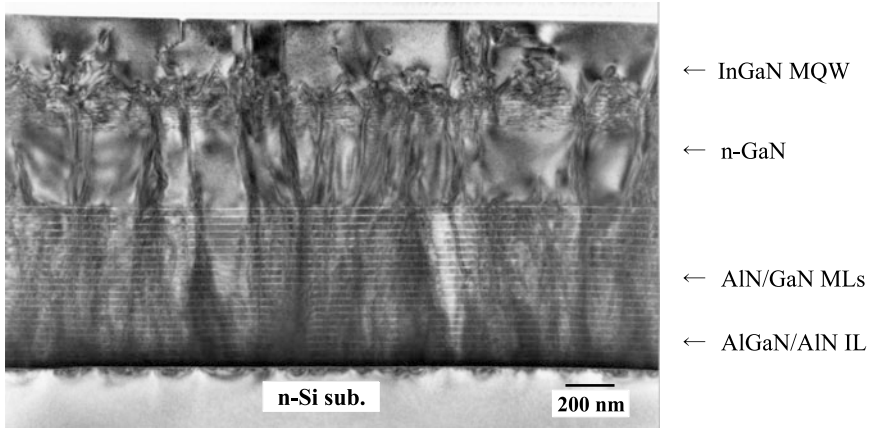
(a)
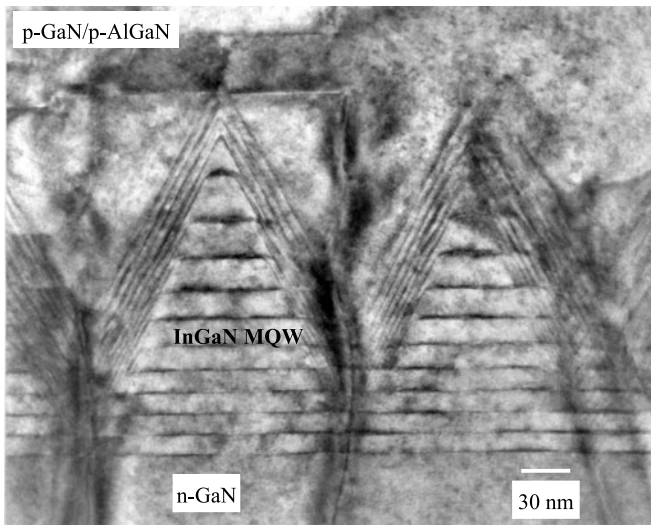
(b)

Fig. 3.5 Cross-sectional TEM images for (**a**) the overall structure and (**b**) the near active layer of InGaN MQW LED on Si with the thin AlN layer (reprinted from Ref. [77], copyright 2005 The Institute of Electrical and Electronics Engineers, Inc. and from Ref. [99] with permission, copyright 2010 IOP Publ.)

$In_{0.08}Ga_{0.92}N$: Si barrier in the active layer and a 10 nm GaN cap layer. Reactor temperature was decreased to 800 °C during the active layer growth.

All samples grown were crack-free and showed no melt back-etching. Samples with the $Al_{0.06}Ga_{0.94}N$/GaN SLS interlayer showed a specular surface for the whole wafer. However, the sample with the GaN interlayer was cloudy at the wafer edge due to the thick GaN layer.
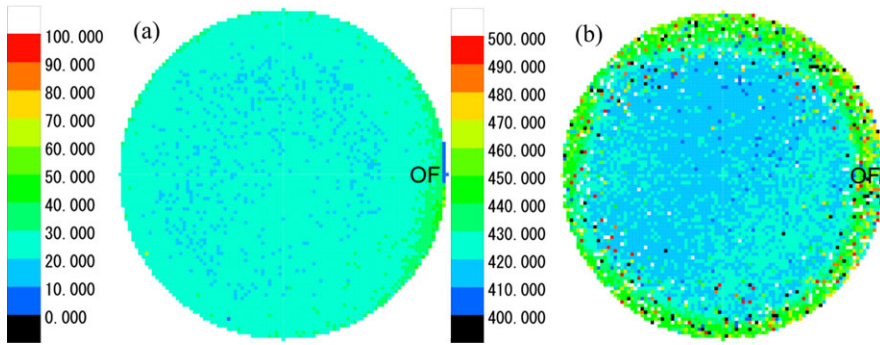
Fig. 3.6 PL surface mapping showing the emission peak wavelength FWHM distribution for wafers with (**a**) an $Al_{0.06}Ga_{0.94}N$/GaN SLS interlayer and with (**b**) a GaN interlayer (reprinted from Ref. [98] with permission, copyright 2010 Japan Society of Applied Physics)
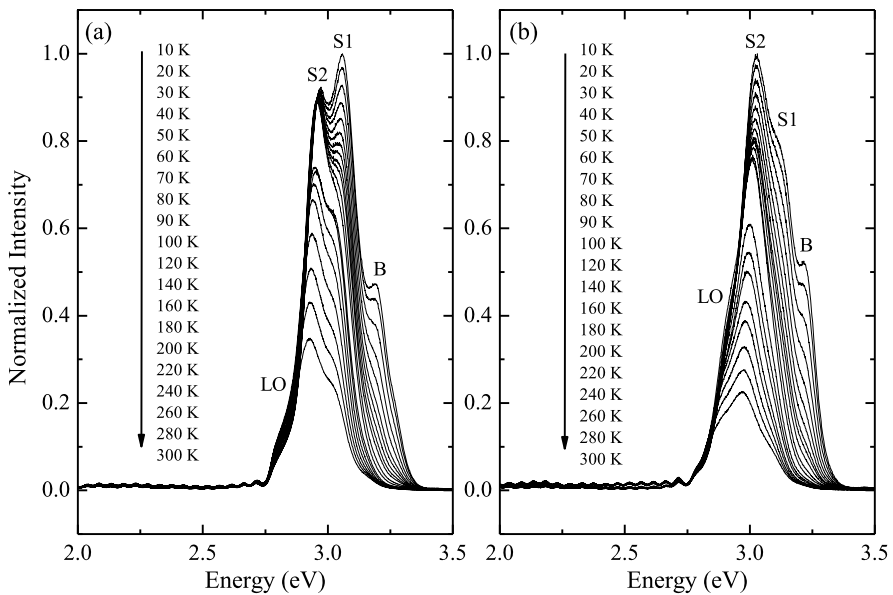
Fig. 3.7 PL spectra of (**a**) a sample with an $Al_{0.06}Ga_{0.94}N$/GaN SLS interlayer and (**b**) a sample with a GaN interlayer recorded in 10 K to 300 K temperature range (reprinted from Ref. [98] with permission, copyright 2010 Japan Society of Applied Physics)

The PL spectrum has been studied and we found several features of the effect of the insertion of SLS structures.

(i) The uniformity of the MQW emission was improved compared with that without SLS structures from 11.1 % to 2.5 % (Fig. 3.6).

(ii) It shows the main peak at 3.05 eV (S1) while without SLS, 3.02 eV (S2) is the main peak at low temperature as shown in Fig. 3.7. The sample with SLS al-
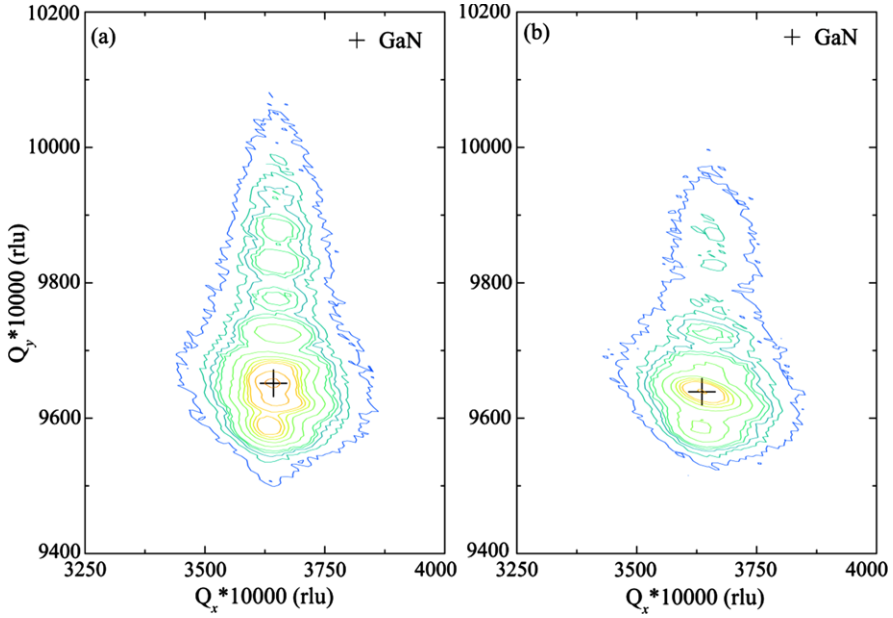
Fig. 3.8 RSM around $(10\bar{1}5)$ diffraction plane of (**a**) a sample with an $Al_{0.06}Ga_{0.94}N/GaN$ inter-layer and (**b**) a sample with a GaN interlayer. The GaN position is marked in each figure (reprinted from Ref. [98] with permission, copyright 2010 Japan Society of Applied Physics)

<div style="margin-left:2em">

ways demonstrates a higher intensity than that without SLS because the 1.5 nm thin MQW layer gives minimal effect to electron-hole separation in the MQW.

(iii) With the temperature increase, the sample with SLS layers shows a smaller emission peak shift than that without SLS layers.

(iv) The activation energy is 151.1 meV for the sample with SLS which suggests that thermal quenching in that sample is attributed by the thermal escape of electrons and/or holes from the InGaN wells into the InGaN barriers [198].

 (v) Internal quantum efficiency (IQE) η_{int} was estimated from integrated photolu-minescence (PL) intensity ratio of 300 K to 10 K ($I_{300\,K}/I_{10\,K}$), assuming that η_{int} at 10 K is 100 %. The estimated η_{int} in the sample with SLS and the sample only with the GaN layer was 29.4 and 20.6 %, respectively. The higher IQE agrees well with the mechanism of thermal quenching in InGaN MQW [97].

</div>

All these features are due to the crystal quality improvement by the insertion of SLS layers. In fact, we have evaluated the crystal quality by TEM observation and reciprocal space mapping (RSM) measurement (Fig. 3.8) and found that SLS layers have an effect to bend threading dislocations from AlN/GaN MLs and to block screw dislocations at the ML/SLS interface. The dislocation density was thus reduced to the level of $3\sim6 \times 10^9$ cm^{-2} at the upper edge of SLS layers.

Ni/Au

Ni/Au (Semitransparent)

| P$^+$-GaN, 200nm, 1180℃ |
| p-Al$_{0.15}$Ga$_{0.85}$N, 20nm, 760℃ |
| InGaN MQW |
| n-GaN, 800nm, 1180℃ |
| n-Al$_{0.27}$Ga$_{0.73}$N, 380nm, 1180℃ |
| AlN, 120nm, 1180℃ |
| n$^+$-Si(111) substrate |

AuSb/Au

(a)

| P$^+$-GaN, 200nm, 1180℃ |
| p-Al$_{0.15}$Ga$_{0.85}$N, 20nm, 760℃ |
| InGaN MQW |
| n-GaN, 4μm, 1180℃, 4×10^{18}cm^{-3} |
| GaN Buffer Layer, 25nm, 500℃ |
| Sapphire(0001) substrate |

Ti/Al

(b)
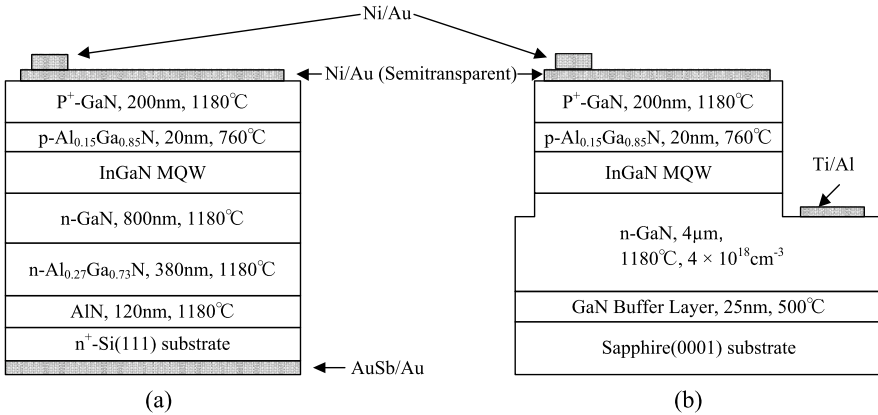
Fig. 3.9 (a) MQW LED structure on Si and (b) MQW LED structure on sapphire (reprinted from Ref. [65] with permission, copyright 2002 The Institute of Electronics, Information and Communication Engineers)

3.3 Fabrication of LEDs and Their Performances

3.3.1 Device Characteristics of LED Structures with HT AlN/AlGaN Intermediate Layers [62–66]

Since we first could grow specular mirror-like surface using HT AlN/AlGaN intermediate layers as described in Sect. 3.2.4, we have fabricated LEDs with MQW structures on them and evaluated their device performances. Ni/Au thin transparent metals and Ni/Au (12/100 nm) *p*-type Ohmic metals were fabricated on top of the GaN/Si LED structures and the *n*-type Ohmic contact was made from the backside through Si substrate using AuSb/Au (18/100 nm) (Fig. 3.9(a)). In the case of LEDs on sapphire substrates fabricated for comparison, both Ohmic contacts were made on the top side with a mesa structure formed by reactive ion etching (Fig. 3.9(b)). The operating voltage of 7.0 V and the series resistance of 100 Ω were obtained for GaN/Si LEDs with the optical output power of 20 μW (the detector was set 10 mm above the surface of LED chip of 10 × 10 mm^2 size for the measurement. The value is more than two orders magnitude lower that the value measured for a molded LED using an integral sphere detector) at the forward current of 20 mA, with a peak emission wavelength of 505 nm, with a full width at half maximum of 33 nm. The LED also exhibited a stable operation over 500 h under the automatic current control (20 mA) condition at 27 °C. However, the optical output power was still half compared with LEDs on sapphire substrates and the operating voltage and the series resistance were higher compared with GaN/sapphire LEDs with 4.5 V and 40 Ω, respectively. This may be due to the higher resistance at the AlN/Si interface.

Fig. 3.10 Comparison of $I–V$ characteristics of LEDs on Si with the different thickness of AlN layer (reprinted from Ref. [99] with permission, copyright 2010 IOP Publ.)
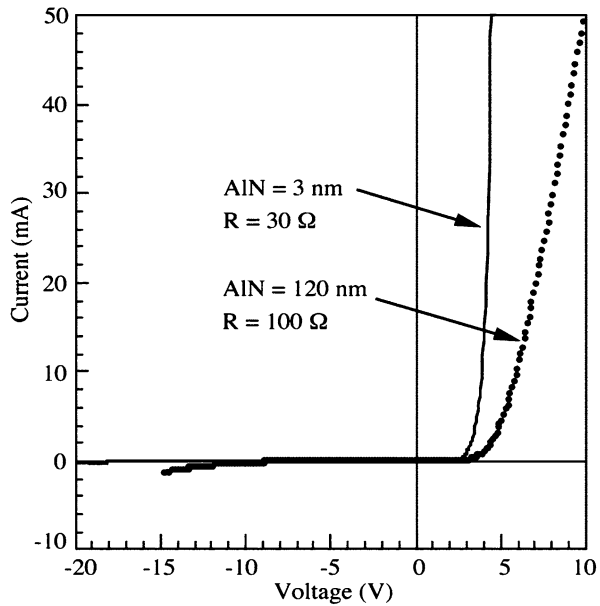
3.3.2 Effect of Thin AlN Intermediate Layers and AlN/GaN MLs [35, 71–78]

In order to reduce the operating voltage and the series resistance to the level of LEDs on sapphire substrates, we have developed a structure with a thin AlN intermediate layer with AlN/GaN MLs, as the crystal growth and the characterization of the grown crystal described in Sect. 3.2.5. We have fabricated LEDs with MQW structures on them and LEDs with the same active layer structure on c-plane sapphire substrates. Ni/Au thin transparent metals and Ni/Au (12/100 nm) p-type Ohmic metals were fabricated on top of the GaN/Si LED structures and the n-type Ohmic contact was made from the backside through Si substrate using AuSb/Au (18/100 nm). In the case of LEDs on sapphire substrates, both Ohmic contacts were made on the top side with a mesa structure formed by reactive ion etching (RIE).

Figure 3.10 shows the $I–V$ characteristics of blue LEDs grown on Si. Compared with the operating voltage of 7.0 V and the series resistance of 100 Ω at the forward current of 20 mA for LEDs described in Sect. 3.3.1, the operating voltage and the series resistance were reduced to 3.8∼4.1 V and 30 Ω, respectively, when the thin AlN layer of 2.5–3 nm together with AlN/GaN MLs was applied. The same active layer structure LED grown on sapphire substrates showed an operating voltage of 3.8 V and a series resistance of 25 Ω at the forward current of 20 mA. GaN/Si LEDs thus showed the similar results as in the case of LEDs on sapphire.

The band diagram at the AlN/Si interface was evaluated using the X-ray photoelectron spectroscopy (XPS) measurement. As shown in the Fig. 3.11, the valence- and conduction-band discontinuities at the AlN/Si interface were evaluated to be

Fig. 3.11 Schematic energy band diagram at the AlN/Si interface (reprinted from Ref. [199] with permission, copyright 2003 Japan Society of Applied Physics)

AlN   Si

E_C^{AlN}

ΔE_C
2.3±0.4 eV

E_C^{Si}

6.2 eV   1.1eV

E_V^{Si}

ΔE_V
2.8±0.4 eV

E_V^{AlN}

70.4±0.2 eV   97.9±0.2 eV

E_{CL}^{AlN}

$\Delta E_B(AlN\text{-}Si)$
24.7±0.2 eV

E_{CL}^{Si}
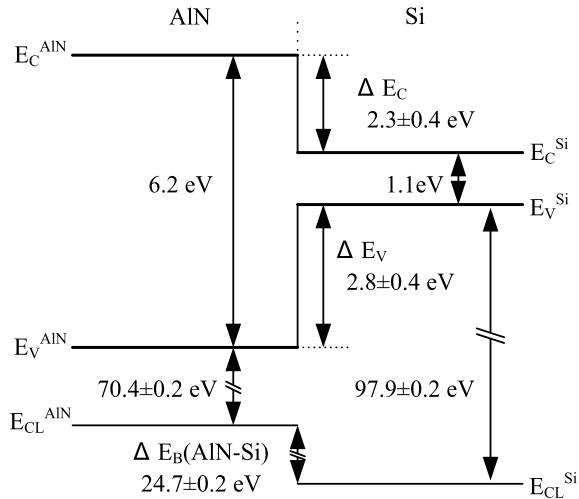
2.8 ± 0.4 and 2.3 ± 0.4 eV, respectively [199]. Moreover, the AlN layer is an insulator with a band gap of 6.2 eV. When an *n*-contact is made on the backside of *n*-Si substrate, electrons are injected into the active layer from the *n*-Si substrate side through the AlN layer. There is a large barrier to the electron injection due to the presence of thick AlN layer. On the contrary, when the AlN layer is thin below a critical value, the operating voltage of the LED can be largely reduced as mentioned above probably because of the formation of the tunnel junction between the *n*-AlGaN layer and the *n*-Si substrate.

Figure 3.12 shows the comparison of *L*−*I* characteristics of the LEDs on sapphire and Si substrates under the highly injected current. The wavelengths at 20 mA were 470 nm and 475 nm for the LEDs on sapphire and Si substrates, respectively. The output power from the LED on sapphire increased with the injected current density up to 100 A/cm^2, and then decreased with increasing the injected current density. Since the thermal conductivity of sapphire is as low as 0.35 W/cmK, the output power of the LED on sapphire is saturated due to the heating effect. The output power from the LED on Si was lower than that on sapphire below the 120 A/cm^2 because of the absorption of emission by Si substrate. However, the output power from the LED on Si did not saturate till the injected current density of 150 A/cm^2 because of higher thermal conductivity of Si substrate (1.5 W/cmK). Thus, the LED on Si exhibited the better characteristic than that on sapphire under highly injected current. Some LED chips were molded by epoxy resin. The optical output power of the molded blue LEDs on Si were as high as 1.5∼2 mW at 20 mA. Based on this technology, very luminous LEDs up to 200 mcd have been commercialized [200]. Taking into account of the absorption of partial emission by Si substrate, it can be concluded that the high-quality active layer is grown on Si substrate in spite of its highly mismatched system. Figure 3.13 shows the emission morphology of a large area green (505 nm) LED on Si with a thin AlN layer under the injected current of 800 mA (120 mA/cm^2). Although the injected current density was as low as

Fig. 3.12 Comparison of
L–I characteristics of LEDs
on sapphire and Si substrates
under highly injected current.
The detector was set 10 mm
above the surface of LED
chip (reprinted from Ref. [74]
with permission, copyright
2003 The Institute of
Electronics, Information and
Communication Engineers)
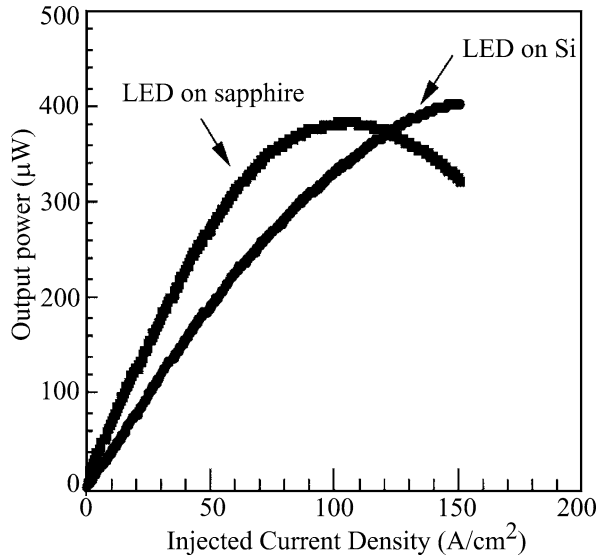
Fig. 3.13 Demonstration of
the light emission from a
large area green LED on Si
with a thin AlN layer under
injected current of 800 mA,
corresponding to the current
density as low as
120 mA/cm^2 (reprinted from
Ref. [99] with permission,
copyright 2010 IOP Publ.)
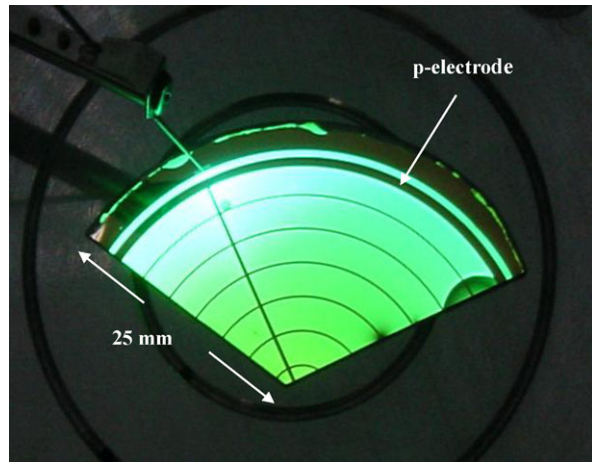
120 mA/cm^2, a uniform emission has been observed for the large area. The onset of emission was observed at 12 mA, corresponding to the injected current density as low as 1.8 mA/cm^2. In Fig. 3.14, blue and green LEDs thus fabricated are shown.

In order to study the reliability of the LED on Si, the aging tests were performed at 27 and 80 °C under automatic current control (ACC). Figure 3.15 shows the reliability results of the blue LED on Si with the thin AlN layer. The output power was constant during the 1000 h aging at 27 °C. At 80 °C. There was a decrease in the output power at the initial stage, which was as low as 6 %, but after this small decrease the output power of LED almost remained constant.

Fig. 3.14 Photographs of the blue and green LEDs on Si emitting at 478 nm and 505 nm (reprinted from Ref. [71] with permission, copyright 2002 Japan Society of Applied Physics)

Fig. 3.15 Variation of the optical output power as a function of aging time measured at 27 and 80 °C (reprinted from Ref. [99] with permission, copyright 2010 IOP Publ.)
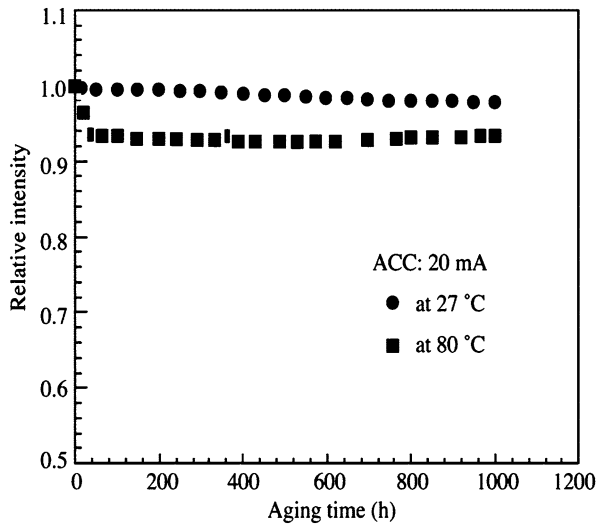
3.3.3 Wafer Bonding and Lift-Off [79]

The problem of GaN/Si LEDs is the partial absorption of the emission by Si substrate. However, the combination of the selective etching of Si and the metal-to-metal bonding was effective in improving the LED characteristics, such as 49 % increase in the optical output power and the reduction of the operating voltage to 3.6 V [79].

In this technology, a 10 nm Ni was deposited and annealed on the top surface of the epitaxially grown structure. The LED structure is the same as described in Sect. 3.3.2. In order to reflect the light emitting from the copper carrier side,

Fig. 3.16 Selective lift-off process flow: (**a**) formation of the *p*-type Ohmic contact and the deposition of a high-reflectivity metal reflector (Al/Au) onto the epitaxial surface of the LED epitaxial wafer, (**b**) bonding of a metal/InGaN MQW LED/Si wafer onto the copper carrier, (**c**) thin Si substrate by mechanical polishing and (**d**) selective removal of the Si substrate by wet-chemical etching. The inset shows the image of the SLO process (reprinted from Ref. [79] with permission, copyright 2005 American Institute of Physics)
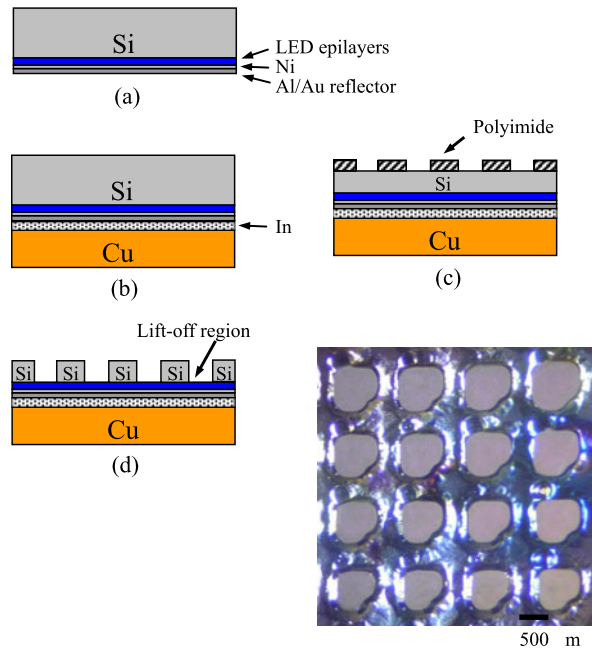
a highly reflective Al/Au metal reflector was deposited on the thin annealed Ni layer. The InGaN MQW LED epitaxial wafer with the metal reflector was then bonded onto an Au plated copper carrier heated at 200 °C using indium as an adhesive. After the substrate was thinned down to about 60 μm by mechanical polishing, the Si substrate was selectively removed by wet-chemical etching in HNA solution (HF:HNO$_3$:CH$_3$COOH = 1:1:1) using a polyimide mask. This selective lift-off (SLO) process flow is shown in Fig. 3.16. Fabrication of LEDs on the substrate removed areas was carried out as follows. First, the AlN/AlGaN buffer layers and AlN/GaN multilayer on the substrate removed areas were removed by RIE in order to further reduce the operating voltage and the series resistance. Then, the exposed *n*-GaN layer was partially etched to the *p*-GaN by RIE. Subsequently, thin Ti/Au (5/5 nm) metal patterns were deposited on the exposed *n*-GaN as a transparent electrode for the *n*-GaN side up LED. Finally, *n*-type contact metal Al/Au (45/60 nm) patterns were formed.

Figure 3.17 shows the $I{-}V$ characteristics of the LED before and after substrate removal. In order to avoid the strain relaxation in the thin nitride epilayer at a high temperature, the *n*-type contact of the LED after substrate removal was formed without any heat treatment. In spite of this, the operating voltage at 20 mA decreases from 4.2 to 3.6 V, and the series resistance decreases from 42 to 27 Ω for the LED after substrate removal. These results are comparable with that of the LEDs on sapphire substrate. The significant improvement is due to the elimination of the large band offset between the AlN buffer layer and Si substrate after substrate removal. The removal of the AlN intermediate layer and AlN/GaN multilayer was also help-

Fig. 3.17 *I–V* characteristics of the LEDs before (*dashed line*) and after (*solid line*) substrate removal. The inset shows an emission image of the LED fabricated on the substrate removal region. The image was taken at 0.5 mA under room light and microscope light conditions. The emission wavelength is about 518 nm (reprinted from Ref. [79] with permission, copyright 2005 American Institute of Physics)
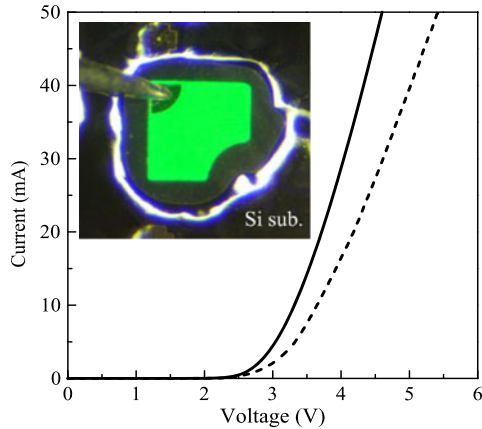
Fig. 3.18 *L–I* characteristics of the LEDs before (*dashed line*) and after (*solid line*) substrate removal using an on-wafer testing configuration. The detector was set 10 mm above the surface of LED chip of $10 \times 10 \ mm^2$ size (reprinted from Ref. [79] with permission, copyright 2005 American Institute of Physics)
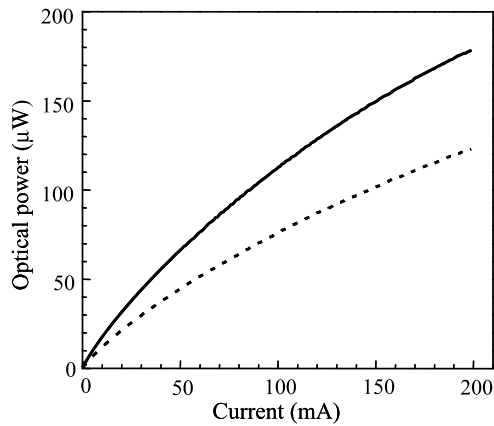
ful to decrease the operating voltage and the series resistance. An emission image of the LED fabricated on the substrate removal region is shown in the inset. The image was taken at 0.5 mA under room light and microscope light conditions. The emission wavelength was about 518 nm. It showed crack-free, high-brightness, and uniform emission.

Figure 3.18 illustrates the light output power versus injection current (*L–I*) characteristics of the LED before and after substrate removal. Since the LED fabricated on the substrate removal region is *n*-GaN layer side up, the 0.2 μm-thick *n*-GaN layer is too thin to spread the current. A transparent conducting metal is needed for this kind of LED. However, unlike the conventional *p*-GaN side up LED, there is no suitable metal which can form Ohmic contact with *n*-GaN layer and also has a high transmittance. We therefore employed Ti/Au (5/5 nm) as a transparent metal for the LED after substrate removal. The transmittance of the Ti/Au layers is only 52 % at the wavelength of 518 nm. In spite of this, the optical power at 20 mA in-

creased from 21 to 31.3 μW (the detector was set 10 mm above the surface of LED chip of 10×10 mm^2 size for the measurement. The value is more than two orders magnitude lower that the value measured for a molded LED using an integral sphere detector) for the LED after substrate removal, corresponding to a 49 % increase in optical power compared to the LED before substrate removal. The increase in the output power is attributed to the removal of the absorptive substrate and the Al/Au metal reflector. The light emitting to copper carrier side was reflected to the top side by the metal reflector. The improved $I-V$ characteristics as shown in Fig. 3.17, and the better thermal conductivity of copper carrier were another reason for the output power enhancement.

3.3.4 Effect of the Insertion of SLS Layers [97–99]

The optimized LED structure with the insertion of the $Al_{0.06}Ga_{0.94}N$/GaN SLS interlayer was grown using the MOCVD method as mentioned in Sect. 3.2.6. LED fabrication was performed on the epitaxially grown structure with and without SLS, without substrate removal process. Ni/Au thin transparent metals and Ni/Au (12/100 nm) p-type Ohmic metals deposited on the top p-GaN contact layer. The n-type Ohmic contact was fabricated both from top on mesa-etched n-GaN contact layer using Ti/Al/Ni/Au (15/60/12/60 nm) and from the backside on the Si substrate using AuSb/Au (18/100 nm).

For these samples, current-voltage ($I-V$), light output power and electroluminescence (EL) spectra were measured. Optical characteristics were measured at room temperature inside an integrated sphere. MQW light emission characteristics were evaluated by temperature-dependent PL.

Figure 3.19(a) shows the light output power versus the injected current density for the optimized LED structures, with and without the $Al_{0.06}Ga_{0.94}N$/GaN SLS interlayer. TT indicates the current injection from the p-GaN top layer to the n-GaN contact layer, while TB indicates the current injection from the p-GaN top layer to the backside of the n-type Si substrate. The results clearly shows that the LED structure with the $Al_{0.06}Ga_{0.94}N$/GaN SLS interlayer yields a higher intensity and a higher saturation current in comparison to the conventional LED structure without the SLS interlayer. At 20 A/cm^2 current injection, the light intensity from the LED structure with the $Al_{0.06}Ga_{0.94}N$/GaN SLS interlayer was 34 % higher than that of the conventional structure. It was shown that a better current spreading yielded an improved heat distribution in the LED chip at a higher saturation current when the current is injected from the backside of the n-type Si substrate (TB). Figure 3.19(b) shows a great improvement of the current-voltage characteristics in LED with the $Al_{0.06}Ga_{0.94}N$/GaN SLS interlayer structure than that of conventional structure. The operating voltage measured from the p-GaN top layer to the n-GaN contact layer at 20 mA for LED with the $Al_{0.06}Ga_{0.94}N$/GaN SLS interlayer and the conventional structure was 3.2 and 3.6 V, respectively, with the series resistance of 16 and 29 Ω, in each respective structure. The reduction of threading-dislocations by
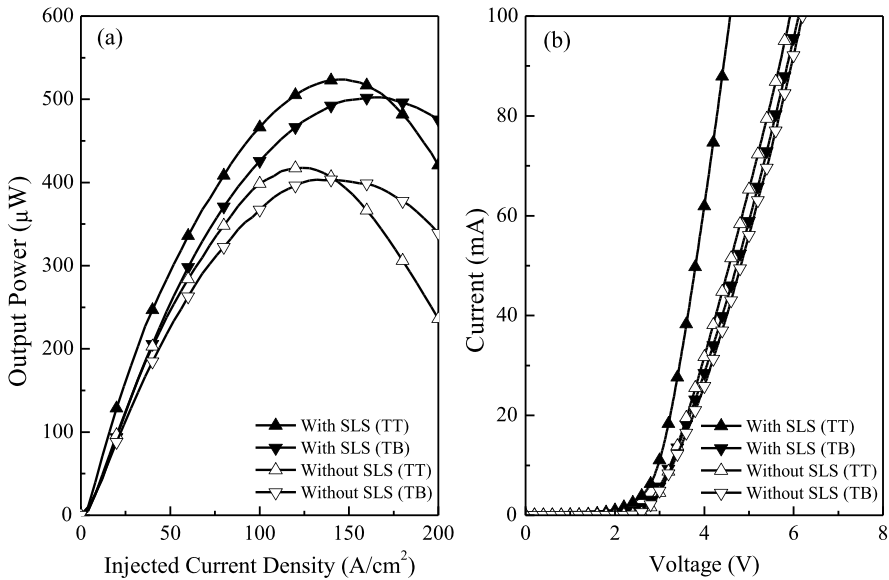
Fig. 3.19 Comparison of (**a**) light output power vs. injected current density and (**b**) injected current vs. voltage, each for optimized LED structure with/without $Al_{0.06}Ga_{0.94}N$/GaN SLS interlayer. The measurement was performed in an integrated sphere for 10×10 mm^2 size area (reprinted from Ref. [99] with permission, copyright 2010 IOP Publ.)

the $Al_{0.06}Ga_{0.94}N$/GaN SLS interlayer in the LED structure contributes to significant improvement in lateral carrier spreading, thus greatly improved the electrical characteristics in this LED structure. Typical EL characteristics for respective samples are shown in Fig. 3.20, for each 20 A/cm^2 increment of the injected current. The LED structure with the $Al_{0.06}Ga_{0.94}N$/GaN SLS interlayer shows a narrower FWHM spectrum peak of 27 nm, compared to that of 35 nm in the conventional structure. The result agrees well with the improved InGaN emission uniformity that can be achieved by the insertion of the $Al_{0.06}Ga_{0.94}N$/GaN SLS interlayer [98].

3.3.5 Other Structures

In order to solve the problem of the light absorption by Si substrates, we have developed other technologies such as DBR reflectors, through-holes and photonic crystals (PhCs).

We have first developed a DBR structure [158, 159] in order to increase the reflectivity. The various stack DBR layers were grown on AlN/AlGaN intermediate layer and AlN/GaN multilayers (Sect. 3.2.5) and found that the output power is increased with the number of DBR stacks. It was found that at three stacks, the output power was doubled because of the reflection of the emission light by the DBR layer.
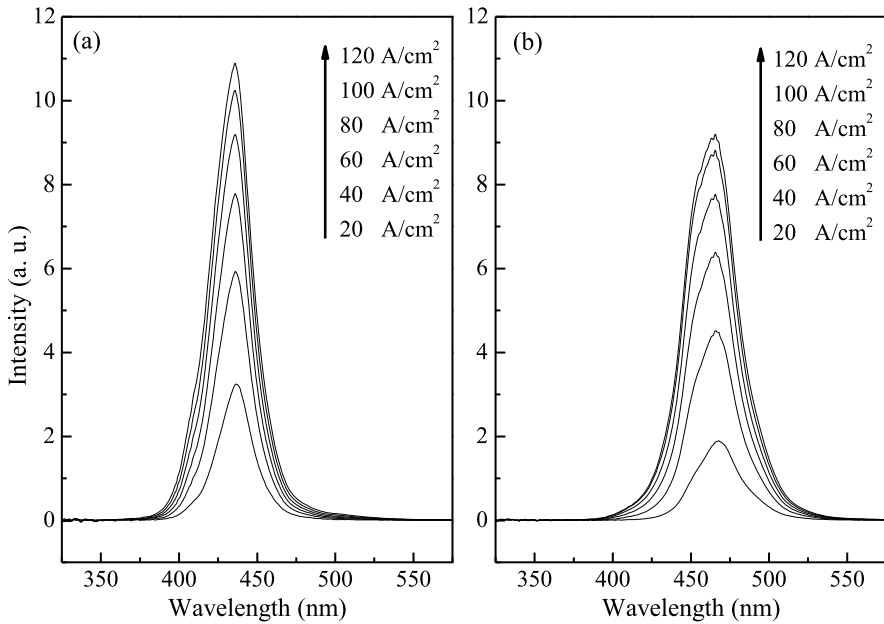
Fig. 3.20 EL characteristics for optimized LED structure (**a**) with an $Al_{0.06}Ga_{0.94}N/GaN$ SLS interlayer and (**b**) without an $Al_{0.06}Ga_{0.94}N/GaN$ SLS interlayer (reprinted from Ref. [99] with permission, copyright 2010 IOP Publ.)

The disadvantage is in that the cracking occurs when the DBR number is increased. The other disadvantage is that the series resistance is increased as a function of the stack number, which is disadvantageous for vertical structure LEDs.

We also have developed vertical LEDs using the LED structure of AlN/AlGaN intermediate layer and AlN/GaN multilayers (Sect. 3.2.5), via a through-hole device process as shown in Fig. 3.21. The through-hole has been fabricated via selective Inductively Coupled Plasma-RIE (ICP-RIE) up to the Si substrate. Cr (20 nm)/Pd (20 nm)/Au (200 nm) metals were used to connect the *n*-GaN and Si substrate. The conventional structure without thorough holes was also fabricated for comparison. AuSb (20 nm)/Au (200 nm) metals were deposited on the back side of the Si substrate after lapping and polishing the substrate down to 150 μm to compare the horizontal and vertical structure. It was found that the series resistance and the operating voltage at 20 mA were reduced from 26 to 22.5 Ω and from 4.4 to 4.0 V respectively for the vertical configuration by using the through-holes. This is due to the current pass via the through-hole electrode biasing the high resistive AlN layers.

We also have developed LEDs with photonic crystals (PhCs) [201]. Using a Si substrate as a mold for forming the PhCs. GaN-based epitaxial layers were grown on 2-D grooved Si substrates and they were bonded onto a highly reflective substrate and the Si substrate was removed. The resultant LEDs with PhC showed the 80 % higher optical output power due to the light extraction effect by PhCs. It is also noteworthy that the lateral epitaxial growth using grooved Si substrates decreased
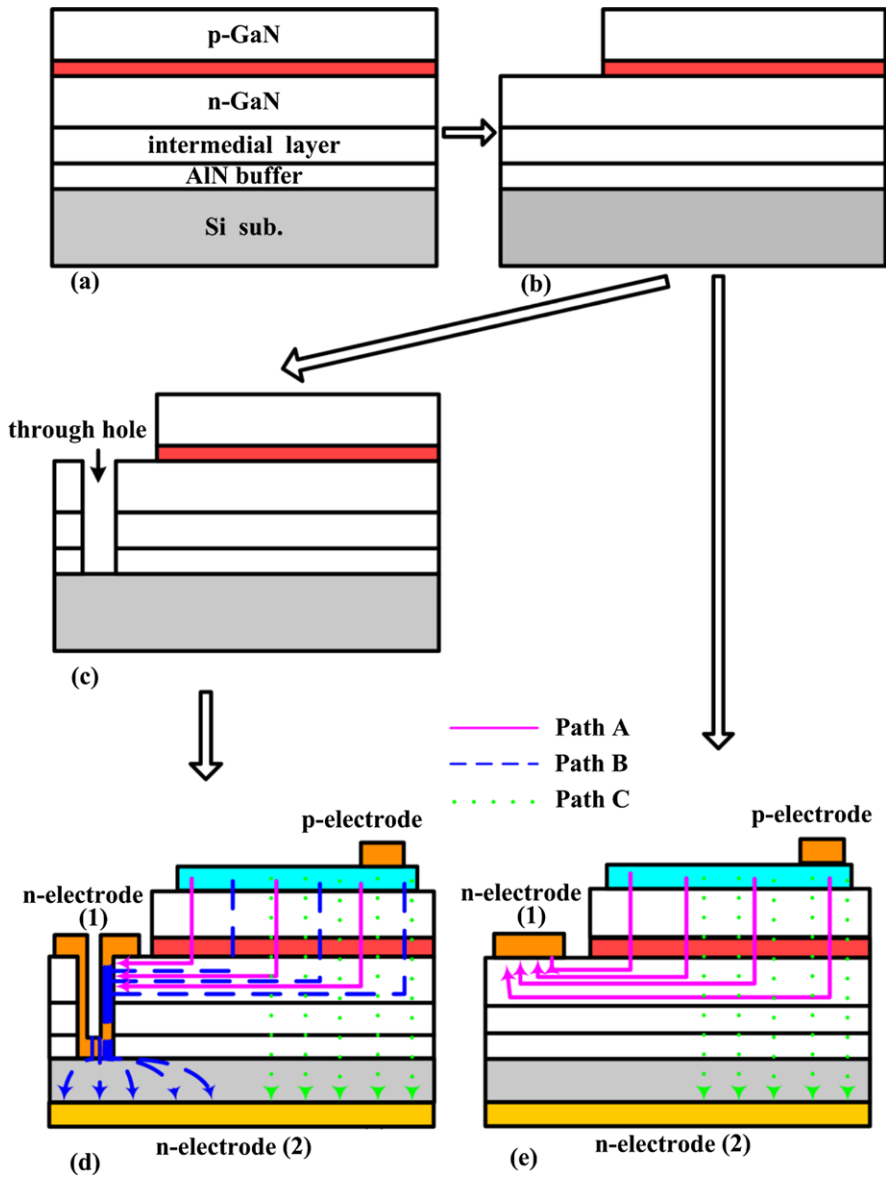
Fig. 3.21 Fabrication process of a novel LED structure with through holes and a conventional LED structure: (**a**) epitaxial layer grown on silicon substrate, (**b**) dry etching to expose the *n*-GaN layer, (**c**) formation of through-holes, (**d**) and (**e**) deposition of metals as electrodes for the LED structure with through-holes and for the conventional LED structure (reprinted from Ref. [165] with permission, copyright 2010 Japan Society of Applied Physics)

the dislocation densities and also affected the increase of the internal quantum efficiency.

3.4 Conclusion

We have developed high temperature AlN/AlGaN intermediate layers, AlN/GaN multilayers and AlGaN/AlN SLSs in order to prevent the cracking and to improve the surface morphology of GaN materials on Si substrates and could succeed in preparing high quality nitride layers on large diameter Si substrates. We could grow mirrored surface epitaxial layers to the diameter up to 100 mm. We also found that LEDs fabricated on these high quality epitaxial wafers showed the LED performances similar to LEDs on sapphire substrates. This is mainly due to the reduction of the series resistance by the formation of tunnel junction between the n-AlGaN layer and n-Si substrate with decreasing the high-temperature AlN layer thickness to 3 nm, and also due to the dislocation density reduction by the pyramid-shaped (quantum-dot-like) structure of MQW layers on AlN/GaN multilayers. We also showed the advantage of GaN/Si LEDs compared with LEDs on sapphire is that the optical output power can be increased much higher compared with LEDs on sapphire because of a better thermal conductivity of Si than sapphire. The LEDs also exhibited the good reliability and a uniform emission for a large size wafer. GaN/Si LEDs are thus very promising for future general lighting because it is clear that the fabrication cost will be highly reduced, up to 75–80 % of conventional LEDs based on sapphire and SiC substrates. There also have many kinds of structure modifications to improve the disadvantage of GaN/Si so that GaN/Si LEDs have a promising perspective to become the main LEDs in the near future.

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Chapter 4
Epitaxy Part B. Epitaxial Growth of GaN on Patterned Sapphire Substrates

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Abstract This chapter describes the Patterned Sapphire Substrate (PSS). Firstly, the properties and the fabrication process of the PSS are described. Then, the mechanism to decrease the dislocation density in the crystal growth of GaN layer on the PSS is elaborated. Thirdly, the principle of the improvement in the LEE of the GaN-LEDs fabricated on the PSS is explained. Finally, the novel application of the PSS to grow nonpolar and semipolar GaN layers is introduced.

4.1 Introduction

Low-carbon society and sustainable development of society are becoming the most important issue for survival of human beings. In this meaning, high-efficiency solid state lighting technology is one of the significant energy saving technologies. Recently, white light-emitting diodes (LEDs) with high luminous efficacy (η_L) of 249 lm/W and high luminous flux (Φ_V) of 14.4 lm at a forward-bias current of 20 mA were developed by using yellow phosphors $\{(Y_{1-a}Gd_a)_3(Al_{1-b}Ga_b)_5O_{12}:Ce^{3+}$ (YAG)$\}$ and blue GaN-based LEDs (GaN-LEDs) dies fabricated on the break-though technique of "patterned sapphire substrate (PSS)" [1]. The highest luminous efficacy of commercialized white LEDs is 170 lm/W, which is the highest luminous efficacy among all kinds of white light sources.

Briefly, the PSS is the sapphire substrate with micrometer-sized or sub-micrometer-sized artificial structures on its surface, which contributes to improve the internal quantum efficiency (IQE) and the light extraction efficiency (LEE) of the GaN-based LEDs (GaN-LEDs) [2]. On the one hand, the IQE is improved by the reduction of threading dislocation density by selective-area growth (SAG) and by epitaxial lateral overgrowth (ELO) of GaN layers on the PSS. On the other hand, the LEE is improved by the photon scattering at the uneven GaN/sapphire interface.

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Fig. 4.1 History of the improvement of EQE of GaN-LEDs
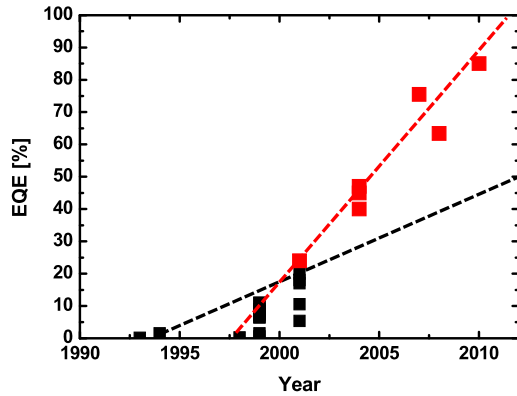
Figure 4.1 shows the history of the external-quantum efficiency (EQE) of the GaN-LEDs [1–12]. The improvement rate of the EQE after 2001 is larger than that of the years before, because almost all GaN-LEDs are fabricated on the PSS after the year when the PSS was published [2].

In this chapter, the properties and the fabrication process of the PSS are described firstly in Sect. 4.2. Then, the mechanism to decrease the dislocation density in the crystal growth of GaN layer on the PSS is elaborated in Sect. 4.3. Thirdly, the principle of the improvement in the LEE of the GaN-LEDs fabricated on the PSS is explained in Sect. 4.4. Finally, the novel application of the PSS to grow nonpolar and semipolar GaN layers is introduced in Sect. 4.5.

4.2 Properties and Fabrication of PSSs

The PSS technique was derived from the ELO technique, which leads to grow the GaN with lower dislocation density on the PSS. Furthermore, the PSS improves the LEE of GaN-LEDs. The advantages of the GaN layer and the GaN-LEDs on PSS are summarized in Table 4.1 in comparison with the GaN layer (GaN-LEDs) by ELO techniques. The substantial advantage of the PSS technique is that the device structure is grown with a single metal-organic vapor phase epitaxy (MOVPE) growth process without any mask, which cannot be avoided in the conventional ELO process. Therefore, the PSS technique is contamination-free from the mask material such as SiO_2. Additionally, a single growth process gives economical advantage because of shorter growth time. The ELO technique itself has no positive effect to improve the LEE of the GaN-LEDs.

The PSS is usually fabricated through standard photolithography, inductively-coupled plasma reactive ion etching system (ICP-RIE), and water cleaning process. Figure 4.2 shows the sequence of the process thereof. Majority of lithography exposure equipment in manufacturing process is projection stepper or high resolution mask aligner. The photoresist pattern is transferred to the surface structure of the sapphire substrate by ICP-RIE as shown in Fig. 4.2(d). The etching gases, such
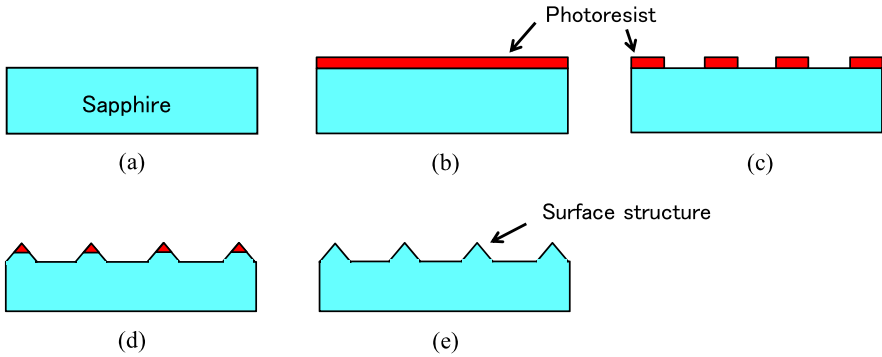
Fig. 4.2 PSS manufacturing process. (**a**) Sapphire substrate preparation, (**b**) Photoresist coating by spinner, (**c**) Photolithography by stepper and developer, (**d**) Dry etching by ICP-RIE, (**e**) Rinse and drying

Table 4.1 Advantages of GaN layer and GaN-LEDs on PSS in comparison with those by ELO technique

Item	PSS technique	ELO technique
Growth	• A single growth process of MOVPE • Without any mask	• More than two times growth process of MOVPE • Mask is vital
Production Cost	• Shorter growth time	• Longer occupation time of MOVPE • Masking process and machines
Impurity	• No mask material contamination	• Possible contamination from mask material
Dislocation Density	• Dislocation density: On average 1×10^8 cm^{-2}	• Dislocation density: Partially 10^4–10^5 cm^{-2}
Availability	• Applied to almost all GaN-LEDs	• Applied to GaN substrate for Laser Diodes

as BCl$_3$ and/or Cl$_2$, and inert gas, such as Ar, are used. The typical etching rate is about 100 nm/min, but the rate depends on the etching conditions and type of etchers. The shape of the structure on the PSS is controlled by the sapphire-etching rate and the selective ratio (etching rate of photoresist/etching rate of sapphire substrate) depending on the kind of photoresist, baking conditions, and etching conditions.

The residual photoresist and foreign materials deposited during etching process are removed using water cleaner. Water cleaning process is; first step: to immerse

(a)   (b)   (c)
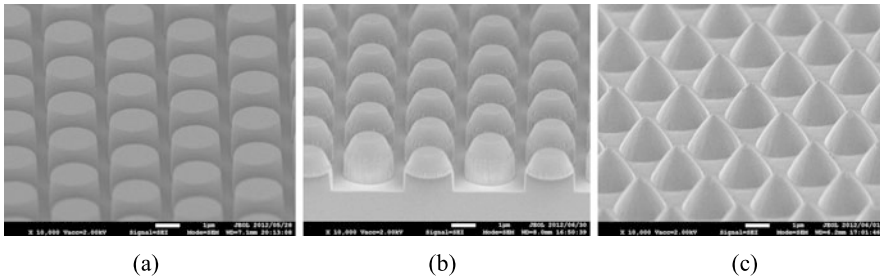
Fig. 4.3 Bird's-eye view SEM images of every manufacturing stage of the PSS. (**a**) The sapphire substrate with columnar-shaped photoresist arranged hexagonal lattice pattern, (**b**) that of the cone-shaped PSS during ICP-RIE, and (**c**) that of the PSS after water cleaning

the PSS in Piranha solution ($H_2SO_4 + H_2O_2$) at 120 °C for 10 min, second step: to immerse the same in SC-2 (Standard Clean 2) solution ($HCl + H_2O_2 + H_2O$) at 40 °C for 10 min, final step: to rinse with deionized water and drying with 2-propanol steam.

Figure 4.3 shows the bird's-eye view scanning electron microscope (SEM) images of every manufacturing stage of the cone-shaped PSS. Typical height and diameter of the cones, and distance between the cones are 1.5 µm, 2.0 µm, and 3.0 µm respectively. The typical PSS surface coverage by the cones is approximately 40 %. The LEE of the GaN-LEDs increases in accordance with the increase in the surface coverage up to 70 %. However, it is difficult to grow the high quality GaN layer when the surface coverage exceeds 70 %, because etched-flat c-plane surface is very important to supply the nucleation sites for the initial SAG and to allow subsequent overgrowth of the GaN layer on the PSS. Therefore, the surface coverage of around 60 % is estimated to give the highest LEE.

4.3 Growth of GaN on PSS, and Properties of GaN-LEDs on PSS

4.3.1 SAG and ELO

The PSS technique has been derived from the SAG technique and the ELO technique as mentioned before. In this section, SAG and ELO techniques, which are the origins of the PSS, are reviewed.

The SAG is the technique which fabricates the controlled-structures of GaN such as lines and hexagonal pyramids [13, 14]. Kitamura et al. fabricated the hexagonal pyramids with six {10–11} facets grown on the c-plane GaN/sapphire template covered with the dot-patterned SiO_2 mask.

The original idea of the ELO was proposed by Nishinaga et al. to reduce the threading dislocation density of the GaAs layer grown on GaAs substrate and Si substrate using liquid phase epitaxy (LPE) [15, 16]. The ELO based on the
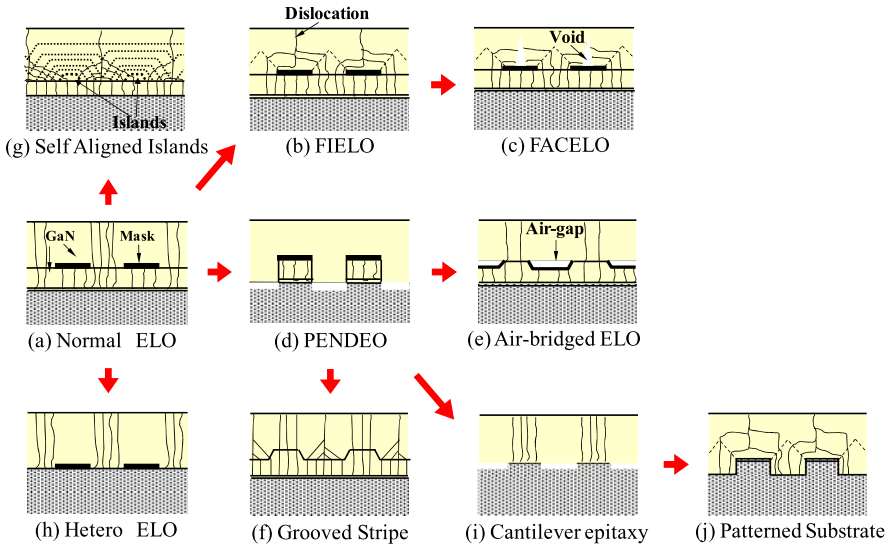
Fig. 4.4 Schematic diagram of various ELO techniques (Reference [17])

SAG has attracted considerable attentions, because the ELO is a useful technique to reduce the threading dislocation density of the GaN layer grown on foreign substrate such as sapphire substrate. The various ELO techniques of the GaN layer developed by various researchers are summarized in Fig. 4.4 [17] and Table 4.2.

Here, two representative examples of the ELO techniques are introduced. Firstly, Usui et al. applied the ELO technique to the GaN epitaxial growth on sapphire substrate (Fig. 4.4(b)) [18, 19]. They developed the break-through technique named facet initiated ELO (FIELO) by hydride vapor phase epitaxy (HVPE). It was revealed that the inclined facet structure formed during FIELO bent the propagation direction of dislocations, resulting in reducing the threading dislocation density. They obtained the GaN layer on SiO_2 mask with a low dislocation density of the order of 10^7 cm^{-2}. Secondly, the ELO using grooved substrate was developed by Detchprohm et al. [20–22] (Fig. 4.4(i)). In the ELO techniques using grooved substrate, GaN layer only grew on the terrace of the grooved substrate and extended the wing on the trench. The dislocation density of the wing area of the GaN were less than 4×10^6 cm^{-2} [21].

Tadatomo et al. developed basic technique of the standard PSS as shown in Fig. 4.4(j). They made the stripe-trench-patterned PSS with the depth of more than 1 μm by ICP-RIE using the photoresist mask. Furthermore, they grew the GaN layer not only on the terrace but also on the bottom of the trench. Then, they have succeeded in fabricating high-efficiency GaN-LEDs on the PSS due to the photon scattering at the uneven GaN/sapphire interface.

Table 4.2 Representative ELO techniques

Figure No.	Name/Affiliation/National	Characteristics	References
(a)	Nishinaga et al. Nagoya University, Japan	• Normal ELO, GaAs on GaAs (Si) substrate, LPE	[15, 16]
(a)	Nam et al. North Carolina State University, The United State of America	• Normal ELO, SiC substrate, MOVPE	[23]
(b)	Usui et al. NEC, Japan	• FIELO (Facet Initiated ELO) • HVPE (Hydride Vapor Phase Epitaxy) • Inclined facet structure during FIELO bent the propagation direction of dislocations • Reduced threading dislocation density	[18, 19]
(c)	Hiramatsu et al. Mie University, Japan	• FACELO (Facet Controlled ELO) technique • FACELO is based on the control of GaN facet by growth conditions. • Void decreases dislocation density because void terminates the laterally bent dislocations	[24–26]
(d)	Linthicum et al. North Carolina State University, The United State of America	• PENDEO (from the Latin: pendeo—to hang, or to be suspended)	[27, 28]
(e)	Kidoguchi et al. Panasonic Corporation, Japan	• Air-bridged ELO (ABLEG) • Laser Diodes were achieved	[29, 30]
(f)	Ishida et al. Panasonic Corporation, Japan	• GaN on sapphire substrate template with grooved stripe • Ultraviolet laser diodes were achieved	[31, 32]
(g)	Lahreche et al. (representative) CNRS, France	• Self-organized ELO • Dislocation Density was reduced • GaN buffer layer grown at low temperature technique is included in this group [33]	[33–35]
(h)	Sakai et al. (representative) Tokushima University, Japan	• Direct ELO using the sapphire substrate with SiO₂ mask	[36, 37]
(i)	Detchprohm et al. Nagoya University, Japan	• Cantilever Epitaxy using Grooved substrate • GaN growth only on the terrace of the grooved substrate • Dislocation Density of the wing area $<4 \times 10^6$ cm^{-2}	[20–22]
(j)	Tadatomo et al. Mitsubishi Cable Industries, Japan	• Basic Patterned Sapphire Substrate (PSS)	[2]

Fig. 4.5 Schematic structure of the GaN-LED on the PSS

ITO p-electrode
p-GaN
p-Al$_{0.2}$Ga$_{0.8}$N
InGaN/GaN MQWs
n-GaN
u-GaN
LT-GaN buffer layer
PSS
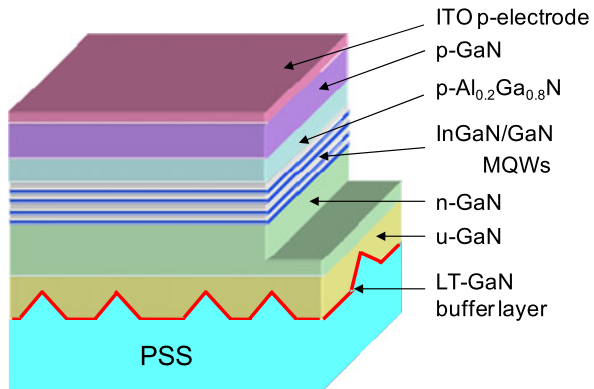
Table 4.3 Detailed LED structure

Layer	Growth temperature	Thickness	Doping
p-type contact layer	1050 °C	80 nm	Mg: 2×10^{19} cm^{-3}
p-type Al$_{0.2}$Ga$_{0.8}$N electron blocking layer	1050 °C	20 nm	Mg: 5×10^{19} cm^{-3}
InGaN/GaN multiple quantum well (MQW)	780–820 °C	InGaN/GaN: 2-nm-thick/12-nm-thick	Si in GaN barrier: 1×10^{18} cm^{-3}
n-type GaN layer	1150 °C	4 µm	Si: 2×10^{18} cm^{-3}
Unintentionally doped GaN layer	1150 °C	3 µm	–
GaN buffer layer grown at low-temperature	480 °C	20 nm	–
Sapphire substrate	–	430 µm	–

4.3.2 GaN Growth on PSS and the Mechanism of Decreasing Dislocation Density by ELO

The commercially available high-efficiency GaN-LEDs is fabricated on the PSS by MOVPE. The growth conditions of the GaN layer on the PSS are almost the same as those of growth on the flat sapphire substrate (FSS). Figure 4.5 depicts a schematic structure of the standard GaN-LEDs on the PSS fabricated. The detailed GaN-LED structure is shown in Table 4.3.

The p-type GaN layer is partially etched until the n-type GaN layer is exposed. A 100-nm-thick transparent indium–tin oxide (ITO) is used as electrode and current-spreading layer to p-type GaN layer. The transmittance of the ITO is more than 95 % at the wavelength range from 400 nm to 800 nm. A Cr/Au (20/200 nm) layer is deposited on exposed n-type GaN layer and on the partial ITO contact as a pad electrodes. Finally, SiO$_2$ passivation layer is deposited by plasma-enhanced chemical vapor deposition (PE-CVD) in order to reduce the surface leakage current.

Fig. 4.6 The bird's-eye-view or the plan-view SEM images of GaN layer on the RC-PSS at various growth times: (**a**) before growth, (**b**) 12 min growth, (**c**) 24 min growth, and (**d**) 36 min growth (Reference [38])

(a)

(b)

(c)

(d)
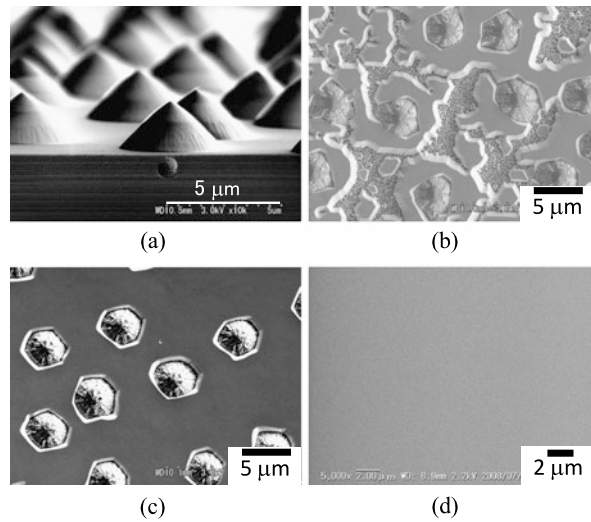
Fig. 4.7 The magnified plan-view SEM image of Fig. 4.6(b) (Reference [38])
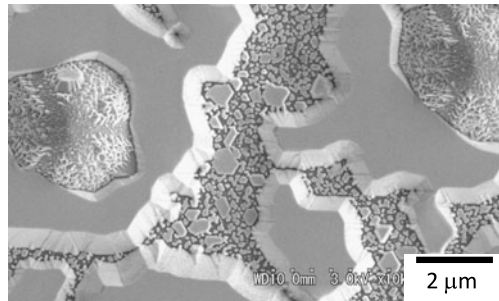
Figure 4.6 shows the plane-view SEM images of the GaN layer grown on the PSS with randomly-arranged-cone (RC-PSS) at the various growth times [38]. Figure 4.6(a) shows the plane-view SEM image of the RC-PSS before the MOVPE growth of GaN layer. Figures 4.6(b) and (c) show the SEM images of the GaN layer grown on the RC-PSS at the growth time of 12 min and 24 min, respectively. Figure 4.6(b) shows that the GaN grows from the etched-flat c-plane surface. Although the GaN deposited at low temperature (LT-GaN) buffer layer is deposited not only on the etched-flat c-plane surface but also on the sidewall of the cones, the LT-GaN buffer on the sidewall evaporated or transported to the etched-flat surface during ramping to the high temperature for the growth of GaN, because c-plane GaN on the etched c-plane sapphire is more stable than that on the sidewall. Therefore, there is no nuclear GaN layer from LT-GaN buffer layer on the sidewall and the growth of GaN on the sidewall at high temperature does not occur. Figure 4.6(c) shows that the cones are embedded by the overgrowth of the GaN layer. Finally, the GaN layer with flat surface is obtained by embedding as shown in Fig. 4.6(d). Figure 4.7 shows the magnified plane-view SEM image of the GaN layer shown in Fig. 4.6(b).
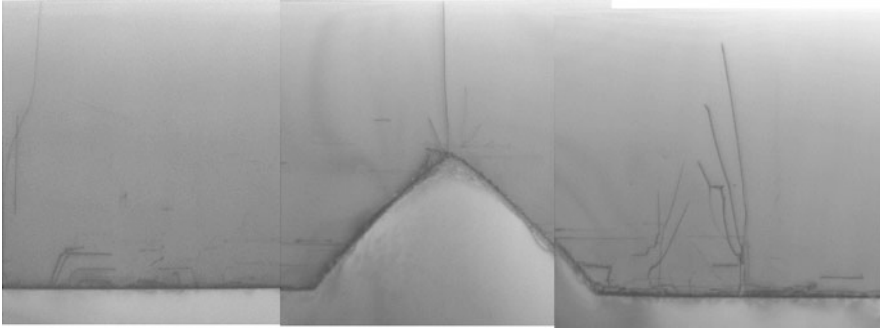
Fig. 4.8 Cross-sectional TEM image of GaN layer grown on the RC-PSS (Reference [38])

Slight growth of the GaN with six-fold symmetry on the sidewall is observed. Although the sidewall has no specific plane to grow the GaN, the GaN may grow on partially exposed planes such as n-plane. The growth rate of the GaN around the cone is higher than that away from the cone. It is estimated that the migration of the ad-atoms such as Ga atoms from the cones contributes to the growth of GaN layer.

Figure 4.8 shows a cross-sectional transmission electron microscope (TEM) image of the GaN layer grown on the RC-PSS. This TEM image reveals the dislocation behavior during the growth of the GaN layer on the PSS. Some dislocations generated at the GaN/sapphire interface are bent laterally and terminated at the inclined sidewall of the cone. No dislocation is generated from the inclined sidewalls because GaN does not grow on the sidewall. Dislocations generated at the top of the cone are observed because the laterally overgrown GaN layer coalesced on the top of the cone.

The growth of GaN layer initiates from the etched-flat and narrow area, which is a kind of the SAG. In conclusion, when considering the dislocation behaviors and the growth modes, the decrease of the dislocation density of the GaN layer on the PSS is due to the combination of SAG and ELO. On the other hand, in the etched-flat region away from the cones, the dislocation density of the GaN layer is same as that on the FSS. Therefore, the average dislocation density of the GaN layer on the PSS is lower than that on the FSS.

4.3.3 Characteristics of LEDs Grown on PSS

The output power of the GaN-LEDs fabricated on the PSS is usually 30–50 % larger than that of GaN-LEDs on the FSS. For example, the output power of the packaged GaN-LEDs on the RC-PSS with the wavelength of the approximate 420 nm is 1.34 times larger than that on the FSS with the same LED structure at a forward-bias current of 20 mA. The output power and the EQE of the packaged GaN-LEDs measured by an integrating sphere systems are 29.6 mW and 49.5 % at forward-bias

current of 20 mA, respectively. The EQE decreases markedly with increasing injection current, which is well known as "efficiency droop phenomena". The maximum EQE of 75 % is observed at a forward-bias current of 0.5 mA.

The EQE is a product of the IQE and the LEE as described in the following formula (4.1).

$$EQE = LEE \times IQE \tag{4.1}$$

Although, it is not easy to measure the LEE and the IQE directly, the IQE can be estimated by the relationship between excitation power density and temperature-dependent photoluminescence (PL) [39]. The GaN-LEDs on the RC-PSS is characterized under selective excitation conditions using a dye laser with a peak wavelength of 385 nm [38]. As a result, the IQE is estimated to be approximately 57 %, which is the maximum PL quantum efficiency at room temperature, assuming that the PL quantum efficiency at a cryogenic temperature of 7 K is 100 % [39]. At this excitation condition, the photo-excited carrier density is estimated to be nearly equal to the injected-carrier density at 20 mA [39]. The LEE is estimated to be 86.8 %, which is EQE (49.5 %) divided by IQE (57 %), due to the formula (4.1).

4.4 The Principle of Light Extraction Efficiency Improvement of GaN-Based LEDs by Patterned Sapphire Substrate

4.4.1 Impact of Surface Structure of LEDs on Light Extraction Efficiency Improvement

It has been well known that there is a significant gap between the IQE and the EQE of the LEDs since the research and development of GaAs-based LEDs (GaAs-LEDs) have started [40]. For example, the IQE of the GaAs-LEDs with double hetero-structures exceeds 99 %, but the EQE is only a few percent. The reason for this gap is the narrow escape cone for photons emitted in the LEDs due to high refractive index of the semiconductor. The escape cone is defined by the critical angle (ϕ_c) of the total internal reflection at the semiconductor/surrounding material interface. The critical angle of the total internal reflection is described in formula (4.2) by Snell's Law, where n_1: the refractive index of the semiconductor, and n_2: that of surrounding material.

$$\phi_c = \sin^{-1}\left(\frac{n_2}{n_1}\right) \tag{4.2}$$

Because the area of escape cone, where photons can escape from a semiconductor, is $2\pi r^2(1 - \cos\phi c)$ as shown in Fig. 4.9 [41], the ratio of the escaped photons to all the generated photons at the surface, i.e. the escape ratio (η_c), is described as following formula (4.3).

$$\eta_c = \frac{(1 - \cos\phi_c)}{2} \tag{4.3}$$
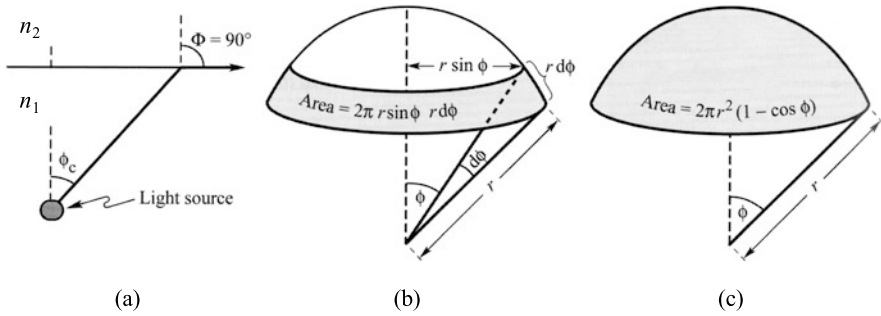
Fig. 4.9 Schematic drawings of escape cone [41]: (**a**) Definition of the critical angle ϕ_c of the total internal reflection phenomena. (**b**) Area element dA. (**c**) Area of dome-shaped section of the sphere defined by radius r and angle ϕ
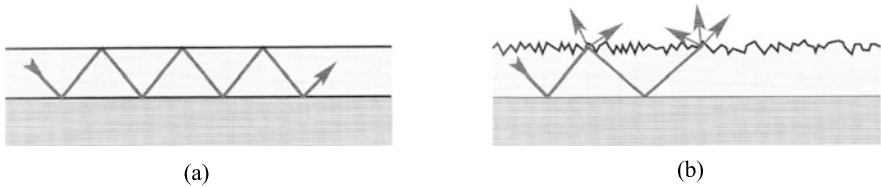
Fig. 4.10 Illustrated photon trajectories: (**a**) In the planar layer and (**b**) in the layer with surface structure [40, 41]

For example, in the case of the GaAs ($n_1 = 3.5$) in the air ($n_2 = 1.0$), the critical angle (ϕ_c) of the total internal reflection and the escape ratio (η_c) at the surface are 16° and 0.02, respectively.

Due to the small escape ratio, many photons are trapped in the planar semiconductor layer as shown in Fig. 4.10(a). When the semiconductor layer has textured surface, the photons emitted out of the escape cone originally scatter at the textured surface to get the opportunities to enter into the escape cone as shown in Fig. 4.10(b). Thus, the surface structure enhances the LEE of the semiconductor layer. It is essential to destruct the cavity structure trapping photons and to give photons multiple chances to escape from the layer. Therefore, a lot of studies such as the surface-textured LEDs [40, 42], the hemispherical-domed LEDs [43], truncated-inverted-pyramid LEDs [44], etc. have been carried out.

4.4.2 The Principle of Light Extraction Efficiency Improvement of GaN-Based LEDs by Patterned Sapphire Substrate

In the case of GaN-LEDs fabricated on a FSS, the critical angle (ϕ_c) and the escape ratio (η_c) at air ($n_2 = 1.0$)/GaN layer ($n_1 = 2.5$) interface by formulae (4.2) and (4.3) are 24° and 0.042, respectively. On the other hand, the critical angle
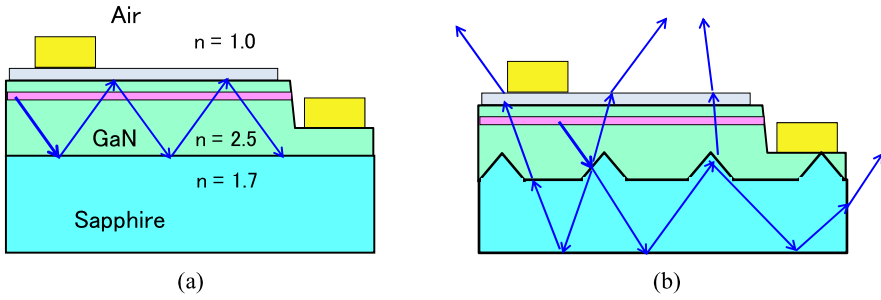
Fig. 4.11 The schematic cross-sectional LED structure and photon trajectories: (**a**) LED fabricated on a FSS. (**b**) LED fabricated on a PSS

and the escape ratio at the GaN layer ($n_1 = 2.5$)/sapphire ($n_2 = 1.7$) are 43° and 0.13, respectively. Because the GaN-LEDs have five air/GaN layer interfaces with escape ratio of 0.042 and one GaN layer/sapphire interface with escape ratio of 0.13, the LEE is estimated to be 0.34 by simple calculation. Although, the critical angles and the escape ratios at air/GaN layer interface and GaN layer/sapphire interface are larger than those at GaAs/air interface, more than 60 % of the emitted photons in the InGaN wells are still trapped in the GaN layer as shown in Fig. 4.11(a)) [1].

The surface structure of the PSS, such as cone-shaped, plays the same role of the surface structure described in the preceding section. To put it more concrete, the structure on the PSS effectively scatters photons emitted out of escape cone and gives them multiple chances to enter the escape cone, and the surface structure with inclined sidewall facilitates the penetration of photon at the interface as shown in Fig. 4.11(b). Due to the low refractive index of a sapphire substrate, photons escape from the sapphire easily. Thus, the PSS improves the LEE of the GaN-LEDs.

Figure 4.12 shows the schematic sample structure for ray-tracing simulation and the results of simulation. The layer structure and the size of the samples are same as GaN-LED chip. One sample is designed on a FSS and the other on the stripe-trench-patterned PSS. The light source is set in the center of MQW only and the sample is set in the air in order to simplify as shown in Fig. 4.12(a). Many of photons emitted from MQW are trapped in the GaN layer grown on the FSS and only the photons emitted within the escape cone escape from the sample as shown in Fig. 4.12(b). Figure 4.12(c) shows many of photons escape through the PSS into the air, resulting in the improvement of the LEE.

4.4.3 Development of PSS with Micrometer-Sized Structures

The PSS was firstly reported by Tadatomo et al. in 2001 as mentioned before [2]. The reported GaN-LEDs were fabricated on the stripe-trench-patterned PSS, which
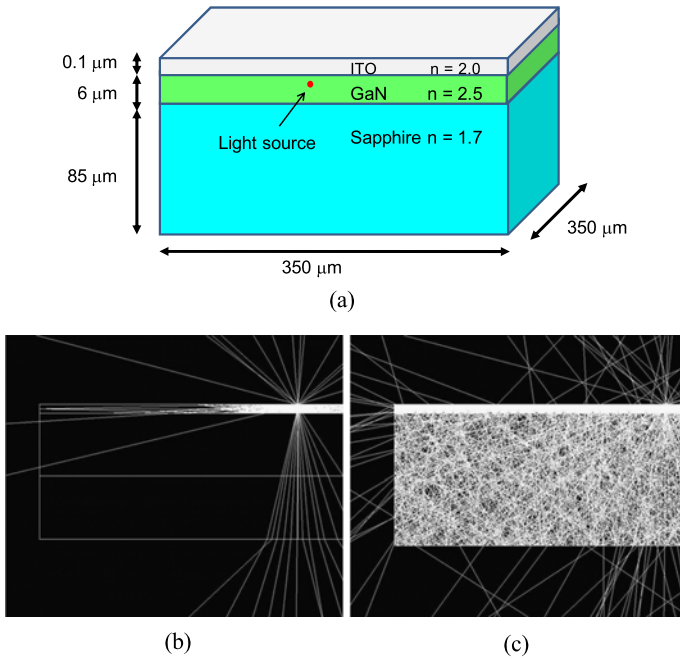
Fig. 4.12 The ray-tracing simulation: (**a**) Schematic sample structure for simulation, (**b**) Result of GaN layer on a FSS, (**c**) Results of GaN layer on the stripe-trench-patterned PSS

direction was along m-axis of sapphire. The EQE of the GaN-LEDs, which were mounted on flip chip package, was 24 % at the peak wavelength of 382 nm [2, 45] and 43 % at 403 nm [46, 47] at forward-bias current of 20 mA. The dislocation density of the GaN layer grown on the PSS estimated by counting dark spots in the cathodeluminescence (CL) is approximately 1.5×10^8 cm^{-2}, which is approximately one third of that on the FSS.

After 2001, there has been paid great attention in the PSS technique, and the EQE has been improved drastically. Yamada et al. developed hexagonal-patterned PSS [48]. The EQE of the blue GaN-LEDs with the wavelength of 460 nm fabricated on the hexagonal PSS was 35.5 % at forward-bias current of 20 mA. The investigation of surface structures with holes, semi-spheres, pyramidal structures, cone array, random-cone, etc followed by others [49–51]. New lithography technique using polystyrene microspheres instead of photolithography, so called "natural lithography", was also reported to fabricate random-cone-shaped PSS [38]. Wet etching process instead of ICP-RIE was reported as well to fabricate the PSSs [50–54]. Figure 4.13 shows the typical PSSs with micrometer-sized-structures investigated in this trend. The method and properties of the representative PSSs are also summarized in Table 4.4.
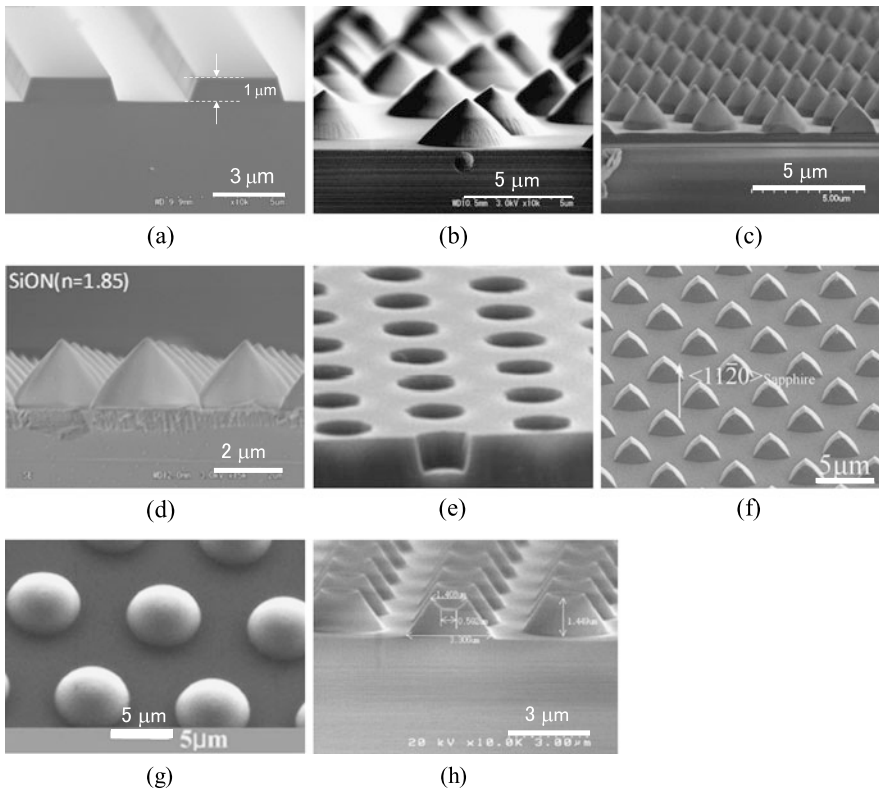
Fig. 4.13 Bird's-eye-view SEM images of representative PSSs: (**a**) stripe-trench-patterned PSS, (**b**) randomly-shaped-cone PSS, (**c**) standard cone PSS, (**d**) refractive-index-controlled PSS, (**e**) hole-arrayed PSS [56], (**f**) pyramidal PSS by wet etching [53], (**g**) hemispherical PSS [51], and volcano type PSS [58]. The SEM images from (**a**) to (**d**) are provided by courtiers of Yamaguchi University

4.4.4 Development of PSS with Sub-micrometer-Sized Structures

As mentioned in the proceedings, the PSS with micrometer-sized surface structure improves the LEE of the GaN-LEDs due to extensive scattering of the photons trapped in the GaN layer by the total internal reflection phenomena. When the size and period of the lattice array structure on the sapphire substrate decreases to smaller level of the sub-micrometer, so called "nano-structure", the LEE will be expected to increase more due to diffraction or photonic crystal (PhC) effect instead of photon scattering effect.

Generally, a PhC is fabricated by the artificial periodic structure with a different refractive index from that of the surrounding material. In the PhC, a band structure of photon energy, which is called photonic bandgap, is formed due to the artificial periodic structure. In two dimensional PhC (2D-PhC), the photons having the energy within the photonic bandgap does not propagate along the 2D-PhC. Therefore, if the

Table 4.4 Representative sub-micrometer-sized PSSs (Methods and properties of the PSS are summarized from papers published by those affiliations)

Year	Affiliation	Name	Methods and properties	References
2001	Mitsubishi Cable Industries, Japan	K. Tadatomo et al.	Photolithography & ICP-RIE Stripe-trench-patterned PSS Improvement of EQE = 24 % at 382 nm	[2, 45]
2002	Nichia Corporation, Japan	M. Yamada et al.	Photolithography & ICP-RIE Hexagonal-patterned PSS EQE = 35.5 % (at 20 mA, 460 nm)	[48]
2003	Mitsubishi Cable Industries, Japan	H. Kudo et al.	Photolithography & ICP-RIE Stripe-trench-patterned PSS EQE > 43 % (at 20 mA, 405 nm)	[46]
2003	Mitsubishi Cable Industries, Japan	K. Tadatomo et al.	Stripe-trench-patterned PSS Improved LEE	[47]
2003	National Cheng Kung Univ. Taiwan	S.J. Chang	Stripe-PSS Reduced dislocation density	[55]
2005	National Chung Hsing Univ. Taiwan	W.K. Wang et al.	Photolithography & ICP-RIE Stripe-trench-patterned PSS	[56]
2005	National Chiao Tung Univ. Taiwan	Y.J. Lee et al.	Photolithography & ICP-RIE Hole arrayed PSS	[49]
2006	Nichia Corporation, Japan	Y. Narukawa et al.	Photolithography & ICP-RIE PSS and ITO, EQE > 63 % (at 20 mA, blue)	[10]
2008	Yamaguchi Univ. Japan	K. Hoshino et al.	Photolithography and ICP-RIE Stripe-trench-patterned PSS, Thick SiO_2 mask Low dislocation density	[57]
2008	Chinese Academy of Science, China	H. Gao et al.	Photolithography (Self-assembled Ni mask) & Wet etching, Pyramidal PSS EQE of the LED on nano-PSS larger than that on micro-PSS	[53, 54]
2009	Yamaguchi Univ. Japan	N. Okada et al.	Natural lithography (Polystyrene sphere) Random-cone PSS EQE = 50 % (at 420 nm, 20 mA)	[38]
2010	National Chung Hsing Univ. Taiwan	J.H. Cheng et al.	Photolithography and Wet etching Triangle pyramidal array PSS	[50]
2010	National Central Univ. Taiwan	Y.J. Chen et al.	Mask free wet etching Pyramidal PSS	[52]
2010	Chonbuk National Univ. Korea	S.M. Jeong et al.	Photolithography and ICP-RIE Lens-PSS Improved LEE	[51]
2010	Korea Photonics Tech. Inst. Korea	J.B. Kim et al.	Photolithography and ICP-RIE Volcano-PSS	[58]
2010	Nichia Corporation, Japan	Y. Narukawa et al.	Photolithography & ICP-RIE PSS and ITO, EQE > 84 % (at 20 mA, blue)	[1]

Table 4.5 Representative sub-micrometer-sized structure PSSs (Fabrication methods and properties of PSS summarized from papers published by those affiliations)

Year	Affiliation	Name	Method and property	Reference
2004	Panasonic Corporation, Japan	K. Orita et al.	Electron Beam Lithography and RIE PhC patterned on *p*-GaN Surface for Blue LED	[59]
2006	Univ. California Santa Barbara, USA	A. David, et al.	Electron Beam Lithography and RIE Surface Patterning	[60, 61]
2008	National Cheng Kung Univ. Taiwan	Y.K. Su et al.	Polystyrene nano-sphere, followed by Ni evaporation, lift-off, and ICP-RIE	[65]
2009	National Central Univ. Taiwan	C.H. Chan et al.	SiO_2 Nano-sphere and Wet Etching	[66]
2009	Seoul National Univ., Korea	J. Lee et al.	Laser Holography 2D PhC-PSS Lattice constant/diameter/height = 500 nm/230 nm/125 nm	[62, 67]
2010	National Cheng Kung Univ. Taiwan	C.C. Kao et al.	Nano-imprint Lithography Nano-PSS Lattice constant/diameter/height (typical) = 500 nm/200 nm/200 nm	[63]
2011	Rensselaer Polytechnic Inst. USA	Y. Li et al.	Nano-imprint Lithography Nano-PSS Triple output power	[64]

2D-PhC is fabricated in the LEDs, the photons which propagate only perpendicular to the 2D-PhC are emitted by the electron-hole radiative recombination, because photons cannot propagate along the 2D-PhC. To put it simple, the photons originally emit in the escape cone, then the LEE of the GaN-LEDs with 2D-PhC is expected to be improved.

To realize the 2D-PhC and/or to improve the LEE, several researches of the GaN-LEDs with sub-micrometer-sized structure PSS have been carried out. Table 4.5 summarizes the typical results of the application of the sub-micrometer-sized structure PSS to the GaN-LEDs. The majority of the researches incorporate with the surface structures on the chips, such as *p*-type GaN layer or transparent ITO contact, which were fabricated by electron beam lithography and RIE [59–61]. Some researches have achieved sub-micrometer-sized structure PSS using laser holography [62] or nano-imprint lithography (NIL) [63, 64] and ICP-RIE, and have tried to apply the PhC-LEDs. But, PhC-PSS is currently still under investigation.

Figure 4.14(a) shows the bird's eye view of SEM image of nano-PSS. The sub-micrometer-sized structure PSS, which has the hexagonally arranged pillars with the diameter of 200 nm, was fabricated by NIL and ICP-RIE. The period, the diameter, and the height of the arranged pillar structure are 400 nm, 200 nm, 200 nm, respectively. The far field pattern of the GaN-LEDs fabricated on the sub-micrometer-sized structure PSS at a forward-bias current of 20 mA before chip dicing is the six-fold symmetric diffraction pattern as shown in Fig. 4.14.
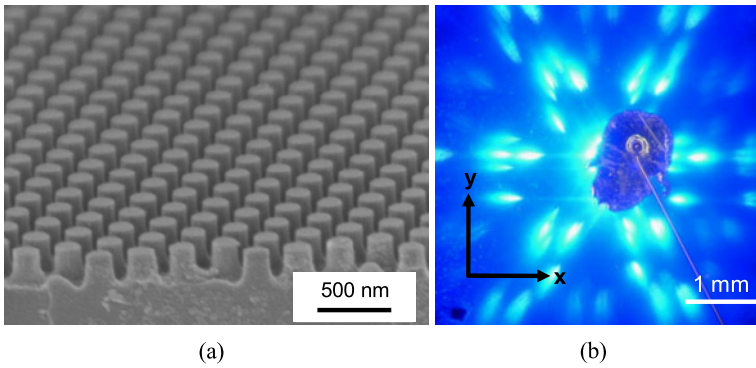
Fig. 4.14 (**a**) Bird's-eye-view SEM image of nano-PSS. (**b**) Far field pattern of electroluminescence from GaN-LED on sub-micrometer-sized PSS (Provided by courtiers of Yamaguchi University)

The output power toward the upper side of the GaN-LEDs fabricated on the sub-micrometer-sized PSS measured by probing system is two times larger than that of the GaN-LEDs on the micrometer-sized PSS. But the total radiant flux of the GaN-LED dies fabricated on the sub-micrometer-sized PSS structure measured by integral sphere system is nearly equal to that on micrometer-sized PSS. It is eagerly expected to optimize the sub-micrometer-sized structure PSS and to develop the PSS with 2D-PhC.

4.5 Novel Application of PSS to Growth of Nonpolar or Semipolar GaN

The commercialized GaN-LEDs fabricated on a polar c-plane GaN layer are under the quantum-confined Stark effect (QCSE) due to the polarization-induced large electric field such as piezoelectric field. The QCSE causes band bending in In-GaN quantum well (QW) in multiple quantum wells (MQWs) and results in the electron-hole spatial separation. The QCSE is one of the reasons to decrease the IQE especially at long wavelength region and to cause so-called "green-gap problem" [68]. Usually, thin-InGaN-QW, around 3-nm-thick, is fabricated in the GaN-LEDs in order to enhance the overlap of the electron and hole wave-functions. This thin-InGaN-QW may be one of the possible suspected origins of "efficiency droop phenomenon" at the high injection current density [69, 70].

To overcome these problems, the GaN-LEDs fabricated on nonpolar GaN layer, such as m-plane, or semipolar GaN layer, such as {11–22} plane, have been investigated. The nonpolar and/or semipolar GaN have possibility to resolve the green-gap problem due to reduced QCSE in InGaN-QW. Furthermore, thick-InGaN-QWs are available due to the reduced QCSE, which will result in reducing the efficiency droop due to lowered the carrier density in the thick-InGaN-QWs [71–73].
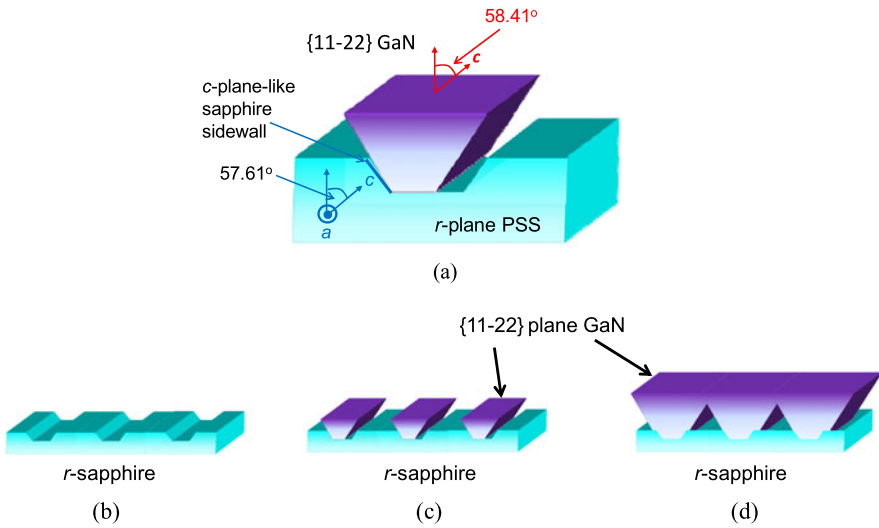
Fig. 4.15 Principle of the growth of semipolar {11–22} GaN on r-plane PSS: (**a**) Relationship between crystal orientations of GaN layer and sapphire substrate. (**b**) Stripe-trench-patterned r-plane PSS. (**c**) Initial stage of semipolar {11–22} GaN growth on the r-plane PSS. (**d**) Semipolar {11–22} GaN layer grown on the r-plane PSS

The high quality c-plane GaN bulk crystal is fabricated by hydride vapor phase epitaxy (HVPE) [74]. The Na-flux solution method [75] and ammonothermal growth [76, 77] have been studied intensively to obtain the GaN bulk crystal. On the other hand, semipolar or nonpolar GaN substrates are usually obtained by slicing these GaN bulk crystals. Thus, these GaN substrates are small in size and expensive.

Nonpolar and semipolar GaN layers grown on the patterned Si substrates have been developed by SAG from the Si sidewall [78, 79]. However, there are many problems, such as the low transmittance of Si for the visible light. Thus, the development of large diameter, low-cost, transparent, and high-quality nonpolar and semipolar GaN substrate has been eagerly awaited.

The novel growth technique of nonpolar or semipolar GaN layers with large diameter and high quality on transparent PSS has been investigated. The nonpolar or semipolar GaN layer is grown from the sapphire sidewall of the PSS using SAG and ELO [80–86].

Figure 4.15 schematically shows the growth principle of nonpolar or semipolar GaN layer on the PSS. As an example, the growth of semipolar {11–22} GaN layer on the stripe-trench-patterned r-plane PSS is described in Fig. 4.15. The stripe-trench-pattern is formed perpendicular to the c-axis of the r-plane sapphire by conventional photolithography and the ICP-RIE. The exposed sidewall of the trench is c-plane-like, because the etched-sapphire-sidewall is not the exact c-plane. Because the angle of 57.61° between r-plane and the c-plane of a sapphire is similar to that of 58.41° between {11–22} plane and c-plane of GaN, c-plane GaN with {11–22} facets grows from c-plane-like sidewall of r-plane PSS, resulting in the growth of

Fig. 4.16 Cross-sectional
SEM image of {11–22} GaN
layer grown on *r*-plane PSS
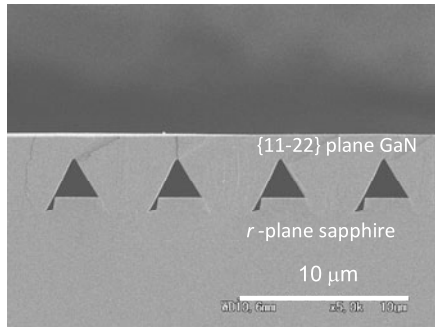
semipolar {11–22} GaN on *r*-plane PSS as shown in Fig. 4.15. The significant point of this technique is that *c*-plane GaN layer grows on *c*-plane-like sapphire sidewall and both *c*-axes are exactly parallel.

Furthermore, the {11–22} GaN with perfect flat surface grows on the 0.8°-off-angle *r*-plane PSS, because the {11–22} plane of the GaN layer is exactly parallel to the surface of 0.8°-off-angle *r*-plane sapphire substrate. The typical trench depth, the width, and the period are 1 µm, 3 µm, and 6 µm, respectively. In the case of {11–22} GaN growth on *r*-plane PSS, SiO$_2$ mask is not necessary on the terrace region of the PSS, because the growth rate of *c*-plane GaN on *c*-plane-like sapphire sidewall is larger than that of *a*-plane GaN on *r*-plane terrace region of *r*-plane PSS. The SiO$_2$ mask may be used to prevent the GaN growth on the terrace region of the PSS, when *n*-plane PSS on which c-GaN grows, is used.

The cross-sectional SEM image of semipolar {11–22} GaN layer grown on *r*-plane PSS is shown in Fig. 4.16. The GaN layer is selectively and laterally grown from the *c*-plane-like-sapphire sidewall only, and forms a facet structure with {11–22} and -*c* (000-1) plane [81]. As seen in this Fig. 4.16, the GaN layer coalesced perfectly. Okada et al. reports that the average threading dislocation density of {11–22} GaN layer is estimated to be approximately 2×10^8 cm^{-2} by counting dark spots in the cathodeluminescence (CL) images.

Using this principle, *m*-plane, *a*-plane, {1–101} plane, and {11–22} plane GaN layers are grown on the *a*-plane, *m*-plane, *n*-plane, *r*-plane PSSs, respectively [80–82, 87], as shown in Fig. 4.17 and {20–21} plane GaN is grown on {11–23} plane PSS [88]. Each PSS has a small off-angle to obtain the GaN layer with the lattice plane, such as {11–22}, parallel to the surface of sapphire substrate. In principle, the GaN layer with any lattice plane is able to grown on the PSS by this procedure.

4.6 Summary

In this chapter, the fabrication technique, the history, future technique "nano-PSS", principle of decreasing the dislocation, principle of increasing light extraction efficiency of GaN-based LED, and novel application to nonpolar and/or semipolar GaN
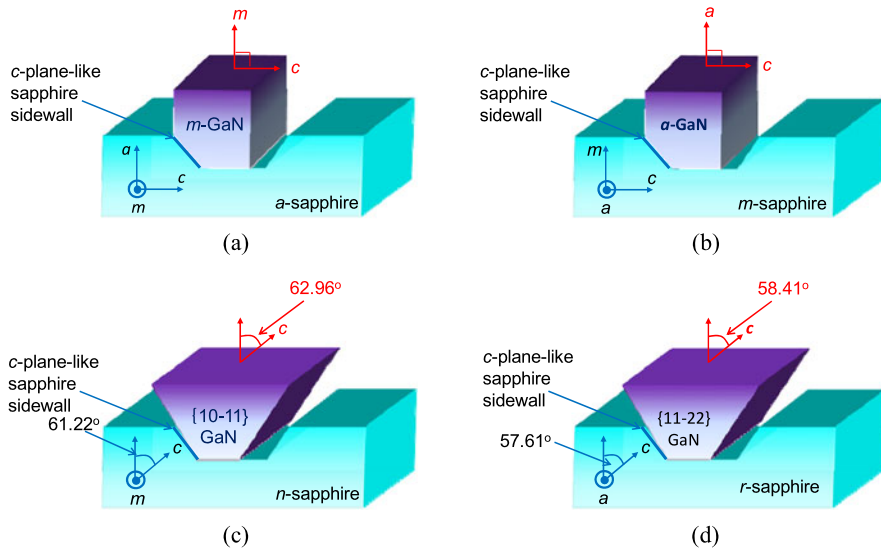
Fig. 4.17 Relationship between crystal orientations of the GaN layer and sapphire substrate. (**a**) *m*-plane GaN is grown on *a*-plane PSS. (**b**) *a*-plane GaN is grown on *m*-plane PSS. (**c**) {10–11} plane GaN is grown on *n*-plane PSS. (**d**) {11–22} plane GaN is grown on *r*-plane PSS

of the Patterned Sapphire Substrate (PSS) are elaborated. Advanced PSS enables us to obtain high-efficiency GaN-LEDs due to both increasing the Internal Quantum Efficiency (IQE) and increasing the Light Extraction Efficiency (LEE) of the GaN-LEDs.

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Chapter 5
Growth and Optical Properties of GaN-Based Non- and Semipolar LEDs

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Abstract Light emitting diodes (LEDs) based on the (In,Al,Ga)N material system offer the possibility to generate light in the entire visible wavelength range, extending into the ultraviolet and the infrared regions. The widely tunable bandgap makes nitride based LEDs suitable devices for applications such as general energy efficient lighting, water purification, UV curing and medical applications. Conventionally, all group III-nitride based devices have been grown epitaxially on the polar (0001) *c*-plane of the wurtzite crystal structure. This leads to the formation of strong polarization fields pointing along the [0001] *c*-axis. These fields reduce the device efficiency through the quantum confined Stark effect (QCSE) and also cause other detrimental effects like wavelength-shifts and efficiency droop with increasing current densities. By growing InAlGaN heterostructures on non- and semipolar growth planes, these fields can be significantly reduced or even eliminated. In addition, due to the reduction of in-plane symmetry, a number of new heterostructure design options emerge to control the optoelectronic properties of non- and semipolar light emitters. Among these are the occurrence of anisotropic strain with the consequence of an anisotropic valence band structure and the possibility to generate strongly polarized light emission from LEDs. In this chapter we will discuss the origin of the polarization fields in III-nitrides and their control by growth on non- and semipolar crystal planes. Different approaches for the homo- and heteroepitaxial growth of non- and semipolar nitride heterostructures as well as structural properties, such as surface morphologies and indium incorporation efficiencies will be discussed. The influence of the crystal plane and the indium content on the valence band structure

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and the polarization state of the emitted light will be presented and the state-of-the-art device characteristics of non- and semipolar LEDs will be reviewed.

5.1 Introduction

The development of smooth (0001) GaN films on c-plane sapphire [1] and the activation of p-dopants in GaN [2] led very quickly to the realization of high brightness InGaN LEDs on c-plane sapphire substrates [3, 4]. Already at the end of the last century blue and green LEDs with tens and hundreds of milli-Watt output power levels were demonstrated. Today, blue InGaN LEDs boast record external quantum efficiencies exceeding 80 % and the emission wavelength of c-plane InGaN quantum well (QW) LEDs has been pushed into the yellow and even red spectral range. Although the performance characteristics of c-plane LEDs seem excellent, the strong polarization fields at In-GaN/GaN heterointerfaces can lead to a significant deterioration of the device performance. This polarization field is suppressed or reduced in LEDs with InGaN/GaN heterointerfaces of nonpolar or semipolar orientation, respectively. Triggered by the first demonstration of nonpolar GaN quantum wells grown on LiAlO$_2$ by Waltereit and colleagues in 2000 [5], impressive advancements in the field of non- and semipolar nitride semiconductors and devices have been achieved. Today, a large variety of heterostructures free of polarization fields exhibiting exceptional optical properties have been demonstrated, and the fundamental understanding of polar, semipolar, and nonpolar nitrides has made significant leaps forward. This chapter is intended to provide an overview on the epitaxial growth and optical properties of group III-nitride LEDs on non- and semipolar surface orientations [6]. After introducing the physical origins of piezoelectric and spontaneous polarization effects in group-III nitrides, different approaches for the heteroepitaxial growth of low defect density non- and semipolar (Al,In)GaN layers and (Al,In)GaN/GaN heterointerfaces are presented, followed by a discussion of the effect of surface orientation on the indium incorporation efficiency in InGaN layers and quantum wells. In the third section the polarized light emission characteristics and the optical properties of non- and semipolar InGaN QWs are discussed and finally the performance characteristics of non- and semipolar LEDs are presented including the effects on droop, wavelength shift and external quantum efficiencies of state-of-the-art devices.

5.2 Piezoelectric and Spontaneous Polarization in Group-III Nitrides

The stable crystal structure of the group-III nitrides AlN, GaN, and InN and their ternary compounds AlGaN, InGaN, and AlInN is the hexagonal wurtzite structure. The high electronegativity of the nitrogen atom creates a strong dipole in the metal–nitrogen bond which is the origin of the strong polar character of the group-III nitrides. In their real crystal structure the bond lengths and bond angles deviate slightly

from the ideal tetrahedral configuration of the wurtzite structure, without loosing the hexagonal crystal symmetry C_{6v}. This is the point group of highest symmetry without a center of inversion [7]. Because of this distinction between "up" and "down", the crystal has two polar surfaces, one terminated with metal atoms Al, Ga, or In, the other N-face. By convention, the [0001] direction points from a metal atom to the nearest-neighbor nitrogen atom in the direction of highest symmetry, which is also the optical axis of the uniaxial crystal, or the c-axis.

Because of the missing inversion symmetry, the material is pyroelectric and consequently exhibits spontaneous and piezoelectric polarization [8]. This leads to sheet charge layers at surfaces and interfaces. Surface charges screen most of the fields in thick layers or bulk material. However, at internal interfaces the spontaneous and piezoelectric polarization causes strong band bending and can induce two-dimensional electron gases (2DEG) of high charge carrier density. These are the basic mechanisms enabling GaN-based high electron mobility transistors (HEMT) [9]. In InGaN quantum wells (QW) of optoelectronic devices, the polarization field has a strong influence on radiative recombination. In standard c-plane LEDs or laser structures, the active region consists of several few nanometer thick InGaN QWs between GaN or InGaN barriers. Because of the increasing in-plane lattice constants of InGaN with indium content, the QW experiences biaxial compressive strain. The thickness of the QWs must remain below the critical thickness to suppress the formation of new dislocations. In this regime of pseudomorphic growth, the in-plane strain ε_{xx} for the QW is given by the difference of the a-lattice constant a of the QW and a_0 of the barrier:

$$\varepsilon_{xx} = \frac{a - a_0}{a_0}. \tag{5.1}$$

Stress and strain are related through the elastic tensor with elastic constants c_{ij}. For the polar QW with biaxial strain the stress component σ_1 in the basal plane can be calculated by

$$\sigma_1 = \varepsilon_{xx}\left(c_{11} + c_{12} - 2\frac{c_{13}^2}{c_{33}}\right). \tag{5.2}$$

Strain and piezoelectric polarization are related through the piezoelectric coefficients d_{ij}. For the hexagonal crystal of C_{6v} symmetry the piezoelectric polarization P_i in i direction is given by

$$P_i = \sum_j d_{ij}\sigma_j, \quad i = 1, 2, 3, \ j = 1, \ldots, 6. \tag{5.3}$$

For the polar QW the piezoelectric polarization in the direction z perpendicular to the QW resulting from biaxial in-plane strain ε_{xx} is then given by [8]

$$P_z = 2d_{31}\varepsilon_{xx}\left(c_{11} + c_{12} - 2\frac{c_{13}^2}{c_{33}}\right). \tag{5.4}$$

The resulting internal field in an InGaN QW is of the order of 0.5 MV/cm to 3 MV/cm for a standard LED structure. This large internal field causes a strong
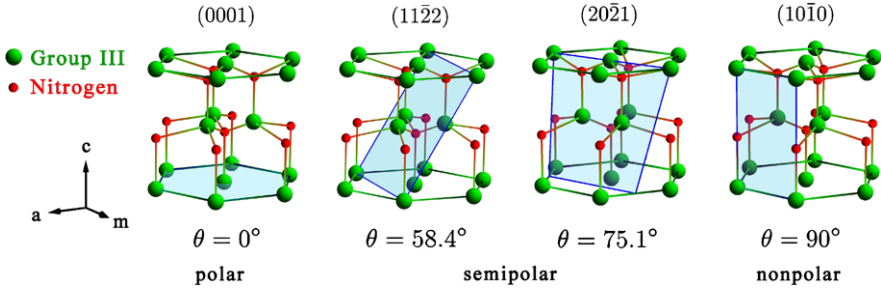
Fig. 5.1 Orientations of c-plane, semipolar $(11\bar{2}2)$- and $(20\bar{2}1)$-planes and nonpolar $(10\bar{1}0)$-plane (*shaded polygons*) with inclination angle θ to the c-plane [12] (Color figure online)

bending of the band edge in the active region. It severely affects recombination rates, wavelength of emission, and transport.

Radiative recombination is proportional to the absolute square of the overlap integral of electron and hole wave functions. The field inside the QW causes a spatial separation of the electron and hole wave function, resulting in a spatial indirect transition which reduces the radiative recombination rate. This shifts the balance from radiative to non-radiative recombination, which is less affected by the internal field, and therefore reduces the internal quantum efficiency of the LED.

The wavelength of emission is shifted due to the so-called quantum confined Stark effect (QCSE): Band bending in the active region shifts the transition between bound states of the QW to longer wavelengths. Because the internal fields are partially screened with increasing carrier density, the wavelength of a polar InGaN LED is shifting towards shorter wavelength with increasing current density. This effect is most pronounced for LEDs emitting in the green spectral region where the strain and consequently the field inside the QW is largest. Carrier transport is affected because the polarization field causes additional barriers in the active region which affect forward voltage and current injection efficiency [10, 11].

A way to avoid these limitations is to use different lattice orientations than the polar c-plane. For planes perpendicular to the c-plane the spontaneous and piezoelectric field perpendicular to the QW is zero because of crystal symmetry. These planes are therefore called nonpolar planes. In other planes with an inclination between polar and nonpolar plane the polarization fields can be expected to be reduced. Prominent polar, semipolar and nonpolar planes which are being used for LED growth are shown in Fig. 5.1. Each plane is associated with a specific angle θ between the axis perpendicular to the QW and the c-axis of GaN. This angle ranges from $\theta = 0°$ for the polar QW to $\theta = 90°$ for the nonpolar case.

A simple projection of the piezoelectric field pointing in c-direction as described by Eq. (5.3) on the respective plane will not give correct results for the internal field. Instead, one has to first calculate the strain for the QW under the assumption of pseudomorphic growth in the respective orientation. From the strain components the piezoelectric polarization is then calculated with the piezoelectric coefficients. A semipolar QW has non-vanishing strain components ε_{xx}, ε_{yy}, ε_{zz}, and ε_{xz}. In a

Fig. 5.2 Strain tensor components as function of crystal angle for a fully strained $In_{0.25}Ga_{0.75}N$ layer grown on free-standing GaN substrate. The *dashed lines* mark the semipolar ($11\bar{2}2$)- and the ($20\bar{2}1$)-planes [12]
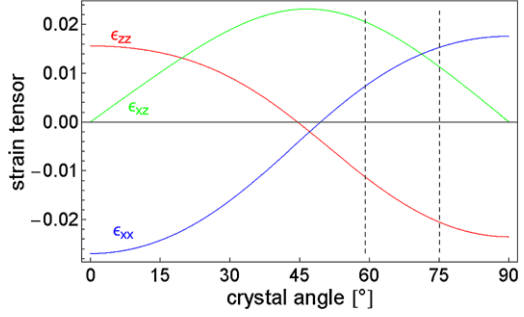
strain model developed by Park and Chuang, the strain components as function of the inclination θ are given by

$$\varepsilon_{xx} = \varepsilon_{xx}^{(0)} + \varepsilon_{xz} \tan \theta, \tag{5.5}$$

$$\varepsilon_{yy} = \varepsilon_{yy}^{(0)} = \varepsilon_{xx}^{(0)}, \tag{5.6}$$

$$\varepsilon_{zz} = \varepsilon_{zz}^{(0)} + \varepsilon_{xz} \cot \theta, \tag{5.7}$$

$$\varepsilon_{xz} = -\frac{((c_{11} + c_{12} + c_{13}\frac{\varepsilon_{zz}^{(0)}}{\varepsilon_{xx}^{(0)}}) \sin^2 \theta + (2c_{13} + c_{33}\frac{\varepsilon_{zz}^{(0)}}{\varepsilon_{xx}^{(0)}}) \cos^2 \theta)\epsilon_{xx}^{(0)} \cos \theta \sin \theta}{c_{11} \sin^4 \theta + 2(c_{13} + 2c_{44}) \sin^2 \theta \cos^2 \theta + c_{33} \cos^4 \theta}, \tag{5.8}$$

where $\epsilon_{xx}^{(0)} = (a_s - a_e)/a_e$ and $\epsilon_{zz}^{(0)} = (c_s - c_e)/c_e$ and a_s, c_s, a_e, c_e are the lattice constants of the substrate [13]. The fact that the shear strain ε_{xz} is non-vanishing in a semipolar QW has major consequences for the band structure and polarization of the optical emission, as will be discussed in Sect. 5.4. The dependency of the different components of the strain tensor on θ is shown in Fig. 5.2.

The polarization discontinuity $\Delta P_{z'}$ between QW and barrier is then given through the strain tensor and piezoelectric coefficients d_{ij}:

$$\Delta P_{z'} = \left(\left[d_{31}(c_{11} + c_{12}) + d_{33}c_{13} \right](\varepsilon_{xx} + \varepsilon_{yy}) + (2d_{31}c_{13} + d_{33}c_{33})\varepsilon_{zz} \right) \cos \theta$$
$$+ (2d_{15}c_{44}\varepsilon_{xz}) \sin \theta + \left(P_{sp}^{(QW)} - P_{sp}^{(b)} \right) \cos \theta. \tag{5.9}$$

Here, z' points in growth direction and perpendicular to the QW. $P_{sp}^{(QW)}$ and $P_{sp}^{(b)}$ are the spontaneous polarization in the quantum well and in the barrier, respectively. Equation (5.9) holds for all angles θ. The resulting polarization discontinuity is plotted in Fig. 5.3 together with the wave function overlap for a 3 nm wide $In_{0.25}Ga_{0.75}N$ QW. The polarization discontinuity can be implemented as sheet charge layers at the interfaces between QWs and barriers in a drift-diffusion model. This allows the calculation of the internal field as well as carrier transport as function of external applied bias voltage within the accuracy of the drift-diffusion model. An estimation for the internal field can be given by assuming appropriate boundary conditions and employing Gauss' law $\vec{\nabla} \cdot (\varepsilon \varepsilon_0 \vec{E} + \vec{P}) = 0$, resulting in [12]:

$$\varepsilon^{(QW)} E_z^{(QW)} - \varepsilon^{(b)} E_z^{(b)} = -\frac{\Delta P_{z'}}{\varepsilon_0}. \tag{5.10}$$

Fig. 5.3 Polarization discontinuity $\Delta P_{z'}$ along growth direction and wave function overlap as function of crystal angle for a 3 nm wide $In_{0.25}Ga_{0.75}N$ quantum well. The *dashed lines* mark the semipolar $(11\bar{2}2)$- and the $(20\bar{2}1)$-planes [12]
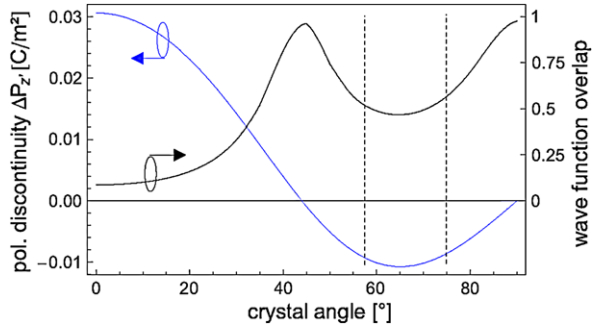
Here, ε_0 is the permittivity of vacuum, $\varepsilon^{(b)}$ and $\varepsilon^{(QW)}$ are the static dielectric constants for the barrier layers and the quantum well, respectively.

For an inclination of about $\theta = 50°$ the different components of the piezoelectric field compensate each other, resulting in a zero polarization discontinuity. For some time the existence of this zero transition of the polarization field was heavily discussed, as it depends critically on the parameters used in the strain model on the theoretical side and on the strain in a real QW when the internal field is determined experimentally [14–16]. However, currently the parameters determining strain and piezoelectric field are converging and there is a general agreement on the existence of the zero crossing for the internal field. Because this zero crossing is close to the angle corresponding to the $(11\bar{2}2)$ lattice plane, this plane, which also shows stable growth, is preferred for semipolar LEDs. However, considering also optical properties and transport, other planes may be better candidates for semipolar LEDs and laser diodes. We will later see that the effect of polarization switching is most prominent for the $(11\bar{2}2)$ orientation, which is relevant for semipolar optoelectronic devices emitting in the green spectral region.

5.3 Growth of GaN and InGaN on Different Non- and Semipolar Surface Orientations

The synthesis of GaN layers with non- or semipolar surface orientations can be achieved by different means. The most straightforward approach is the growth of single crystal GaN boules, from which GaN wafer slabs can be cut out at basically any angle and therefore crystal orientation of choice. Another advantage of this approach is that the defect densities in the semi- and nonpolar bulk GaN substrates are similar to the defect densities of the original GaN boule. Since GaN boules can be realized with very low defect densities the resulting semi- and nonpolar bulk GaN provide an excellent growth substrate with very low defect densities. This type of approach has already been successfully demonstrated by a number of research groups [17–21]. In most cases the bulk GaN crystal boules have been grown by hydride vapor phase epitaxy (HVPE) or ammonothermal growth along the (0001)

direction. In order to obtain GaN wafers with non- or semipolar surface orienta-
tions the GaN boules have to be cut perpendicular to the c-plane surface for nonpo-
lar orientations or at the appropriate off-axis angle corresponding to the respective
semipolar surface plane. The challenge with this approach is, that in order to ob-
tain larger non- and semipolar substrate sizes, the c-plane GaN boules have to be
large in height as well as in diameter. This is still a significant challenge, since the
GaN growth rates, even with HVPE, are relatively low and the built-up of strain and
parasitic deposits during growth limits the maximum heights that can be achieved
without fracturing the boule. In addition the fabrication costs for bulk GaN boules
are relatively high, which also translates into high costs for the semi- and nonpolar
bulk GaN substrates. Therefore the heteroepitaxial growth of non- and semipolar
GaN layers on readily available sapphire, silicon, SiC and other substrates is still of
great interest and importance. All of the heteroepitaxial substrates are available with
large diameters (e.g. $6''$ is already standard for sapphire substrates and $8''$ and $12''$
sapphire are under development) and the costs of these substrate are low to moder-
ate. In the past decade a large number of substrate materials and surface orientations
have been explored in order to realize heteroepitaxial GaN layers with different non-
and semipolar surface orientations. A detailed discussion of the different approaches
is given in the following paragraph.

5.3.1 Heteroepitaxial Growth of Non- and Semipolar GaN on Sapphire, Silicon, Spinel, and LiAlO$_2$ Substrates

Growth of non- and semipolar GaN layers has been demonstrated on sapphire, sil-
icon, SiC, spinel, as well as LiAlO$_2$ substrates and on a number of different sub-
strate orientations. Initially most of the experiments have focused on heteroepitaxial
growth on planar substrates [5, 22–27]. In these instances the orientation of the
grown GaN layers is strongly correlated with the crystal orientation of the host sub-
strate, although in some cases several semipolar orientations can be obtained for the
same growth substrates (e.g. growth on $(10\bar{1}0)$ m-plane sapphire can yield $(11\bar{2}2)$
and $(10\bar{1}3)$ GaN layers). An overview of the different heteroepitaxial substrates
and the orientations of the resulting non- and semipolar GaN layers is provided in
Fig. 5.4. Although the figure does not attempt to provide a complete list, all of the
main approaches are included newline.

   In addition the heteroepitaxial growth of non- and semipolar GaN has also been
demonstrated on patterned host substrates. Growth of GaN on stripe-patterned sub-
strates allows access to different surface orientations depending on the angle of the
exposed growth facet and simultaneously is a method to reduce the defect density
in the overgrown GaN layers. Figure 5.5 shows schematically the heteroepitaxial
growth on planar host substrate and on a stripe patterned sapphire substrate. In or-
der to prepare the stripe pattern several micron wide mesa stripes oriented along
the a-direction are dry-etched into the r-plane sapphire substrate. To prevent GaN
growth on the $(11\bar{2}3)$ n-plane mesa terraces a 100 nm thin SiO$_2$ layer is deposited
on top of the mesa. The subsequent GaN growth is nucleated at the inclined etched
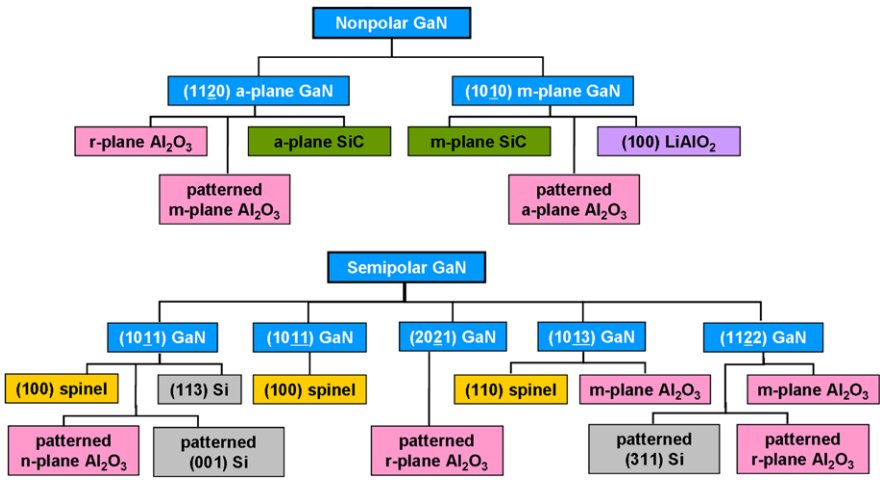
Fig. 5.4 Overview showing the relationship between different substrate orientations and materials and the orientation of the resulting GaN layer. The examples include heteroepitaxial growth of GaN layers on planar substrates as well as growth of non- and semipolar GaN on patterned sapphire and silicon substrates
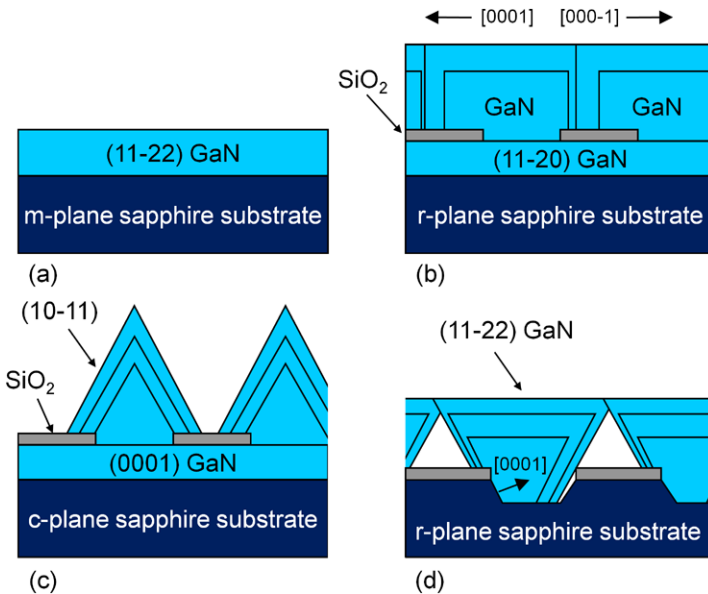
Fig. 5.5 Schematic illustration of (**a**) heteroepitaxial growth of semipolar $(11\bar{2}2)$ GaN on planar $(10\bar{1}0)$ m-plane sapphire substrate, (**b**) defect reduction by epitaxial lateral overgrowth (ELO) for $(11\bar{2}0)$ GaN layers on $(10\bar{1}2)$ r-plane sapphire, (**c**) $(10\bar{1}1)$ GaN facet on 3-dimensional stripes on c-plane sapphire and (**d**) heteroepitaxial growth of semipolar $(11\bar{2}2)$ GaN on a stripe patterned $(10\bar{1}2)$ r-plane sapphire substrate

mesa facet, which is close to parallel to the sapphire and GaN c-planes. The GaN growth from these facets continues along the c-direction and a planar $(10\bar{1}1)$ GaN surface is formed when the adjacent GaN growth stripes coalesce. This method has been pioneered by Tadatomo et al. [28, 29] and Sawaki et al. [30, 31] and has since been demonstrated for a number of different patterned sapphire as well as silicon substrates as can be seen from the genealogy displayed in Fig. 5.4. The patterned substrate approach provides access to a number of GaN surface orientation, e.g. by adjusting the angle of the mesa facets or by utilizing different crystal orientations. Using the patterned sapphire substrate approach m-plane GaN layers on patterned a-plane sapphire, a-plane GaN on m-plane or c-plane sapphire, $(10\bar{1}1)$ GaN on n-plane sapphire, and $(20\bar{2}1)$ and $(10\bar{1}3)$ GaN on r-plane sapphire have been demonstrated. Similarly $(10\bar{1}1)$ GaN layers have been realized on patterned (001) silicon substrates and $(11\bar{2}2)$ GaN on patterned (311) silicon substrates. Most recently the group of Krost et al. have shown the growth of $(1\bar{1}06)$ oriented GaN films on (112) Si as well as $(1\bar{1}05)$ and $(1\bar{1}04)$ GaN layers on (113) Si substrates [32]. Again, this list is certainly not complete and the number of possible orientations is expected to grow as more and more substrate materials, stripe patterns, and orientations are being explored.

5.3.2 Surface Morphologies and Strutural Defects of Non- and Semipolar GaN Films

The control of the surface morphology and defect densities of heteroepitaxially grown semipolar and nonpolar GaN layers on planar substrates is very challenging. Wernicke et al. have investigated $(11\bar{2}0)$ GaN on r-plane sapphire, $(10\bar{1}0)$ GaN on LiAlO$_2$, and $(11\bar{2}2)$ GaN on $(10\bar{1}0)$ sapphire substrates [34] and found strongly varying surface morphologies. For example, one problem with the growth of a-plane GaN was the creation of surface pits that are formed by $(10\bar{1}0)$, $(10\bar{1}1)$ and $(000\bar{1})$ facets, whereas semipolar $(11\bar{2}2)$ GaN layers exhibited a strong tendency to roughen and show arrowhead like features as depicted in Fig. 5.6(a) [33, 35].

Distinct surface morphologies develop even for non- and semipolar GaN layers grown on low defect density bulk GaN substrates as shown in Fig. 5.6(b). The scanning white light interferometry (SWLI) images of homoepitaxially grown m-plane, $(10\bar{1}1)$, $(10\bar{1}2)$, and $(11\bar{2}2)$ GaN layers show very distinct short range surface features for each surface orientation. While the $(11\bar{2}2)$ surface seems relatively smooth, the m-plane and the $(10\bar{1}1)$ exhibit very distinct microscopic features, which are partly due to faceting. The surface morphologies are affected by the growth conditions, in particular growth temperature and reactor pressure. Figure 5.7 shows a series of Normarski contrast microscope images of semipolar $(10\bar{1}1)$, $(10\bar{1}2)$, and $(11\bar{2}2)$ GaN layers grown homoepitaxially with different MOVPE reactor pressures on bulk GaN substrates. While under optimized growth conditions the $(10\bar{1}2)$ and $(11\bar{2}2)$ GaN layers showed a relatively smooth surface morphology, the $(10\bar{1}0)$ and $(10\bar{1}1)$ GaN surfaces exhibited macroscopic pyramids, which originated from screw dislocations [36].
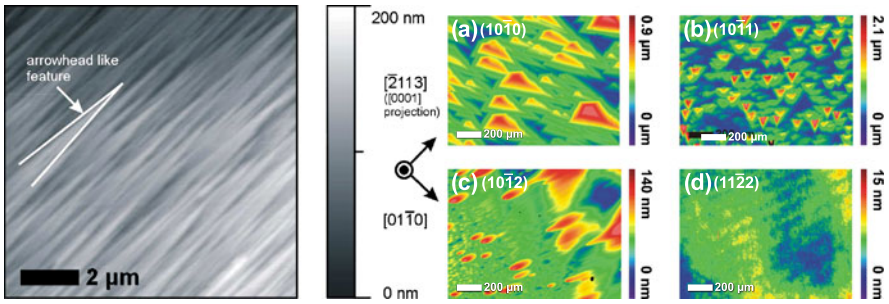
Fig. 5.6 (**a**) Atomic force microscopy (AFM) image of a (11$\bar{2}$2) GaN layer grown on an *m*-plane sapphire substrate [33]. Clearly visible are arrowhead like features on the surface, which are oriented along the c' [11$\bar{2}$3] direction. (**b**) Scanning white light interferometry (SWLI) images of *m*-plane GaN, (10$\bar{1}$1), (10$\bar{1}$2), and (11$\bar{2}$2) GaN layers homoepitaxially grown on bulk GaN substrates [34] (Color figure online)

Fig. 5.7 Normarski contrast microscope images of semipolar (10$\bar{1}$1), (10$\bar{1}$2), and (11$\bar{2}$2) GaN layers grown homoepitaxially with different reactor pressures on bulk GaN [34]. The growth temperature was kept constant at 980 °C
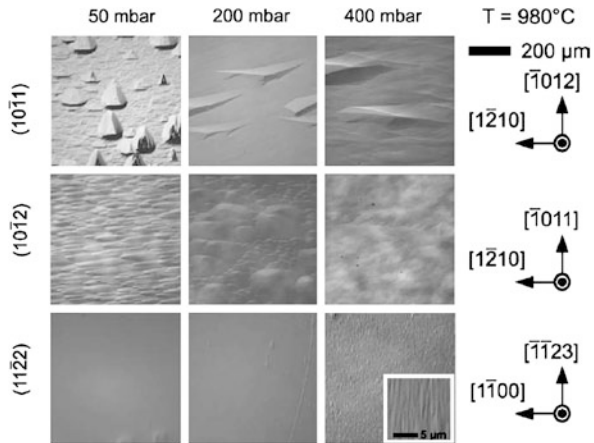
Even though the surface morphologies of non- and semipolar GaN layers can be significantly improved by optimization of the growth parameters, the reduction of the defect densities, especially for heteroepitaxially grown layers is much more challenging. Despite refined growth processes, the edge dislocation densities of heteroepitaxially grown non- and semi-polar GaN layers range typically from 10^9 to 10^{10} cm^{-2} and the basal plane stacking faults (BSF) density is about 10^6 cm^{-1}. This poses serious limitations to the device performance, since extended defects act as non-radiative recombination centers and are thus detrimental to the internal quantum efficiencies of light emitting diodes. In order to reduce the dislocation densities a number of different techniques have been explored. The most intensively explored is lateral epitaxial overgrowth (ELOG) [37]. Originally the ELOG method has been applied to reduce the defect density in planar semi- and nonpolar GaN layers. For example, thin silicon dioxide stripes were patterned on nonpolar (11$\bar{2}$0) GaN layers on sapphire which enabled a selective re-growth of *a*-plane GaN in the

Table 5.1 Cathodoluminescence (CL) dark spot density of different semipolar GaN layers grown on patterned sapphire substrates

GaN orientation	Patterned sapphire orientation	Dark spot density [cm^{-2}]	Reference
$(11\bar{2}2)$	$(10\bar{1}2)$	$2 \cdot 10^8$	[28]
$(10\bar{1}1)$	$(11\bar{2}3)$	$1.2 \cdot 10^8$	[28]
$(20\bar{2}1)$	$(22\bar{4}3)$	$5.6 \cdot 10^6$	[29]
$(10\bar{1}0)$	$(11\bar{2}0)$	not specified	[42]

mask openings [38]. By aligning the 100 nm thick SiO_2 stripe mask parallel to the $[10\bar{1}0]$ direction, TEM investigations and other studies have shown that the BSF and threading dislocation densities can be reduced by several orders of magnitudes in the laterally overgrown GaN areas [39–41]. The improved structural quality also translates into much improved optical properties of the nonpolar GaN films. For example, cathodoluminescence (CL) investigations of ELOG GaN films show sharp donor bound excitonic emission from the defect-reduced laterally overgrown areas whereas without defect reduction the CL is dominated by BSF luminescence [40]. An alternative approach to defect reduction is the growth on patterned substrates, as already described in Sect. 5.3.1. Since the semi- and nonpolar layers are selectively nucleated on c-plane like facets very few basal plane stacking faults are generated in the grown GaN layers and the threading dislocations are mostly concentrated in the coalesced regions [28, 29]. Depending on the semipolar orientation of the overgrown GaN layer and the patterned sapphire substrate, the threading dislocation densities and BSF densities can vary significantly. Table 5.1 summarizes some of the recent results. As can be seen from Table 5.1, very low CL dark spot densities in the 10^6 cm^{-2} range can be achieved. These values are close to numbers obtained from growth of non- and semipolar layers on bulk GaN substrates, indicating that heteroepitaxial growth has the potential to provide low defect density templates for high efficiency non- and semipolar light emitters.

Recently Ravash et al. have shown that high silicon doping [32] can lead to a significant reduction of threading dislocation density and BSF density in $(1\bar{1}06)$ and $(1\bar{1}04)$ GaN layers grown on (112) and (113) Si substrates, respectively. Similarly the insertion of low-temperature (LT) AlN interlayers [43] has also led to a significant reduction of threading dislocation and BSF density in $(1\bar{1}06)$ and $(1\bar{1}04)$ GaN. In both cases the reduction of BSFs is most likely due to the generation of a- and c-type misfit dislocations at the LT-AlN/GaN interface, which are preferential to the generation of new stacking faults, especially for semipolar GaN with a low-inclination angle.

A third approach to realize InGaN light emitters on semipolar GaN surfaces is the selective growth of three-dimensional (3D) triangular-shaped GaN structures on stripe patterned c-plane GaN/sapphire substrates [44]. A schematic of this approach is shown in Fig. 5.5(c). Depending on the growth conditions, the resulting triangular-shaped GaN stripes can either exhibit $\{1\bar{1}01\}$ or $\{11\bar{2}2\}$ facets. One advantage of

Fig. 5.8 Emission wavelength and indium incorporation efficiency of InGaN quantum well structures grown on *c*-plane and different semipolar and nonpolar GaN substrates [46]
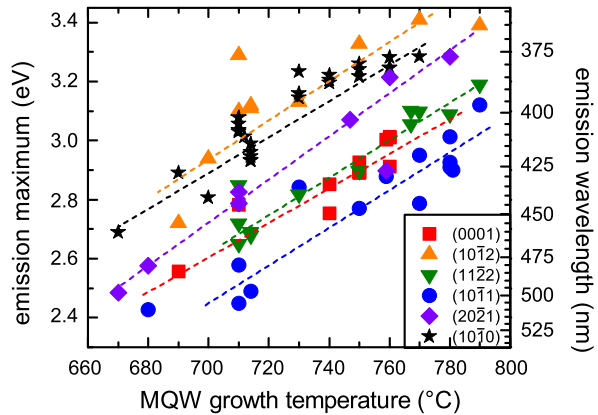
this approach is that it can be realized on large area and low cost (0001) sapphire substrates. Since the growth originates from the *c*-plane, the semipolar surfaces are mostly free of stacking faults and exhibit fairly low threading dislocation densities. The downside is that the 3-dimensional growth makes the device fabrication and electrical contacting more difficult. In addition indium incorporation efficiency depends strongly on the position on the pyramid leading to large scale fluctuations of the emission wavelength.

5.3.3 Indium Incorporation in InGaN Layers and Quantum Wells on Different Semipolar and Nonpolar Surfaces

For the realization of high efficiency blue, green, and yellow LEDs indium incorporation efficiency in InGaN layers and quantum wells is very critical. Theoretical studies based on first principles calculations by Northrup et al. [45] have already shown that the indium incorporation efficiency on ($11\bar{2}2$) surfaces is expected to be higher than for ($10\bar{1}0$) surfaces. For the ($11\bar{2}2$) and ($10\bar{1}0$) surfaces this was explained by the strain-induced repulsive interaction between the relatively large indium atoms at the surface. Recently a systematic experimental study by Wernicke et al. [46] has shown that the indium incorporation in InGaN multiple quantum well structures varies indeed significantly for the different semi- and nonpolar surfaces. A fundamental aspect of these experiments was to compare the growth of InGaN QWs on low defect density bulk GaN substrates with different surface orientations, in particular ($10\bar{1}0$), ($10\bar{1}1$), ($10\bar{1}2$), ($11\bar{2}2$), and ($20\bar{2}1$) GaN surfaces for a series of growth temperatures. The experiments showed that the indium incorporation efficiency in InGaN QWs was consistently highest on the ($10\bar{1}1$) surfaces and very similar to (0001) GaN surfaces for growth on ($11\bar{2}2$) and ($20\bar{2}1$) oriented surfaces. In contrast, InGaN QWs grown on the ($10\bar{1}2$) and ($10\bar{1}0$) surfaces showed much lower indium mole fractions than QWs grown on the conventional (0001) GaN surfaces. Figure 5.8 summarizes these results for a number of different non- and

semipolar orientations as well as different growth temperatures. These results indicate that growth on $(11\bar{2}2)$ and $(20\bar{2}1)$ oriented GaN may be preferable, especially for long-wavelength light emitters that require high In mole fraction InGaN QWs. Overall these investigations demonstrates that the growth on semipolar GaN constitutes a very interesting approach for high efficiency green, yellow, and even longer wavelength light emitting diodes.

5.4 Polarization of the Light Emission from Non- and Semipolar InGaN QWs

In contrast to LEDs with c-plane QWs which emit unpolarized light in the direction perpendicular to the surface, a planar LED structure with QWs of semipolar or nonpolar growth direction emits light which is partially polarized. This polarization is a direct consequence of the lower symmetry of the system. While a polar QW is symmetric within the QW plane, a semipolar or nonpolar QW has two distinct directions, parallel and perpendicular to the projection of the c-axis of the crystal onto the plane of the QW. Standard notation is to introduce a Cartesian coordinate system (x', y', z') with z' perpendicular to the QW, x' and y' within the QW plane; x' and y' are parallel and perpendicular to the projection of the c-axis onto the QW plane, respectively. x' is also called c'. The degree of polarization is then defined as

$$P = \frac{I_{y'} - I_{x'}}{I_{y'} + I_{x'}} \tag{5.11}$$

and determined from intensities $I_{y'}$ and $I_{x'}$ measured using linear polarizers oriented in y'- and x'-direction. $P = \pm 1$ corresponds to fully linear polarized light, $P = 0$ to unpolarized light. Positive and negative P stands for a dominant polarization perpendicular (y') and parallel (x') to the projection of the c-axis onto the QW plane, respectively. One has to be aware that this definition of the degree of polarization does not distinguish between partially linear polarized light and elliptically or circular polarized light. The latter may arise in InGaN LEDs due to birefringence or reflection from tilted surfaces. The full state of polarization in that case can be described with Stokes parameters.

Emission of polarized light was observed in electroluminescence [47, 48] and photoluminescence [49]. For nonpolar QWs in m-direction polarization ratios as high as 0.91 [48, 50] have been observed by electroluminescence at room temperature. Light emitted from nonpolar QWs is linearly polarized, perpendicular to the c-axis. Semipolar QWs generally emit light with a lower degree of polarization. Ueda et al. observed that the direction of the predominant polarization changes from perpendicular to the c-axis to parallel to the c-axis with increasing indium content for QWs of $(11\bar{2}2)$ orientation [51]. This phenomenon was termed "polarization switching" and was reproduced in semipolar InGaN QWs of different lattice orientations [49].

Generally, the degree of polarization and the dominant direction of the linearly polarized light depend both on the structure of the LED, which is determined during

growth, and on measurement conditions. During growth, QW inclination, composition (In or Al content), thickness, and strain are fixed. Upon observation, temperature and excitation density are free parameters. The first set of parameters determines the band structure. The symmetry of the band structure and consequently the selection rules for intraband transitions are the physical origin of the emission of polarized light. The second set of parameters determines the occupation of the subbands, such selecting those states which contribute to light emission. The situation is complicated by the inhomogeneities of the ternary InGaN or AlGaN QWs. Composition, thickness, and strain vary within the QW plane. As discussed in the previous section, semipolar QWs tend to form micro-facets of different lattice orientations with different physical properties. Dislocations—in particular stacking faults intersecting nonpolar QWs—modify strain and are a competing source of polarized light. Last but not least, a precise measurement of the degree of polarization is complex. The experimental setup may either scramble polarization (e.g. with a microscope objective of high numerical aperture), leading to the observation of a lower degree of polarization, or artificially polarize the emitted light by reflection from a tilted surface of an optical component (e.g. a beam splitter). Still, the main effect causing polarized light emission is the band structure. Therefore in the following the band structure of polar, nonpolar and semipolar InGaN QWs will be discussed, proceeding from the simplest to the most complex situation.

The crystal and band structure of bulk GaN around the Γ-point ($k_{x'} = k_{y'} = k_{z'} = 0$) are shown in Fig. 5.9. The spin degenerate heavy-hole (HH), light-hole (LH), and crystal-field split-off (CH) bands form the valence band. They are composed from wave functions of p-character with angular momentum $l = 1$. The angular momentum eigenfunctions of heavy-hole and light-hole bands are $\frac{1}{\sqrt{2}}|X - iY\rangle$ and $-\frac{1}{\sqrt{2}}|X + iY\rangle$ at the Γ-point. For the crystal-field split-off band the eigenfunction is $|Z\rangle$. Here, we neglect the spin degree of freedom. A full description of the 6×6 Hamiltonian in $k \cdot p$ approximation can be found e.g. in Refs. [13, 52, 53]. The eigenfunction of the conduction band has s-character with angular momentum $l = 0$ and is spherical symmetric. The angular momentum eigenfunction for all bands are shown as insets in Fig. 5.9. As the intensity of an optical transition is proportional to the transition matrix element, the degree of polarization for an individual subband m can be written in the following way [54]:

$$P_m = \frac{|M_{y'}^m|^2 - |M_{x'}^m|^2}{|M_{y'}^m|^2 + |M_{x'}^m|^2}. \tag{5.12}$$

The transitions matrix elements $M_{x'}^m$ and $M_{y'}^m$ between conduction and valence band states are calculated in dipole approximation from the wavefunctions of the individual subband as calculated in $k \cdot p$ approximation. Technical details of these calculations are discussed e.g. in [53–55].

For light emitted in c-direction, the polarization vector is perpendicular to the c-axis of the crystal. The transition matrix element is large and independent of the in-plane direction for transitions between conduction band and both heavy-hole and light-hole bands. Therefore light emitted from both bands in c-direction is unpolarized. This can be seen from the shape of the angular momentum eigenfunctions
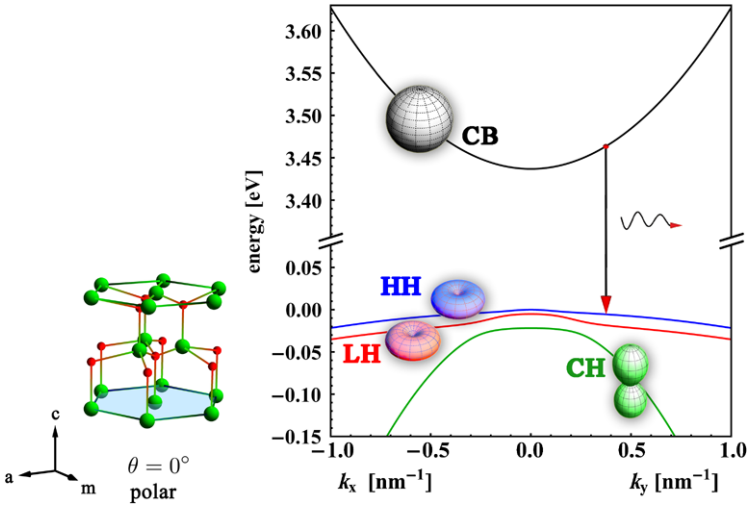
Fig. 5.9 *Left*: Orientation of the polar (0001) lattice plane with respect to the GaN lattice structure. *Right*: Band structure of a bulk GaN. The *insets* show the angular momentum eigenfunctions of the individual bands. The *arrow* marks one possible optical transition [12]

as shown in Fig. 5.9. The crystal-field split-off band does not contribute to emission in c-direction, because its angular momentum eigenfunction has neither $|X\rangle$ nor $|Y\rangle$ component. The situation changes for light emitted in a direction perpendicular to the c-axis. From the orientation of the angular momentum eigenfunctions it follows that photons stemming from the transition between conduction band and heavy-hole or light-hole bands are polarized perpendicular to the c-axis, while the transition from conduction band to crystal-field split-off band is polarized parallel to the c-axis. The energy splitting between heavy-hole and light-hole band is smaller than $k_B T$ at room temperature, therefore the thermal occupation of both states is similar. In unstrained bulk GaN, the crystal-field split-off band is separated by about 22 meV from the band edge, therefore contributing to a lesser extend to light emission. Therefore light emitted perpendicular to the c-axis in a bulk GaN crystal is partial linearly polarized. This is also the reason for the higher gain of the TE mode when compared to the TM mode in an in-plane (Al,In)GaN laser diode.

The bound states in an InGaN QW result in subbands of the band structure. Anticrossing between these bands results in a modification of the transition matrix elements. For a c-plane QW, however, the in-plane symmetry is conserved. Therefore the polarization characteristics are those of the bulk GaN crystal, i.e. unpolarized light is emitted perpendicular to the QW. Light propagating along the QW plane remains linearly polarized, resulting in a larger TE gain.

In addition, the separation of the crystal-field split-off band is increased in an InGaN QW by strain and reaches several $k_B T$. Therefore the discussion of the electroluminescence and photoluminescence can be limited to heavy hole and light hole bands.
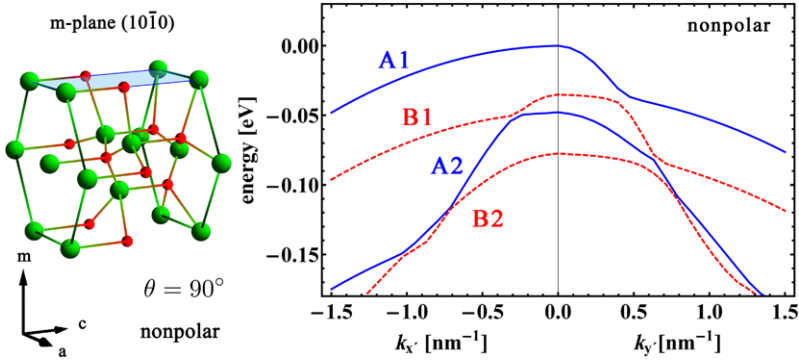
Fig. 5.10 Topmost valence bands of a 3 nm thick nonpolar $In_{0.2}Ga_{0.8}N$ QW at a carrier density of 1×10^{18} cm^{-3} [53]

5.4.1 Light Emission from Nonpolar InGaN QWs

This symmetry is broken in a nonpolar QW. Even in an unstrained thick InGaN or GaN layer, light emitted in a direction perpendicular to the c-axis is polarized because of the symmetry of the crystal structure, as discussed above. In addition to this effect, the band structure is strongly modified in a nonpolar QW by anisotropic strain. This anisotropic strain is a consequence of the different lattice constants in the direction parallel and perpendicular to the c-axis. It modifies the band structure through the deformation potential, which is described in $k \cdot p$ approximation by six parameters D_1 to D_6. The resulting band structure for a 3 nm thick nonpolar $In_{0.2}Ga_{0.8}N$ QW at a carrier density of 1×10^{18} cm^{-3} is shown in Fig. 5.10. The bands are labeled with capital letters A, B for the valence bands and a number m for the respective subband corresponding to the bound states of the QW. For the nonpolar QW the eigenstates are fully polarized at the Γ-point. For the A and B bands the angular momentum eigenfunctions are approximately $|Y'\rangle = |Y\rangle$, $|X'\rangle = |Z\rangle$ respectively.

In contrast to the polar case (cf. Fig. 5.9), the band structure in a nonpolar QW is different in k'_x and k'_y directions, resulting in an anisotropy of the effective mass and density of states [56]. The energy distance between $A1$ and $B1$ band in a nonpolar QW is considerably larger than the splitting between heavy-hole and light-hole band in a polar QW. The band structure in Fig. 5.10 was calculated with standard parameters and strain model, resulting in an $A1$–$B1$ splitting of approximately $k_B T$. Experiments usually report a wider splitting, which is probably caused by a higher strain in the QW. The large energy difference results in a small occupation of the $B1$ band and consequently in a high degree of polarization even at room temperature.

For a typical QW thickness of a few nanometers, the confinement energy is similar to the splitting between A and B subbands. This results in anti-crossing of the $B1$ and $A2$ bands close to the Γ-point, as shown in Fig. 5.10. Transitions from conduction band to the $A2$ band are forbidden by symmetry in the nonpolar QW. However, the $A2$ band contributes to the density of states. Also, the subbands exchange

Fig. 5.11 Polarization
dependent spectra measured
for a nonpolar (10$\bar{1}$0) QW
sample at room temperature.
The *spectra* were measured
with a linear polarizer as
analyzer which was rotated in
5° steps between each
spectrum
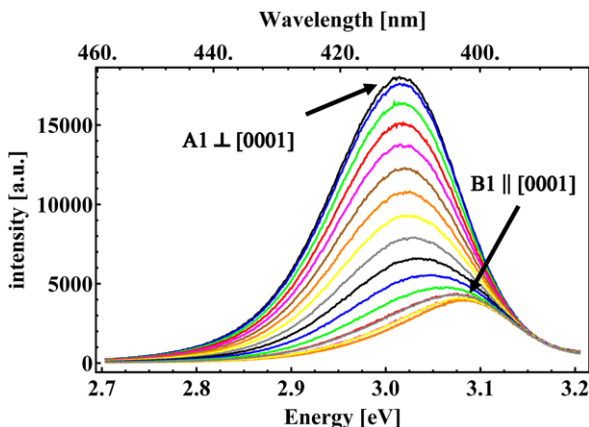
their dominant polarization at these anti-crossings. This means that transitions from
the conduction band to the $B1$ band are x'-polarized at the Γ-point. However, for
$k'_x > 200$ cm^{-1} and for $k'_y > 600$ cm^{-1} the emitted photon will be y'-polarized.

The strain model used in the $k \cdot p$ Hamiltonian is symmetric with respect to a tilt
of the QW towards m- or a-direction. So from the theoretical point of view no differ-
ence between a-plane and m-plane nonpolar QWs would be expected. However, the
microscopic growth conditions are different on the individual lattice plane, as was
discussed above. Therefore a-plane and m-plane QWs grown in a single growth run
may be different not only in terms of indium incorporation and thickness but also
with respect to their microstructure and anisotropic strain within the QW, modifying
both band structure and optical properties.

The measured photoluminescence spectra of a nonpolar (10$\bar{1}$0) QW sample are
shown in Fig. 5.11. The spectra were taken with a linear polarizer, rotated by 5°
between each spectrum. The emitted light is strongly polarized in y'-direction, i.e.
perpendicular to the c-axis. This is the emission from the transition from conduction
band to the $A1$ valence band. With the polarizer rotated by 90°, the emission of the
$B1$ band can be measured. It is much weaker and blue shifted by 78 meV. The
spectrally integrated degree of polarization is $P = 0.68$. This high value at room
temperature is in quantitative agreement with the large energetic spacing of the $A1$
and $B1$ band [57].

5.4.2 Light Emission from Semipolar InGaN QWs

For semipolar orientation, the eigenstates of each of the valence bands are com-
posed of all three angular momentum eigenfunctions $|Y\rangle$, $|X\rangle$, and $|Z\rangle$ even at the
Γ-point. Therefore, partially polarized light is emitted from the transition of the
conduction band to the uppermost valence band already at low temperature. Still,
one of the angular momentum eigenfunctions dominates at the Γ-point. This pre-
dominant component of the hole wave function is indicated for the semipolar planes
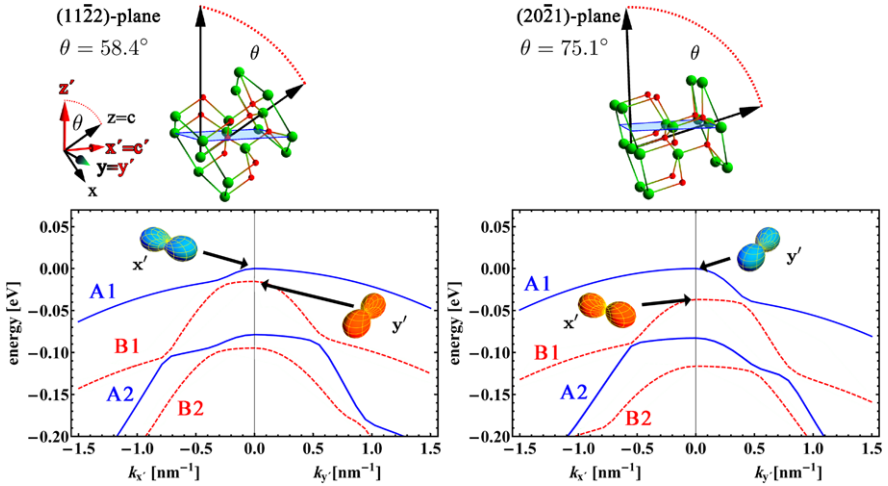
Fig. 5.12 *Top left*: Orientation of the (11$\bar{2}$2) lattice plane (*light blue rectangle*) with respect to the GaN lattice structure. Indicated are the crystal coordinate system x, y, z, and the growth coordinate system x', y', z'. The angle between the surface normal and the crystal c-axis is $\theta = 58°$. *Top right*: Orientation of the (20$\bar{2}$1) crystal plane; this image has to be compared to the one on the *top left*. *Below* are the valence band structures for a 3 nm In$_{0.35}$Ga$_{0.65}$N QW ($n = 10^{18}$ cm^{-3}) of the respective orientation. The angular momentum of the in-plane component of the hole wave function at the Γ-point is indicated for the two topmost valence bands [12]

(11$\bar{2}$2) and (20$\bar{2}$1) in Fig. 5.12. In the example of 3 nm thick In$_{0.35}$Ga$_{0.65}$N QWs, the polarization of the transition to the uppermost valence band changes from parallel to the projection of the c-axis for the (11$\bar{2}$2) plane to perpendicular polarization for the (20$\bar{2}$1) plane. The band ordering depends on the inclination of the lattice plane, and a crossing of the A1 and B1 bands occurs between $\theta = 58°$ and $\theta = 75°$. It is worth noting that the band structure of the (20$\bar{2}$1) plane is very similar to that of the nonpolar plane, indicating little change of the band structure between $\theta = 75°$ and $\theta = 90°$.

For the (11$\bar{2}$2) band structure the energy spacing of the uppermost bands A1 and B1 is relatively small, and anti-crossing occurs close to the Γ-point. Both effects reduce the degree of polarization. The small energy shift results in a similar thermal occupation of both bands. The anti-crossing indicates a strong mixing of the angular momentum eigenfunctions.

Ueda et al. first observed a dependency of the polarization on the indium content of the semipolar (11$\bar{2}$2) In$_x$Ga$_{1-x}$N QW [51]. For an indium content $x \leq 0.22$ the polarization degree was positive, for $x \geq 0.3$ it was negative (see Fig. 5.13). Also the energy difference between the two uppermost valence bands changed accordingly. They termed this behavior "polarization switching" and interpreted it in terms of band ordering. This behavior was confirmed by electroluminescence measurements for (11$\bar{2}$2) LEDs [58] as well as for other lattice planes [49].

The mechanism causing polarization switching is illustrated by the band structure for (11$\bar{2}$2) In$_x$Ga$_{1-x}$N QW with increasing indium content $x = 0.15, 0.25$ and
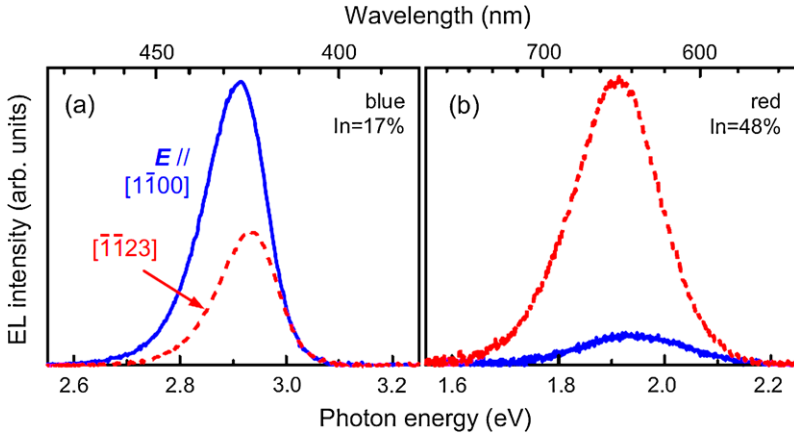
Fig. 5.13 Polarization resolved electroluminescence spectra of a $(11\bar{2}2)$ blue and red LED with indium content of 17 % and 48 %, respectively. The *solid blue* and *broken red line* correspond to y' and x' polarization, respectively. The LED was driven at 0.5 mA [51]

0.35 in Fig. 5.14. As the indium mole fraction increases, the point of anti-crossing moves from $k_{y'}$ direction to $k_{x'}$ direction, changing the symmetry at the Γ-point and shifting the dominant contribution to the hole wave function from $|Y'\rangle$ to $|X'\rangle$ [59]. The polarization P_m of the $A1$ and $B1$ subbands is plotted in Fig. 5.14 below the respective band structure. Going from $k_{x'}$ to $k_{y'}$, the polarization of the $A1$ subband changes from $P_m = +1$ to $P_m = -1$. The k-value where the optical polarization becomes zero is a function of indium content. For an indium content of about 20–25 %, polarization P_m is zero at the Γ-point. This corresponds to the situation where polarization switches from y' to x'. Looking at Fig. 5.14 it becomes obvious that this is a smooth transition, which also depends on the occupation of states in k-space and therefore on carrier density. Quantum confinement has only slight effect on ordering and cannot explain the observed polarization switching [60].

The main cause for the switching of the uppermost valence bands is the anisotropic strain in the semipolar InGaN QW. Ueda et al. pointed out that a high value of the deformation potential D_6 for InN is necessary to explain the observed polarization switching. From his measurements he derived a value $D_6 = -8.8$ eV [51]. To calculate the band structure shown in Fig. 5.14 we used the value $D_6 = -7.1$ eV [59] and the strain model as proposed by Park and Chuang [13]. Other models for the strain in semipolar InGaN QWs have been developed [61, 62], but lead to similar results regarding strain and band structure. The large value of D_6 is however in contradiction to first-principles calculations which predict $D_6 = -3.95$ eV for GaN and $D_6 = -3.02$ eV for InN [63]. Using the standard strain models, these lower values of D_6 would not be able to explain the observed polarization switching. Yan et al. proposed that a partial strain relaxation would cause polarization switching of a semipolar InGaN QW even for these low values of D_6. The morphology of semipolar InGaN QWs supports this idea of an
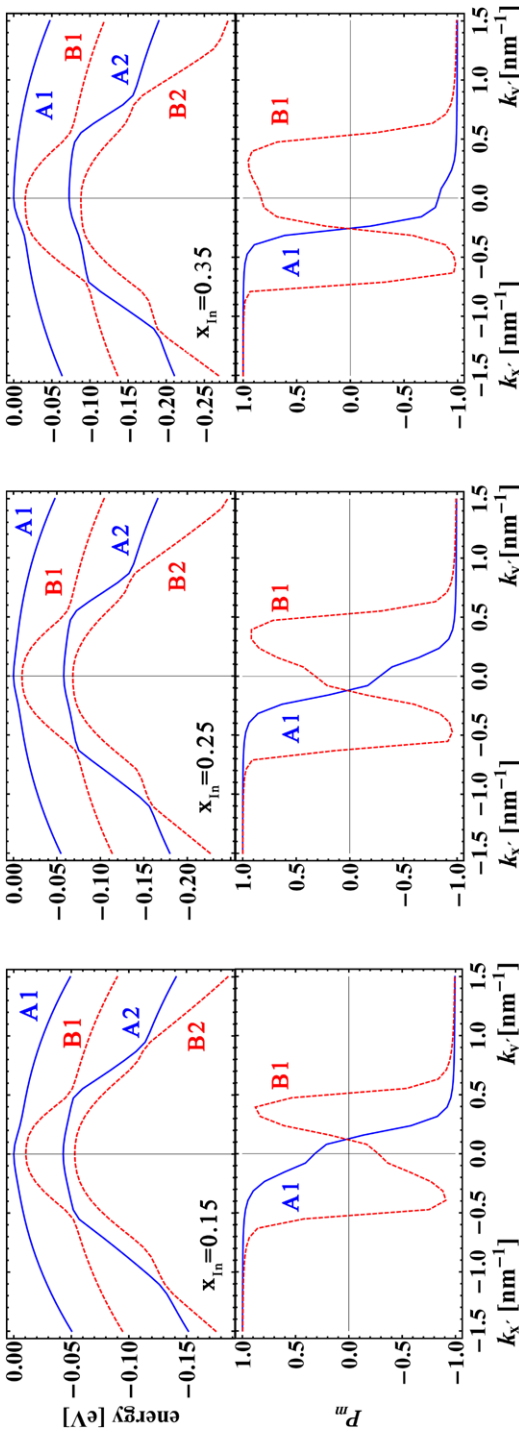
Fig. 5.14 Band structure and polarization of the two topmost valence bands $A1$ (*continuous line*) and $B1$ (*dashed line*) for an $(11\bar{2}2)$ QW ($n = 10^{18}$ cm^{-3}). The indium content is 15 %, 25 %, and 35 %, as indicated. The polarization P_m corresponding to the $A1$ and $B1$ subbands is plotted below the respective band structure plots. Higher order valence bands are drawn for reasons of clarity
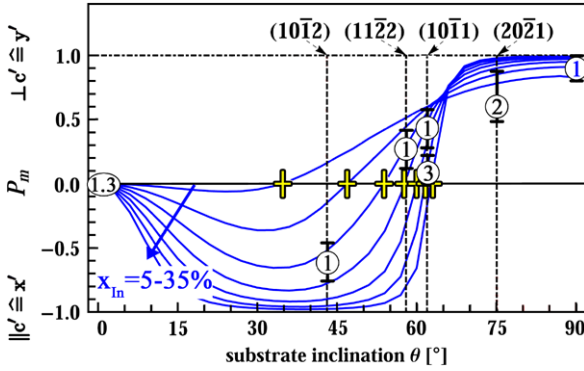
Fig. 5.15 Degree of polarization P_m for transitions from the conduction band to the $A1$ band at the Γ-point. The indium content was varied between 5 % and 35 % in steps of 5 %. QW thickness was 3.5 nm and carrier density was $n = 10^{18}$ cm^{-3}. The numbered markers (*1*) to (*3*) represent measured polarizations at $T = 10$ K. (*1*): photoluminescence from violet light emitting QW structures; (*2*): electroluminescence from violet ($20\bar{2}1$) LED; (*3*): photoluminescence from blue-green light emitting QW structures. The *crosses mark* the critical angle θ_c for the different indium contents [49]

anisotropic strain relaxation [58]. Yet, XRD and TEM measurements demonstrate pseudomorphic growth of an ($11\bar{2}2$) InGaN QW between GaN barriers [62].

Polarization switching is not a phenomenon of the individual ($11\bar{2}2$) lattice plane, but rather a general property of InGaN QWs which can be explained for all lattice orientations and indium compositions in a unified picture [49]. This can be seen from the plot of polarization for transitions to the two uppermost valence bands, as shown in Fig. 5.15. For this figure the band structures for 3.5 nm wide QWs with indium contents between $x = 0.05$ and $x = 0.35$ have been calculated in $k \cdot p$ approximation. The family of curves shows the degree of polarization for the $A1$ subband. For semipolar QW orientation $\theta < 30°$, the polarization is negative, i.e. the polarization is parallel to the projection of the c-axis on the QW, the x'-direction. For large angles $\theta > 70°$ up to nonpolar case $\theta = 90°$ the emission is y' polarized, perpendicular to the c-axis. For intermediate planes, the direction of polarization switches from negative to positive. The switching point therefore depends on the indium content. The yellow crosses mark the critical angle θ_c where polarization switching occurs for a given indium concentration. The maximum critical angle is $\theta_{c,\max} = 69.4°$ for InN and the chosen set of parameters [49]. For angles smaller than $\theta_{c,\max}$ the degree of polarization P_m decreases with increasing indium content, for angles larger than $\theta_{c,\max}$ it increases.

Electroluminescence and photoluminescence measurements of LED structures on different polar, semipolar and nonpolar orientations are indicated as markers in Fig. 5.15. Also the polarization switching as observed by Ueda et al. [51] and the decreasing P_m with increasing indium content observed by Masui et al. for ($11\bar{2}2$) LEDs [58] are in agreement with this unified picture of polarization switching. However, the values measured for the ($20\bar{2}1$) orientation by several groups [49, 58, 64]
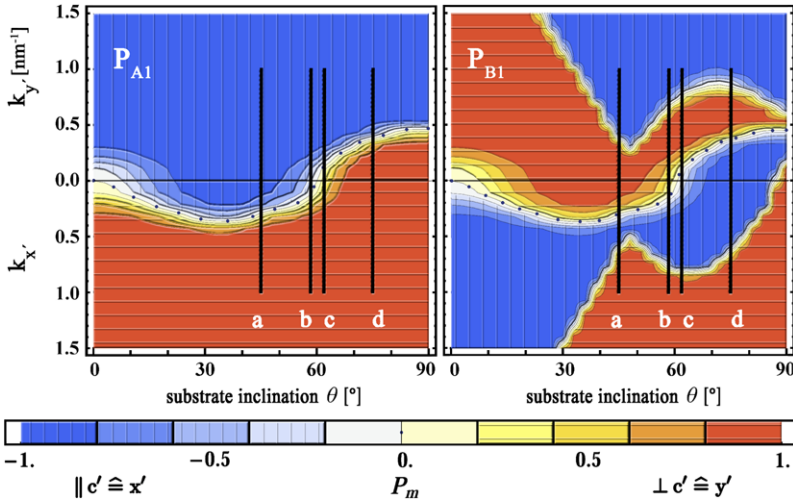
Fig. 5.16 Degree of polarization P_m for transitions from the conduction band to the $A1$ (*left*) and the $B1$ (*right*) valence bands as function of substrate inclination θ in k_x'–k_y' space for a 3.5 nm wide In$_{0.3}$Ga$_{0.7}$N QW. The *dotted line* represents unpolarized emission ($P_m = 0$). The *vertical lines a* to *d* mark the lattice planes $(10\bar{1}2)$, $(11\bar{2}2)$, $(10\bar{1}1)$, and $(20\bar{2}1)$ [49] (Color figure online)

are consistently lower than predicted by Fig. 5.15. Still, the increase of the polarization ratio with increasing indium content was confirmed by experiments for the $(11\bar{2}2)$ plane [58].

Zhao et al. studied the polarization ratio of semipolar LEDs in the blue and green wavelength region grown on the $(20\bar{2}1)$- and $(20\bar{2}\bar{1})$-facet. Both of these planes are tilted by $\pm15°$ from the m-plane and differ in their surface configuration [65]. The experimental work showed that the LEDs on $(20\bar{2}\bar{1})$ have a higher degree of polarization and a larger band splitting than the $(20\bar{2}1)$-devices. This was attributed to indium interdiffusion in $(20\bar{2}1)$ QWs and consequently a modification of the valence sub band structure.

As pointed out above, the polarization is not only important at the Γ-point, but has to be considered in k-space, as band mixing occurs in semipolar QWs in the valence band close to the Γ-point. Therefore the degree of polarization is plotted in Fig. 5.16 for the $A1$ and $B1$ band of a 3.5 nm thick In$_{0.3}$Ga$_{0.7}$N QW as function of substrate inclination θ and hole momentum $k_{x'}$ and $k_{y'}$. Horizontal and vertical lines (red and blue shading, color online) indicate the sign of P_m. In the red shaded area the polarization is perpendicular to the c-axis. For nonpolar orientation ($\theta = 90°$) the range of positive P_m extends from the Γ-point both in $k_{x'}$ and $k_{y'}$ direction. The dotted line and white shading indicates the unpolarized case $P_m = 0$. The polar QW with $\theta = 0°$ is unpolarized. The dotted line crosses the $k = 0$ axis again between the black vertical lines marking the lattice planes of $(11\bar{2}2)$ and $(10\bar{1}1)$ orientation. This is the signature of polarization switching of the $(11\bar{2}2)$ QW at the indium composition of about $x = 0.20$–0.25. The polarization map of the $B1$ band is complementary

to that of the $A1$ band in the center of the Brillouin zone. At higher k-values mixing with higher bands modifies the polarization of the $B1$ band.

The map of P_m shows clearly that polarization switching is not an abrupt phenomenon, but rather a transition of a QW through a range of low degree of polarization as function of either inclination θ or indium content. The map also tells that high degrees of polarization can only be expected far from regions of polarization switching. This explains why large values of P_m are observed for nonpolar QWs. The experimental fact that $(20\bar{2}1)$ QWs up to now do not show the expected high polarization ratios may either suggest some strain relaxation mechanisms or faceting during the growth of these QWs, or that the deformation potential parameters are still not correct.

Masui et al. observed a decrease of polarization with current injection in a nonpolar LED [66]. They suggested that the filling of states according to Fermi statistics is responsible for this reduction of P_m. With increasing carrier density the quasi-Fermi level comes close to the top of the valence band, resulting in a similar occupation probability of the $A1$ and $B1$ bands. Because of the complementary polarization of both bands, the ratio of polarization decreases. Another effect is that states further away from the Γ-point are occupied, which also decreases the overall polarization ratio (see Fig. 5.16). It should be noted that Kyono et al. did not observe a decreasing polarization ratio with increasing current for a semipolar QW [64]. However, for a nonpolar LED structure the decrease of the polarization ratio with increasing carrier density was confirmed by Schade et al. [57] and could be explained quantitatively by state filling using the Fermi-Dirac distribution.

5.5 Performance Characteristics of Non- and Semipolar InGaN QW Light Emitting Diodes

5.5.1 Wavelength Shift

Light emitting diodes based on the (Al,In,Ga)N material system show a strong dependency of the emission wavelength λ on the carrier concentration n in the quantum well (QW) active region and hence on the injected current I. This behaviour is undesired since this means a dependency of spectral characteristic on the light output power, which is unacceptable for many applications such as lighting, sensing and consumer electronics systems. There are four main effects that influence the emission wavelength.

- Due to ohmic heating at high operation currents, the bandgap E_g decreases and hence the wavelength is red-shifted to longer values. This effect is present in all semiconductor materials and cannot be avoided completely. Thermal management, heat sinking and the reduction of the current density by increasing the

device area can reduce the effect. The dependency of the bandgap on the temperature T is phenomenologically described by the Varshni model:

$$E_g(T) = E_g(T = 0) - \frac{\alpha T^2}{T + \beta} \tag{5.13}$$

where α and β are the Varshni parameters (e.g. [67]).

- At high injection currents I the increased carrier density n leads to the so called band-filling where lower energy states are filled and the emission moves to higher excited states in the quantum well. The consequence is a blue-shift in emission wavelength. In order to avoid this the volume of the active region can be increased, thus reducing the carrier density n. This can be done either by an increase in the number of quantum wells or by an increase of the thickness and volume of each quantum well. The reduction of the carrier density also reduces the droop (see Sect. 5.5.2).
- If the active region exhibits fluctuations in the thickness of the quantum well or the indium content, the emission spectra broaden. Especially the indium fluctuations are an important technological challenge and the strength of the fluctuation increases with the total amount of indium in the active region, making the growth of green and yellow emitters more challenging. Upon carrier injection the so-called band-tail states are filled first since they have the lowest band gap energy. When the injection current increases, emission from material regions with lower indium content and hence shorter emission wavelength occurs, shifting the emission wavelength to the blue.
- If polarization fields are present in the active region, the emission wavelength is originally red-shifted by the QCSE (see Sect. 5.2). This effect is reduced in semipolar emitters and vanishes for nonpolar emitters. If carriers are injected into the active region then the polarization charges at the interfaces of the quantum well are partially screened by the free carriers, compensating the initial red-shift. The consequence is a strong blue shift of the emission wavelength which is stronger in c-plane devices than in semipolar or nonpolar devices. This effect therefore dominates the emission characteristics in c-plane devices while it is not present in nonpolar devices and reduced in semipolar devices.

All of the above mentioned effects occur at the same time. The design of the active region, the crystal orientation and the material quality determine which effect is the dominating part. If the crystal quality is low and the indium fluctuations are large, the filling of band-tail states dominates.

In Fig 5.17 a spectrum of an m-plane MQW-LED in cw-mode is shown and only a very small shift is observed. The blueshift in pulsed mode is compared to an LED with the same epitaxial structure on (0001) c-plane. At low current densities j the initial shift of the m-plane LED is stronger than for the c-plane LED which is attributed to filling of states caused by indium fluctuations. When j is increased, the shift of the m-plane LED is smaller than that of the c-plane LED, since here no polarization fields occur and hence the only cause is band filling, while strong polarization fields in the c-plane LED are partially screened, causing a stronger blue shift.
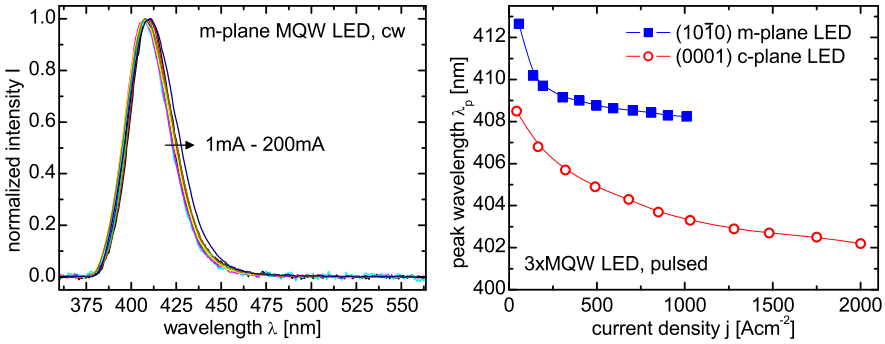
Fig. 5.17 The comparison of current dependent emission wavelength in nonpolar and polar MQW LEDs shows a strongly reduced blue shift in nonpolar LEDs due to the absence of polarization fields

Similar results have been found by Kuokstis et al. [68]. Schmidt et al. reported a very small wavelength shift of less than 1 nm for a 407 nm *m*-plane LED on bulk GaN in the range of 1 mA to 20 mA [69]. LEDs on $(20\overline{2}1)$ GaN that have been reported by Zhao et al. also show a very small wavelength shift, caused by the strongly reduced polarization fields [70].

5.5.2 Droop

The term "droop" describes the fact that the external quantum efficiency *EQE* which is defined as the ratio of extracted photons divided by the injected carriers is not a constant, but is influenced by the carrier concentration *n* in the active region and therefore the injected current *I*. The easiest way to describe this phenomenon is to treat the recombination mechanism by a third order polynomial model, the ABC-model. Since the different contributions to the overall recombination rate are proportional to *n* (non-radiative Shockley Read Hall SRH recombination, e.g. at defects), to n^2 (radiative recombination) and to n^3 (Auger-like recombination), the overall efficiency can be described as

$$EQE = \frac{r_{\mathrm{rad}}}{r_{\mathrm{nrad}} + r_{\mathrm{rad}} + r_{\mathrm{Aug}}} = \frac{Bn^2}{An + Bn^2 + Cn^3} \tag{5.14}$$

A, *B* and *C* are the corresponding recombination parameters. The consequence of this droop model is that the maximum EQE is found at a carrier concentration $n_{\mathrm{max}} = \sqrt{A/C}$. Beyond this density, the efficiency decreases, making it favorable to drive the LED at low currents. There has been a strong debate regarding the actual cause for the droop [71], since theory predicts a *C*-coefficient that is so low for nitrides, that droop shouldn't occur at normal carrier concentrations. The fact that many groups observed strong droop was explained by higher than expected *C*-coefficients [72], electron leakage [73], non-constant *ABC*-coefficients

[74], density-activated defect recombination (DADR) [75] or indirect Auger recombination which was mediated by electron-phonon coupling and alloy scattering [76].

In order to reduce the unfavorable droop, the carrier concentration in the active region of the LED should be small, and as close as possible to $n_{\max}$. This however contradicts the demand for high-power devices and requires an increase in the device area, which in turn increases the device costs. In semipolar and nonpolar LEDs the droop in principle should be smaller than in c-plane devices due to the absence of polarization fields. Since the QCSE is reduced or eliminated, the radiative recombination rate and the B-coefficient are much larger, increasing the device efficiency.

Recent studies on the droop give hints for fundamental advantages of QWs with reduced fields [77, 78]. According to them, Auger recombination has the same linear connection with the absolute squared overlap integral of the wave functions in the conduction and the valence bands as the radiative recombination has. Consequently, there is a lower charge carrier density in a semipolar compared to a similar polar QW for a given current density. The droop maximum is shifted effectively to a higher current density.

In c-plane devices the quantum wells are very thin in order to limit the effect of spatial separation of the electron- and hole wavefunctions and thus the reduction in oscillator strength. This limitation in the design of the active region is lifted for non- and semipolar LEDs and hence the quantum wells can be made much thicker, thus reducing the carrier concentration n at a given injection current I.

Pan and Zhao et al. studied the droop in blue LEDs on $(20\overline{2}1)$ GaN for multiple 3 nm thin quantum wells (MQW) [70] and one 12 nm thick single quantum well (SQW) [79]. The EQE reduced from 52.6 % at 35 A cm^{-2} current density to 45.3 % at 200 A cm^{-2} for the MQW LED which is very low compared to c-plane devices [80, 81]. Similar results were found for the SQW-LEDs.

5.5.3 Polarization and Light Extraction

It is important to note that the optical polarization of the emitted light is a crucial factor for the light extraction. Only photons which are emitted within a narrow angle ϑ with respect to the surface normal can escape due to total internal reflection. The refractive index n_r of the GaN and the surrounding air together define the escape cone. Only photons with a $\vec{k}$-vector within this cone can escape. Since $\vec{E} \perp \vec{k}$, this means that mostly TE-polarized light is emitted. A change in the optical polarization for semipolar and nonpolar emitters, as discussed in Sect. 5.4, therefore strongly affects the light extraction efficiency.

There has been much work on the increase in light extraction efficiency by surface roughening techniques and attempts to increase the angle ϑ of the escape cone. The mechanisms behind this is the random and multiple reflection of photons at the rough surface which changes the propagation direction and hence allows the photons to escape from the semiconductor. The drawback of this method is that the

linear polarization, which is desirable for many applications such as liquid crystal display (LCD) back lighting, is lost during the scattering process.

In order to increase the extraction rate and maintain the polarization, Matioli et al. employed photonic crystals (PhC) tailored to the wavelength and dominant polarization direction of *m*-plane oriented blue LEDs [82]. By using one-dimensional air-gap PhCs, the extraction rate was significantly increased compared to planar devices.

Regarding optical properties, one should also be aware that InN, GaN, and AlN are birefringent. This is of particular importance to semipolar (Al,In)GaN laser diodes, where waveguide modes have to be categorized as TE/TM or ordinary/extraordinary, depending on the waveguide orientation [53, 83–86]. In an LED the effect of birefringence is of lesser importance, as light usually passes only through a thin layer of semiconductor. Still, one has to be aware that the thickness of a $\lambda/4$ plate at 470 nm made of *a*-plane GaN would be of about 4.5 µm which is of the order of the *n*-GaN layer of a typical thin-film LED. For a standard planar nonpolar LED the polarization is perpendicular to the *c*-axis which is the optical axis of the crystal. Therefore the layer does not convert the emitted linearly polarized light to circular polarized light. However, in any structure with out-of-plane geometry or photonic structures, one needs to consider birefringence also in GaN based LEDs.

5.5.4 3D-Semipolar LEDs on c-Plane Sapphire

One of the main obstacles in the realization of nonpolar and semipolar LEDs is the challenge of supplying large area and cheap substrates of high quality. While many groups experimented with heteroepitaxial growth on various crystal planes on sapphire, spinel or silicon (see Sect. 5.3.1), another approach is to use high quality semipolar surfaces on conventional planar *c*-plane GaN-on-sapphire substrates. Selective area growth of stripes along the [11$\bar{2}$0]-direction results in triangular pyramidal stripes with {10$\bar{1}$1} side facets. The quality of these samples exceeds the one known from other methods since the growth starts from a high quality *c*-plane GaN template [44]. Although the surface of the overgrown wafer is not planar anymore, devices have been processed using this technique (layout depicted in Fig. 5.18). First LEDs with continuous wave emission at 425 nm with 0.7 mW at 20 mA have been shown in 2008 by Wunderer et al. [87]. In 2010 Scholz et al. demonstrated 495 nm LEDs on these stripes with an output power of 0.24 mW at 20 mA driving current [88].

5.5.5 State-of-the-Art of Non- and Semipolar Blue, Green, and White LEDs

During the past ten years several groups worldwide have investigated the properties of nonpolar and semipolar LEDs, and tremendous progress has been achieved. In

Fig. 5.18 The selective growth of triangular stripes with semipolar side facets allows the realization of semipolar LEDs on high quality (0001) GaN templates [89]
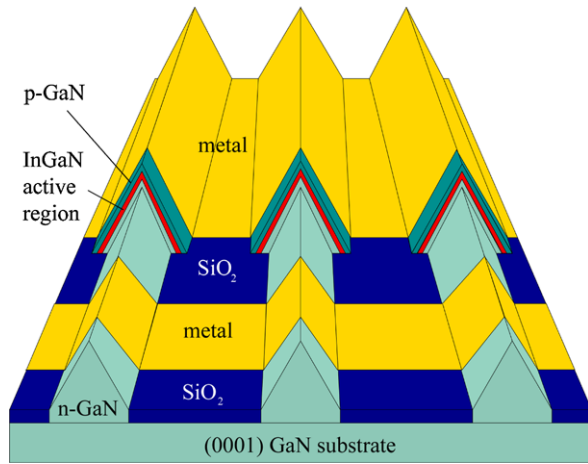
the beginning, the limiting factor was the availability of large area and high quality substrates, and hence most LEDs were grown heteroepitaxially on foreign substrates with a high density of threading dislocations (TDD) and basal plane stacking faults (BSF). Among the most favored substrates were sapphire overgrown by HVPE-GaN, and in 2012 Jung et al. demonstrated a violet LED on $(11\bar{2}0)$-GaN with 0.24 mW output power at a dc current of 20 mA [90]. By using epitaxial lateral overgrowth (ELO or LEO) for defect reduction, Chakraborty et al. realized a blue nonpolar LED on a sapphire substrate with 7.5 mW emission power [91]. Furthermore, LEDs with 439 nm emission wavelength were demonstrated on semipolar $(10\bar{1}3)$- and $(10\bar{1}1)$-GaN orientations grown on (100) and (110) spinel $MgAl_2O_4$ substrates [92]. Due to its large available size, low cost and the compatibility to existing processing procedures silicon has attracted high interest. In 2008 Hikosaka et al. realized LEDs with blue-violet emission on patterned Si on the semipolar $(11\bar{2}2)$ and $(10\bar{1}1)$ orientation [93].

Due to the large defect densities present in all heteroepitaxially grown LEDs, many groups focused on homoepitaxial growth on quasi-bulk substrates cut from HVPE-grown boules. Although the size and price of these substrates is a limiting factor, LEDs with emission from the violet, blue and green up the yellow wavelength region have been demonstrated. For a long time the most commonly used orientations were the nonpolar a- and m-planes and the semipolar $(11\bar{2}2)$-plane. In 2009 impressive progress was shown based on the newly studied $(20\bar{2}1)$-plane [112, 113], and many groups have explored LEDs on this plane since then [46, 49, 65, 114–118]. Other planes such as the $(30\bar{3}1)$-plane have also been investigated for the use in LEDs and laser diodes [119].

In the blue region external quantum efficiencies (EQE) of more than 50 % have been reported on the $(10\bar{1}1)$-plane [97] and on the $(20\bar{2}1)$-plane by Zhao et al. [70]. Green emitters with high EQE values were realized on the semipolar $(20\bar{2}1)$-plane with 516 and 552 nm wavelength and 19.1 and 11.6 % external quantum efficiency,

Table 5.2 Overview over performance of nonpolar and semipolar LEDs at 20 mA dc current

Growth plane	Substrate	Wavelength [nm]	Power [mW]	EQE [%]	Reference
$(11\bar{2}2)$	GaN	425	1.76	3.0	[94]
$(11\bar{2}2)$	GaN	525	1.91	4.1	[94]
$(11\bar{2}2)$	GaN	585	0.54	1.3	[94]
$(11\bar{2}2)$	GaN	516	5.0	10.5	[95]
$(11\bar{2}2)$	GaN	562.7	5.9 (pulsed)	13.4	[96]
$(11\bar{2}2)$	Patterned Si	439			[93]
$(10\bar{1}1)$	Patterned Si	419			[93]
$(10\bar{1}\bar{1})$	GaN	455	31.1	54.7	[97]
$(10\bar{1}\bar{1})$	GaN	444	16.21 (pulsed)	29	[98]
$(20\bar{2}1)$	GaN	516.3	9.2	19.1	[99]
$(20\bar{2}1)$	GaN	552.3	5.2	11.6	[99]
$(11\bar{2}0)$	PLOG-sapphire	460	7.5		[90]
$(11\bar{2}0)$	HVPE-LEO	413	0.24		[91]
$(10\bar{1}0)$	GaN	435	1.79	3.1	[100]
$(10\bar{1}0)$	GaN	407	23.7	38.9	[69]
$(10\bar{1}0)$	GaN	457	0.24		[101]
$(10\bar{1}0)$	GaN	452	0.6	1.09	[102]
$(10\bar{1}0)$	GaN	402	28 (pulsed)	45.4	[103]
$(20\bar{2}\bar{1})$	GaN	423	30.6	52	[70]
$(10\bar{1}\bar{1})$	MgAl$_2$O$_4$	439	0.01	0.02	[92]
$(10\bar{1}3)$	MgAl$_2$O$_4$	439	0.19	0.35	[92]

respectively [99]. Furthermore, on the $(11\bar{2}2)$-plane a 562.7 nm LED was shown with 13.4 % EQE under pulsed conditions [96].

Figure 5.19 shows the emission power P and the external quantum efficiency (EQE) of semipolar and nonpolar LEDs. For blue and green emitters *EQE*-values of more than 50 % and 20 % have been achieved, respectively.

5.5.6 Towards Yellow LEDs and Beyond

The initial driving force behind the increase in wavelength of nitride-based light emitting diodes was the aim to close the "green gap", thus making the realization of emitters for the green wavelength region the most profitable and also most challenging goal. Since the InGaN-system covers the whole visible spectrum from UV to IR, even longer wavelength such as yellow and orange seem possible. This is even more challenging due to the higher indium content of the longer wavelength active region resulting in problems such as material decomposition, indium inhomogeneities and strain due to the increased lattice mismatch. The increased strain also increases the
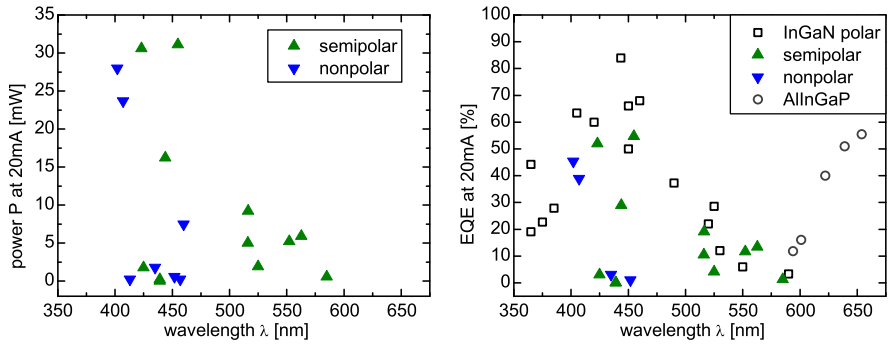
Fig. 5.19 Emission power *P* (*left*) and external quantum efficiency *EQE* (*right*) of polar, nonpolar and semipolar InGaN based LEDs and AlInGaP showing the "green gap". For reference see Table 5.2 and [80, 104–111]

polarization fields, making semipolar and nonpolar crystal orientations the natural choice for this application.

The interest for the realization of GaN-based yellow light emitters is not as strong as for green, though, since the AlInGaP-system covers this wavelength range and the wavelength is on the long-wavelength edge of the "green gap'". In 2008 Sato et al. reported on yellow semipolar (11$\bar{2}$2) LEDs and compared them to AlInGaP-based devices [96]. They showed that for nitride-based LEDs the dependency of output power and EQE onto the ambient temperature was lower than for the phosphide-based devices. This behaviour was attributed to carrier overflow due to the smaller energy offset between quantum well and barriers in the AlInGaP-LEDs.

5.6 Summary and Outlook

Despite the short time period in which the growth of light emitters on non- and semipolar surfaces has been explored, non- and semipolar InGaN QW LEDs already show great promise. This is impressively demonstrated by a number of performance indicators. For example, external quantum efficiencies of blue, green and yellow non- and semipolar light emitters are close or already exceeding those of conventional *c*-plane InGaN LEDs. Other important parameters like droop and wavelength stability with drive current are also showing great advances compared to LEDs grown on polar surfaces. However, many of these improvements and records have been realized on the relatively costly bulk GaN substrates. Therefore the questions remains whether non- and semipolar InGaN QW LEDs can also be produced cost-effectively, e.g. on sapphire or silicon substrates. Some of the possible approaches have been outlined in the previous chapter and indicate a number of pathways to realize low-cost and large volume production of high-efficiency non- and semipolar LEDs. Young start-up companies, like Soraa Inc. (Fremont, USA) are already trying to seize these opportunities and have introduced the first LED lamps based on

non- and semipolar technology that generate more than 2300 candela of white light with only 12 Watt of electric input power [120]. Of course this is just the start. Only time will tell whether non- and semipolar LEDs will have a lasting impact on future lighting technology. Considering the short time span in which non- and semipolar light emitters have been explored and the astonishing advances that have already been demonstrated, non- and semipolar LEDs are certainly serious contenders.

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Chapter 6
Active Region Part A. Internal Quantum Efficiency in LEDs

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In this chapter, we present different techniques used to assess the internal quantum efficiency (IQE) in light emitting diodes (LEDs). The commonly used technique based on temperature-dependent photoluminescence relies in strong assumptions which are discussed in this chapter. We introduce an alternative method to determine IQE based on electroluminescence, in which the external quantum efficiency (EQE) is measured from a single facet of the LED, where the light emission can be calculated with good accuracy. The IQE is ultimately obtained from the ratio of the EQE and the calculated light extraction efficiency. We develop an optical model of the light emission in a multilayered LED structure, from which we derive and validate an approximate model to easily calculate the extraction efficiency through the top facet of any LED structure. We address the various assumptions made to calculate the direct emission model through a single facet and evaluate the effect of photon recycling in the quantum wells. We also calculate the sensitivity of the model to the LED parameters and surface roughness. Finally, we apply this technique to calculate the IQE of both a state-of-the-art and a low performance GaN-based LEDs, highlighting the particular features in each structure.

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6.1 Introduction

The development of the next generation of high efficiency light emitting diodes (LEDs) for solid state lighting requires a quantitative determination of intrinsic device parameters to further performance. A common metric of optoelectronic devices is their output power emitted externally to the device (P_{out}) measured in an integrating sphere. From there, two quantities define the efficiency of the LEDs: the wall-plug efficiency (WPE) η_{wp} i.e., the ratio of electrical input power to optical output power, and the external quantum efficiency (EQE) η_{EQE}, ratio between the number of electrically injected carriers and externally observed photons. The WPE is related to EQE by the voltage drop V at the device due to the diode forward voltage and series resistance, as $\eta_{\text{wp}} = P_{\text{out}}/VI$, where I is the injected current. To optimize it for a given EQE, one requires structures with low contact resistance and high conductivity materials, as well as efficient heat sink to maintain high performance under all operating conditions [1]. η_{EQE} is easily assessed by

$$\eta_{\text{EQE}} = P_{\text{out}}/(\hbar\omega)(I/q)$$

but it only reveals the combination of non-easily separable key parameters, such as carrier injection efficiency η_{inj}, internal quantum efficiency (IQE) η_{IQE}, and light extraction efficiency (LEE) η_{extr} of the LED structure, ratio between the externally emitted photons and the internally generated photons in the active region.

What determines these efficiencies rely on intrinsic and extrinsic properties of materials and architecture of the LEDs. η_{extr} is mainly determined by the LED architecture (such as chip shaping [2], use of patterned substrates [3], photonic crystals [4], surface roughening [5, 6], etc.), with however some dependence on materials parameters (materials absorption, reflection, etc.) (see the chapter by Lalau Keraly et al. in this book); η_{IQE} is mainly connected to the quality of the active layer which is determined by growth conditions such as growth temperature, pressure, quality and flow of precursors, impurity incorporation, as well as growth technique and reactor used, and by the choice of substrate, which affects the crystalline quality, doping profile, defect density of the material, uniformity and surface morphology. It is therefore of great importance to have a precise evaluation of the materials quality as given by η_{IQE} to assess growth quality. This quantity also serves to separately evaluate η_{extr} from η_{EQE} and compare on an absolute scale the various techniques to enhance light extraction.

In addition, the IQE defined by the ratio between the electrically injected carriers and the internally emitted photons, is itself the product of the electron injection efficiency η_{inj}, ratio of carriers injected in the device to those reaching the light emitting regions (usually quantum wells (QWs)) and of the radiative efficiency, η_{rad}, ratio of injected electron-hole (e-h) pairs that recombine radiatively to generate photons and the total number of injected e-h pairs. η_{inj} is mostly determined by the LED heterostructure design, with carrier overflow and QWs uniform injection being examples of phenomena to be controlled.

Equations (6.1) and (6.2) express the relations between these various quantities.

$$P_{\text{out}} = \frac{\hbar\omega}{q} I \eta_{\text{inj}}\eta_{\text{rad}}\eta_{\text{extr}} \tag{6.1}$$

$$\eta_{\text{IQE}} = \eta_{\text{inj}}\eta_{\text{rad}} \tag{6.2}$$

where $\hbar\omega$ is the photon energy and q is the electron charge.

The variations in the measured IQE are expected to be mainly due to η_{rad}, which is directly linked to materials quality, allowing to assess growth quality in a routine fashion, to optimize epitaxial design and to check run to run reproducibility. From Eq. (6.1), the IQE in an LED structure can be written as

$$\eta_{\text{IQE}} = \frac{P_{\text{out}}}{\frac{\hbar\omega}{q} I \eta_{\text{extr}}} = \frac{\eta_{\text{EQE}}}{\eta_{\text{extr}}} \tag{6.3}$$

In the following sections, we present two methods to determine IQE, the first based on photoluminescence (PL) and the second on electroluminescence (EL). To obtain η_{IQE} from the η_{EQE} measurement using Eq. (6.3) a good knowledge of η_{extr} is required, which is not known in a complex LED structure usually containing light extraction features which complicates the precise determination of η_{extr}. The PL-based method circumvents this issue by considering a reference point at low temperature (LT) where η_{IQE} is assumed to be unity, and assuming that η_{extr} is the same at room temperature (RT), any variation between LT and RT is due to the change in η_{IQE}. Thus the η_{IQE} at RT is just the ratio of the η_{EQE} measurements at RT and LT. The EL-based method considers a LED structure with a well defined geometry and without light extractors, such that η_{extr} is known a priori. Then the η_{IQE} is determined from η_{EQE} as $\eta_{\text{EQE}}/\eta_{\text{extr}}$.

6.2 Assessment of IQE from Photoluminescence Measurements

The IQE is widely estimated by temperature-dependent photoluminescence (PL), in which the intensity of the emitted light at a certain range of PL excitation is measured at low temperature (LT), usually below $\sim$10 K and at room temperature (RT). Under the assumption that the IQE at LT is 100 % and that η_{extr} does not change with T, the IQE is estimated from the ratio of the peak PL intensities I_{RT} at RT and I_{LT} at LT.

$$\eta_{\text{IQE}} = \frac{I_{\text{RT}}}{I_{\text{LT}}} \tag{6.4}$$

The assumption that IQE is unity at low temperatures is often made regardless the excitation power density. However the IQE and the PL intensity are themselves dependent on excitation power density [7, 8] at all temperatures, being dominated by Shockley-Read-Hall (SRH), or non-radiative recombination mechanisms at low

Fig. 6.1 [After Watanabe et al. [7]] Relative PL emission normalized by the peak intensity at LT as a function of excitation power density for two different devices at 8 K and 300 K. Note that the PL emission is largely dependent on excitation density and the peak of PL emission occurs at different excitation power densities at LT and RT
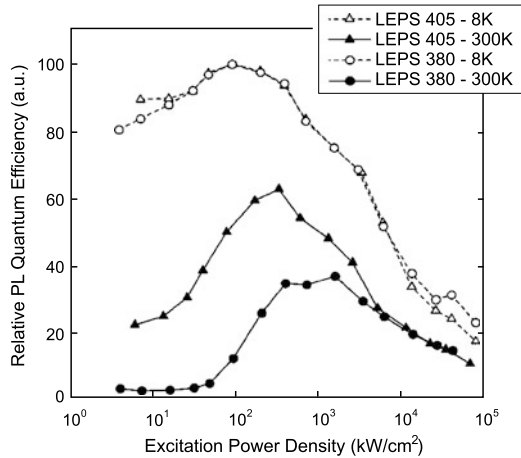
excitation density and by Auger recombinations at high excitations [9] (see Fig. 6.1 from Ref. [7]).

A careful procedure is therefore required for estimating the IQE using the temperature-dependent PL method. A large range of excitation power densities, over several orders of magnitude, is needed to correctly determine the peak IQE at low temperature, otherwise the base level that normalizes the low temperature PL intensity at 100 % IQE is incorrect. This procedure is often times not mentioned in the literature [10].

The assumption of unity IQE considers that non-radiative recombination mechanisms are eliminated at low temperatures, which would be justified if only thermally-activated defect-induced recombinations were the dominant non-radiative recombination mechanism. However, Fig. 6.1 shows that non-radiative mechanisms are still present at low excitation density even at low temperatures. The assumption that the peak PL at LT corresponds to 100 % IQE lacks fundamental support.

Another issue is that the peak PL emission occurs at different excitation power densities at LT and RT, and hence to different carrier density in the active region, which makes it difficult to compare these quantities since the competition between SRH, radiative and auger recombinations depend on carrier density.

The IQE determined from PL-based methods also assumes similar carrier injection to the active region compared to electroluminescence. It neglects the effects of applied bias in the internal electric field in the QWs, and also neglects the different carrier injection in individual wells of a MQW structure under PL and EL excitation. In a MQW structure under PL excitation, electron-hole pairs are generated in all QWs, whereas under EL excitation, the carrier distribution in the MQW is, in general, inhomogeneous and dependent on temperature and current density [11, 12]. Laubsch et al. demonstrated a similar behavior of the emission under PL and EL excitation for a single QW structure [8], however the majority of the LEDs fabricated today are based on MQW structures.

These strong assumptions used in the temperature-dependent PL method justify the search for other techniques based on similar operation condition and injection mechanisms as the operating LED.

6.3 Principle of IQE Assessment from Electroluminescence Measurements

As given by Eq. (6.3), P_{out}, or equivalently η_{EQE}, can be measured at a given current, and η_{extr} can be precisely calculated for a simple enough LED geometry. This section is divided in two parts, the first covering the results of calculations of the η_{extr} in a simple LED geometry, with details to be found in Ref. [13] or in the Appendix A of this chapter, and the second covering the fabrication details of a simple LED structure adapted for this method [14], and experimental results.

The small extraction efficiency of planar LED structures originates from the fact that only a small fraction of the photons generated within the active region is directly emitted out of the device at their encounter with interfaces, which corresponds to the fraction of light emitted with its wavevector within the air cone. The majority of the light is emitted outside the air cone and is reflected back in the device. A fraction of such photons is dissipated by material defects, free carriers in doped regions, and absorbing materials (such as metal contacts). Another fraction is absorbed by the active material (light emitting material) in the device which can be re-emitted and may re-attempt to escape the LED structure. This mechanism, called photon recycling is an efficient way to significantly increase extraction efficiency in the LED when both η_{IQE} and the fraction of light recycled are large. This is treated in more details in Sect. 6.5 of this chapter. Photons can also escape the LED through the light cone of any exposed facets after a few bounces within the structure. This mechanism can be very efficient when dissipation mechanisms, including QW reabsorption, are very small along the sometimes long trajectories required to bounce back in the escape angle: suffices to remark that in GaAs LEDs with 2 % direct extraction, tens of bounces are required to extract most of the light [1]. It is the variety of fates for emitted photons that makes the extraction efficiency difficult to quantify, and highly dependent on the LED chip architecture. The light extraction efficiency based on the use of extracting features such as truncated pyramids, roughened surfaces, patterned sapphire substrates, surface photonic crystals is in a way difficult to quantify, as it will depend on the exact description of the LED geometry, including its sidewalls, and on the loss mechanisms within the LED for the various materials and regions. Detailed simulations rely on ray tracing techniques to explore the different trajectories (see the chapter by Lalau Keraly et al. in this book), or on solving full Maxwell's equations by finite-difference time-domain (FDTD) or plane wave expansion [15].

The way to circumvent the difficult analysis of η_{extr} in commonly used LED structures is to rely on a simple LED geometry, with a computable direct light ex-
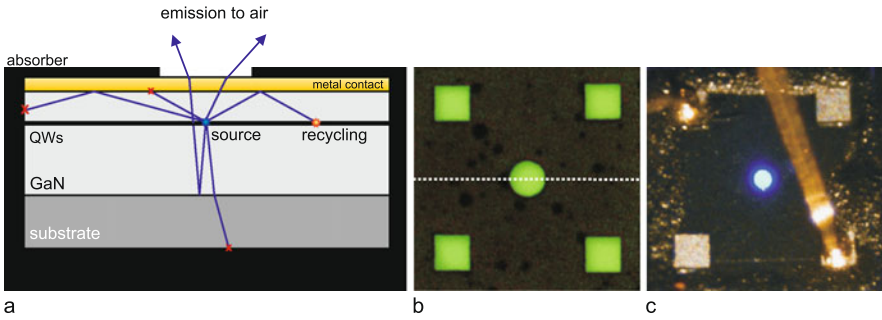
Fig. 6.2 (**a**) Schematic of a simple geometry LED (contact pads are not shown) surrounded by light absorbing material, along with possible trajectories of the light emitted by the QWs. This view corresponds to the cross-section view at the *dashed-line* in part (**b**). A small circular aperture allows directly emitted light from the QWs to escape as well as the small fraction reflected off the substrate hetero-interface. Light attempting to escape from sidewalls and backside of the device is dissipated and therefore eliminated from the collected light through the aperture (indicated by *red crosses*). (**b**) Picture of the absorbing material covering the LED. The *squares* are openings for electric contacts and the *circle* is the aperture for light emission. (**c**) Picture of the LED covered with absorbing material under bias. Only light emitted through the aperture is observed in air [After Ref. [14]] (Color figure online)

traction, where all other extraction paths are suppressed. This is achieved by covering the LED with a perfectly absorbing material and leaving only a small well-defined aperture of area a on the LED top surface. This is illustrated in Fig. 6.2a, where all trajectories originating from sidewalls, backside of the device, contact pads (not shown), etc., are dissipated by this surrounding material and therefore eliminated from the collected light. Only photons emitted directly from the QW region, plus a small fraction reflected off the substrate hetero-interface (in the case of hetero-structures, for example GaN on sapphire) go through the aperture, and thus reliable estimates of the extraction efficiency η_{extr} can be made. Some care is required when defining the aperture in the absorber material, such as a large spatial separation from the sidewalls and contacts, and a much larger aperture size compared to the GaN thickness L (both roughly larger than $10\,L$).

The η_{EQE} is measured using an integrating sphere and subsequently corrected by the ratio between the aperture surface a and the total LED top surface A, corresponding to the surface of current injection, as $\eta_{\text{EQE}} = \eta_{\text{EQE}}^{\text{meas}} A/a$, where $\eta_{\text{EQE}}^{\text{meas}}$ is the measured EQE in the integrating sphere. Bringing this technique into practice requires effort in two separate areas. First, we must calculate with good accuracy the light extraction for this device, and evaluate the precision of such model. Second, a device with only direct light extraction as described above must be fabricated.

One possible difficulty is that the distribution of light emission from multi-QWs is somewhat involved. QWs in LEDs, usually more than one, are distributed over a finite thickness within the device, and are each few nanometers wide, spaced apart by barrier layers tens of nanometers thick. One might assume that each well is

equally excited and thus emits an equal number of photons in a spherically symmetric pattern. However, it is just as plausible to assume that under EL conditions, either the first well that carriers encounter, the last well (perhaps immediately preceding an electron-barrier layer) or one of the center wells (being most equally accessible to both electrons and holes) would have a disproportionately larger share of the emission. David et al. [12] examined this question with regard to GaN/InGaN LEDs, and reported that nearly all of the emission from an LED comes from the QW nearest to the p-doped side of the device. The predominant emission from the top QW was attributed to the poor hole transfer between QWs, and occurs regardless of the number of quantum wells, often requiring some special design such as a double heterostructure to modify the carrier distribution [12]. That determination was made possible due to the fact that the various QWs have different emission patterns due to their different distances to the interface with an Ag mirror. In our structures comprising materials with only low index contrast, the exact position of emitting QWs will not make any difference as long as they are sufficiently away from the device interfaces with non-nitride materials (see discussion in Appendix B).

For the sake of completeness, Kivisaari et al. [16] proposed an alternative EL-based method to measure the IQE based on extracting the radiative part as the quadratic term of the A-B-C model [17] from the measurements of EQE. The A-B-C model uses as parameter the carrier density in the active region. However, in practice, it is not possible to assess this parameter from the current density injected in the device, especially in the case of multiple quantum wells: this method assumed that all QWs were equally injected with an injection efficiency of 100 % which has been demonstrated to be inaccurate in the nitride system [12]. Their attempt to correct for the uneven injection in each QW, which directly influences the carrier density in each well, was to deliberately consider a thinner active region, replacing the nominal 25 nm by 10 nm without clear theoretical or experimental support, which in turn might lead to an inaccurate estimation of the absolute IQE. Another issue is to assign the quadratic term to radiative recombination only. This is a critical point to the method and this assumption might fail in cases where the injection efficiency is not unity or varies with applied current density. This technique however could have some merit as a relative measurement between batches of similar chips, to assess the origin of variations in efficiency on growth and fabrication quality.

A simple LED geometry on a foreign substrate (the case of LEDs in bulk GaN substrates is treated in Sect. 6.3.2) can be described from an optical point of view as consisting of 3 distinct homogeneous layers: the LED active layer (L), substrate (s), and a thin metal contact (m) in the top surface, as shown in Fig. 6.3. The outer medium is air (a). The structure is surrounded by light absorbing material (abs) which removes all contributions to E_{out} other than the direct propagation of the $E_{\text{source}}^{(\gamma,\rho)}$ upward and its reflection at the LED/substrate interface. Here γ indicates the dipole orientation and ρ the polarization TE and TM. In the analysis presented in Ref. [13] and in Appendix A, we derive a general expression of the electric field in a structure without metal contacts and separately calculate a transmission function corresponding to the absorbing metal, since the electric field is much more sensitive

Fig. 6.3 Schematics of the simple LED geometry consisting of 3 distinct homogeneous layers: the LED (L), substrate (s), and a thin metal contact (m) in the top surface
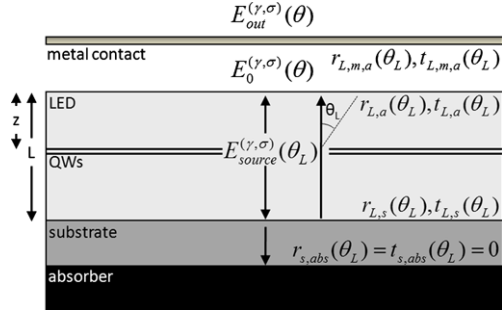
to the metal parameters, such as thickness and refractive index, than to the LED structure parameters—this is actually the core of our analysis.

The external electric field $E_{\text{out}}^{(\gamma,\rho)}$ is then calculated from the metal transmission function $\mathcal{T}_m^\rho$ and the external electric field of a structure without metal contact $E_0^{(\gamma,\rho)}$, benefiting from a general property of light propagating in stratified planar media [18, 19], described by a matrix formalism, which allows us to separate the effect of the metal contact from the LED as [13] $|E_{\text{out}}^{(\gamma,\rho)}|^2 = \mathcal{T}_m^\rho |E_0^{(\gamma,\rho)}|^2$. From the expression of the electric field emitted to air, we calculate the external power per solid angle $dP_{\text{out}}^{(\gamma,\rho)}(\theta,\lambda)/d\Omega\,dS$ and the exact extraction efficiency. The complete derivation of these expressions is described in Appendix A and Ref. [13]. The extraction efficiency was then simplified by the following approximations (see Appendix A):

Variable-Incidence (VI) approximation:

$$\eta_{\text{extr}} \simeq \langle \mathcal{T}_m^{\text{TE}} \rangle_{(\lambda,\theta)} \eta_{\text{extr}}^{0,\text{TE}} + \langle \mathcal{T}_m^{\text{TM}} \rangle_{(\lambda,\theta)} \eta_{\text{extr}}^{0,\text{TM}} \tag{6.5}$$

or the *Normal-Incidence (NI) approximation*:

$$\eta_{\text{extr}} \simeq \langle \mathcal{T}_m(0°) \rangle_\lambda \eta_{\text{extr}}^0 \tag{6.6}$$

where $\langle \mathcal{T}_m^\rho \rangle_{(\lambda,\theta)}$ is the metal transmission function averaged in both λ and θ, $\langle \mathcal{T}_m(0°) \rangle_\lambda$ is the metal transmission function averaged over the LED emission wavelengths and η_{extr}^0 is the extraction efficiency of the LED with metal contacts.

6.3.1 Calculation of Light Extraction Efficiency in a Simple GaN-Based LED

The LED structure considered here is the same used in the experimental section of this chapter. The LED material is GaN on a sapphire substrate. The LED thickness is $L = 4869$ nm, the distance z of the top surface to the QW is 140 nm. The central emission wavelength from the QW is 445 nm and the measured refractive
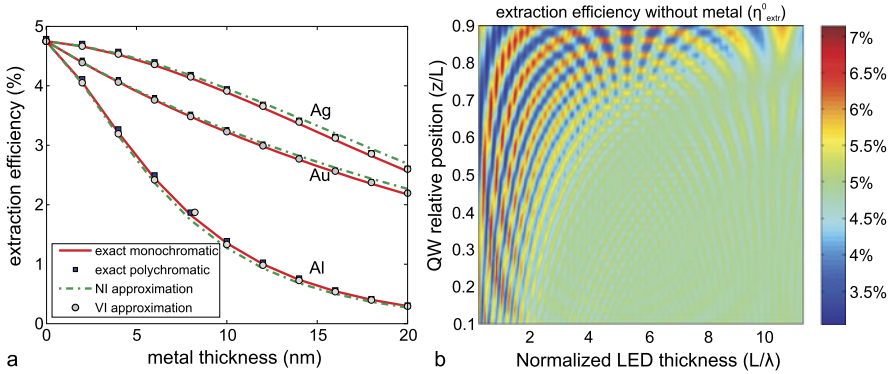
Fig. 6.4 (**a**) Extraction efficiency of a GaN-based LED on sapphire substrate versus metal thickness for Au, Ag and Al, using the analytical expressions for monochromatic (*solid*) and polychromatic (*squares*) emission, VI (*circle*) and NI (*dashdoted*) approximations. The *center* wavelength was $\lambda = 445$ nm and for the polychromatic emission $\sigma^2 = 25$ nm. (**b**) Extraction efficiency to air as a function of LED thickness and QW position in the LED (z is relative to the LED top surface)

indexes of the materials at this wavelength are $n_L = 2.475$ and $n_s = 1.779$ for the LED and substrate respectively. The lineshape of the QW was approximated by a symmetric Gaussian function centered at 445 nm with variance $\sigma^2 = 25$ nm, which is typical for InGaN/GaN QWs. We calculated the extraction efficiency versus metal thickness for the common metals: Au, Ag and Al, whose optical constants are $n_{Au} = 1.544 + i1.896$, $n_{Ag} = 0.155 + i2.421$ and $n_{Al} = 0.616 + i5.394$ at $\lambda = 445$ nm were obtained from the literature [20]. We calculated the exact extraction efficiency from this structure considering both mono- and polychromatic radiation from the QWs and compared these results to the VI and NI approximative models, shown in Fig. 6.4a. The approximative models show an excellent agreement to the analytical and polychromatic results for these three different metals presenting very different optical constants, which further validates our approximate models. The details on the evaluation of the approximative model given by Eqs. (6.5) and (6.6) and validation with the exact calculation of the extraction efficiency for a monochromatic and polychromatic light emission are presented in Appendix A.

Figure 6.4b shows the effect of the LED thickness and position of the QWs relative to the top surface of the LED on the calculated extraction efficiency without metal η_{extr}^0. The extraction efficiency oscillates significantly for an LED structure thinner than 1 μm and tends to a fixed value for thicker structures ($L/\lambda > 5$) and for the QWs located far from the bottom interface with the sapphire substrate (this is the case for most of the GaN-on-sapphire LEDs). The fast oscillating behavior of light extraction with LED thickness (also shown in Fig. 6.13a) is averaged in practical LEDs due to thickness fluctuations and polychromatic emission from the source which yields a much smoother function almost independent on LED thickness.

Therefore, the extraction efficiency results shown in Fig. 6.4a can be generally used to assess the IQE for a large range of LED structures other than the one considered in this section. In fact, the calculation of the extraction efficiency is mostly, if not solely, affected by the metal contact properties. The slope of the curves in Fig. 6.4 reflects the sensitivity of these results to thickness variations of the metal contact, where Au and Ag seems to be a better choice than Al. A more detailed investigation of the sensitivity of our model to different metal contacts, presented in Appendix B, shows that the appropriate choice of metal contacts, such as Au for example (see Fig. 6.13 in Appendix B), reduces significantly the dependence of the modelling results on the LED parameters.

6.3.2 Application to LEDs Grown on Bulk GaN Substrates, Complex LED Structures and Lasers

Let us calculate here the extraction efficiency to air η_{extr}^0 for bulk GaN LEDs, which can then be applied to determine the IQE in such structures by using the VI or NI approximations. The external electric field in this case is written in a simple expression, due to the absence of the interface with the substrate, $|E_0^{(\gamma,\rho)}| = |E_{\text{source}}^{(\gamma,\rho)}||t_{L,a}^\rho|$, which also eliminates the oscillations due to the LED cavity on the extraction efficiency. Figure 6.5a shows the η_{extr}^0 for a bulk GaN LED (solid-red) for QWs placed at 140 nm from the top surface. In this case, the absence of the reflection from the substrate reduces slightly the extraction efficiency through the top surface, considering that the bottom interface is covered with absorbing material.

The $\eta_{\text{extr}}^0(\lambda)$ can be easily calculated for bulk LED from a quadratic fit of this curve:

$$\eta_{\text{extr}}^0(\lambda) = -1.493 \times 10^{11}\lambda^2 + 1.860 \times 10^5\lambda - 5.817 \times 10^{-3} \qquad (6.7)$$

For an easy comparison, we plot in the same figure the η_{extr}^0 for a GaN LED on sapphire (dashed-blue), previously calculated.

The extraction efficiency through a simple facet is often approximated by the fraction of solid angle of the air cone, given by $[1 - \cos(\theta_c(\lambda))]/2$, where $\theta_c(\lambda)$ is the critical angle of total internal reflection inside GaN. For comparison to the exact results, we plot the solid angle approximation (dashdoted-green) in Fig. 6.5a, which shows that this simple approximation has a significant error of at least 10 %, which is due to the non-isotropic radiation from the dipole source.

Turning to the impact of various additional structures commonly found in LEDs, we investigate the two most important: superlattices, often grown above the nucleation layer at the substrate interface to improve growth quality, and the confining layers, needed in most laser structures. We leave aside the thin electron blocking layers, commonly used in LEDs, as it would have a smaller optical effect than these two dominant structures. Figure 6.5b shows the effect of the superlattices: even for 40 period superlattices, the relative modification to light extraction remains below
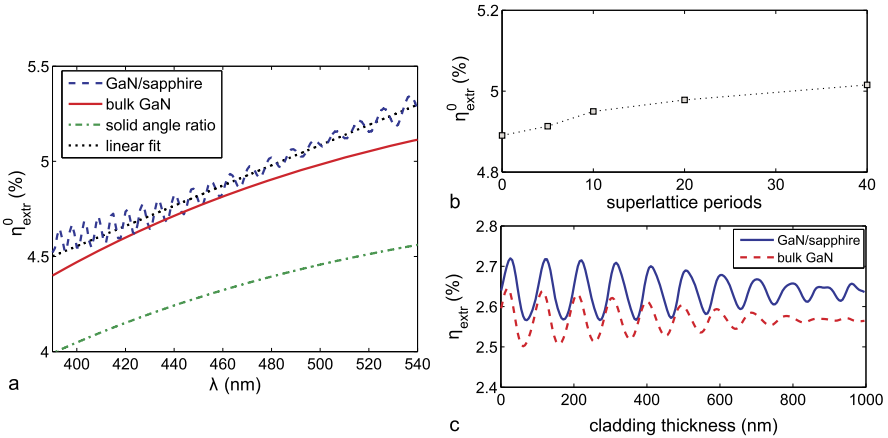
Fig. 6.5 (**a**) Extraction efficiency η_{extr}^0 for bulk GaN LED (*solid-red*) and for the GaN LED on sapphire previously calculated (*dashed-blue*), for QWs placed at 140 nm from the top surface, along with the solid angle ratio approximation for the extraction efficiency (*dashdotted-green*). (**b**) Effect of embedded superlattices on the light extraction efficiency to air. We considered a 2 nm-InGaN/2 nm-GaN superlattice below the active region with 0, 5, 10, 20 and 40 periods. (**c**) Impact of AlGaN cladding layers thickness on the light extraction efficiency using a 15.4 nm-thick Ni–Au contact and Al content of 15 % in the cladding layer. We considered an active layer (quantum wells, barriers, guiding layers) of fixed 240 nm thickness, sandwiched by two equal cladding layers, for a GaN structure grown on a sapphire substrate (*solid-blue*) or on a bulk GaN (*dashed-red*) (Color figure online)

2 %. Figure 6.5c shows the impact of adding $Al_{0.15}Ga_{0.85}N$ cladding layers of variable thicknesses to an active layer (quantum wells, barriers, guiding layers) of fixed 240 nm thickness over a 4 μm-thick GaN slab on a sapphire substrate (solid-blue) or on a bulk GaN (dashed-red). The extraction efficiency remains fairly constant in both cases. The rigid downwards shift of 3 % observed when compared to heterostructure is due to the absence of reflection at a substrate interface.

Therefore, the results presented in Fig. 6.4 can be extended to more complex nitride-based LED structures as well as lasers. In the following section we present the combination of this theoretical results with experiments to determine the internal quantum efficiency of LEDs.

6.4 Experimental Assessment of IQE

In the present technique, the LED must be coated with a perfectly light absorbing material containing only a small well-defined aperture on the LED top surface [14] (illustrated in Fig. 6.2a). The aperture diameters varied from 5 to 300 μm to assess the dependence of the total collected light on the aperture size, which is used to feed the theoretical model. The P_{out} from each aperture, for a given current, was measured using an integrating sphere and subsequently corrected by the ratio between

the aperture surface a and the total LED top surface A, corresponding to the surface of current injection. Current was injected by a semi-transparent thin metal layer on the top surface of the LED, through large metallic pads placed far (tens to hundreds of microns) from the circular aperture where the light is collected, to isolate the light measured from optical features with uncertain properties. The IQE was then determined from the following expression:

$$\eta_{\text{IQE}} = \frac{\eta_{\text{EQE}}}{\eta_{\text{extr}}} = \frac{q\,P_{\text{out}}}{\hbar\omega I\,\eta_{\text{extr}}}\frac{A}{a} \tag{6.8}$$

where $\hbar\omega$ is the photon energy, q is the electron charge and η_{extr} is the extraction efficiency for a single facet of the LED as calculated in the previous section.

Next sections show the application of such method to a state-of-the-art GaN-based LED as well as to a relatively poor performance device [14].

6.4.1 IQE Measurement of a State-of-the-Art LED

The LED investigated was a state-of-the-art device from Seoul Semiconductors, grown by metal-organic chemical vapor deposition (MOCVD) on a sapphire substrate. The LED structure consisted of a ∼4670 nm-thick n-GaN followed by 60 nm-thick layer of QWs emitting at $\lambda = 445$ nm and a 140 nm-thick p-GaN layer. The LED top surface was extremely smooth (RMS roughness ∼0.247 nm for a $5 \times 5\ \mu m^2$ scan measured by atomic force microscope (AFM)), which is an important requirement of the technique. Homogeneous current injection was assured by an annealed semitransparent Ni–Au contact (5/10 nm-thick) of area $A = 484.5 \times 484.5\ \mu m^2$ over the LED top surface. The complex refractive index of the Ni–Au alloy ($n_{\text{alloy}} = 1.4649 + 1.6485i$) was measured from co-deposited and co-annealed films on a sapphire substrate, using variable angle spectroscopic ellipsometry (VASE) [21].

The light absorbing material used was an equal volume mixture of AZ4210 photoresist (PR) and PureBlack carbon black particles (Superior Graphite Corp.). Apertures of 50, 100, 200, and 300 μm diameter were patterned into the absorber by lift-off, where a bilayer of PMGI SF 15 PR followed by AZ4210 was spun onto the surface of the processed LED structure [21].[1] The deposited absorber presented

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[1]The carbon black particles prevent proper exposure of the PR to UV light, hence it is not possible to pattern this layer using standard photolithography techniques. Instead, a lift-off method was applied, where a bilayer of PMGI SF 15 PR followed by AZ4210 was spun onto the sample surface, on which complete LED structures had already been processed. The AZ4210 was then patterned using standard optical lithography, which acted as a mask for the deep-UV exposure of the PMGI SF 15 layer. This exposed layer was then developed, re-exposed, and developed again, creating a deep undercut in the lift-off mask profile. The mixture of PR and carbon black particles was spun over this patterned lift-off mask and then baked at 200 °C for 1 hour. The baking process caused the mixture to adhere strongly to the sample surface and did not affect the PMGI SF 15, which is robust to temperatures up to 250 °C. The lift-off was completed by placing the sample in 1165 PR stripper at 80 °C for 1 hour with strong sonication.
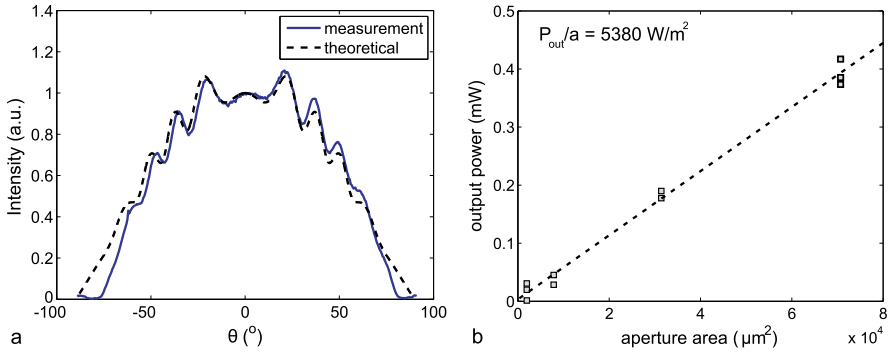
Fig. 6.6 (**a**) Theoretical (*dashed*) and experimental (*solid*) angle-resolved pattern of the light emission from the LED. (**b**) Output power emitted from aperture versus aperture area

optical transmissivity of 0.04 % and reflectivity of 0.8 % at normal incidence. The light absorbing material for the backside and sidewalls of the device was flat black paint (Rustoleum). The measured specular reflection at normal incidence for this material was less than 1.5 %, and the total scattered reflection was less than 3 % of incident light from an absorber-air interface, which are suitable for the application envisaged in this work. After coating with absorber, samples were diced, mounted to headers, and wire-bonded for integrating sphere measurements.

As shown in the previous section, the theoretical η_{extr} is readily available for the common metals, such as Ag, Au and Al (Fig. 6.4). The calculation for other metal contacts requires the accurate knowledge of the refractive index of the semitransparent contact $n_{\text{alloy}} = 1.4649 + 1.6485i$, which in this case was done by VASE. The calculated η_{extr} for this device, using the model described in the Appendix A, was 2.7 %. This low value is justified by the rather thick Ni-Au current spreading layer that considerably absorbs the light going to air. As shown in Fig. 6.4a, the calculated η_{extr} is fairly constant for a GaN layer thicker than 1.5 μm, and does not depend on the QW position in the present structure. For the nominal GaN thickness and QW relative position of our structure, the maximum variation in η_{extr} is at most 1.6 % (Appendix B).

As a confirmation of the theoretical predictions, we compared the angular pattern of the light emitted to air from the theoretical model (Eq. (A.18)) to the experimental results (Fig. 6.6). The angular emission from the LED was assessed using an angle-resolved setup, where the far field spectrum was collected at all angles θ from $-90°$ to $90°$. The oscillations observed correspond to Fabry-Perot interferences from the GaN interfaces with the substrate and the metal. The excellent match between the experimental and theoretical curves in Fig. 6.6a is an indication of the correct theoretical model used to predict the LED light emission. As explained in Appendix C, a small damping corresponding to an isotropic emission of 1 % of the total output power emitted from the LED ($F_{\text{rough}} = 1$ %) was considered in the theoretical curve. While this improved the agreement between the experimental and theoretical curves, it accounted for light scattering from any small roughness present in

the metal contact. In the following section, we show that the damping included in LEDs with much rougher surfaces (rms roughness of ∼10 nm for a $5 \times 5 \ \mu m^2$ scan measured by AFM) is as high as 70 %, which corroborates the smoothness of the interfaces in the present sample.

The output power of the light emitted from the aperture on the LEDs was measured in an integrating sphere under pulsed excitation, with duty cycle of 0.1 %, to eliminate the degrading effect of heating on the IQE. The measured output power for 50, 100, 200, and 300 μm diameter apertures varied linearly with the aperture area as presented in Fig. 6.6b. The linear variation of the output power versus aperture size validates the use of a homogeneously distributed set of dipoles replacing the QWs in the theoretical model. The power per unit area P_{out}/a deduced from its slope at a nominal input current density of 8.52 A/cm^2, averaged for several devices, was 5380±380 W/m^2. This current density corresponds to the peak IQE, which was first determined by measuring the P_{out} versus current density. Using this value in Eq. (6.8) combined to the theoretical η_{extr} of 2.7 % resulted:

$$\eta_{IQE} = 83.8 \ \%$$

6.4.2 EL-Based IQE Measurement of a Poor Performing LED: Effect of Surface Roughness

We also applied the present technique in a poor performing LED. The GaN-based LED used in this study was grown by MOCVD on a sapphire substrate, with 5x-InGaN QWs emitting at 445 nm. The LED was processed with a semitransparent Ni–Au contact to the p-GaN layer. The LED structure was composed of a 4570 nm-thick GaN over a sapphire substrate. The QWs, emitting at $\lambda = 445$ nm, were below a 295 nm-thick p-GaN layer. The index of refraction and effective thickness of the Ni/Au alloy, measured by VASE, was $n_{alloy} = 1.544 + i1.016$ at $\lambda = 445$ nm and 18.4 nm, respectively. The sapphire substrate was considered infinitely thick because of the assumption that the light going downwards in the sapphire substrate is completely absorbed at the interface of the absorbing material and sapphire. The calculated η_{extr} for this device was 3.03 %, which was different than the previous sample due to differences on the index of refraction and thickness of the Ni/Au alloy. This result also considered a device with a smooth material-to-air interface. The output power varied linearly with aperture area indicating uniform emission across the device surface. The slope of the curve in Fig. 6.7a yields a power per unit area of 2900 W/m^2 at an injection current density of 7.9 A/cm^2, where the peak EQE occurred. Using Eq. (6.8), the peak IQE of these devices was estimated to be 43 % at a current density of 7.9 A/cm^2. The IQE value obtained must be carefully considered at this point.

The theoretical model assumes a perfectly flat and smooth surface, however, the device material used had a RMS surface roughness of 10 nm measured by AFM, due to the low temperature growth used for the p-type GaN. This rough surface destroys
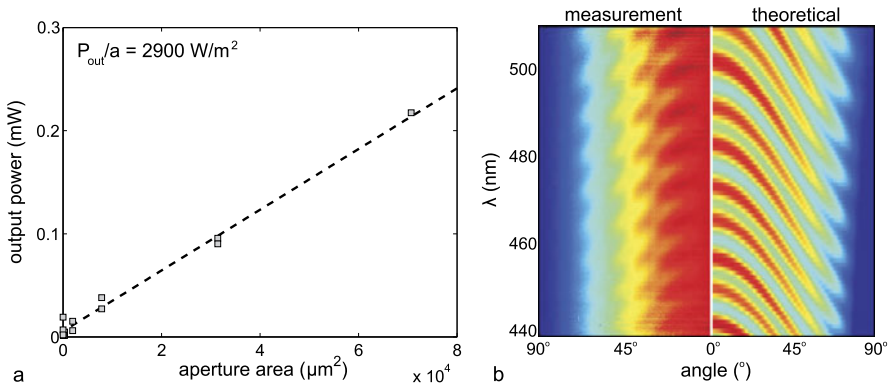
Fig. 6.7 (**a**) Measured output power versus aperture surface. (**b**) Angle-resolved measurement on the sample with rough surface (*left*) compared to the calculated emission using the dipole model (*right*). Notice the blurry Fabry Perot fringes from the current sample indicating a pronounced effect of the roughened surface

the constructive interferences in the LED interfaces and enhances the extraction efficiency of the device compared to the calculated value. The ratio between guided light and directly extracted light is not maintained which partially invalidates the results from the theoretical model. Figure 6.7b shows the angle-resolved measurement in this rough LED and the calculated radiation from a similar LED with flat surfaces. The Fabry-Perot fringes are much less pronounced when the LED surface is rough.

To account for this effect, we modelled the effect of the surface roughness as a Lambertian light source emitting simultaneously with the dipoles inside the flat structure, as described in details in the Appendix C. This is justified by considering that the light impinging the rough surface is randomly scattered, corresponding to an isotropic source inside the structure. Let us use this model to account for the surface roughness in our sample. The comparison of the angle-resolved emission between the measurements and dipole model is shown in Fig. 6.8. The blue curve corresponds to the angle-resolved measurement of the LED at $\lambda = 460$ nm, which is simply a cross section of the measurement shown in Fig. 6.7b. The red curve corresponds to the calculated emission of the dipoles inside the structure with a perfectly flat surface for the same wavelength, which is far from matching the experimental result shown in blue. However, when the Lambertian emission due to the rough surface, represented by the black curve in Fig. 6.8, is added to the theoretical Fabry-Perot oscillations, it results in the green curve which matches very well the experimental curve. The addition of an amount of Lambertian emission related to the surface roughness to the calculated angular pattern resulted in a excellent agreement to the measurements, validating the dipole model used.

The angle-resolved measurement is therefore a very useful technique to assess the contribution of light randomly scattered to the measured EQE which ultimately can be used to validate the model requirements of flat interfaces. The pronounced oscillations in such measurements fade away proportionally to the presence of light
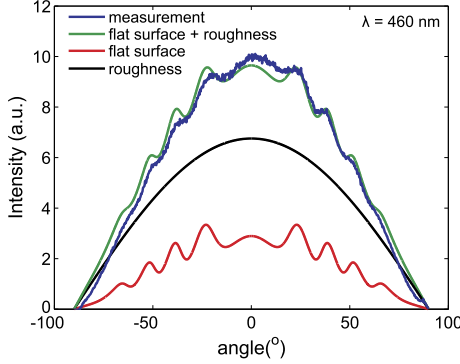
Fig. 6.8 Cross section of the angle-resolved measurement at $\lambda = 460$ nm (*blue curve*). The *red curve* represents the calculated emission of the dipoles inside the structure with a perfectly flat surface for the same wavelength. The *black curve* is the Lambertian emission due to the rough surface and the *green curve* shows the Lambertian emission added to the theoretical Fabry-Perot oscillations which matches the experimental curve (Color figure online)

scatterers in the LED and the accuracy of this model is largely reduced with the presence of such scatterers. The intensity of the Lambertian emission here corresponds to 70 % of the intensity from the measurement. To roughly estimate the IQE, we consider that only the remainder 30 % of the measured P_{out} is due to the emission from dipoles in a flat surface, for which our theoretical extraction efficiency is valid, which results in an estimated IQE of 13 %. However, this should be considered just a rough estimation to account for the effect of surface roughness on the present model, a precise determination of the IQE requires flat smooth interfaces.

6.5 Model for Photon Recycling

In this section we present a simple model to estimate the effect of photon recycling on the extraction efficiency of an LED (Refs. [1, 13, 22]). This mechanism consists of a sequence of re-absorption of the guided light in the LED structure and re-emission by the QWs, that ultimately can play an important role in extracting photons that have not been directly emitted to the air cone in the first pass. The schematic in Fig. 6.9a illustrates the infinite iterations of this mechanism. The η_{EQE} can be written as $\eta_{EQE} = \eta_{IQE}\eta_{extr} + \eta_{IQE}^2(1 - \eta_{extr})F_{rec}\eta_{extr} + \cdots$, therefore, the extraction efficiency with photon recycling is:

$$\eta_{extr}^{PR} = \eta_{extr}\left[1 + \frac{\eta_{IQE}(1 - \eta_{extr})F_{rec}}{1 - \eta_{IQE}(1 - \eta_{extr})F_{rec}}\right] \qquad (6.9)$$

where $F_{rec} = \frac{F_{QW}}{F_{QW}+F_{diss}}$ is the ratio of the LED guided light absorbed by the QWs (F_{QW}) to the total dissipated light ($F_{diss} + F_{QW}$).
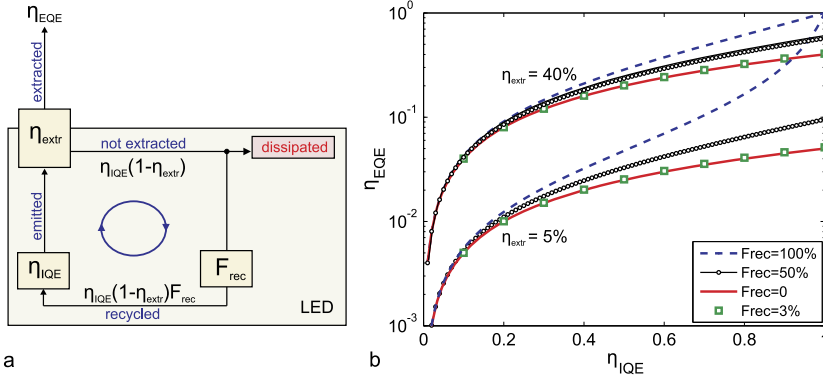
Fig. 6.9 (**a**) Schematic of the light emitted within the LED structure. (**b**) Calculated $\eta_{\text{extr}}^{\text{PR}}$ versus η_{IQE} for $F_{\text{rec}} = 0$, 3 %, 50 % and 100 % and for $\eta_{\text{extr}} = 5$ % and 40 %

In the case of nitride-based LEDs, the absorption by QW is spectrally shifted (large Stokes shift) from its emission edge, and only the tail of the absorption curve overlaps the emission spectrum, thus the absorption coefficient is quite small (experimentally estimated in Ref. [23] as 10^3 cm^{-1}). The volume ratio of the QWs to the LED is approximatively 1.5 %. The absorption coefficient of the metal layer, which is the most significant absorption mechanism in the LED, is approximatively $\alpha_m = (4\pi/\lambda)k_m = 5 \times 10^5$ cm^{-1} and the volume fraction penetrated by the guided modes in the metal layer is about 0.2 %. Thus, as a rough estimation, $F_{\text{rec}} \approx 3$ %.

Figure 6.9 shows the plot of $\eta_{\text{extr}}^{\text{PR}}$ versus η_{IQE} for the cases without, with 50 % and 100 % photon recycling, as well as the estimated $F_{\text{rec}} = 3$ % for a nitride-based LED. Two different extraction efficiencies are considered which illustrated both the case of a simple LED geometry ($\eta_{\text{extr}} = 5$ %) as well as a extreme case of an ultra thin microcavity LED ($\eta_{\text{extr}} = 40$ %). In neither of these cases was $\eta_{\text{extr}}^{\text{PR}}$ modified by the photon recycling for nitride-based LEDs, due to the very small F_{rec}.

In other material systems, where there is a larger overlap between the absorption and emission energy-edges of the QWs, F_{rec} is much higher and can be close to 100 % [1]. In this case, photon recycling is very efficient in increasing the effective extraction efficiency of the LED [24]. This effect is more pronounced when the direct extraction efficiency (without photon recycling) is small. Equation (6.9) provides a corrective factor for the extraction efficiency for material systems such as GaAs where photon recycling is an effective mechanism.

6.6 Conclusions

In this chapter we presented an overview of a few techniques to assess the internal quantum efficiency in LEDs. The IQE is widely estimated by temperature-dependent photoluminescence which relies on strong assumptions, such as that

non-radiative mechanisms are totally eliminated at low temperatures and also similar carrier injection to the active region compared to electroluminescence, neglecting effects of applied bias on the internal electric field in the QWs and different carrier injection in individual wells of a MQW structure under PL and EL excitation.

We presented a technique based on electroluminescence to measure IQE in GaN-based LEDs, which relies on similar operation conditions and injection mechanisms as the operating LED. We derived a model to determine light extraction efficiency, LEE, through a single facet of simple LED structures. The model relies on covering the LED with light absorbing material, which eliminates the difficult-to-model, indirectly extracted light. Thus, the output power, or η_{EQE}, measured in an absolute manner by integrated sphere is only due to directly extracted light, an easily calculated quantity. Applied to GaN structures, the model predicts that the LEE is a product of a bare GaN LED LEE times the transmission function of the top metal contact. General values of the bare LED LEE as $\eta_{\mathrm{extr}}^{0}(\lambda) = 5.31 \times 10^4 \lambda + 2.43 \times 10^{-2}$ were given, which were quite independent on the details of the LED structure such as exact value of GaN thickness, presence of superlattices or confinement layers, etc. The same is true for LEDs grown over GaN free-standing substrate, with a small modification of $\eta_{\mathrm{extr}}^{0}(\lambda)$ as one misses the power reflected from the GaN/sapphire interface into the escape cone. The effect of different contact metals on the LEE was investigated, leading to a conclusion that Au-based metal contacts are the best choice, among commonly used metals, for the application of the present technique in terms of LEE robustness to LED parameters.

This model was extended to predict the angular light emission in LEDs, which can be used to compare the theoretical and experimental results as well as to identify the presence of surface roughness which could be detrimental to this model. Photon recycling was evaluated and shown to be negligible in nitride structures, mainly because of the large Stokes shift between emission and absorption.

We also presented the experimental details to fabricate a suitable structure for the application of the present method, consisting of patterning a well-defined aperture in a light absorbing material that surrounds the device. The application of this method to a state-of-the-art GaN LED yielded a peak IQE of 83.8 %. The angle-resolved prediction of the light emitted in the LED from the theoretical model was validated with experimental data and the excellent match between theory and experiments corroborates the model developed.

Appendix A: Theoretical Model of Light Emission in LEDs: QW Emission Described by Classical Dipoles

QWs in LEDs are usually more than one, distributed over a finite thickness within the device, as each is a few nanometers wide and spaced apart by barrier layers of tens of nanometers thick. One might assume that each well is equally excited and thus emits an equal number of photons in a spherically symmetric pattern. However, it is just as plausible to assume that under EL excitation, either the first well that carriers encounter, the last well (perhaps immediately preceding an electron-barrier layer) or one of the center wells (being most equally accessible to both electrons and holes) would have a disproportionately larger share of the emission. David et al. [12] examined this question with regard to GaN/InGaN LEDs, and reported that nearly all of the emission from an LED comes from the QW nearest to the p-doped side of the device. The predominant emission from the top QW was attributed to the poor hole transfer between QWs, and occurs regardless of the number of quantum wells, often requiring some special design such as a double heterostructure to modify the carrier distribution [12]. That determination was made possible due to the fact that the various QWs have different emission patterns due to their different distances to the strongly reflecting Ag/GaN interface.

To address the question of calculating light emission per QW, we refer to Benisty et al. [25], who used dipole emitters as a photon source term in LED emission models. Benisty et al. model the emission from the dipoles as a discontinuity in the scattering matrix propagation technique [26], as we describe below. The use of dipole emitters is justified by the similarity between the normalized expressions of the rate of spontaneous emission in QWs and the power emitted by classic electric dipoles. The rate of spontaneous transitions of e-h pairs between the conduction and valence bands in a QW, given by Fermi's golden rule, is proportional to $\mathcal{M}_{c-v} = |\langle \psi_c | \hat{e}\hat{p} | \psi_v \rangle|^2$, where $\hat{e}$ is the light polarization, $\hat{p}$ is the momentum operator, ψ_c and ψ_v are the conduction and valence band wavefunctions, respectively [27, 28]. The angular dependence of $\mathcal{M}_{c-v}$ is well described by combinations of horizontal and vertical dipole-like terms [29]. The normalized radiation patterns, given by the power per unit of solid angle, of vertical (v) and horizontal (h) dipoles, for both TE and TM polarizations are given by [25]:

$$\frac{dP_{source}^{(v,TM)}}{d\Omega}(\theta_L) = \frac{3}{8\pi} \sin^2 \theta_L \tag{A.1a}$$

$$\frac{dP_{source}^{(h,TE)}}{d\Omega}(\theta_L) = \frac{3}{16\pi} \tag{A.1b}$$

$$\frac{dP_{source}^{(h,TM)}}{d\Omega}(\theta_L) = \frac{3}{16\pi} \cos^2 \theta_L \tag{A.1c}$$

where θ_L is the angle with respect to the vertical direction.

The replacement of the electron-hole excitons by a uniform distribution of dipoles, whose emission is propagated through the multilayered structure by

transfer-matrix formalism, allows the calculation of the electric field in any layer of the structure by propagating the electric field from the dipole layer, which are determined from Eqs. (A.1a)–(A.1c) as

$$E_{\text{source}}^{(\gamma,\rho)}(\theta_L) = \sqrt{\frac{d\,P_{\text{source}}^{(\gamma,\rho)}}{d\Omega}(\theta_L)} \qquad (A.2)$$

for a γ dipole orientation and a ρ polarization.

A.1 Analytical Model for Light Extraction Efficiency

The theoretical assessment of the light extraction efficiency η_{extr} in an LED structure is based on the fraction of the integrated emitted power that exits the LED structure and propagates to air. This is determined from the calculation of the electric field radiated from the QWs, based on the propagation of the dipole electric field throughout the structure. The external power per solid angle $d\,P_{\text{out}}^{(\gamma,\rho)}/d\Omega dS(\theta)$, in the external direction θ, is given by the flux of the Poynting vector emitted from the dipoles, transmitted through the structure and corrected by the change in solid angle when the medium is changed [25]:

$$\frac{d\,P_{\text{out}}^{(\gamma,\rho)}}{d\Omega dS}(\theta) = \underbrace{\left(\left|E_{\text{out}}^{(\gamma,\rho)}(\theta)\right|^2 \frac{n_{\text{out}}\cos(\theta)}{n_L\cos(\theta_L)}\right)}_{\text{transmitted power from dipoles}}\ \underbrace{\left(\frac{n_{\text{out}}^2\cos(\theta)}{n_L^2\cos(\theta_L)}\right)}_{\text{change in solid angle}} \qquad (A.3)$$

where $E_{\text{out}}^{(\gamma,\rho)}$ is the external electric field for a γ dipole orientation and a ρ polarization, n_{out} and n_L are the refractive indexes of the external and LED media, respectively. θ and θ_L are respectively the external and internal angles with respect to the vertical direction. The angles in each medium θ_i, between the vertical and the propagation direction of emitted light, are determined from Snell's law $n_i\cos(\theta_i) = n_j\cos(\theta_j)$. The first term in Eq. (A.3) corresponds to the transmitted power from the dipole source to the external medium, given by the Fresnel transmission coefficient and the second term corresponds to the changes in solid angle from different media, obtained from the derivative of Snell's law.

The light extraction efficiency of the LED through a single facet is given by the ratio of the total output power, calculated from the integration of Eq. (A.3) for θ from 0 to $\pi/2$, to the total emitted power. The total emitted power by the normalized dipole source is unity for dipoles in bulk material. However, the total radiation by the dipole source inside an LED heterostructure is modified by the Purcell factor, which corresponds to the relative change in spontaneous emission from a source within an optical cavity compared to the same source in a bulk material [30]. The Purcell factor depends largely on the position of the QWs relative to the interface between the metallic contact and GaN, as well as on the choice of metal (or more specifically, on

the metal reflectivity), which is treated in more details in Appendix B. Figure 6.13 in Appendix B shows the deviation from unity of the total emitted power by the dipoles for different metals (Al, Au and Ag) versus the distance of the QW to their interface with GaN. A judicious choice of contact metal, such as Au, can largely reduce this dependence and for most of the practical cases as well as for the simple geometry structure treated here, the Purcell factor is close to one and the total emitted power by the dipoles can be approximated to unity (deviation from unity of 2.8 % at most). Figure 6.13 of Appendix B also offers a correction factor in case a different metal is used as contacts.

Therefore, the light extraction efficiency of the LED through its top facet is given by

$$\eta_{\text{extr}} = \int_0^{\frac{\pi}{2}} 2\pi \frac{dP_{\text{out}}^{(\gamma,\rho)}(\theta)}{d\Omega\, dS} \sin(\theta) d\theta \tag{A.4}$$

summed for $\rho = $ TE and TM, and $\gamma = v$ and h. The e-HH recombinations in the QWs, in both TE and TM polarizations, can be well described by horizontal electric dipoles. The e-LH recombinations can only be partially described by the combination of horizontal and vertical dipoles and will be neglected at low current injections due to the nondegeneracy of the light hole band at the minimum of energy and to the lower density of states of this energy band [29]. Thus $\gamma = h$. Therefore, the only requirement to determine η_{extr} is to calculate the external electric field $E_{\text{out}}^{(\gamma,\rho)}$, which is treated in the following section.

A.2 Exact Calculation of the Electric Field in a Multilayer Structure

The calculation of the electric field in a multilayer structure is based on the transfer matrix formalism, where a wave $E_\uparrow e^{+ik_z z} + E_\downarrow e^{-ik_z z}$ is represented by $\begin{pmatrix} E_\uparrow \\ E_\downarrow \end{pmatrix}$, which is the electric field of the upward and downward plane waves, respectively, and $k_z^i = n_i k_0 \cos(\theta_i)$ in the medium i.

In a multilayer structure, the source electric field is propagated through the structure by multiplying it to propagation M_{prop} matrices in the homogeneous layers and interface M_{interf} matrices in the interface between two different media. A convenient property of this method is that an equivalent matrix $\mathcal{M}$ for the entire structure can be simply obtained by multiplying all the interface and propagation matrices corresponding to the structure (Fig. 6.10), as

$$\mathcal{M} = \cdots M_{\text{interf}}^{i+1,i} M_{\text{prop}}^i M_{\text{interf}}^{i,i-1} \cdots \tag{A.5}$$

The propagation matrix from z_1 to z_2 in a homogeneous medium is:

$$M_{\text{prop}} = \begin{bmatrix} e^{i(k_z(z_2-z_1))} & 0 \\ 0 & e^{-i(k_z(z_2-z_1))} \end{bmatrix} \tag{A.6}$$
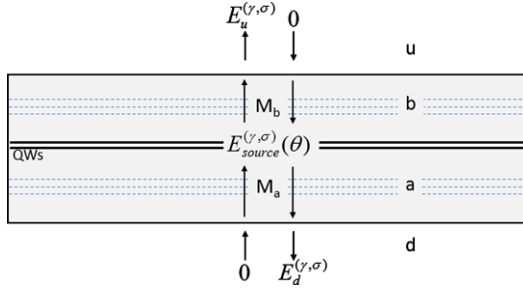
Fig. 6.10 Schematic of a the propagation of the electric fields in a multilayer structure. The external electric fields E_u and E_d, from the top and bottom media, are propagated to the source, where the discontinuity from the source electric field is applied. The matrices M_a and M_b correspond to the transfer matrix of the *bottom* (a) and *top* (b) halves of the LED structure

The interface matrices for TE and TM polarizations are given by

$$
M_{\text{interf}}^{\text{TE}} =
\begin{bmatrix}
\dfrac{k_z^{(2)}+k_z^{(1)}}{2k_z^{(2)}} & \dfrac{k_z^{(2)}-k_z^{(1)}}{2k_z^{(2)}} \\[2mm]
\dfrac{k_z^{(2)}-k_z^{(1)}}{2k_z^{(2)}} & \dfrac{k_z^{(2)}+k_z^{(1)}}{2k_z^{(2)}}
\end{bmatrix},
\qquad
M_{\text{interf}}^{\text{TM}} =
\begin{bmatrix}
\dfrac{n_2^2 k_z^{(1)}+n_1^2 k_z^{(2)}}{2n_1 n_2 k_z^{(2)}} & \dfrac{-n_2^2 k_z^{(1)}+n_1^2 k_z^{(2)}}{2n_1 n_2 k_z^{(2)}} \\[2mm]
\dfrac{-n_2^2 k_z^{(1)}+n_1^2 k_z^{(2)}}{2n_1 n_2 k_z^{(2)}} & \dfrac{n_2^2 k_z^{(1)}+n_1^2 k_z^{(2)}}{2n_1 n_2 k_z^{(2)}}
\end{bmatrix}
\tag{A.7}
$$

To obtain analytical expressions for the electric field in the outer media, we impose, as boundary condition of this problem, that the electric field of the incoming waves in the outer media are zero (Fig. 6.10). The external electric fields, from the top and bottom media, are then propagated to the source, where the discontinuity from the source is applied as

$$
\mathcal{M}_b
\begin{pmatrix} E_u^{(\gamma,\rho)} \\ 0 \end{pmatrix}
- \mathcal{M}_a
\begin{pmatrix} 0 \\ E_d^{(\gamma,\rho)} \end{pmatrix}
=
\begin{pmatrix} E_{\text{source}}^{(\gamma,\rho)}(\theta) \\ E_{\text{source}}^{(\gamma,\rho)}(\theta) \end{pmatrix}
\tag{A.8}
$$

where

$$
\mathcal{M}_a =
\begin{bmatrix} a_{11} & a_{12} \\ a_{21} & a_{22} \end{bmatrix}
\quad \text{and} \quad
\mathcal{M}_b =
\begin{bmatrix} b_{11} & b_{12} \\ b_{21} & b_{22} \end{bmatrix}
\tag{A.9}
$$

are calculated from Eq. (A.5) for the bottom (a) and top (b) halves of the LED structure (Fig. 6.10). The electric fields in the top (u) and bottom (d) outer media are

$$
E_u^{(\gamma,\rho)}(\theta) = E_{\text{source}}^{(\gamma,\rho)} \frac{\frac{1}{b_{11}}\left(1+\frac{a_{12}}{a_{22}}\right)}{1-\frac{b_{21}}{b_{11}}\frac{a_{12}}{a_{22}}}
\quad \text{and} \quad
E_d^{(\gamma,\rho)}(\theta) = E_{\text{source}}^{(\gamma,\rho)} \frac{\frac{1}{a_{22}}\left(1+\frac{b_{21}}{b_{11}}\right)}{1-\frac{b_{21}}{b_{11}}\frac{a_{12}}{a_{22}}}
\tag{A.10}
$$

In the case of the simple geometry considered in this chapter, the matrix terms are

$$
\frac{1}{b_{11}} = \frac{2k_z^b}{k_z^b + k_z^u} e^{i k_z^b L_b}
\quad \text{and} \quad
\frac{1}{a_{22}} = \frac{2k_z^a}{k_z^a + k_z^d} e^{i k_z^a L_a}
\tag{A.11}
$$

which correspond to transmission coefficients and

$$\frac{b_{21}}{b_{11}} = \frac{k_z^b - k_z^u}{k_z^b + k_z^u} e^{2ik_z^b L_b} \quad \text{and} \quad \frac{a_{12}}{a_{22}} = \frac{k_z^a - k_z^d}{k_z^a + k_z^d} e^{2ik_z^a L_a} \tag{A.12}$$

corresponding to reflection coefficients.

The electric field after the metal layer can be written as

$$
\begin{aligned}
E_{\text{out}}^{(\gamma,\rho)} &= \frac{E_{\text{source}}^{(\gamma,\rho)} t_{L,m,a}^\rho e^{i\phi'} (1 + r_{L,s}^\rho e^{i2(\phi-\phi')})}{1 - r_{m,L,a}^\rho r_{L,s}^\rho e^{i2\phi}} \\
&= \frac{E_{\text{source}}^{(\gamma,\rho)} t_{L,a}^\rho e^{i\phi'} (1 + r_{L,s}^\rho e^{i2(\phi-\phi')})}{1 - r_{L,a}^\rho r_{L,s}^\rho e^{i2\phi}} \frac{1 - r_{L,a}^\rho r_{L,s}^\rho e^{i2\phi} t_{L,m,a}^\rho}{1 - r_{m,L,a}^\rho r_{L,s}^\rho e^{i2\phi} t_{L,a}^\rho}
\end{aligned}
\tag{A.13}
$$

The first term is the external electric field without metal contacts and the second is the metal transmission function. Using the general property: $x, y \in \mathbb{C}$, $|xy|^2 = |x|^2 |y|^2$, we obtain [13]:

$$\left| E_{\text{out}}^{(\gamma,\rho)} \right|^2 = \mathcal{T}_m^\rho \left| E_0^{(\gamma,\rho)} \right|^2 \tag{A.14}$$

Let us first calculate the external electric field $E_0^{(\gamma,\rho)}$ in a structure without metal contact. Let the total LED thickness be L and the position of the dipole source within the LED be z. The corresponding phase shifts in the LED are $\phi(\theta_L) = n_L k_0 L \cos(\theta_L)$ and $\phi'(\theta_L) = n_L k_0 z \cos(\theta_L)$, where $k_0 = 2\pi/\lambda$, λ is the wavelength of the emitted light, n_L and θ_L are the refractive index and angle in the LED layer respectively. The transmission and reflection coefficients for a polarization ρ, at each interface from a medium i to j, are given respectively by $t_{i,j}^\rho$ and $r_{i,j}^\rho$.

The external electric field can be determined from the propagation of the dipole electric field $E_{\text{source}}^{(\gamma,\rho)}$ [18]:

$$
\begin{aligned}
E_0^{(\gamma,\rho)}(\theta) &= E_{\text{source}}^{(\gamma,\rho)} e^{i\phi'} t_{L,a}^\rho \left(1 + r_{L,a}^\rho r_{L,s}^\rho e^{2i\phi} + \cdots \right) \\
&\quad + E_{\text{source}}^{(\gamma,\rho)} e^{i(2\phi-\phi')} r_{L,s}^\rho t_{L,a}^\rho \left(1 + r_{L,a}^\rho r_{L,s}^\rho e^{2i\phi} + \cdots \right) \\
E_0^{(\gamma,\rho)}(\theta) &= \frac{E_{\text{source}}^{(\gamma,\rho)} t_{L,a}^\rho e^{i\phi'} (1 + r_{L,s}^\rho e^{i2(\phi-\phi')})}{1 - r_{L,a}^\rho r_{L,s}^\rho e^{i2\phi}}
\end{aligned}
\tag{A.15}
$$

We recall the transmission and reflection expressions for both TE and TM polarizations for a plane wave going from a medium i to j:

$$r_{i,j}^{\text{TE}} = \frac{n_i \cos(\theta_i) - n_j \cos(\theta_j)}{n_i \cos(\theta_i) + n_j \cos(\theta_j)}, \qquad r_{i,j}^{\text{TM}} = \frac{n_i \cos(\theta_j) - n_j \cos(\theta_i)}{n_i \cos(\theta_j) + n_j \cos(\theta_i)}$$

$$t_{i,j}^{\text{TE}} = \frac{2n_i \cos(\theta_i)}{n_i \cos(\theta_i) + n_j \cos(\theta_j)}, \qquad t_{i,j}^{\text{TM}} = \frac{2n_i \cos(\theta_i)}{n_j \cos(\theta_j) + n_i \cos(\theta_j)}$$

The transmission function $\mathcal{T}_m^{\rho}$ for the metal contact is a simple replacement of the transmission and reflection coefficients of a simple LED/air interface (L, a) with those of an LED/metal/air interface (L, m, a). It contains all the metal parameters separated from the more general expression of $E_0^{(\gamma,\rho)}(\theta)$ and it is written as

$$\mathcal{T}_m^{\rho} = \left| \frac{1 - r_{L,a}^{\rho} r_{L,s}^{\rho} e^{i2\phi}}{1 - r_{m,L,a}^{\rho} r_{L,s}^{\rho} e^{i2\phi}} \frac{t_{L,m,a}^{\rho}}{t_{L,a}^{\rho}} \right|^2 \tag{A.16}$$

where now $r_{L,m,a}^{\rho}$ and $t_{L,m,a}^{\rho}$ accounts for the complex refractive index $\tilde{n}_m = n_m + i k_m$ and thickness t_m of the metal layer as in a homogeneous film [18]:

$$r_{L,m,a}^{\rho} = \frac{r_{L,m}^{\rho} + r_{m,a}^{\rho} e^{i\beta}}{1 + r_{L,m}^{\rho} r_{m,a}^{\rho} e^{i2\beta}}, \qquad t_{L,m,a}^{\rho} = \frac{t_{L,m}^{\rho} t_{m,a}^{\rho} e^{i\beta}}{1 + r_{L,m}^{\rho} r_{m,a}^{\rho} e^{i2\beta}} \tag{A.17}$$

where $\beta = k_0 \tilde{n}_m t_m \cos(\theta_m)$. Therefore, the external field after the metal contact is calculated using Eqs. (A.15) and (A.16) in Eq. (A.14).

The angular dependence of the external power per solid angle can be directly calculated as

$$\frac{dP_{\text{out}}^{(\gamma,\rho)}}{d\Omega dS}(\theta) = \mathcal{T}_m^{\rho}(\theta) \left| E_0^{(\gamma,\rho)}(\theta) \right|^2 \frac{n_{\text{out}}^3 \cos(\theta)^2}{n_L^3 \cos(\theta_L)^2} \tag{A.18}$$

This equation predicts the angular emission of the LED which is used to corroborate the theoretical and experimental results, as shown in later sections. The light extraction efficiency from the top surface in such structure is then calculated using Eq. (A.4), which can be easily done numerically.

A.3 Model for Light Extraction in a Simple LED Geometry

The analytical model presented in the previous section considered a monochromatic emission from the dipoles, but the QW emission in practical LEDs has a broad spectrum (typically 5–10 % of the central wavelength). In this section, the spectral broadening of the source is considered and, while the results are not much different from the monochromatic model, it allows us to simplify the analytical model for η_{extr}.

The effect of the QW lineshape is included in the previous model by averaging the extraction efficiency at different wavelengths with the normalized spectral emission $s(\lambda)$ from the QWs as $\eta_{\text{extr}}^{\text{poly}} = \int_0^{\infty} s(\lambda) \eta_{\text{extr}}(\lambda) d\lambda$.

The QW spectral emission can be approximated by a Gaussian function $s(\lambda) = \frac{1}{\sqrt{2\pi\sigma^2}} e^{-\frac{(\lambda - \lambda_c)^2}{2\sigma^2}}$ and the asymmetry of its lineshape can be taken into account by combining two Gaussian functions, with different variances σ^2, at their central
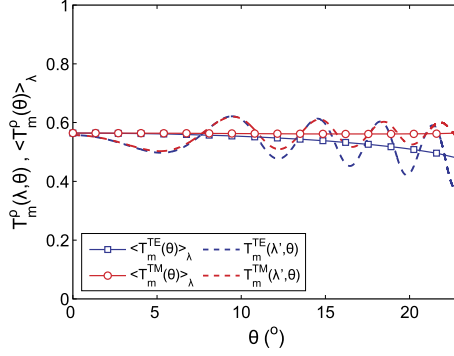
Fig. 6.11 Plot of $\mathcal{T}_m^\rho(\theta)$ for θ within the air cone ($\theta < \theta_c$) and $\lambda' = 445$ nm for TE (*dashdoted*) and TM (*dotted*), along with the plot of the averaged $\langle \mathcal{T}_m^\rho(\theta) \rangle_\lambda$ for TE (*squares*) and TM (*circles*) polarizations. The structure is composed of a 4869 nm-thick GaN-based LED over a sapphire substrate, and the QWs are 140 nm below the top surface. The 15.4 nm-thick metal contact is Ni–Au alloy with experimentally determined optical properties [14]

wavelength. The polychromatic extraction efficiency is thus explicitly written as

$$\eta_{\text{extr}}^{\text{poly}} = \sum_{\rho=\text{TE,TM}} \int_0^\infty s(\lambda) \int_0^{\frac{\pi}{2}} 2\pi \mathcal{T}_m^\rho(\theta,\lambda) \frac{dP_0^{(\gamma,\rho)}(\theta,\lambda)}{d\Omega dS} \sin(\theta) d\theta d\lambda \quad \text{(A.19)}$$

where $dP_0^{(\gamma,\rho)}(\theta,\lambda)/d\Omega dS$ is the external power per unit of solid angle of a structure without metal contacts and its dependence with λ is due to Fabry-Perot interferences on the interfaces of the LED layer, which is generally not very pronounced in most of LED structures (thick LED structures), allowing us to make the following approximation [13]:

$$\eta_{\text{extr}}^{\text{poly}} \simeq \sum_{\rho=\text{TE,TM}} \int_0^{\frac{\pi}{2}} \underbrace{\left[\int_0^\infty s(\lambda) \mathcal{T}_m^\rho(\theta,\lambda) d\lambda \right]}_{\langle \mathcal{T}_m^\rho(\theta) \rangle_\lambda} 2\pi \frac{dP_0^{(\gamma,\rho)}(\theta,\overline{\lambda})}{d\Omega dS} \sin(\theta) d\theta \quad \text{(A.20)}$$

for a fixed $\overline{\lambda}$ and only the metal transmission function is averaged with respect to λ. The averaged function $\langle \mathcal{T}_m^\rho(\theta) \rangle_\lambda$ varies very slowly with θ for both TE and TM polarizations (an example is shown in Fig. 6.11). This is because the reflection and transmission coefficients do not vary much for θ_L inside the air cone, which is small in the case of high refractive index semiconductors, such as GaN and GaAs. Thus we can approximate Eq. (A.20) by a *Variable-Incidence (VI) approximation*:

$$\eta_{\text{extr}} \simeq \langle \mathcal{T}_m^{\text{TE}} \rangle_{(\lambda,\theta)} \eta_{\text{extr}}^{0,\text{TE}} + \langle \mathcal{T}_m^{\text{TM}} \rangle_{(\lambda,\theta)} \eta_{\text{extr}}^{0,\text{TM}} \quad \text{(A.21)}$$

where $\langle \mathcal{T}_m^\rho \rangle_{(\lambda,\theta)}$ is the metal transmission function averaged in both λ and θ.

A simpler expression can be obtained by noticing that the averaged function $\langle \mathcal{T}_m^\rho(\theta) \rangle_\lambda$ is nearly constant inside the air cone for both TE and TM polariza-

tions (Fig. 6.11). Moreover, its value at $\theta = 0°$ is the same for both polarizations $\langle \mathcal{T}_m^{\mathrm{TE}}(0°) \rangle_\lambda = \langle \mathcal{T}_m^{\mathrm{TM}}(0°) \rangle_\lambda = \langle \mathcal{T}_m(0°) \rangle_\lambda$ (reflection and transmission coefficients are equal for both polarizations at normal incidence), thus we obtain the *Normal-Incidence (NI) approximation*:

$$\eta_{\mathrm{extr}} \simeq \langle \mathcal{T}_m(0°) \rangle_\lambda \, \eta_{\mathrm{extr}}^0 \qquad (A.22)$$

Notice that we started calculating the polychromatic extraction efficiency but ended with expressions for the monochromatic extraction efficiency. The spectrally averaged transmission function $\langle \mathcal{T}_m(0°) \rangle_\lambda$ is numerically calculated as

$$\langle \mathcal{T}_m(0°) \rangle_\lambda = \sum_{\lambda = \lambda_1}^{\lambda_2} \mathcal{T}_m(0°, \lambda) \frac{1}{\sqrt{2\pi\sigma^2}} e^{-\frac{(\lambda - \lambda_c)^2}{2\sigma^2}} \Delta\lambda \qquad (A.23)$$

where $\Delta\lambda = \frac{(\lambda_2 - \lambda_1)}{n-1}$, n is the number of discrete values of λ and the interval $[\lambda_1, \lambda_2]$ needs to be at least a few times larger than σ^2.

The determination of the extraction efficiency of any LED structure is very simple using one of the two approximative models above. The monochromatic extraction efficiency for the LED structure with metal contacts for any wavelength is simply obtained from the multiplication of the averaged transmission function of the metal, averaged also in the air cone or at $\theta = 0°$, and the monochromatic extraction efficiency of the LED without metal contact (η_{extr}^0).

η_{extr}^0 is not very sensitive to the LED parameters in this simple LED geometry (treated in Appendix B), which can be generally calculated for any LED structure for a given material. The most sensitive parameters in these models are those from the metal, which are contained solely in the metal transmission function $\mathcal{T}_m^\rho$ and can be separately calculated using Eq. (A.16). In the following section, we apply these models to the case of a GaN-based LED.

A.4 Determination of the Extraction Efficiency: Evaluation of η_{extr}, η_{extr}^0, $\mathcal{T}_m^\rho(\theta, \lambda)$ and $\langle \mathcal{T}_m(0°) \rangle_\lambda$

Let us evaluate the terms $\mathcal{T}_m^\rho(\theta, \lambda)$, $\langle \mathcal{T}_m(0°) \rangle_\lambda$ and η_{extr}^0 and compare the results of the extraction efficiency η_{extr} from the approximative models (Eqs. (A.21) and (A.22)) with the analytical model (Eq. (A.18)) to validate our approximations.

The transmission function $\mathcal{T}_m(\theta, \lambda)$ was averaged using Eq. (A.23) to obtain the approximative models (Eqs. (A.21) and (A.22)), for which the main assumption was that $\langle \mathcal{T}_m^\rho(\theta) \rangle_\lambda$ varied slowly with θ for both TE and TM polarizations. The evaluation of such function for the case of the GaN LED, for both polarizations, is presented in Fig. 6.11, which shows the plot of $\mathcal{T}_m^\rho(\theta)$ for a fixed $\lambda' = 445$ nm for TE (dashdoted) and TM (dotted), as well as the averaged $\langle \mathcal{T}_m^\rho(\theta) \rangle_\lambda$ for TE (squares) and TM (circles) polarizations, for θ within the air cone. The oscillations with respect
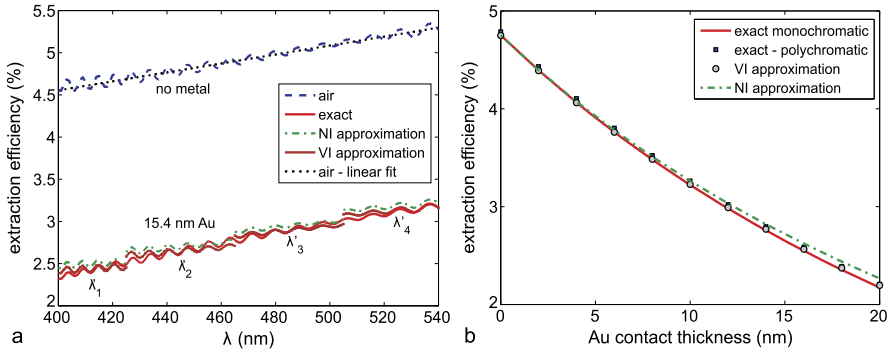
Fig. 6.12 (**a**) Extraction efficiency of the GaN LED structure versus wavelength using the analytical (*solid-red*), variable-incidence (VI) (*cross-brown*) and normal-incidence (NI) (*dashdotted-green*) approximations along with the extraction efficiency to air (*dotted-blue*) and its linear fit (*dashed-black*). The structure is composed of a 4869 nm-thick GaN-based LED over a sapphire substrate, and the QWs are 140 nm below the top surface. The 15.4 nm-thick metal contact is Ni–Au alloy with optical properties determined experimentally. Four different emission wavelengths were considered, $\lambda_1' = 405$ nm, $\lambda_2' = 445$ nm, $\lambda_3' = 485$ nm and $\lambda_4' = 525$ nm in the function $s(\lambda)$ to evaluate the approximative models. (**b**) Extraction efficiency of the GaN LED structure versus metal thickness using the analytical with a monochromatic (*solid*) and polychromatic (*squares*) emission, VI (*circle*) and NI (*dashdoted*) approximations (Color figure online)

to λ observed in $\mathcal{T}_m^\rho(\theta)$ are completely eliminated when this function is averaged with the lineshape of the QWs. As a matter of fact, the averaged function is nearly constant for both polarizations, within the air cone, which supports both approximations made to obtain Eqs. (A.21) and (A.22). This is due to the weak variation of the reflection and transmission coefficients inside the air cone which, in its turn, is small in the case of high refractive index semiconductors, such as GaN and GaAs.

Let us now calculate the extraction efficiency to air η_{extr}^0 (without metal contact) for the GaN LED treated here, which is shown in Fig. 6.12a (dashed line). The small oscillations in extraction efficiency to air with λ for a monochromatic emission average out in real devices with polychromatic (poly) emission—note that the oscillation period in this example is around 7 nm, which is substantially less than the typical 25 nm EL linewidth for blue GaN-based LEDs. As shown later, the extraction efficiency to air, through the top surface of the LED, depends weakly on the LED 'fine' structure (for a simple LED geometry), therefore we can determine a general extraction efficiency to air for GaN LEDs grown in sapphire substrates by a linear fit of η_{extr}^0, which is represented by the dashed line in Fig. 6.12a:

$$\eta_{\text{extr}}^0(\lambda) = 5.31 \times 10^4 \lambda + 2.43 \times 10^{-2} \tag{A.24}$$

To validate the models, we calculate the extraction efficiency of the GaN LED structure using the analytical model (Eq. (A.18)) and compare the results to the ones from the approximative models.

The extraction efficiency of the GaN LED structure after the metal contact (η_{extr}) versus wavelength calculated using the analytical model is shown in the solid-red

curve in Fig. 6.12a, along with the results of the VI (cross-brown) and the NI (dashdoted-green) approximations. Four different emission wavelengths were considered in this plot, $\lambda' = 405, 445, 485$ and 525 nm, which were used in the function $s(\lambda)$ to evaluate the approximative models.

The application of both models for same structure when the metal thickness is varied is shown in Fig. 6.12b, where the circles correspond to the VI and the dash-doted line corresponds to the NI approximations. Again, there is a very good agreement with the analytical results (solid line).

These results were compared to the polychromatic extraction efficiency (η_{extr}^{poly}) taking into account the lineshape of the QWs, which was calculated analytically from Eq. (A.19) and represented in Fig. 6.12b by the full squares. The polychromatic η_{extr}^{poly} is very similar to the monochromatic results which is due to the peaked QW emission at the center wavelength (variance σ^2 is small compared to the center wavelength). The polychromatic extraction efficiency to air $\eta_{extr}^{poly,0}$ (corresponding to 0 nm of metal contact) agrees well with the monochromatic η_{extr}^0 calculated at the center wavelength of the QW linewidth.

Appendix B: Sensitivity of Model to LED Parameters

We present in this section, the investigation of the sensitivity of the calculated extraction efficiency to LED parameters, revealing that for large range of LED thicknesses and QW positions within the LED, the extraction efficiency through the top facet does not change significantly. Indeed the estimation of the extraction efficiency is mostly, if not solely, affected by the metal contact properties. Moreover, the sensitivity of our model to different metal contacts is presented, which shows that the appropriate choice of metal contacts reduces significantly the dependence of the modelling results on the LED parameters.

Let us first present the effect of the LED thickness on the calculated extraction efficiency η_{extr}. Figure 6.13a shows the η_{extr} of a GaN LED on sapphire substrate for $\lambda = 445$ nm, where the QW is positioned at a fixed 140 nm distance from the top surface. Four cases of metal contacts are considered: no metal, Au, Ag and Al, where the metal thickness is 15.4 nm. In all cases, the extraction efficiency oscillates significantly for an LED structure thinner than 1 µm, and tends to a fixed value for thicker structures. In practical LEDs, the fast oscillating behavior shown in Fig. 6.13a is averaged due to thickness fluctuations and polychromatic emission from the source which yields a much smoother function almost independent on LED thickness. Suffice to notice that the half period of such oscillations is ~50 nm for a fixed $\lambda = 445$ nm, which is much shorter than the thickness fluctuations present in real devices. A more general map of the extraction efficiency is shown in Fig. 6.4b, where an LED structure without metal contacts was considered, showing that the extraction efficiency η_{extr}^0 is fairly constant for thick LED structures.

Let us investigate the effect of the LED parameters on the total power emitted from the source inside the LED cavity. In the models presented in this chapter, the
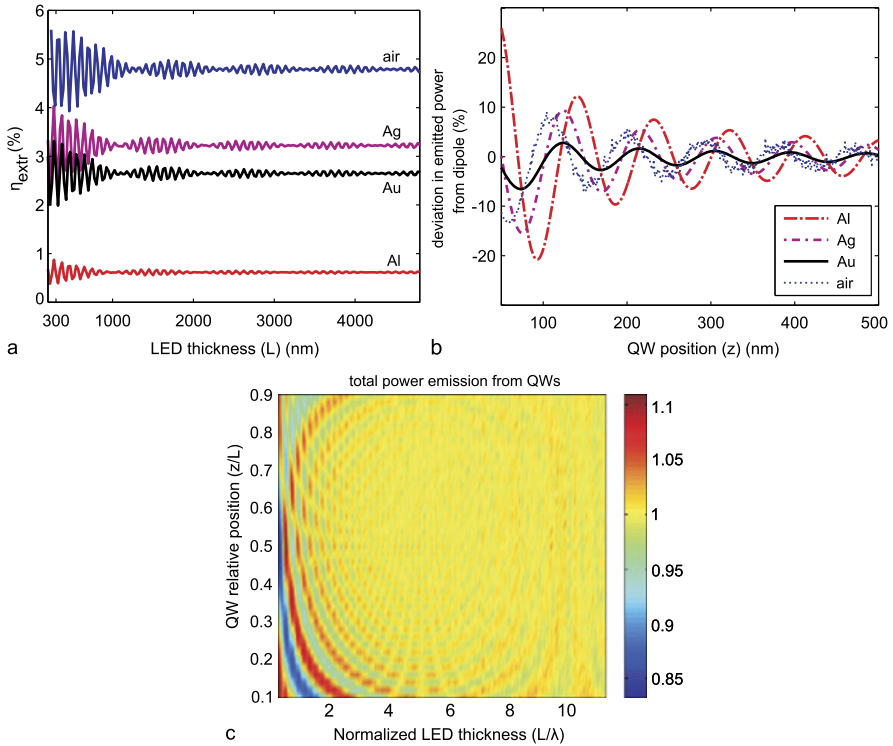
Fig. 6.13 (**a**) Extraction efficiency as a function of LED thickness for a QW positioned at a fixed 140 nm from the top surface for four cases of metal contacts: no metal, Au, Ag and Al and all metals are 15.4 nm-thick. (**b**) Deviation from unity of total radiated power by the source within the LED as a function of the QW position for the same metals. (**c**) Total radiated power by the dipole source inside the LED as a function of LED thickness and QW position

dipole terms were normalized to yield a unity emitted power in a homogeneous medium. When these dipoles are placed inside the LED heterogeneous medium, or in an optical cavity, the total power emitted from the dipole source may vary because its amplitude is kept constant but the optical medium is modified. This effect is related to a change in radiative emission rate from a source within an optical cavity, or Purcell effect [30]. One assumption made in our models was that the total emitted power from the dipole sources inside the LED would be considered unity due to a negligible Purcell effect in common LED structures.

Here, we test this assumption by calculating the deviation from unity of the total emitted power by the dipole sources inside the LED cavity. Let us consider the effect of the most sensitive LED parameter which is the distance of the QWs to the top surface. The deviation of the total emitted power from unity by the dipole sources inside the LED structure calculated as a function of the QW distance to the top surface z for a total LED thickness of 4729 nm is shown in Fig. 6.13b. The same four cases of metal contacts are considered: no metal, Au, Ag and Al, where the

metal thickness is 15.4 nm and it is interesting to notice that the metal contact plays a significant role in this case.

The deviation from unity in total emitted power due to a cavity effect can be significantly reduced by judiciously choosing the metal contacts, for example in the case of Au, it is at most 2 % for realistic LED structures ($z > 150$ nm). In case other metals such as Al or Ag are used as contacts, the deviation given in this plot can be used as a corrective factor for the theoretical extraction efficiency. The total emitted power tends to unity as the QW is placed farther from the LED top surface. While the oscillations observed in Fig. 6.13b are only slightly averaged to the polychromatic emission from the source, the use of several QWs or thick active regions, as well as thickness fluctuations of the LED structure, average these oscillations resulting in a smaller deviation of the emitted power from unity.

A more general map of the deviation from unity of the power emitted from the dipole sources inside the LED is shown in Fig. 6.13c, where an LED structure without metal contacts was considered. As can be seen, the total emission from the dipole source is fairly constant for a thick LED structure ($(L/\lambda) > 5$).

Therefore, under appropriate choice of metal contacts, the technique presented in this chapter is robust to LED parameters and the results presented in Sect. 6.3.1, or more specifically shown in Fig. 6.4 can be generally applied to a large range of LED configurations.

Appendix C: Modelling the Angle-Resolved Emission from LEDs: Accounting for Surface Roughness

The model presented in this chapter (details in Appendix A) can be used to predict the angular emission of an LED structure, as shown in Fig. 6.6. In particular, it is useful to check whether light emission occurs according to theoretical predictions. The power per solid angle of an LED can be experimentally measured using an angle-resolved setup [31] and compared to the theoretical angle-resolved emission, as demonstrated in Ref. [14]. The light radiated to air from the LED structure corresponds to $dP_{\text{out}}^{(h,\text{total})}/d\Omega dS(\theta) = dP_{\text{out}}^{(h,\text{TE})}/d\Omega dS + dP_{\text{out}}^{(h,\text{TM})}/d\Omega dS$. To compare this angular power flow to the corresponding angle-resolved measurement, a correction term $\cos(\theta)$ needs to be used to take into account the projection of the power flow (perpendicular to the LED top surface) into the plane perpendicular to the rotating detector.

Figure 6.14a shows the radiation from the LED (total power per solid angle) versus θ and λ, where the oscillations observed correspond to the Fabry-Perot interferences on the interfaces of the LED structure. While experimental results for LEDs with flat surface match such angular emission (Fig. 6.6a), LEDs with rough surface or pits present diminished oscillations in their angular diagram. The effect of surface roughness can be considered as a damping on the Fabry-Perot oscillations through the assumption that the roughness randomizes the angular distribution of the power flow. Therefore, a fraction F_{rough} of the original power flow is now randomly emitted in all angles, which can be approximated by an isotropic emission
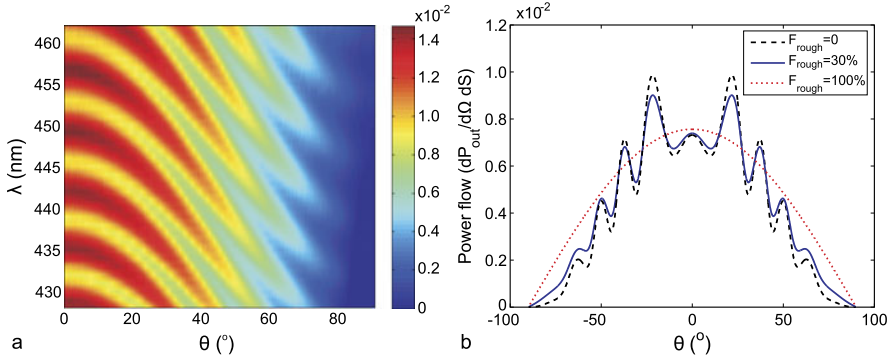
Fig. 6.14 (**a**) Theoretical angle-resolved radiation ($dP_{out}^{(h,\text{total})}/d\Omega dS$) versus θ (for θ from 0° to 90°) and λ showing the Fabry-Perot interferences at the interfaces of the LED structure. The LED structure consisted of a 4.87 μm-thick GaN with QW embedded at 140 nm below the top surface. (**b**) Effect of surface roughness on the angular power flow, normalized by the total power emitted to air, for a smooth LED with $F_{\text{rough}} = 0$ (*dashed line*), for $F_{\text{rough}} = 30$ % (*solid line*) and for $F_{\text{rough}} = 100$ % (*dotted line*))

normalized by the total power emitted to air η_{extr}, as $R = \eta_{\text{extr}}/2\pi$. A more rigorous investigation of the effect of rough surfaces is presented in Refs. [32, 33]. The normalized angular output power flow due to the rough surface is approximated by

$$\frac{F_{\text{rough}}R + (1 - F_{\text{rough}})dP_{out}^{(h,\text{total})}/d\Omega dS}{\eta_{\text{extr}}} \tag{C.1}$$

Fig. 6.14b illustrates this effect, where the dashed line is the corrected angular power flow, normalized by the total power emitted to air, for a smooth LED with $F_{\text{rough}} = 0$. The solid line corresponds to $F_{\text{rough}} = 30$ % and the dotted line is for $F_{\text{rough}} = 100$ %. The angle-resolved measurement is a useful technique to assess the contribution of light randomly scattered at rough surfaces to the measured EQE, which ultimately can be used to validate the model requirements of flat interfaces. The pronounced Fabry-Perot oscillations in this measurement fade with increase of F_{rough} which largely reduces the accuracy of this model.

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Chapter 7
Active Region Part B.
Internal Quantum Efficiency

Jong-In Shim

Abstract In the first part, we introduced experimental results of the internal quantum efficiency (IQE) droop depending on temperature in both the electroluminescence and the resonant photoluminescence. The IQE droop mechanisms ever reported are reviewed. An inherent origin of the efficiency droop is suggested to be the saturation of the radiative recombination rate in the InGaN quantum well at low current and subsequent increase in the nonradiative recombination rates at high current. The degree of saturation is largely influenced by operating temperature and the effective active volume. Although the saturation of the radiative recombination rate is the common origin of the IQE droop, the shapes of the IQE versus current, i.e. the IQE droop curve, vary with the dominant nonradiative recombination process. In the second part, we have reviewed the IQE measurement methods theoretically as well as experimentally. A simple IQE estimation method based on the constant ABC model in the carrier rate equation is introduced in terms of its convenience and application limitation. Other methods have been also reviewed by focusing on all-optical methods such as the temperature-dependent photoluminescence (TDPL) and the temperature-dependent time-resolved photoluminescence (TD-TRPL) methods.

7.1 LED Efficiency

InGaN light-emitting diodes (LEDs) have been the focus of intense research in recent years. Their applications are now in many areas such as traffic lights, mobile phones, automotives, display units, and general lighting. The application expansion capacity depends on improvements of both the efficiency and the reliability, especially for the areas requiring high brightness LEDs like general lighting.

Usually, a LED is electrically driven by a battery producing the voltage of V and the forward current of I and the total electrical power supplied is VI. The LED operation is more clearly understood when we express VI as $(qV)(I/q)$ where q is the unit elementary charge of 1.6×10^{-19} [C], I/q is the total number of electrons

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injected into the active layer per second, and qV is an electrical potential energy of each electron. In the ideal case, each electron energized by a battery emits one photon without any energy loss so that both quantum particles should have the same energy of qV. In a real case, however, there are many sorts of electrical and optical energy loss mechanisms during the electrical-to-optical energy conversion process.

The overall efficiency of a LED is characterized by the wall-plug efficiency ($\eta_{\text{wall-plug}}$), which is defined by the ratio of the optical power emitted into the free space from the LED to the electrical power provided to the LED [1]. Then, the wall-plug efficiency can be rewritten in detail by utilizing the voltage efficiency (η_{VTG}), the injection efficiency (η_{INJ}), the radiative efficiency (η_{RAD}), and the light extraction efficiency (η_{LEE}) in terms of energy conversion processes from a battery to a detector in free space as

$$\eta_{\text{wall-plug}} = \eta_{\text{VTG}} \times \eta_{\text{INJ}} \times \eta_{\text{RAD}} \times \eta_{\text{LEE}}.$$

η_{VTG}, η_{INJ}, η_{RAD}, and η_{LEE} are defined as follows:

$$\eta_{\text{VTG}} = \frac{\text{average photon energy emitted from the active region}}{\text{average electron energy supplied by the battery}} = \frac{h\bar{\nu}}{qV} \tag{7.1}$$

$$\eta_{\text{INJ}} = \frac{\text{\# of electrons injected into the active region per second}}{\text{\# of electrons injected into a LED per second}} = \frac{I_{\text{active}}/q}{I/q} \tag{7.2}$$

$$\eta_{\text{RAD}} = \frac{\text{\# of photons emitted from the active region per second}}{\text{\# of electrons injected into the active region per second}} = \frac{P_{\text{active}}/(h\bar{\nu})}{I_{\text{active}}/q} \tag{7.3}$$

$$\eta_{\text{LEE}} = \frac{\text{\# of photons emitted into the free space per second}}{\text{\# of photons emitted from the active region per second}} = \frac{P_{\text{out}}/(h\bar{\nu})}{P_{\text{active}}/(h\bar{\nu})} \tag{7.4}$$

where P_{out}, V, I, $h\bar{\nu}$, I_{active}, and P_{active} are the optical power emitted from a LED to the free space, the applied voltage, the forward current, the average energy of a photon, the current injected into the active region, and the radiated optical power from the active region, respectively.

The voltage efficiency (η_{VTG}) is a measure of the electrical potential energy loss during electron transport from a battery to the active region of a LED. The average energy of one photon emitted from a LED is almost equal to the bandgap energy of the active region of a LED. The losses in electrical potential energy mainly results from the Ohmic losses occurring at the metal and semiconductor interface, semiconductor hetero-interfaces, and semiconductor bulk materials. In addition, the electrical potential loss is inevitable at the quantum-well (QW) active region during the carrier capture process from the three-dimensional bulk region to the two-dimensional QW. In order to improve the voltage efficiency, both a low forward turn-on voltage and a small series resistance are required. The voltage efficiency is experimentally measurable so that the value can be improved by a sort of feedbacks to fabrication processes and LED chip designs.

The injection efficiency (η_{INJ}) is a measure of how many electrons recombine in the active QW region compared to the total injected electrons into a LED. It depends on the current level as well as the LED structure itself. The efficiency is determined by nonradiative recombination rates occurring outside the active QW

region. Possible leakage currents are the semiconductor surface current, the defect-related tunneling current, and the electron overflow from the QW active region to the p-clad region [2–4]. The surface leakage current is observed at around zero bias voltage and shows a very symmetric current-voltage curve for forward and reverse biases. Special surface treatment and passivation techniques have been utilized to suppress the surface leakage current. The electron blocking layer in the epitaxial growth has been introduced to reduce the electron overflow from the active region to the p-type clad layer at high-level current injection [5]. Current spreading is important to improve the current injection efficiency via reducing the current density over an entire LED surface. The nonuniform current injection increases both the Joule heating and the nonradiative recombination processes and eventually lowers the wall-plug efficiency [6, 7].

The radiative efficiency (η_{RAD}) is defined as the ratio of the radiative recombination rate to the total recombination rate, i.e. the sum of the radiative and nonradiative recombination rates, in the active multiple-quantum-well (MQW) layers. In order to improve the radiative efficiency, it is necessary to increase the radiative recombination rate and decrease the nonradiative recombination rate. High-quality epitaxial growth with low defects is essential to reduce the nonradiative recombination rate. The radiative recombination rate is determined basically by the "Fermi's golden rule" and the total number of recombining electron-hole pairs [8]. Elimination of the piezoelectric field is very effective to increase the recombination probability of an electron-hole pair and the large active volume is one of the best ways to increase the total number of carriers contributing to the radiative recombination.

The light extraction efficiency (LEE, η_{LEE}) is a measure of the photon losses during the propagation from the active region into the free space. In an ideal LED, all photons emitted by the active region should escape from the LED die. However, in a real LED, not all the photon emitted from the active region is emitted into the free space. This is due to the phenomenon of total internal reflection and several possible optical loss mechanisms. The total internal reflection of GaN-based LEDs results from the large difference in refractive index between a GaN film ($n = 2.4$) and surrounding air ($n = 1.0$). The light reaching the surface beyond the critical angle will undergo the total internal reflection and will continue to be reflected within the LEDs until it is absorbed. Thus, much of the light generated by the conventional LEDs will continue to be reflected until it is absorbed. The light can be absorbed by the substrate, epitaxial layers, and the metallic layer via free-carrier absorption or band-to-band absorption. One method of reducing the percentage of total internal reflection of light is to generate light-scattering centers inside or outside the LED chip. The methods improving the light extraction efficiency are known as angular randomizations by scattering of photons via random texturing of the LED's surfaces, artificial voids inside the epitaxial layers, micro patterning of the sapphire substrate or epitaxial layers, chip shaping of multiple escape cones, and the coupling of photons to surface plasmons [9, 10].

The internal quantum efficiency (IQE, η_{IQE}) is defined as the product of the injection efficiency and the radiative efficiency and is the ratio of the number of the photons emitted from the active region to the number of electrons injected into a

LED. The IQE and the LEE are two principal parameters for improving the efficiency of LEDs; both of which can be improved through the improvement of crystal quality and the modification of the structure of the epitaxial layers. The external quantum efficiency (EQE, η_{EQE}) is defined as the product of the IQE and the LEE and is the ratio of the number of useful photons to the number of injected electrons.

$$\eta_{IQE} = \frac{\text{\# of photons emitted from the active region per second}}{\text{\# of electrons injected into the LED per second}}$$

$$= \left(\frac{I_{active}/q}{I/q}\right)\left(\frac{P_{active}/h\bar{\nu}}{I_{active}/q}\right) = \left(\frac{P_{active}/h\bar{\nu}}{I/q}\right) = \eta_{INJ} \times \eta_{RAD} \quad (7.5)$$

$$\eta_{EQE} = \frac{\text{\# of photons emitted into the free space per second}}{\text{\# of electrons injected into the LED per second}}$$

$$= \left(\frac{I_{active}/q}{I/q}\right)\left(\frac{P_{active}/h\bar{\nu}}{I_{active}/q}\right)\left(\frac{P_{out}/h\bar{\nu}}{P_{active}/h\bar{\nu}}\right)$$

$$= \left(\frac{P_{out}/h\bar{\nu}}{I/q}\right) = \eta_{INJ} \times \eta_{RAD} \times \eta_{LEE} \quad (7.6)$$

We summarized seven types of efficiencies used in the LED society, i.e. the wall-plug efficiency ($\eta_{wall\text{-}plug}$), the voltage efficiency (η_{VTG}), the injection efficiency (η_{INJ}), the radiative efficiency (η_{RAD}), the light extraction efficiency (η_{LEE}), the internal quantum efficiency (η_{IQE}), and the external quantum efficiency (η_{EQE}). Among these seven efficiencies, however, only three efficiencies, $\eta_{wall\text{-}plug}$, η_{VTG}, and η_{EQE} are experimentally measurable but others, η_{INJ}, η_{RAD}, η_{IQE}, and η_{LEE} are hardly measurable by experiments. Among these efficiencies, the IQE is the most important because it is directly related to the crystal growth technology. Once we know the IQE value, then the LEE can be known from the measured EQE. In the following sections, we first review the IQE characteristics focusing on the droop and then present the IQE measurement methods.

7.2 Efficiency Droop Mechanisms

7.2.1 Efficiency Droop Overview

A major challenge for III-nitride LEDs is delivering the highest efficiency performance at the current density and temperatures relevant to high-power operation. The EQE efficiency breakthroughs are due to improvements in both the IQE and the LEE. Since the LEE is nearly independent of the operational environments, the variation of the EQE is attributed to that of the IQE. Typically, the IQE shows a maximum at low injection current and then decreases as current is further increased, the so-called efficiency (or IQE) droop. Experimentally, the IQEs are strong functions of emission wavelength, current density, and temperature as sketched in Fig. 7.1. InGaN-based LEDs operating especially at blue-green wavelengths suffer from very large efficiency droop. The IQE droop is currently a limiting factor to more efficient

Fig. 7.1 IQEs of red, green, blue, and near-UV LEDs as a function of current at room temperature. All LEDs are nearly the same sizes of $300 \times 300\ \mu m^2$. IQEs are estimated by an analysis method based on the carrier rate equation
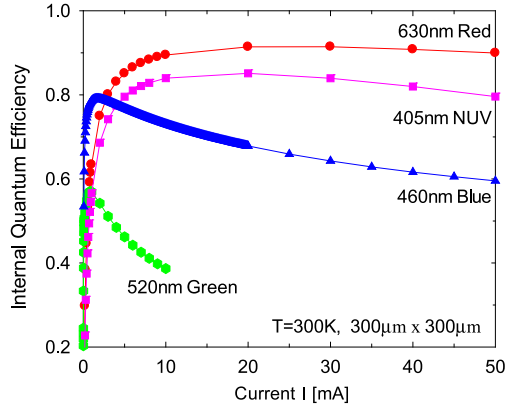
LEDs under high current. Understanding origins of the IQE droop is an imperative to reduce the droop and finding the IQE measurement method for the future of general lighting.

The following experimental IQE behaviors are commonly observed: The IQE droop of III-nitride LED is observed (1) in both blue and green LEDs [11–13] (2) in both resonant photoluminescence (PL) and electroluminescence (EL) measurements with very similar excitation dependences [12, 25] (3) under pulsed, as well as continuous-wave (CW), current injections [14, 15], (4) in polar, non-polar, and semi-polar QW structures [16–19], (5) for single QW, MQW, and bulk active layers [13, 16, 20], (6) at all temperatures from 4 to 450 K [11–18], and (7) less severe in short wavelengths (near-UV) and more pronounced at longer wavelengths (green-yellow) [21].

The IQE defined in Eq. (7.5) can be rewritten by the ratio of the radiative recombination current I_r in the active region to the total injection current I as shown in Eq. (7.7). The total current I consists of the radiative recombination current and the nonradiative recombination current I_{nr}. At room temperature, electrons and holes in the active region can be considered as free carriers and they recombine radiatively according to the Fermi's golden rule. There are three types of recombination currents around the active region as shown in Fig. 7.2: (1) the electron overflow current recombining radiatively and nonradiatively in p-GaN, (2) the electron tunneling leakage current from the active QW to the p-side, and (3) the recombination currents in the active QW via the Shockley-Read-Hall (SRH), the radiative, and the Auger recombination processes.

$$\eta_{IQE} = \eta_{INJ} \times \eta_{RAD} = \frac{I_r}{I} = \frac{I_r}{I_r + I_{nr}} = \frac{1}{1 + I_{nr}/I_r} \qquad (7.7)$$

where $I_{active} = \eta_{INJ} I$ and $I_r = \eta_{RAD} I_{active} = \eta_{RAD} \eta_{INJ} I = \eta_{IQE} I$.

In general, the optical power P_{active} from the active QW is a monotonically increasing function of total current I and P_{active} is linearly proportional to the radiative current I_r. Thus, we know from Eq. (7.7) that the IQE droop happens only when I_{nr} increases faster than I_r with increasing total current I. This implies that the amount

Fig. 7.2 Schematic illustration of current components in an InGaN-based LED. The radiative current in *p*-GaN stands for the electron recombination current through defects resulting from the Mg impurity doping
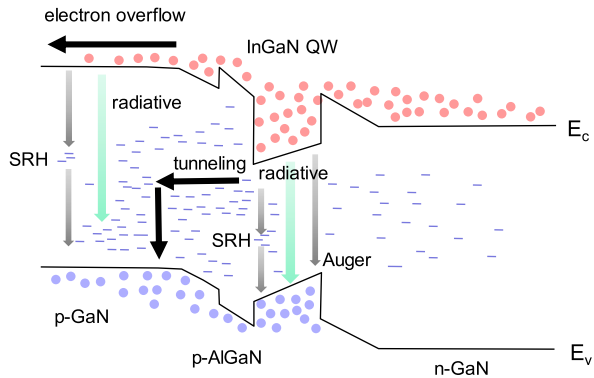
of IQE droop is mainly determined by I_{nr} rather than I_r, which is the reason why the nonradiative recombination processes has been considered as a dominant mechanism of the IQE droop.

Many possible explanations have been proposed to explain the efficiency droop, including local junction heating [14], carrier overflow enhanced by internal polarization field [15], poor hole injection [22, 47], or asymmetric electron and hole distributions in the active region [23], carrier delocalization from indium-rich regions and nonradiative recombination at high defect sites [4, 24], and Auger recombination [10–13, 25–27]. All previous IQE droop explanations were based on additional increases in the nonradiative recombination rates with increasing current. More recently, saturation of the radiative recombination rate and subsequent increase of the nonradiative recombination rate was proposed by us in order to explain the IQE droop phenomena more comprehensively including their temperature dependences [28, 29]. Nevertheless, a consensus on the mechanism behind the IQE droop remains elusive [30–33]. Here, we first review the proposed IQE droop mechanisms by comparing experimental IQE behaviors and then explain the importance of the radiative recombination rate on the IQE behaviors depending on current as well as temperature.

7.2.2 Auger Nonradiative Recombination

To find the origin of the EL efficiency droop, the PL efficiency is frequently utilized with different operating temperatures and optical pumping powers. In the resonant PL experiments, the photon energy of the optical pumping source is between the bandgap energy of QWs and that of barriers to selectively excite the carriers only in the QWs. The experimentally observed PL efficiency droop shows the dependences on the carrier density and the operating temperature quite similar to those of the EL efficiency droop [11–13, 15, 34]. This observation is consistent with intrinsic recombination mechanisms in the active QW itself, such as Auger recombination

and carrier escape from the QWs, because the carrier transport issue is not involved in the resonant PL experiments.

Nonradiative electron-hole recombination processes transfer the excess electron energy to other particles except photons. In case of direct Auger recombination, these other particles are electrons or holes that are excited into higher energy levels within the same band. The theoretical probability of this Auger process decreases exponentially with increasing bandgap energy E_g and decreasing operating temperature T, i.e. $C \propto \exp(-E_g/k_B T)$ where C is the Auger coefficient and k_B is the Boltzmann constant [8]. Therefore, the importance of Auger recombination in III-nitride semiconductors has been in doubt for a long time. However, similar efficiency droop curves in the resonant PL and EL experiments invoked Auger process as the dominant nonradiative mechanism leading to efficiency loss at high carrier density. Efforts to verify the Auger recombination have been carried out both theoretically and experimentally.

Several theoretical groups calculated the III-nitride Auger coefficients [25–27]. Hader et al. computed a very small Auger coefficient C of 3.5×10^{-34} cm^6/s using the 8×8 $\vec{k} \cdot \vec{p}$ band model for the direct band-to-band Auger process and concluded that this value is too small to explain the experimentally observed droop [25]. Delaney et al. calculated a peak Auger coefficient of 2×10^{-30} cm^6/s in bulk InGaN with a 2.5 eV bandgap ($\lambda_g = 495$ nm) using first-principle density-functional and many-body-perturbation theory and concluded that an interband Auger may be responsible for droop in InGaN LEDs [26]. But the probability of this interband Auger process decreases rapidly with changing bandgap and it seems to be insufficient to explain the efficiency droop observed across a wide wavelength range. In addition, Pasenow et al. proposed the phonon-assisted Auger recombination as one of the candidates for explaining the droop [27]. But some people still argued that the phonon-assisted Auger process is typically not strong enough to generally explain the droop [24, 30].

The influence of Auger recombination is typically estimated by using the constant ABC model in the semiconductor carrier rate equation. The C coefficient in the rate equation stands for the nonradiative recombination rate due to Auger processes. Table 7.1 gives an overview of recombination coefficients for InGaN QWs extracted from measurements using the ABC model [31, 32]. Reported Auger coefficients at room temperature in InGaN-based LEDs range from 3.5×10^{-31} cm^6/s to 3.2×10^{-29} cm^6/s. These experimental values are generally larger than theoretical values of about 10^{-34}–10^{-30} cm^6/s. In fact, it is not easy to obtain exact values of these three fit parameters simultaneously by only one fit process. Thus, most of works fit the A and B coefficients in the low current injection regime and the C coefficient in the high current injection regime. It should be noted that all these measurements neglected the influence of carrier leakage, i.e. $\eta_{\text{INJ}} = 1$.

Recently, we successfully developed an analysis method of the IQE and each recombination current at arbitrary current based on the constant ABC model [39]. A similar method was also proposed in the reference [37]. We managed to cut the number of fitting parameters from three to one by manipulating the rate equations so that the fitting errors can be minimized. Our method is applied to InGaN-based blue and green LEDs and their IQE curves are relatively well-fitted by the

Table 7.1 Carrier recombination A, B, and C coefficients of InGaN-based LEDs

Wavelength λ [nm]	A [1/s]	B [cm^3/s]	C [cm^6/s]	Reference
440	5.4×10^7	2.0×10^{-11}	2.0×10^{-30}	Shen et al. [25]
407	1.0×10^7	2.0×10^{-11}	1.5×10^{-30}	Zhang et al. [35]
450	2.3×10^7	1.0×10^{-11}	1.0×10^{-30}	Meneghini et al. [36]
523	0.47×10^7	1.2×10^{-12}	3.5×10^{-31}	Laubsch et al. [13]
444	–	1.0×10^{-10}	8.0×10^{-29}	Dai et al. [37]
445	1.0×10^7	1.9×10^{-10}	3.2×10^{-29}	Özgür et al. [38]
440	$(8.13\text{–}15.6) \times 10^7$	1.0×10^{-10}	–	Schubert et al. [44]

constant ABC model. However, the estimated range of C coefficients is around 10^{-28}–10^{-25} cm^6/s, which is at least 100 times higher than the reported C coefficients of 10^{-30} cm^6/s. To date, there are no theories explaining such large C coefficients. Thus, it is plausible that the carrier loss process with a N^3-dependence, where N is the carrier density, could represent other processes than the Auger processes.

7.2.3 Defect-Related Nonradiative Recombination

The InGaN material system has lots of defects compared to other III-V semiconductors. Since the early stage, attention has been paid to the defect-related recombination mechanisms as another origin of the efficiency droop. They include carrier delocalization from local in-plane potential minima and nonradiative recombination at high defect sites and defect-assisted tunneling of carriers out of the active region [4, 24, 40–43].

The idea of surprisingly high radiative recombination rate at low current is based on the early observation of randomly distributed potential minima, possibly nonuniform distributions of Indium-rich clusters and/or micro-stresses, inside InGaN QWs. Potential minima are associated with a lower bandgap and therefore lead to carrier localization. At low current and low QW carrier density, potential minima then keep carriers away from structural defects. With higher current, however, carriers gradually escape from these potential minima due to carrier-carrier screening and increasingly recombine nonradiatively at defects. This IQE droop mechanism can be modeled theoretically by assuming that the SRH recombination coefficient A itself exhibits a superlinear rise with the QW carrier density N [24]. On the other hand, the IQE droop due to the defect-assisted tunneling of carriers out of the active region should be modeled by the injection efficiency varying with the carrier density.

Similar to Auger recombination processes, however, additional increase of nonradiative recombination rate due to defect-related nonradiative recombination

mechanisms in the MQW active region cannot fully explain the temperature-dependent characteristics of the IQE droop [11, 16, 28, 29]. As the operating temperature is reduced, the amount of droop in both the resonant PL efficiency and the EL efficiency is observed to increase. But it is expected that the nonradiative recombination at defects working as the nonradiative recombination centers (NRCs) in the MQW active region decreases as temperature decreases. Thus, larger efficiency droop at lower temperature is hard to be explained only by the defect-related droop mechanisms without the aids of other temperature-dependent recombination mechanisms.

7.2.4 Transport-Related Nonradiative Recombination

Unbalanced carrier transport mechanisms have also been suspected as a dominant cause resulting in the poor current injection efficiency and consequently the IQE droop [22, 23, 38, 44–48]. It is still controversial whether electron leakage out of the active region or ineffective hole injection into the active region is the cause of eventual electron leakages from the active region to the p-side region. There are two cores in this model. One is a parasitic nonradiative carrier recombination on the p-side of the device. The other is that the carrier capture process from the bulk region to the QW active region is assumed to be insufficient. The piezoelectric field, low hole concentration in the p-side, asymmetry in electron and hole distributions, and quasi-ballistic electron transport are proposed as the background deterioration mechanisms of the carrier injection efficiency defined in Eq. (7.2). It was reported that the IQE droops were alleviated by introducing special LED structures such as the electron blocking layer (EBL) [48, 49] and the polarization controlled MQW active region [45, 46]. Although the carrier-transport-related droop model cannot completely explain all the experimental droop phenomena, it seems evident that the model is useful in improving the IQE droop [45–59].

While there is a difference in the degree of efficiency droop depending on temperature, the EL efficiency droop is observed not only in InGaN-based blue and green LEDs, but also in AlGaInP-based red LEDs grown on GaAs substrates [60]. It is noticeable that the AlGaInP/GaAs material system is quite different from the InGaN-based material system in that (1) the AlGaInP epitaxial system is perfectly lattice-matched to the GaAs substrate, (2) the active volume of the AlGaInP red LED is more than 10 times larger than that of the InGaN blue or green LED, and (3) the AlGaInP material is not subject to the internal electric field caused by the spontaneous and piezoelectric polarizations. Moreover, it was reported that InAs/GaAs quantum-dot (QD) structures grown on the GaAs substrate show temperature-dependent EL behaviors at 1.3 μm, which are very similar to those of InGaN-based blue and green LEDs [61]. The efficiency droop also increases with decreasing temperature or QD size. From these results, we suspect that the efficiency droop is not a peculiar phenomenon existing only in the InGaN-based material system. Thus, it is plausible that the carrier recombination mechanism governing the efficiency droop is not the

unique one that is possible only in a special epitaxial structure or material system. Consequently, it is natural that the increase in nonradiative recombination rate assisted by the piezoelectric field is not a key mechanism of the efficiency droop because there is almost no piezoelectric field in both AlGaInP and InAs/GaAs material systems.

On the other hand, as the operating temperature is reduced, the hole concentration in GaN-based LEDs drastically decreases due to high activation energy of the hole, leading to poor hole transport in the MQWs [23, 60, 62]. Therefore, carriers typically recombine in MQWs near the *p*-GaN clad. The asymmetric carrier distribution originating from the poor hole transport may promote the carrier overflow from the MQW region [68–70]. In this sense, the UV LEDs incorporating the *p*-type AlGaN are expected to show more carrier asymmetry than the *p*-GaN used in blue and green LEDs. However, the UV LEDs show less efficiency droop than the blue and green LEDs. Moreover, LEDs with a single QW or bulk active layer also show the efficiency droop. This implies that all experimental IQE droop results cannot be consistently explained only by one of the transport-related mechanisms.

Up to now, none of the above droop model is generally accepted. Auger and defect-related processes consider additional increase in the nonradiative recombination process inside the QW active region with sufficient carrier capture rates into the QW active region. On the other hand, carrier-transport-related models suspect additional increase in the nonradiative recombination outside the QW active region such as in the *p*-side and insufficient carrier capture rate into the QW active region due to different reasons. However, all these models focus on faster increases in the nonradiative recombination without taking into account the possible mechanisms of slower increase in the radiative recombination rate. In addition, most of these models attempt to explain the IQE droop behavior just at room temperature without encompassing cryogenic temperatures. Actually, the IQE is greatly dependent on several factors such as the device structure, emission wavelength, current injection level, and operating temperature. Thus, we need to develop an advanced IQE droop model which is able to explain previously reported experimental results not only qualitatively but also quantitatively in a unified manner.

7.2.5 Saturated Radiative Recombination

We know from Eq. (7.7) that the IQE droop at certain temperature can happen just in case that the nonradiative recombination rate increases faster than the radiative recombination rate or the radiative recombination rate increases slower than the nonradiative recombination rate. As a concept for an origin of the IQE droop, the former is different from the latter in the following sense: The former emphasizes additional increase in the nonradiative recombination rate without taking into account the increasing speed in the radiative recombination rate seriously. On the other hand, the latter supposes that saturation in the radiative recombination rate at high carrier density is a motive for the IQE droop and subsequently it accelerates additional

increases in the nonradiative recombination rate. In the former case, for reducing the IQE droop, we need to find important nonradiative recombination processes and eliminate them in the first place. In the latter case, however, we need to increase the radiative recombination rate without its saturation up to high current level. Thus, it is essential to clarify which concept is more suitable to understand the LED efficiency comprehensively.

Here, we attempt to investigate the saturation characteristics in the radiative recombination rate. In addition, we do not specify a dominant nonradiative recombination process because it can vary with operating conditions as well as the LED device structure. Note that the nonradiative recombination processes can take place both inside and outside the active QW region but the radiative recombination process occurs only inside the active QW region.

Recently, there have been some reports investigating the saturation characteristics of the radiative recombination rate in the active QW and relating them to the LED efficiency droop although each proposed origin of the LED droop is different [28, 71–74]. Different parameters are utilized as a measure of the saturation in the radiative recombination rate such as the fall time of the temporal response in an electrical-optical pump-probe technique, the depth of the S-shape of the PL peak energy and the PL efficiency as in the temperature-dependent PL excitation, differential carrier lifetime with current, and the temporal decay times in the time-resolved PL techniques. Here, we propose another technique showing the saturation phenomenon in the radiative recombination rate. The method is a comparative study of the PL efficiency, the EL efficiency, and the open-circuit voltage under different temperatures and excitation powers in InGaN-based blue LEDs.

The structure of the LED used in this work was grown on a c-plane (0001) sapphire substrate, followed by a 30-nm thick buffer layer, and a 2.8-nm thick n-GaN layer (n-doping $= 5 \times 10^{18}$ cm^{-3}). The active region consisted of six 3-nm thick In$_{0.15}$Ga$_{0.85}$N QWs sandwiched by 5-nm thick GaN barriers. On top of the active region were a 40-nm thick p-Al$_{0.18}$Ga$_{0.12}$N EBL and a 60-nm thick p-GaN layer (p-doping $= 8 \times 10^{17}$ cm^{-3}). The device geometry was designed into a rectangular shape of 700×270 μm^2. The device had lateral electrodes and was mounted on an SMD-type package.

The PL efficiency was measured under open-circuit configuration from the room temperature to 12 K as a function of excitation laser power. The device was cooled by a helium closed-cycle cryostat by Advanced Research System. A 405-nm semiconductor diode laser was utilized for resonant optical excitation. The excitation power was varied from 3 mW to 100 mW by controlling the laser diode current. The excitation spot size of the laser on the LED device had a diameter of <80 μm. The PL signal was collected by a spectrometer and the open-circuit voltage was measured simultaneously. Also, the EL signal was collected separately by an optical power meter as a function of injection current.

Figure 7.3 depicts the open-circuit voltage as a function of excitation power from room temperature to cryogenic temperatures. At room temperature, as expected in other photovoltaic devices, the open-circuit voltage increases monotonically from 2.43 V to 2.57 V as the excitation power increases. This experimental result comes

Fig. 7.3 Experimental result
of the open-circuit voltage as
a function of excitation power
from room temperature to
cryogenic temperatures
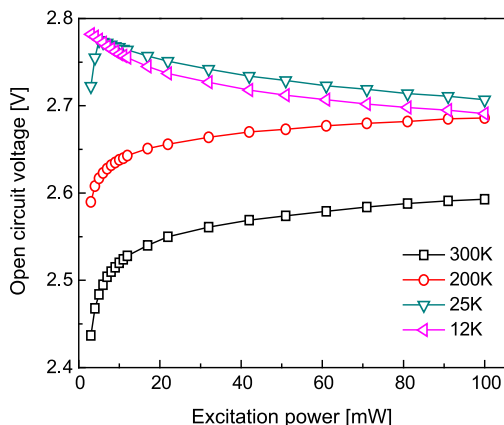
Fig. 7.4 Experimental result
of the PL efficiency as a
function of excitation power
from room temperature to
cryogenic temperatures
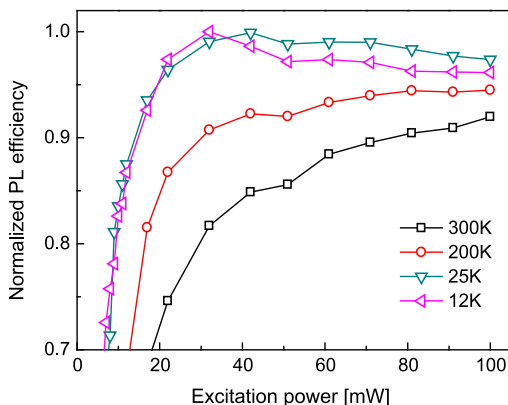
from the difference in quasi-Fermi energy induced by photogenerated charge carriers in MQWs by the resonant excitation. However, at cryogenic temperatures below 100 K, the open-circuit voltage does not increase any more with the excitation power. For example, at 12 K, it increases and then decreases from 2.79 V to 2.69 V with the excitation power. Figure 7.4 shows the PL efficiency as a function of excitation power from room temperature to cryogenic temperatures, in which the PL efficiency is normalized by the maximum PL efficiency at 12 K.

By comparing the open-circuit voltage in Fig. 7.3 with the PL efficiency in Fig. 7.4, we find that there is a very close relationship between their droop behaviors. In other words, the PL efficiency droop and the reduction in open-circuit voltage show a similar trend at the same temperature and become worse with decreasing temperature. We explain these temperature-dependent behaviors as follows. We first consider the case of room temperature. Electron-hole pairs generated by optical pumping recombine via radiative and nonradiative recombination processes in MQWs. Since some of the generated charge carriers escape outside the MQW region (electrons to the *n*-side and holes to the *p*-side) and no net current should flow under

Fig. 7.5 Experimental result
of the EL efficiency as a
function of injection current
from room temperature to
cryogenic temperatures
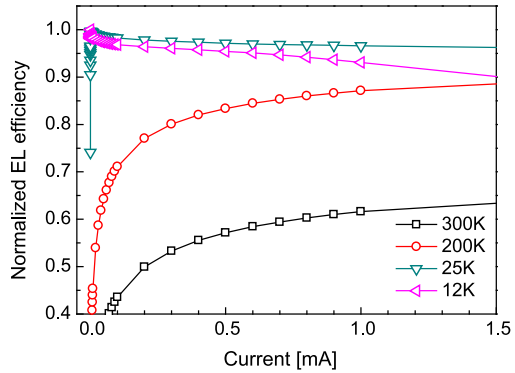
open-circuit configuration, the quasi-Fermi levels in MQW region for the electron
and the hole separate, playing a role of forward bias that cancels the photogenerated
current. This forward bias manifests as the open-circuit voltage. When the excitation
power is increased, more carriers are generated, accumulate, and recombine radia-
tively in MQWs, increasing the PL efficiency and the open-circuit voltage. We next
consider the case of cryogenic temperatures. As the laser excitation power initially
increases, the open-circuit voltage also increases. This is because rapid carrier ac-
cumulation in MQWs occurs due to the elimination of nonradiative recombination
channel at NRCs and early saturation of the radiative recombination. With further
increase of the laser excitation power, now the carriers in MQWs can start escaping
to opposite directions (electrons to the p-side and possibly holes to the n-side), re-
ducing the forward bias for the open-circuit and as a result, the open-circuit voltage.

Figure 7.5 is the EL efficiency normalized by the maximum EL efficiency at
12 K as a function of injection current. As the temperature decreases from room
temperature, overall IQE values initially increase with a similar amount of droop.
Beyond a critical temperature, the IQE curve starts to show the droop. The initiation
of droop depending on the operating temperature is considered to be caused by the
increase of carrier overflow due to the temperature dependence of saturation of the
radiative recombination rate. When the temperature dependence of the open-circuit
voltage and the PL efficiency in Figs. 7.3 and 7.4 are compared with that of the EL
efficiency, we can observe very similar tendencies among three experimental results.
In the resonant PL experiments, the temperature-dependent droop behaviors of both
the PL efficiency and the open-circuit voltage are determined mainly by the carrier
escape process rather than the carrier capture process in the QW. Thus, we think that
the rapid carrier pile-up due to the saturation of radiative recombination rate and
subsequent increase of nonradiative recombination rate due to the carrier leakage
of electron overflow from the MQW region to the p-GaN layer is the dominant
mechanism behind all these droop phenomena at cryogenic temperatures.

There are three important points to be noted from previous discussions. First, car-
riers excited by the resonant laser pumping do escape from the QWs, although the
photon energy of the pump laser is smaller than the bandgap energy of the semicon-
ductor surrounding the active QWs. The carrier escape may result from the carrier

heating process at the early stage of the pumping [72, 75]. There is the same result of carrier escape from the QWs by the resonant optical pumping in reference [76]. Second, the origin of efficiency droops in the PL and EL experiments is the same. Third, carrier overflow from the QW to the *p*-side is dominated by the carrier escape process rather than the carrier capture process. In other words, carriers injected from the outside QWs are mostly captured into the QWs without passing them as long as the total carrier recombination rate, i.e. the sum of the radiative and nonradiative recombination rates, is so fast that the captured carriers are completely consumed.

In summary, by examining the temperature-dependent PL efficiency, EL efficiency, and open-circuit voltage, a comprehensive model of efficiency droop phenomenon has been explained. Through comparative studies, we have suggested that the inherent origin of the efficiency droop is the saturation of the radiative recombination rate in the InGaN QW at low current and subsequent increase in the nonradiative recombination rates at high current. The degree of saturation is determined by the operating temperature and the effective active volume.

7.2.6 Comprehensive Efficiency Droop Model

Now we present an efficiency droop model that can comprehensively explain the experimental phenomena depending on both temperature and excitation levels in InGaN-based MQW blue and green LEDs. As the operating temperature is reduced, the amount of droop in the resonant PL efficiency and the EL efficiency is observed to increase [11, 12, 62]. We believe that this gives an important clue to the origin of the efficiency droop in addition to the saturation characteristics of the radiative recombination rate.

For experiments, InGaN/GaN MQW blue and green LED devices (with indium compositions of ∼20 % and ∼35 %, respectively) were used. Both devices had lateral electrodes and the chip sizes for blue and green LEDs were $350 \times 430\ \mu m^2$ and $400 \times 400\ \mu m^2$, respectively [29]. First, the IQEs were measured from 300 K to 50 K for both samples. From the measured EL intensities at different temperatures, IQEs were obtained. For both blue and green LED samples, the maximum IQEs at 50 K were defined to be 100 %. The maximum current was limited to 50 mA (10 mA for the green LED) to avoid any issue related to heat generation caused by increased forward voltage at low temperature.

Figures 7.6 and 7.7 depict the estimated IQEs as a function of current for blue and green LEDs, respectively. For clarification of the IQE behaviors at low currents, the same data are plotted on a log scale as well. As the temperature is lowered from 300 K, it is seen that the IQE curves for blue and green LEDs show two distinct behaviors, which is clearer on a linear scale. As the temperature is lowered from 300 K, the IQE curves initially shift to higher values overall with similar amounts of efficiency droop. This indicates that the number of defects in QWs causing the nonradiative recombination via the SRH process is reduced rather monotonically with decreasing temperature [44, 63]. Beyond certain temperatures (200 K for blue
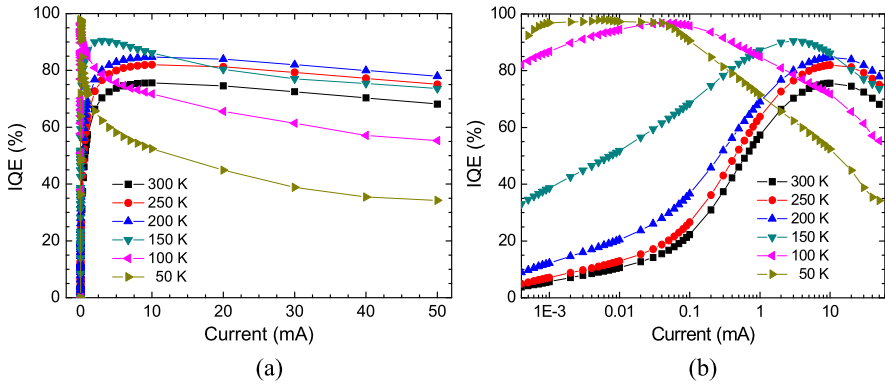
Fig. 7.6 Estimated IQEs for the blue LED plotted on (**a**) linear and (**b**) log scales for current
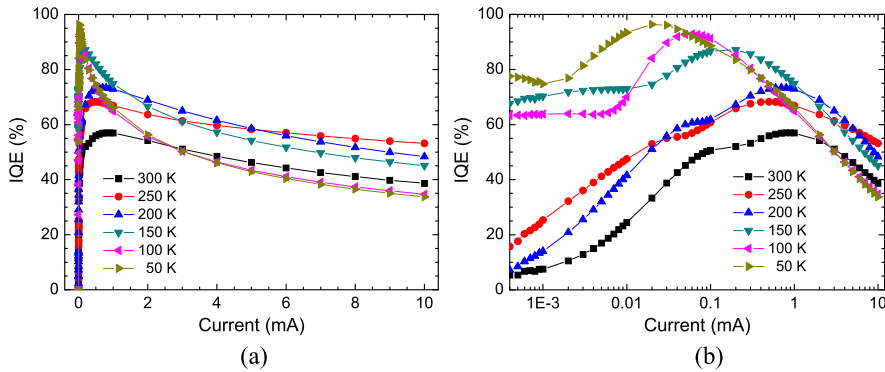
Fig. 7.7 Estimated IQEs for the green LED plotted on (**a**) linear and (**b**) log scales for current

and 250 K for green LEDs), however, the shape of IQE curves changes to show a more pronounced maximum at a lower current level with severer droop. This tendency becomes more distinct as temperature is lowered further. From this characteristics change in the IQE curves beyond certain temperatures, one can guess that the dominant droop mechanism undergoes a qualitative change. Since the nonradiative recombination lifetime in the QW active region becomes longer at lower temperatures [1], the severer droop cannot be dominated by the nonradiative recombination at defects in the QW active region. At the lowest temperature of 50 K, the defects in the QWs causing the nonradiative recombination via the SRH process in the QW active region can almost be ruled out from possible sources of droop. One can also note that the amount of droop at the same current level is much larger for the green LED than for the blue LED.

Figure 7.8(a) and (b) show the EL spectra measured at 10 mA under different temperatures for blue and green LEDs, respectively. As the temperature is lowered from 300 K, a side peak is seen to grow at ∼400 nm. The side peak is thought to
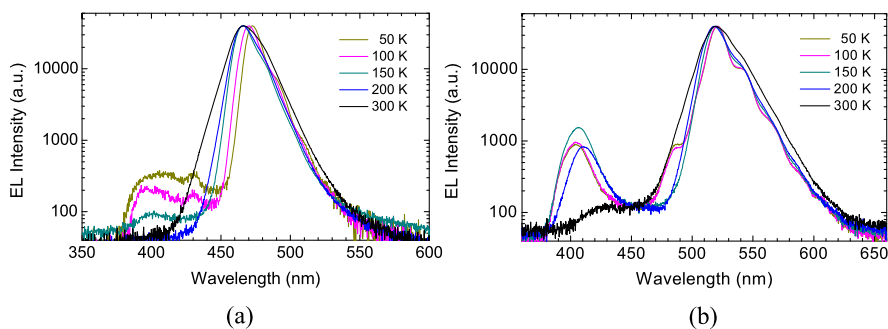
Fig. 7.8 EL spectra at 10 mA for (**a**) blue and (**b**) green LEDs measured at different temperatures. The main peaks are normalized to give the same value

originate from the Mg acceptor level in p-GaN and is caused by the electrons over-flowing to the p-GaN cladding layer [64–67]. At the same current level of 10 mA, this side peak at 400 nm is more pronounced for the green LED, indicating severer electron overflow than for the blue LED. Combining these EL spectra with the IQE curves, one can deduce that the severe IQE droop at low temperatures is mostly caused by the overflow of electrons to the p-GaN clad and the nonradiative recombination there.

A question remains as to why the electron overflow becomes severer at lower temperatures. There are two possibilities. One is the carrier-transport-limited droop model in which the electron overflow is originated partly from less carrier capture rate from the bulk region to the QW active region and partly from an increased carrier asymmetry in electron and hole concentrations in the MQW region. The other is the carrier spill-over from the QW active region to the bulk region due to the saturation in the total carrier recombination rate, especially the radiative recombination rate. As discussed in the previous section, we think that the growing overflow with decreasing temperature originates from the saturation of the radiative recombination rate at a smaller current and also decreasing nonradiative recombination rate rather than the carrier transport process.

The degree of saturation in the radiative recombination rate increases as the effective active volume or the radiative recombination probability per electron-hole pair becomes small. The effective active volume becomes smaller than the nominal one due to two reasons: First, the QWs in the active region are not fully utilized due to the carrier transport problem caused especially by holes in the conventional LED structure [68–70]. Second, even in a fully utilized QW, the effective volume is smaller than the nominal QW volume as indiums and/or stresses in the QW tend to cluster, making the region with potential maxima inefficient in emitting light [28, 40–43, 77]. The fact that the overflow becomes severer at lower temperature can be explained by the first observation. As the hole transport becomes more difficult with lower temperature, combined with a smaller number of activated holes, the number of QWs utilized becomes smaller, reducing the effective active volume. The radiative recombination rate in this reduced effective active volume is eventually sat-

urated at low current (but still high effective current density), and the electrons not recombined in the QW overflow to the p-GaN. As the nonradiative recombination rate via defects becomes smaller at lower temperatures, the tendency of overflow is reinforced. The saturation in the radiative recombination rate is caused fundamentally by the selection rule in k-space [8, 28, 78]. In GaN-based LEDs, this saturation can be exacerbated by the piezoelectric field that induces the quantum-confined Stark effect, making the overlap integral of electron and hole wavefunctions smaller.

The fact that the overflow is severer with the green LED than with the blue LED can be explained by the second observation. With higher indium concentration in the green LED, the QW becomes even more inhomogeneous with severer potential fluctuations possibly produced by indium clusters and/or micro-stresses connected to defects and piezoelectric fields, resulting in more reduced effective active volume than in the blue LED. Also, with higher indium concentration in the QW, the piezoelectric field becomes higher, which negatively impacts the radiative recombination rate in the green LED. The hole transport in the green LED can also be negatively affected by deeper potential in the QW and higher hole effective mass with higher indium composition. Consequently, premature saturation in the radiative recombination rate results at a lower current level than with the blue LED. The temperature at which the different IQE behavior starts to appear is also higher with the green LED than with the blue LED. It is important to note here that the overflow becomes severer with decreasing temperature as the radiative recombination rate saturates at lower current due to the reduced effective active volume. In this way, the overflow and the saturation in the radiative recombination rate are interrelated. One can say that the latter dictates the former, especially at low temperature, rather than the other way around. Although the electron overflow is a dominant nonradiative mechanism in this experiment especially at a low temperature, it should be noted that a dominant nonradiative recombination process can be different with different LED device structures as well as operational conditions.

In summary, by examining the temperature-dependent IQEs and EL spectra of blue and green LEDs, a comprehensive model of efficiency droop phenomena has been explained. As temperature is lowered to the cryogenic region, very severe efficiency droop has been observed in both blue and green LEDs. This severe droop is caused by the saturation of the radiative recombination rate at low current and subsequent increase in the nonradiative recombination rate via the electron overflow as confirmed by the EL spectra. The degree of saturation in the radiative recombination rate increases as the effective carrier density increases due to the decreased effective active volume with the decreasing temperature. It should be noted again that a dominant nonradiative recombination process can vary with operating conditions as well as LED device structure. Figure 7.9 shows a schematic flow chart for the IQE droop model explained so far. We have found from additional works that the shapes of the IQE versus current, i.e. the IQE droop curve, may vary with the dominant nonradiative recombination processes. Table 7.2 explains the relation between them. The concave, linear, and convex shapes in the IQE curves may originate from the electron overflow escaping from the active QW to the p-side, the electron tunneling from the inside QW to the p-side, and the SRH recombination in the QW,
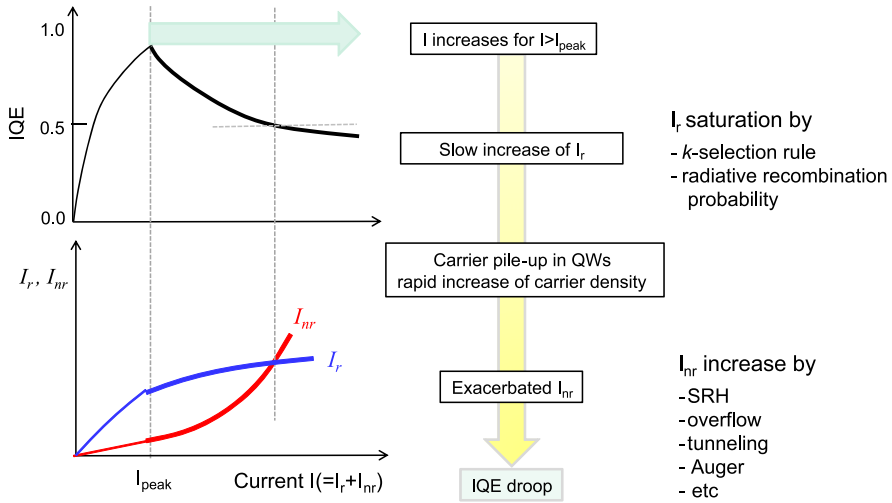
Fig. 7.9 Schematic explanation on the proposed IQE droop model

respectively. In order to reduce the efficiency droop, one should try to suppress the saturation of the radiative recombination rate and reduce the nonradiative recombination rate. Suppression of the saturation in radiative recombination rate requires increased effective active volume via balanced distribution of electrons and holes in the MQW region, small piezoelectric field, less stress localization, and reduced indium clustering. Insights presented here should help design a more efficient LED structure with reduced droop.

7.3 IQE Measurement Methods

The EQE is one of the fundamental parameters for the LED performance evaluation and is given by the product of the IQE and the LEE. The EQE can be estimated directly by measuring the optical power emitted from a LED. The EQE depends on the emitting wavelength, and the highest EQE is obtained in the peak wavelength range of $430 \sim 460$ nm. In recent advanced blue LEDs, EQEs higher than 60 % are typically reported and the highest EQE ever reported is 75.5 % for a blue LED [79]. Since the EQE can be limited by the IQE or the LEE, it is important to evaluate the IQE and the LEE separately and improve both components individually. The measurement of the LEE is very difficult because the LEE is sensitive to micro- and nano-scale structures embedded in the LED chip itself as well as its package. Different IQE measurement methods have been reported but not fully established yet. The reported IQEs and LEEs have exceeded up to 70–80 % in InGaN blue LEDs [79–81].

The discrete and accurate measurement of the IQE in QWs as a function of carrier density is a constant challenge. Different measurement methods of the IQE have

Table 7.2 Three typical IQE curves of concave, linear, and convex shapes. A dominant nonradiative recombination process may be different for each IQE curve shape as designated below
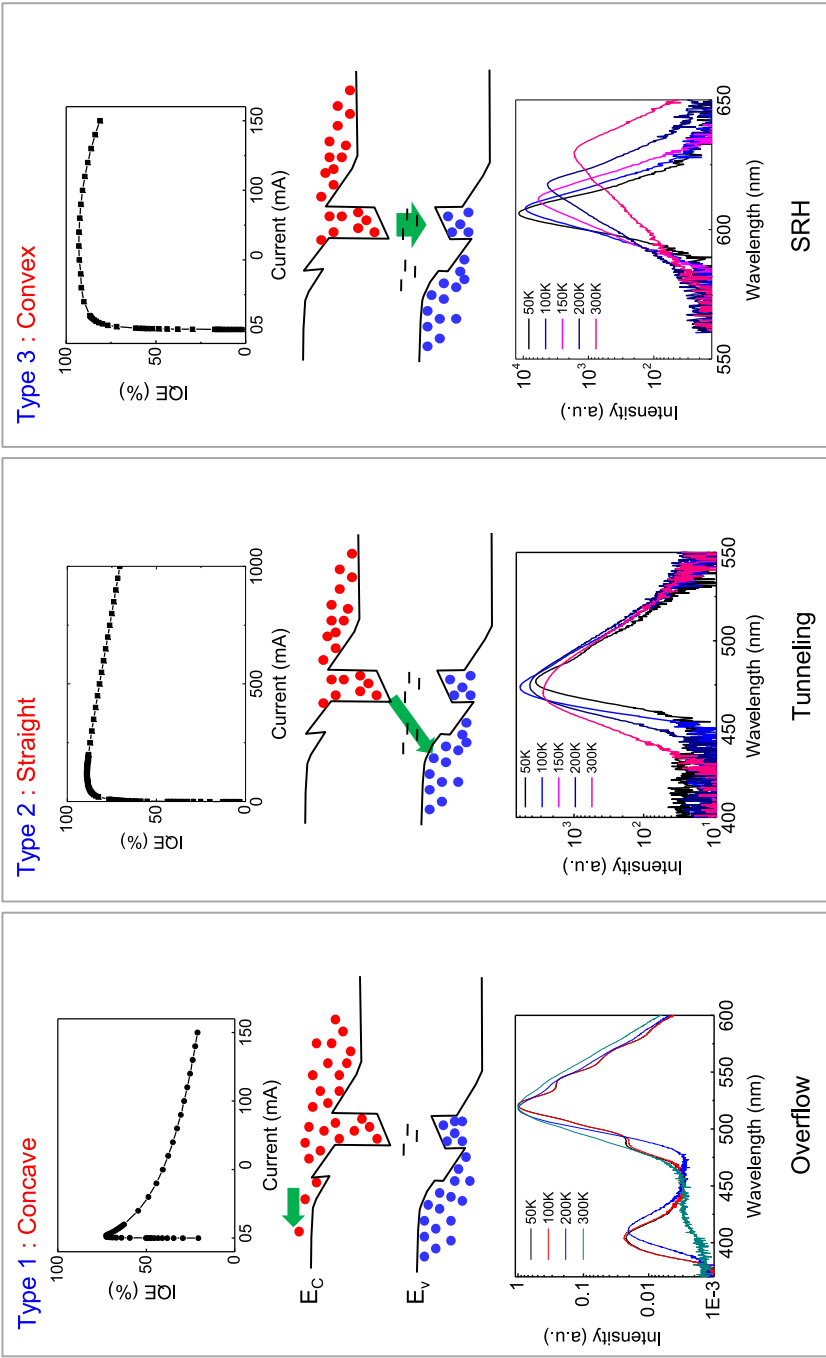
been reported in InGaN-based QWs. The IQE measurement methods are typically divided into three categories: (1) absolute intensity measurements by an integrating sphere with theoretically estimated LEEs, (2) analysis of EL characteristics of LEDs or laser diodes (LDs) operating at the steady state or modulated state, and (3) all optical methods.

Method (1) indirectly measures the IQE by dividing the measured EQE with the known LEE. This method is frequently used due to its simplicity, but some errors can easily occur when estimating the LEE [82–86]. Method (2) is especially suitable for LDs. The IQE above the threshold is extracted from a curve of the external differential efficiency versus the cavity length [87]. The IQE below threshold is evaluated by analyzing the light power-current (P-I) curve based on the carrier rate equation where the current injection efficiency and the recombination coefficients are assumed to be certain constants [88]. In LED chips, two types of IQE measurement methods based on the carrier rate equation have been reported. One method is to analyze the P-I curve of a LED chip operating at the steady state, where the radiative recombination coefficient is calculated theoretically by the parabolic band model [107]. Another method is to measure the EL decay time by applying a small current pulse with an extremely fast falling time as a function of bias current level, where both radiative and nonradiative recombination coefficients are thought to be independent of the bias current level [89]. Here, the experimental set-up for measuring decay time in EL requires complicated configurations such as impedance matching, a high-speed pulse generator, and a detector. One of the limitations included in method (2) is that the fundamental physical parameters, such as recombination coefficients, are treated as known or unknown constants with respect to the carrier density in InGaN-based QWs.

Method (3), the all optical method, is nondestructive and the IQE of the QWs is evaluated on an epitaxial wafer without fabricating actual devices. The most popularly utilized IQE measurement method among these is the continuous-wave (CW) temperature-dependent photoluminescence (TDPL) [90–95]. In this method, the IQE can be directly determined by the ratio of the PL intensity at room temperature to that at a low temperature. An advantage of TDPL is that the IQE can be directly determined experimentally without the aid of any calculations or physical parameter assumptions. However, the TDPL has a bold assumption that the IQE is unity at a low temperature, which has not yet been verified either experimentally or theoretically. Furthermore, we will show that the influence of the optical absorption coefficient on the laser excitation intensity is neglected in the conventional TDPL method. The temperature-dependent time-resolved photoluminescence (TD-TRPL) method is useful because both the temporal and the spectral information are obtained simultaneously [96–102]. The TD-TRPL method can characterize both the IQE and carrier lifetimes at the same time. The conventional TD-TRPL method also has similar assumptions as the TDPL method. Furthermore, it does not have clear definitions of the carrier lifetimes depending on the operational conditions of the steady state and the transient state. Another method for determining the IQEs at room temperature is proposed by fitting the laser pump intensity to powers of the integrated PL intensity [102–106]. The dependence of the PL intensity on laser

pump power is formulated by using photon and carrier rate equations in a semiconductor with the assumption that both the optical absorption coefficient and carrier recombination rate coefficients are independent of the laser pump power. Although this method is useful to the GaAs/AlGaAs system, it has some limitations in InGaN-based QWs where the aforementioned internal physical coefficients strongly depend on the carrier density in a QW due to the internal electric field.

This section reviews some of the popularly utilized IQE measurement methods such as the constant ABC model, TDPL, and TD-TRPL methods both theoretically and experimentally. A measurement method of the IQE at room temperature based on the intensity-dependent PL, i.e. IDPL method, is presented. The proposed method utilizes the laser excitation power as a varying parameter instead of temperature in the conventional TDPL and the TD-TRPL methods. This new method can evaluate the IQE at any temperature without the aid of any calculated parameters.

7.3.1 Constant ABC Model

Assuming that the electron and hole concentrations in the active QW region are nearly equal to each other, the current, I, and the IQE, η_{IQE}, can be written as

$$I = (qV_a/\eta_{\text{INJ}})\left(AN + BN^2 + CN^3\right) \quad \text{and} \quad P = \eta_c h\nu V_a\left(BN^2\right) \quad (7.8)$$

$$\eta_{\text{IQE}} = \eta_{\text{INJ}} \frac{BN^2}{AN + BN^2 + CN^3}, \quad (7.9)$$

where A, B, and C coefficients represent the nonradiative recombination via SRH process, bimolecular radiative recombination, and Auger recombination, respectively. $N, I, P, q, h\nu, \eta_{\text{INJ}}, \eta_c$, and V_a are carrier density, current, measured optical power, elementary charge, photon energy, injection efficiency, and light coupling efficiency between a LED to an optical detector, and the active volume, respectively [39]. With Eqs. (7.8) and (7.9), it is possible to fit the experimental results of the IQE versus current I using the A, B, and C coefficients as fitting parameters. Generally, people try to fit A and B coefficients in the low current injection regime and C coefficient in high current injection regime. Here, we show an analysis method of the IQE and each recombination current at an arbitrary current. We use the peak point in an experimental IQE versus current curve for analysis of the rate equation. This method requires only one fit parameter among three recombination coefficients, i.e., A, B, and C.

In the relation of the IQE versus the current density in typical LEDs, there exists the maximum value of IQE, η_{max} and the current, I_{max} where η_{max} occurs. η_{max} and I_{max} can be obtained from the condition, $d\eta/dI = 0$. Then, η_{max} and I_{max} are expressed as follows:

$$\eta_{\text{max}} = \frac{B}{B + 2\sqrt{AC}}, \qquad I_{\text{max}} = \frac{kA}{C}(B + 2\sqrt{AC}), \quad (7.10)$$

where $k = qV_a/\eta_{INJ}$. By assuming $\eta_{INJ} = 1$, k is determined directly from the active volume. From Eq. (7.10), the A, B, C coefficients are related to each other through η_{max} and I_{max}.

$$B = \frac{4\eta_{max}k}{I_{max}(1 - \eta_{max})^2}A^2, \qquad C = \frac{4k^2}{I_{max}^2(1 - \eta_{max})^2}A^3. \qquad (7.11)$$

Therefore, if one of the A, B, C coefficients is known, remaining coefficients can be obtained by using Eq. (7.11) with known quantities of η_{max}, I_{max}, and k.

Using Eqs. (7.8)–(7.10), η_{IQE} at any current I is expressed as

$$\eta_{IQE} = 1 - \frac{(1 - \eta_{max})}{2}\left(1 + \frac{\eta_{IQE}I}{\eta_{max}I_{max}}\right)\sqrt{\frac{\eta_{IQE}I_{max}}{\eta_{max}I}}. \qquad (7.12)$$

The IQE versus current relation is determined by solving the above equation. Interestingly, the IQE has no explicit dependence on the A, B, C coefficients, and is unambiguously determined only from the I_{max} and η_{max} values. Note that I_{max} is exactly known by the measurement of EQE versus current relation, and η_{max} can be used as a fit parameter and compared with the measurement. Therefore, three fit parameters are reduced to only one in our approach, which makes the fit process much easier and gives accurate results on the IQE analysis of the LED efficiency.

The fitting procedure is as follows: (1) Experimentally measure a relative EQE curve as a function of current I, normalize the experimental EQE curve by the peak EQE value, and find the I_{max}. (2) Simulate curves of η_{IQE} versus I for different η_{max} satisfying Eq. (7.12). (3) Select η_{max} that gives the best fit between the experimental and simulated curves from the step (1) and the step (2), respectively. (4) Calculate $k = qV_a/\eta_{INJ}$ by assuming η_{INJ} of unity and V_a. Finally, estimate $A, B,$ and C coefficients from Eq. (7.11) by assuming one of them, for example A.

From the fitted IQE curve as a function of I, $\eta_{IQE}(I)$, we can calculate each recombination current as

$$I_{SRH} = \left(\frac{1 - \eta_{max}}{2}\sqrt{\frac{\eta_{IQE}I_{max}}{\eta_{max}I}}\right)I = \eta_{SRH}I,$$

$$I_{rad} = \eta_{IQE}I, \qquad (7.13)$$

$$I_{Auger} = \left(\frac{(1 - \eta_{max})}{2}\frac{\eta_{IQE}}{\eta_{max}}\sqrt{\frac{\eta_{IQE}I}{\eta_{max}I_{max}}}\right)I = \eta_{Auger}I.$$

Again, it is interesting to note that the SRH, radiative, and Auger recombination currents have no explicit dependence on the A, B, C coefficients and that they can be expressed as a function of current from known quantities, η_{max}, I_{max}, η_{IQE}, and I. At sufficiently high current where $I \gg I_{max}$, I_{Auger} is much higher than I_{SRH} and the nonradiative current is considered dominated by the Auger recombination current.

However, the fit η_{IQE} from Eq. (7.12) often deviates from the experimental IQE curve, η, which is defined as

$$\eta = \eta_{max}\frac{P/I}{P_{max}/I_{max}}, \qquad (7.14)$$

Fig. 7.10 (**a**) Experimentally measured EQE of a blue LED sample, (**b**) normalized EQE curves (*solid lines*) for different IQE peak value of η_{max}. Good agreement is found between the theoretical and experimental data at $\eta_{max} = 79\%$. (**c**) SRH nonradiative carrier lifetime $\tau_{SRH}(=1/A)$ as a function of the C coefficient calculated from Eq. (7.11)
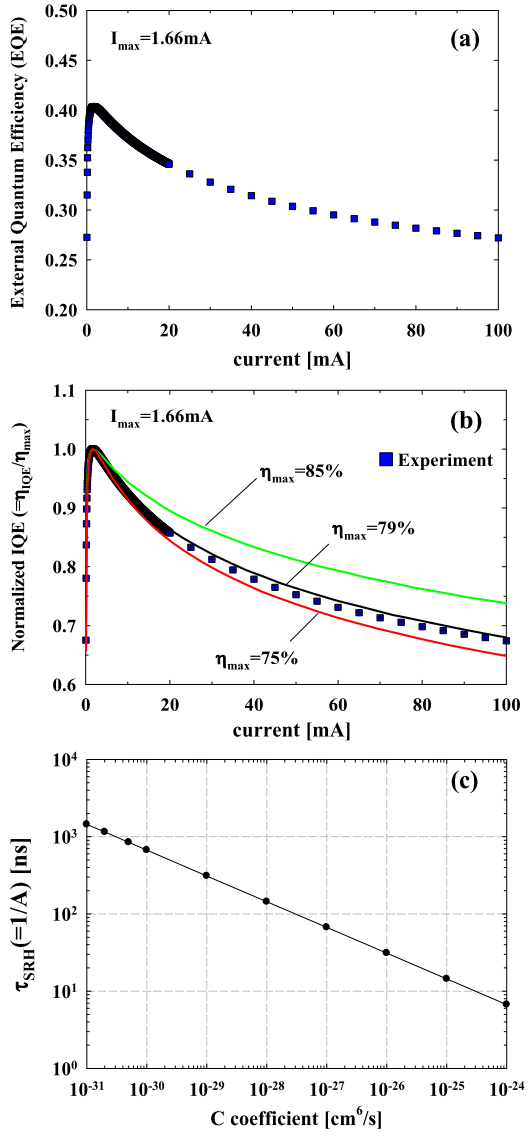
where η_{max} is known from the fit of Eq. (7.12) and the others, I_{max}, P_{max}, I, and P are experimentally measured. We think that the difference of the experimental IQE, η, in Eq. (7.14) and the fitted IQE, η_{IQE}, fitted from Eq. (7.12) results from an additional nonradiative recombination current, $I_{leakage}$, for the very high current range. By using Eq. (7.13), $I_{leakage}$ is calculated as

$$I_{leakge} = I - (\eta_{SRH} + \eta_{IQE} + \eta_{Auger})I. \tag{7.15}$$

Normalized EQE
curves (*solid lines*) and
experimental data
(*rectangular symbols*) for
different IQE peak values as a
parameter in an InGaN blue
LED emitting at 460 nm
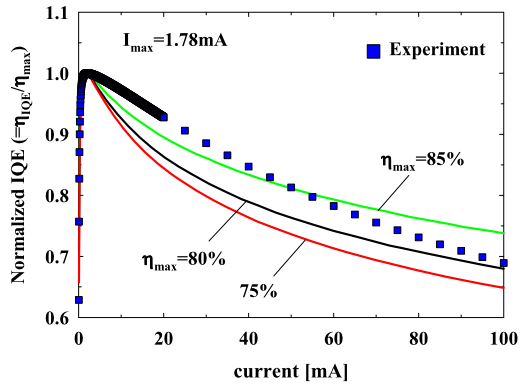
Now the theoretical analysis of the efficiency droop is compared with experimental efficiency curves. We measured the output power of a commercial-grade LED sample as a function of current at room temperature. The active region of the LED consists of two 2.5-nm-thick InGaN QWs with a 10-nm-thick GaN barrier. The peak wavelength of the emission spectrum is 460 nm. Figure 7.10 shows an example to estimate the IQE value following the aforementioned procedure. The I_{max} of this sample is 1.66 mA. The theoretical IQE curves obtained by solving Eq. (7.12) for different η_{max} are overlaid on the measured result of the EQE curve versus current, and good agreement is observed up to 100 mA when η_{max} is 79 %. Using η_{max}, I_{max}, and k of the LED sample, the relation between A and C coefficients can be found from Eq. (7.11). Figure 7.10(c) shows the nonradiative carrier lifetime (τ_{SRH}) that is the inverse of the A coefficient as a function of the C coefficient. Recently reported A coefficient values of InGaN LEDs as shown in Table 7.1 are between 0.5×10^7 s^{-1} and 2×10^8 s^{-1}, which corresponds to τ_{SRH} of 5–100 ns. In Fig. 7.10(c), however, when τ_{SRH} lies between 5 ns to 200 ns, the corresponding range of the C coefficient is around 10^{-28}–10^{-24} cm^6/s. To date, there are no theories explaining such large C coefficients. Thus, it is plausible that a carrier loss process with the N^3-dependence, which is obvious in InGaN-based QW LEDs, may not represent the Auger processes.

Figure 7.11 shows another example of the normalized EQE curves (solid lines) for different IQE peak values as a parameter and an experimental data (rectangular symbols) from a 460-nm InGaN LED. In this sample, we cannot find a η_{max} value giving a good fit between experimental data and the theoretical line. This result implies that there would be a different nonradiative recombination mechanism at high injection current which cannot be described by N^3-dependence.

In summary, we have introduced a very simple fitting method based on the constant ABC model in the carrier rate equation. The method utilizes just two parameters of current at the IQE peak and the IQE peak value itself. It is possible to know the IQE and the carrier recombination currents simultaneously without any knowledge of A, B, and C recombination coefficients. We applied the method to two different InGaN blue LEDs emitting at 460 nm. In one sample, good agreement was obtained between the experimental and theoretical data. However, the C coefficient

was estimated to be a few orders larger than the previously reported Auger coefficients. In the other sample, the experimental data could not be well fitted with the theory. These experimental results imply that the Auger recombination alone could not explain the efficiency droop phenomenon in InGaN LEDs.

7.3.2 Temperature-Dependent Photoluminescence (TDPL) Method

Improvements in the luminous efficacy in blue/green LEDs have been hampered by numerous material defects associated with the indium segregation in InGaN QWs. Even though the self-organized small or dot-like indium-rich InGaN regions may improve the radiative efficiency by localizing carriers and preventing carriers from reaching dislocations, those regions may contain many nonradiative defects due to the compositional disorder and strain. PL studies using resonant optical excitation of the active QW is generally used in order to ensure that transport effects can be avoided. Thus, it has been assumed that in the case of resonant excitation, the injection efficiency is perfect as unity and the radiative efficiency in the InGaN active layer is regarded as the IQE. However, as written in the Sect. 7.2.5, even in the resonant optical excitation, carrier escape from QWs does take place and shows strong dependence upon the excitation and temperature conditions. Nevertheless, all optical methods are still useful to investigate physical mechanisms of epitaxial layers because of their advantages of nondestructive nature and no fitting parameters required between theory and experiment.

One of the popularly utilized methods for determining the IQE at room temperature is the continuous wave (CW) TDPL method [90–95]. The integrated PL intensity (I_{PL}) normalized to the excitation laser intensity (I_L), the so-called PL efficiency (η_{PL}), is dependent on both the temperature T and the excitation level I_L. The PL efficiencies are taken for various temperatures and excitation intensity levels. The maximum PL efficiency $\eta_{PL,max}$ is usually found at very low temperature T_l and a certain excitation level $I_{L,max}$. Then, η_{PL} for any temperature T and excitation intensity I_L is normalized by $\eta_{PL,max}$ and the normalized values are considered as the IQE for any T and I_L, i.e.,

$$\eta_{IQE}(T, I_L)|_{TDPL} = \frac{\eta_{PL}(T, I_L)}{\eta_{PL,max}(T_l, I_{L,max})}. \tag{7.16}$$

First, the validity of Eq. (7.16) is considered qualitatively. In the conventional TDPL, it is assumed that the IQE is linearly proportional to the PL efficiency so that the maximum PL efficiency denotes the maximum IQE of 100 %, i.e., $\eta_{IQE}(T_l, I_{L,max}) = 100$ %. This assumption is usually supported by the following experimental result: The PL efficiency (η_{PL}) for a given excitation laser intensity ($I_{L,c}$) does not change below a certain temperature T_c and drops gradually above it [93, 99, 100]. It is considered that the decrease of the PL intensity for $T > T_c$ results from the increase in density of activated NRCs and that the saturated characteristic of the PL intensity for $T < T_c$ comes from all frozen NRCs. In fact, however, some

epitaxial wafers show that the PL efficiency varies with the excitation laser intensity even for $T < T_c$. If the nonradiative recombination processes are zero at a low temperature, then the PL efficiency should be constant regardless of the excitation level. This observation implies that nonradiative recombination processes may not be negligible even at extremely low temperatures and certain excitation powers. Moreover, the conventional TDPL method does not give any clear IQE difference at room temperature between the epitaxial wafers for green and blue LEDs. This is because $\eta_{IQE}(T_l, I_{L,max})$ is assumed as 100 % in all material systems. But it is well recognized that the crystal quality of InGaN-based QWs decreases with increasing indium composition (or emission wavelength). Combining the facts quoted above, we need a method to estimate the IQE without the assumption of $\eta_{IQE}(T_l, I_{L,max}) = 100$ %.

Now the conventional TDPL method is reviewed theoretically based on the carrier rate equation. Assuming that the carrier concentration generated by optical pumping is small and that they recombine at the generated layer without escape, the rate equation of the carrier density $N(t)$ in the active QW under weak excitation condition is written as

$$\frac{dN}{dt} = -\left(AN + BN^2\right) + (1 - R)\frac{\alpha I_L}{h\nu_L}, \tag{7.17}$$

where A is the nonradiative recombination coefficient through the SRH recombination process and B is the radiative recombination rate coefficient. $R, \alpha, h\nu_L$, and I_L are the reflectivity of the semiconductor surface, the absorption coefficient in the QWs, the photon energy, and the excitation intensity of an excitation laser, respectively [104]. At the steady state $(dN/dt = 0)$, Eq. (7.17) can be rewritten as

$$\frac{I_L}{h\nu_L} = \frac{AN + BN^2}{(1 - R)\alpha}. \tag{7.18}$$

The integrated PL intensity (I_{PL}) detected by a photodetector over the wavelengths is given by

$$\frac{I_{PL}}{h\bar{\nu}} = \left(\eta_{det}\eta_{LEE}\frac{V_a}{A_{det}}\right)BN^2 = \eta_c BN^2, \tag{7.19}$$

where $\eta_{det}, \eta_{LEE}, h\bar{\nu}, V_a$, and A_{det} are the detector coupling efficiency, the light extraction efficiency, the average photon energy, the active volume, and the detector area, respectively, and $\eta_c = \eta_{det}\eta_{LEE}(V_a/A_{det})$. The PL efficiency (η_{PL}) is defined as the ratio of the radiative photon rate to the excitation laser photon rate, i.e.,

$$\eta_{PL} = \frac{I_{PL}/h\bar{\nu}}{I_L/h\nu_L} = \eta_c(1 - R)\alpha\frac{BN^2}{AN + BN^2} = K_{PL}\alpha\eta_{IQE} \tag{7.20}$$

where $K_{PL} = \eta_c(1 - R)$. The PL efficiency expressed by Eq. (7.20) is understandable because the spontaneously radiated light power will increase with the number of photons absorbed in the QWs and also with the number of radiatively recombining carriers at a certain excitation intensity.

The IQE can be also represented by using the carrier lifetimes at the steady state as

$$\eta_{IQE} = \frac{BN^2}{AN + BN^2} = \frac{N/\tau_r}{N/\tau_{nr} + N/\tau_r}. \tag{7.21}$$

Fig. 7.12 PL efficiencies as a function of excitation intensity, measured at 300 K and 8 K
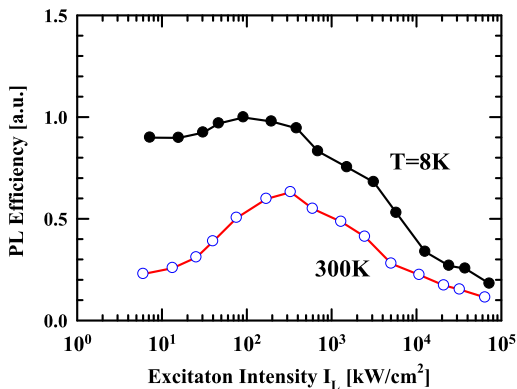
τ_r and τ_{nr} are the radiative and nonradiative carrier lifetimes, respectively. The carrier lifetimes of τ_r and τ_{nr} are related to the recombination coefficients of A and B by

$$\tau_r = \frac{1}{BN} \quad \text{and} \quad \tau_{nr} = \frac{1}{A}. \tag{7.22}$$

The carrier lifetimes at the steady state are different from those at the transient state, as in TRPL experiments. An explicit relation between the IQE in Eq. (7.21) and the IQE based on the conventional TDPL method in Eq. (7.16) is expressed as

$$\eta_{\text{IQE}}(T_h, I_L) = \left[\eta_{\text{IQE}}(T_l, I_{L,\text{max}}) \frac{\alpha(T_l, I_{L,\text{max}})}{\alpha(T_h, I_L)} \right] \eta_{\text{IQE}}(T_h, I_L)|_{\text{TDPL}}. \tag{7.23}$$

Equation (7.23) shows that two conditions should be satisfied simultaneously in order to recognize the IQE measured by the conventional TDPL as a real IQE: (1) IQE at a low temperature and a certain excitation intensity, i.e., $\eta_{\text{IQE}}(T_l, I_{L,\text{max}})$, is equal to 100 %, i.e., $\eta_{\text{IQE}}(T_l, I_{L,\text{max}}) = 100$ % and (2) the optical absorption coefficient α is completely independent of the excitation intensity and temperature.

Figure 7.12 shows an example of the relative η_{PL} normalized by the maximum value at 8 K [91]. The sample used in this study was a near-UV InGaN MQW LED emitting at 406 nm grown on a sapphire substrate. The PL efficiency at 8 K or 300 K increases with an increasing excitation intensity at low excitation levels and shows the peak value at a certain excitation intensity denoted by $I_{L,\text{max}}$. Then, it decreases with further increasing the excitation intensity at high excitation levels. The reasons for these behaviors have not been clearly known, but it is thought that the dominant nonradiative recombination mechanism in both sides of the excitation levels may be different. At 8 K, the carrier escape from the excited layer via tunneling and spillover mechanisms can be considered as possible dominant nonradiative recombination mechanisms at low and high excitation levels around $I_{L,\text{max}}$, respectively. Anyway, it is reasonable to choose the peak efficiency among all investigated conditions of temperature and excitation level as 100 % in the TDPL method. However, this bold assumption of the peak PL efficiency as 100 % for samples without showing the

saturation characteristics in low excitation intensity and temperature could result in an overestimated PL efficiency at room temperature. Therefore, we need a method to estimate the PL efficiency just at room temperature without comparing the PL efficiency at low temperature.

7.3.3 Intensity-Dependent Photoluminescence (IDPL) Method

In this section, we propose a method to estimate the PL efficiency at room temperature only by utilizing relative PL efficiency data as a function of excitation power. In InGaN-based QWs grown on a sapphire substrate along the c-axis, there exist strong internal electric fields due to the piezoelectric and spontaneous polarizations even without an external bias. The internal electric field reduces as the carrier density increases due to the carrier screening of polarizations by the increasing excitation intensity I_L. Thus, the PL efficiency is determined by several physical parameters such as the absorption coefficient α, the SRH nonradiative recombination coefficient A, the radiative recombination coefficient B, and the carrier density N, and eventually the IQE. We suppose that there is a weak excitation intensity of $I_{L,o}$. In this situation, carriers in the active layer are nondegenerate and the nonradiative loss due to N^3-dependency is negligible. Moreover, the electron energy band does not vary with the carrier density due to enough carrier screening of the internal electric fields. As a consequence, the coefficients of α, A, and B can be considered as constants of α_o, A_o, and B_o, respectively, for a small change of the excitation intensity around $I_{L,o}$.

Our method starts with finding the $I_{L,o}$ based on the carrier rate equation at a certain temperature. We first measure the PL intensity as a function of excitation intensity (i.e. the IDPL experiment). Then, a specific excitation intensity of $I_{L,o}$ is found by utilizing two successive excitation intensities of $I_{L,n}$ and $I_{L,n+1}$, where n represents the n-th data point and $I_{L,n} < I_{L,n+1}$. In the excitation range between $I_{L,n}$ and $I_{L,n+1}$ with assumed constants of α_o, B_o, and A_o, the following equations are obtained from Eqs. (7.19) and (7.20):

$$A_o + s\sqrt{I_{\mathrm{PL},n}} = m\frac{I_{L,n}}{\sqrt{I_{\mathrm{PL},n}}} \quad \text{and} \quad A_o + s\sqrt{I_{\mathrm{PL},n+1}} = m\frac{I_{L,n+1}}{\sqrt{I_{\mathrm{PL},n+1}}}, \qquad (7.24)$$

$$B_o N_n = s\sqrt{I_{\mathrm{PL},n}} \quad \text{and} \quad B_o N_{n+1} = s\sqrt{I_{\mathrm{PL},n+1}}, \qquad (7.25)$$

$$A_o = \chi s, \qquad (7.26)$$

$$\chi = \sqrt{I_{\mathrm{PL},n}}\left(\frac{I_{\mathrm{PL},n+1}}{I_{\mathrm{PL},n}} - \frac{I_{L,n+1}}{I_{L,n}}\right)\bigg/\left(\frac{I_{L,n+1}}{I_{L,n}} - \sqrt{\frac{I_{\mathrm{PL},n+1}}{I_{\mathrm{PL},n}}}\right), \qquad (7.27)$$

where $s = \sqrt{B_o/h\bar{\nu}\eta_c}$ and $m = (1 - R)\alpha_o\sqrt{h\bar{\nu}\eta_c B_o}/h\nu_l$. $N_{o,n}$ and $N_{o,n+1}$ are carrier densities in a QW for $I_{L,n}$ and $I_{L,n+1}$, respectively. Equation (7.27) shows the

condition for finding an excitation intensity of $I_{L,n} = I_{L,o}$ where variations of A_o and s become zero

$$\left.\frac{\Delta\chi}{\Delta I_L}\right|_{I_{L,n}=I_{L,o}} = 0. \tag{7.28}$$

Equation (7.28) shows that if a laser intensity of $I_{L,n}$ is found where χ shows a peak value as χ_n, then $I_{L,n}$ corresponds to the laser intensity giving minimum variations of the internal physical parameters (α_o, B_o, and A_o) against laser intensity. In general, χ has a concave shape as a function of I_L.

Then, the IQE at $I_{L,n}$ is obtained as

$$\eta_{\text{IQE},n} = \frac{B_o N_n}{A_o + B_o N_n} = \frac{\sqrt{I_{\text{PL},n}}}{\sqrt{I_{\text{PL},n}} + \chi_n} = \frac{1}{1+\gamma_n}, \tag{7.29}$$

where

$$\gamma_n = \frac{\chi_n}{\sqrt{I_{\text{PL},n}}} = \left(\frac{I_{\text{PL},n+1}}{I_{\text{PL},n}} - \frac{I_{L,n+1}}{I_{L,n}}\right) \bigg/ \left(\frac{I_{L,n+1}}{I_{L,n}} - \sqrt{\frac{I_{\text{PL},n+1}}{I_{\text{PL},n}}}\right). \tag{7.30}$$

It is noticeable that the IQE at $I_{L,n}$ is only determined by the ratios of $I_{L,n}/I_{L,n+1}$ and $I_{\text{PL},n}/I_{\text{PL},n+1}$ without using any assumed values of α, B, and A. Thus, this method generates small errors when determining the IQE compared to previous methods.

The IQE $\eta_{\text{IQE},i\neq n}$ at any excitation intensity $I_{L,i(i=1,2,...)}$ can be found by using a known $\eta_{\text{IQE},i=n}$ in Eq. (7.29). The IQE $\eta_{\text{IQE},i\neq n}$ at $I_{L,i(i=1,2,...)}$ is expressed from Eq. (7.21) as

$$\eta_{\text{IQE},i} = \left.\frac{BN^2}{AN + BN^2}\right|_{I_{L,i}} = \frac{K_\eta}{\alpha_i}\frac{I_{\text{PL},i}}{I_{L,i}}, \quad \text{where } K_\eta = \frac{h\nu_L}{(1-R)h\bar{\nu}\eta_c}. \tag{7.31}$$

In Eq. (7.31), i stands for the i-th excitation intensity used in the experiment, α_i is the optical absorption coefficient at $I_{L,i}$, and K_η is a constant. We can eliminate K_η by dividing $\eta_{\text{IQE},i}$ by $\eta_{\text{IQE},n}$ and express $\eta_{\text{IQE},i}$ by using experimentally measurable data as follows:

$$\eta_{\text{IQE},i} = \frac{\alpha_n}{\alpha_i}\eta_{\text{PL},i}\eta_{\text{IQE},n}, \quad \text{where } \eta_{\text{PL},i} = \frac{I_{\text{PL},i}/I_{L,i}}{I_{\text{PL},n}/I_{L,n}}. \tag{7.32}$$

At a given temperature, the optical absorption coefficient α is a function of the excitation intensity I_L [107]. Therefore, it is not easy to estimate the ratio of α_n/α_i in Eq. (7.32). At this moment, it is assumed that the optical absorption coefficient α_i is the same as α_n.

Figure 7.13 shows the calculated χ's from Eq. (7.27) for the PL efficiency at room temperature plotted in Fig. 7.12. The maximum χ value is obtained at the excitation intensity of 30 kW/cm^2 which corresponds to $I_{L,n}$ at room temperature. In general, χ has a concave shape as shown in Fig. 7.13. From Eq. (7.29), the IQEs at $I_{L,n}$ for 300 K and 10 K are found to be 34.4 % and 56.6 %, respectively. Figure 7.14 shows the IQE values estimated by the proposed IDPL and the conventional TDPL methods at room temperature. Both methods show almost identical IQE values at

Fig. 7.13 Calculated χ's as a function of excitation intensity I_L for PL efficiencies in Fig. 7.12. It is assumed that electron energy bands in InGaN QWs become flat at the excitation intensity of $I_{L,o}$ due to the carrier screening of the internal electric field
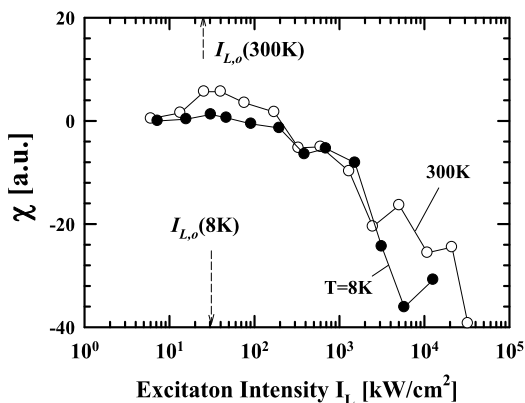
Fig. 7.14 IQEs at room temperature estimated by the conventional TDPL method in Sect. 7.3.2 and the IDPL method in Sect. 7.3.3. Both methods show almost identical IQE values for this sample
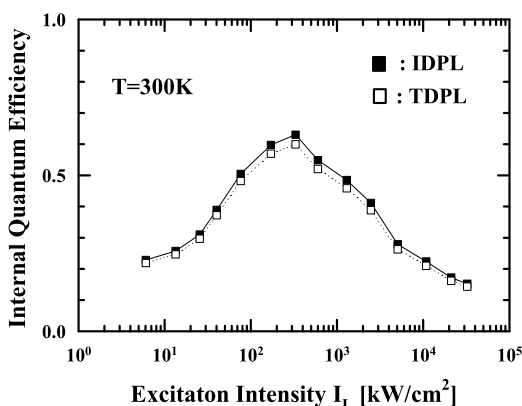
room temperature. The dependency of the PL efficiency indicates that the measured sample does not have a 100 % IQE at 8 K. The maximum IQE at 8 K is estimated as 83 % by the IDPL. In addition to the IDPL method being able to measure the absolute IQE at any temperature, it has advantages that can minimize errors since the integrated PL intensity as a function of excitation intensity at room temperature and their relative ratios are used.

7.3.4 Temperature-Dependent Time-Resolved Photoluminescence (TD-TRPL) Method

The TD-TRPL method is frequently utilized in order to estimate the IQE and the radiative and nonradiative carrier lifetimes simultaneously [96–102]. The short radiative lifetime and long nonradiative lifetime in LEDs are desirable for a reduced carrier density in QWs, high-power operations, and good reliability. The TD-TRPL method utilizes a sort of short pulse train as a photo-excitation source, whose sin-

gle pulse width is about a few hundred femtoseconds and the pulse period is long enough to measure the single pulse response safely without intersymbol interferences. It also measures spectral and temporal responses with a monochromator and a high-speed detector such as a streak camera, respectively.

A transient TRPL response integrated over wavelengths is generally well fitted by the sum of two exponential functions with fast and slow decay times as

$$I(t) = I_o(T) \left\{ \exp\left(-\frac{t}{\tau_{\text{TRPL}}(T)}\right) + C_{\text{slow}} \exp\left(-\frac{t}{\tau_{\text{slow}}(T)}\right) \right\}, \quad (7.33)$$

where C_{slow} is a constant. $\tau_{\text{TRPL}}(T)$ and $\tau_{\text{slow}}(T)$ are the fast and slow decay times of luminescence at the beginning and ending of the PL decay curve, respectively. They are usually a function of temperature T (in K) as well as the excitation laser intensity I_L (in W/m^2). This two-exponential characteristic is often understood by different types of radiatively recombining carriers, including free carriers, free excitons, and bound excitons. It has been explained that the fast decay time is a result of the free carrier recombination in the blue side of the spectrum and under a higher density of excitons. The slow decay time comes from free excitons in the spectral position and bound excitons localized by impurity, an alloy, or well width fluctuations in the red side of the spectrum [108–112]. In fact, an emission spectrum at an instant is observed by the fastest radiative recombination process among all recombination processes. On the other hand, a PL decay curve at that time is determined by the total carrier recombination rate, i.e. the sum of radiative and nonradiative recombination rates. In this work, the carrier dynamics are expected in the high carrier density regime, corresponding to the normal LED or LD operation conditions, especially at room temperature. Therefore, we are only concerned with the fast component for our purpose.

With the single exponential approximation of Eq. (7.33), the luminescence decay $I(t)$ is expressed as

$$I(t) = I_{o,\text{eff}}(T) \exp\left(-\frac{t}{\tau_{\text{TRPL}}(T)}\right), \quad (7.34)$$

where $I_{o,\text{eff}}(T)$ is an effective PL intensity integrated over entire wavelengths and is obtained by integrating both sides of Eq. (7.33) and Eq. (7.34) with respect to time.

$$I_{o,\text{eff}}(T) = I_o\left(1 + C_{\text{slow}}\frac{\tau_{\text{slow}}(T)}{\tau_{\text{TRPL}}(T)}\right). \quad (7.35)$$

One of the most important steps in the TD-TRPL method is to discriminate the radiative and nonradiative decay times at room temperature. The inverse of $\tau_{\text{TRPL}}(T)$ is described by the sum of two components,

$$\frac{1}{\tau_{\text{TRPL}}(T)} = \frac{1}{\tau_{r,\text{PL}}(T)} + \frac{1}{\tau_{\text{nr},\text{PL}}(T)}, \quad (7.36)$$

where $\tau_{r,\text{PL}}(T)$ and $\tau_{\text{nr},\text{PL}}(T)$ are the PL decay times due to the radiative and nonradiative carrier recombination processes, respectively. A relationship between the measured temporal TRPL response and the $\tau_{\text{TRPL}}(T)$ is obtained by

$$I_n(T) = \int_0^\infty \frac{I(t)}{I_{o,\text{eff}}} dt = \tau_{\text{TRPL}}(T) = \frac{\tau_{r,\text{PL}}(T)\tau_{\text{nr},\text{PL}}(T)}{\tau_{r,\text{PL}}(T) + \tau_{\text{nr},\text{PL}}(T)}. \quad (7.37)$$

In some experimental cases, the normalized TRPL intensity $I_n(T)$ defined in Eq. (7.37) does not change below a certain temperature T_c and drops gradually at temperatures above T_c. Similar to the TDPL method, it is assumed that nonradiative recombination is completely negligible and the IQE is unity in this temperature range, i.e., $T < T_c$. This means that the normalized intensity at low temperature T_l ($T_l < T_c$) is approximately equal to the radiative lifetime $\tau_{r(d)}(T_l)$, i.e.,

$$I_n(T_l) \approx \tau_{r,\text{PL}}(T_l). \tag{7.38}$$

If the radiative lifetime is independent of temperature, the expression of the IQE at a high temperature T_h ($T_h > T_c$) can be obtained from Eqs. (7.23), (7.37), and (7.38):

$$
\begin{aligned}
\eta_{\text{IQE}}(T_h) &= \frac{\tau_{\text{nr,PL}}(T_h)}{\tau_{r,\text{PL}}(T_h) + \tau_{\text{nr,PL}}(T_h)} \\
&\approx \left[\frac{\tau_{r,\text{PL}}(T_h)\tau_{\text{nr,PL}}(T_h)}{\tau_{r,\text{PL}}(T_h) + \tau_{\text{nr,PL}}(T_h)} \right] \Big/ \tau_{r,\text{PL}}(T_l) = \frac{I_n(T_h)}{I_n(T_l)}.
\end{aligned} \tag{7.39}
$$

The radiative and nonradiative lifetimes at room temperature are obtained by

$$\tau_{r,\text{PL}}(T_h) = \frac{\tau_{\text{TRPL}}(T_h)}{\eta_{\text{IQE}}(T_h)} = \tau_{\text{TRPL}}(T_h)\frac{I_n(T_l)}{I_n(T_h)}, \tag{7.40}$$

$$\tau_{\text{nr,PL}}(T_h) = \frac{\tau_{\text{TRPL}}(T_h)}{1 - \eta_{\text{IQE}}(T_h)} = \tau_{\text{TRPL}}(T_h)\frac{I_n(T_l)}{I_n(T_l) - I_n(T_l)}. \tag{7.41}$$

Hence, from the measurement of the intensity $I_n(T)$ and lifetime $\tau_{\text{PL}}(T_h)$ with respect to T, radiative and nonradiative lifetimes are derived as a function of T.

We have reviewed the conventional TD-TRPL method, which characterizes the IQE and the radiative and nonradiative carrier lifetimes simultaneously. This method uses numerous assumptions: (1) the zero nonradiative recombination rate and 100 % IQE at a low enough temperature, (2) the excitation intensity is not dependent on carrier lifetimes or the IQE, (3) the radiative lifetime is invariant with temperature, and (4) there are no clear relationships between the PL decay times and the carrier recombination lifetimes through the radiative and nonradiative carrier recombination processes. Assumptions (1) and (2) are investigated in Sect. 7.3.2, and the results show that the IQE cannot be considered as unity and is usually dependent on the excitation intensity. Moreover, the dominant radiative recombination process in InGaN-based QWs changes with both excitation condition and temperature. The radiative carrier lifetime calculated from Eq. (7.40) usually shows a large dependency on temperature since the IQE depends on temperature [99–101]. Therefore, the internal parameters of the QWs should be carefully estimated in the TD-TRPL experiment.

7.3.5 Room-Temperature Time-Resolved Photoluminescence (RT-TRPL) Method

This section proposes a simple method to estimate the IQE and carrier lifetimes at room temperature. We assume a situation that optical pumping intensity is so

Fig. 7.15 Illustration of the
TRPL experiment.
(**a**) Modified optical pump
beam power, (**b**) carrier
density generated by
modified optical pump beam
power P_{pump}, and (**c**) detected
light emission by radiative
recombination of free carriers
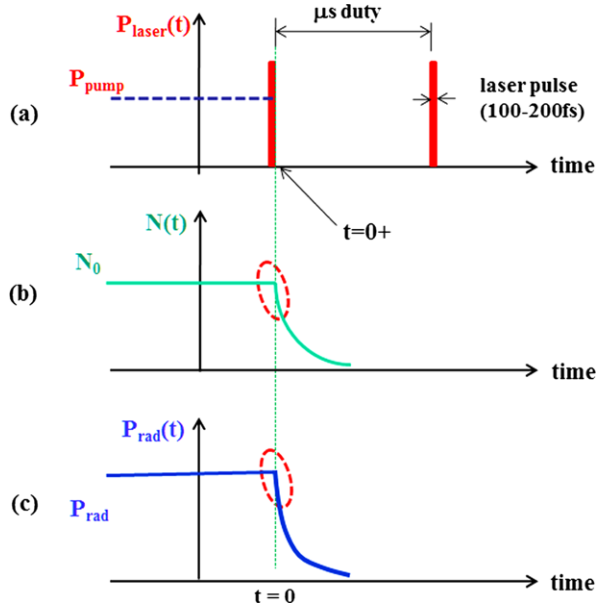
high that excited carriers are recombined as free carriers rather than excitonic ones.
The method is based on room-temperature TRPL (RT-TRPL) measurement and the
analysis of the carrier rate equation depending on the laser excitation intensity. Im-
mediately after the illumination of an extremely short laser pulse of less than a few
hundred femtoseconds in a QW structure, electrons and holes are photo-excited in
the conduction band and valence band, respectively. These excited carriers quickly
relax in the electron momentum and energy spaces through many scattering pro-
cesses within a few tenths of picoseconds. They are distributed into the most sta-
ble states known as the quasi-Fermi distributions. Then, redistributed free carriers
start to recombine with each other in the range of a few nanoseconds to a hundred
nanoseconds and lose their energies radiatively and nonradiatively.

A schematic of the TRPL experiment is shown in Fig. 7.15. At the beginning of
the temporal response in the TRPL method, $t = 0^+$, electrons and holes generated in
a QW by a single femtosecond laser pulse will already be in the quasi-equilibrium
state. This situation can be considered as a pump light source of a certain laser
optical power, i.e., P_{pump}, in the CW PL method as shown in Fig. 7.15(a). Also, the
density of free carriers in the conduction band can be assumed as initially pumped to
the steady state N_0 by P_{pump} and begins its recombination process just after the pulse
has been turned off ($t = 0+$) as shown in Fig. 7.15(b). Here, P_{pump} is assumed to
be sufficiently high so that the recombination of free excitons can be ignored. Since
the spontaneous light emission from LEDs shown in Fig. 7.15(c) is proportional to
the radiative recombination, the detected amount of light power $P_{\text{PL}}(t)$ is given by

$$P_{\text{TRPL}}(t) = (h\bar{\nu}\eta_{\text{det}}\eta_{\text{LEE}}V_a)BN^2(t) \tag{7.42}$$

where $h\bar{\nu}$, η_{det}, η_{LEE}, and V_a are the average photon energy, the detector coupling efficiency, the light extraction efficiency, and the active volume, respectively.

In this model [74, 113], the carrier rate equation at room temperature with a negligible Auger coefficient C in the QW active layer is written as

$$\frac{dN}{dt} = 0 = -\left(AN_o + BN_o^2\right) + (1-R)\frac{\alpha P_{\text{pump}}}{h\nu_L A_L} \quad \text{for } t \leq 0, \tag{7.43}$$

$$\frac{dN}{dt} = -\left(AN + BN^2\right), \quad \text{where } \frac{1}{\tau} = \frac{1}{\tau_{\text{nr}}} + \frac{1}{\tau_r} = A + BN \quad \text{for } t > 0, \tag{7.44}$$

where $R, \alpha, P_{\text{pump}}, h\nu_L$, and A_L are the reflectivity of the semiconductor surface, the absorption coefficient in QWs, time-averaged power of the excitation source, the photon energy of the excitation source, and the net illuminated area at the surface of the LED, respectively. The subscript 'o' represents the steady state. τ is total carrier lifetime which can be separated into nonradiative recombination carrier lifetime (τ_{nr}) and radiative recombination carrier lifetime (τ_r). We consider for all t as an approximation that the nonradiative recombination coefficient A, the radiative recombination coefficient B, and the optical absorption coefficient α as constants regardless of the carrier density $N(t)$.

For $t > 0$, the carrier rate equation is written as

$$\frac{d(\Delta N)}{dt} = -\left(AN_0 + BN_0^2\right) - (A + 2BN_0)\Delta N - B(\Delta N)^2,$$
$$\text{with } N(t) = N_0 + \Delta N(t). \tag{7.45}$$

N_0 is the carrier density for $t < 0$ and $\Delta N(t)$ represents the time-dependent variation of the carrier density for $t > 0$. Then, the temporal responses of the carrier density $N(t)$ in Eq. (7.45) is obtained analytically with the boundary conditions of $N(t=0) = N_o$ and $N(t=\infty) = 0$ as

$$N(t) = \frac{A}{B}\frac{\eta_{\text{IQE}(0)}e^{-At}}{1 - \eta_{\text{IQE}(0)}e^{-At}}, \quad \text{where } \eta_{\text{IQE}(0)} = \frac{BN_0^2}{AN_0 + BN_0^2}. \tag{7.46}$$

The nonradiative lifetime is at the steady state and $\tau_{\text{nr}} = 1/A$ as defined in Eq. (7.44). Then, the detected light power can be calculated from Eqs. (7.42) and (7.46). For simplicity, a temporal response of the integrated PL intensity is normalized by the peak intensity at $t = 0$, i.e., $P_{\text{TRPL}}(t=0)$, and is given by

$$P_{\text{nor}}(t) = \frac{P_{\text{TRPL}}(t)}{P_{\text{TRPL}}(0)} = \left(\frac{A}{BN_0}\right)^2\frac{\eta_{\text{IQE}(0)}^2 e^{-2At}}{(1 - \eta_{\text{IQE}(0)}e^{-At})^2}. \tag{7.47}$$

$P_{\text{nor}}(t)$ is a more complicated function than the single exponential function with respect to time for constant carrier recombination coefficients. It can be approximated as a single exponential function at the initial and final stages as follows:

$$P_{\text{nor}}(t) \approx e^{-t/\tau_{\text{TRPL,initial}}} \quad \text{as } t \to 0, \tag{7.48}$$

$$P_{\text{nor}}(t) \approx \left(\frac{A}{A + BN_0}\right)^2 e^{-t/\tau_{\text{TRPL,final}}} \quad \text{as } t \to \infty. \tag{7.49}$$

In Eqs. (7.48) and (7.49), each lifetimes are given as follows:

$$\frac{1}{\tau_{\text{TRPL,initial}}} = 2(A + BN_0) = 2\left(\frac{1}{\tau_{\text{nr}}} + \frac{1}{\tau_r}\right), \tag{7.50}$$

$$\frac{1}{\tau_{\text{TRPL,final}}} \approx 2A = \frac{2}{\tau_{\text{nr}}}. \tag{7.51}$$

Note that at $t = 0+$, $\tau_{\text{TRPL,initial}}$ is one half of the carrier lifetime τ defined in Eq. (7.44). On the other hand, for the case of the final stage of the TRPL response ($t \gg \tau_{\text{TRPL,initial}}$), $\tau_{\text{TRPL,final}}$ can be approximately given in terms of the nonradiative recombination coefficient only. The nonradiative carrier lifetime τ_{nr} is two times of the carrier lifetime of the final stage of the TRPL response ($\tau_{\text{TRPL,final}}$). Equations (7.50) and (7.51) mean that both the radiative and nonradiative recombination processes determine the lifetime of the TRPL response at the initial stage, whereas the nonradiative recombination process mainly determines the lifetime at the final stage. This can be understood as both radiative and nonradiative recombination processes competing with each other at the initial stage of the PL decay when the radiated power is high. Then, as time has progressed to the final stage of the PL decay, most of the radiative recombination process has taken place and the PL decay curve is mainly governed by the nonradiative recombination lifetime.

From Eqs. (7.50) and (7.51), the IQE at $t = 0$ and the radiative recombination carrier lifetime (τ_r) can be obtained directly from measured $\tau_{\text{TRPL,initial}}$ and $\tau_{\text{TRPL,final}}$ by

$$\eta_{\text{IQE}(0)} = \frac{\tau_{\text{TRPL,final}} - \tau_{\text{TRPL,initial}}}{\tau_{\text{TRPL,final}}}, \tag{7.52}$$

$$\frac{1}{\tau_r} = \frac{1}{2}\left(\frac{1}{\tau_{\text{TRPL,initial}}} - \frac{1}{\tau_{\text{TRPL,final}}}\right), \tag{7.53}$$

$$\tau_{\text{nr}} \approx 2\tau_{\text{TRPL,final}}. \tag{7.54}$$

Thus, one can find the IQE, radiative recombination carrier lifetime (τ_r) and nonradiative recombination carrier lifetime (τ_{nr}) simultaneously by just one measurement of the TRPL response at room temperature.

We applied the proposed method to a vertical blue LED chip with a size of 1 mm × 1 mm. The sample was excited by the Hamamatsu PLP-10, which has an operation pulse width less than 100 ps and a repetition rate up to 100 MHz. The excitation power of the PLP-10 was varied by neutral density filters from 0.04 mW to 0.16 mW. The value of $P_{\text{TRPL}}(t)$ was measured by a Hamamatsu C4780 Picosecond Fluorescence Lifetime Measurement System with a temporal resolution of 5 ps. $P_{\text{TRPL}}(t)$ of each excitation power was measured at every 20 ps until the maximum number of photons counted reached 1000 (inset of Fig. 7.16). In our experimental set-up, an averaged noise level was less than 0.3 % of the maximum $P_{\text{TRPL}}(t)$. After subtracting the background noise, results were normalized to the peak values and they were numerically fitted with a combination of exponential functions and then drawn on a natural logarithmic scale, as shown in Fig. 7.16. In the figure,

Fig. 7.16 Fitted results of the measured excitation power-dependent TRPL data shown in the *inset*. Examples of $\tau_{\text{TRPL,final}}$ and $\tau_{\text{TRPL,initial}}$ are also shown
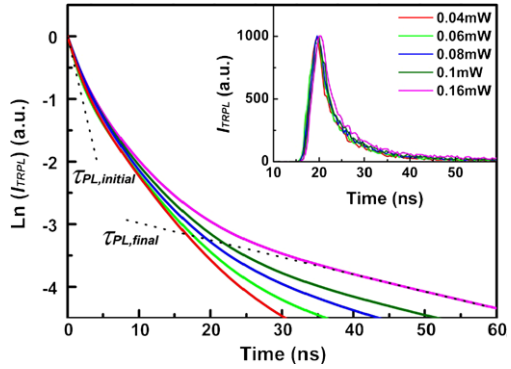
Fig. 7.17 Estimated τ_{nr}, τ_r, and the IQE as a function of excitation power. *Solid lines* are drawn to guide the eyes
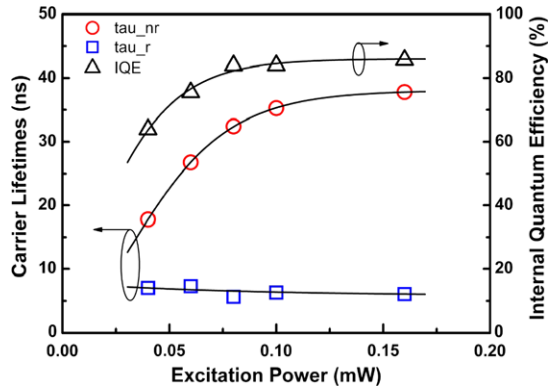
there is a time range during which $P_{\text{TRPL}}(t)$ decays linearly, which means that the response itself can be approximated by a single exponential function with a decay time constant of $\tau_{\text{TRPL,initial}}$ or $\tau_{\text{TRPL,final}}$ on a linear scale. As shown in the figure, generally the fitting range of TRPL lifetimes for $\tau_{\text{TRPL,final}}$ is much wider than that of $\tau_{\text{TRPL,initial}}$, which indicates a smaller estimating error of τ_{nr} compared to that of τ_r.

The estimated results are summarized in Fig. 7.17. Here, τ_{nr} was calculated by using Eq. (7.54). $\tau_{\text{TRPL,final}}$ was estimated at the intensity range between -3.5 and -4 in the y-axis of Fig. 7.16 corresponding to 3 %–1 % of the peak intensity. The results show that τ_{nr} gradually increases with increasing excitation power P_{pump} up to 0.1 mW. It seems clear that τ_{nr} becomes independent for $P_{\text{pump}} > 0.1$ mW. This implies that the nonradiative recombination processes may be different for P_{pump} below and above 0.1 mW. We suppose that there are two defect-related nonradiative recombination processes in a LED chip. The first is the carrier loss via the tunneling transport from the QW active to the p-side as shown in Fig. 7.2. This defect-related tunneling process is easily saturated by increasing carrier density [114], which explains an increasing τ_{nr} for $P_{\text{pump}} < 0.1$ mW. The other is carrier loss via the SRH recombination inside the QWs. It is a well-known fact that the nonradiative carrier

Fig. 7.18 (**a**) Fitted results of the normalized TRPL responses at different temperatures and (**b**) comparison of the estimated τ_{nr} from the TRPL (*left y-axis, filled circles*) and the time-integrated photoluminescence (TIPL) efficiency (*right y-axis, open circles*)
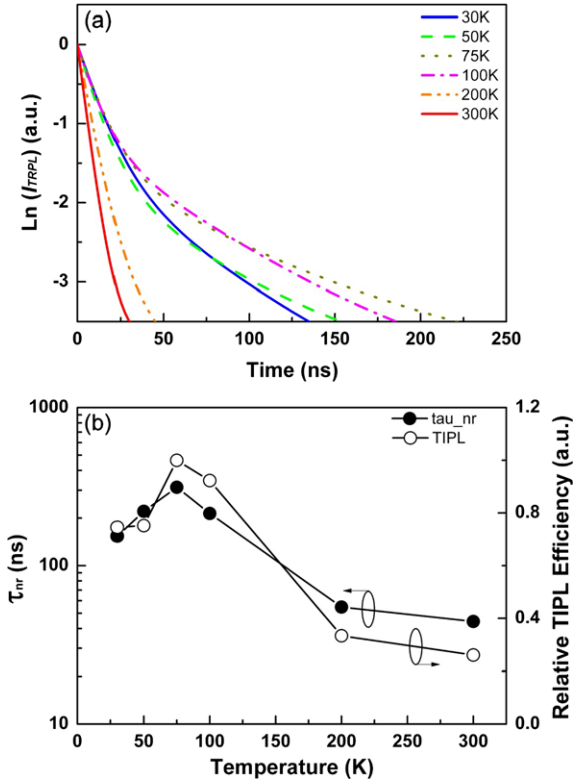
lifetime via the SRH recombination is hardly changed by the carrier density. This is the case for $P_{pump} > 0.1$ mW were τ_{nr} is almost constant over a wide range of excitation powers. On the other hand, $\tau_{PL,initial}$ was determined from 0 to -0.2, where the TRPL intensity decays to ~80 % of its peak intensity. The results show that the measured τ_r slightly decreases with increasing excitation power. This is due to the fact that the carrier density increases with the excitation power and τ_r is inversely proportional to the carrier density.

The TD-TRPL measurement on a commercial LED sample was carried out as the temperature changed from 30 K to 300 K using a closed-cycle helium cryostat. It is well known that the PL efficiency is sensitive to the temperature, T, because the nonradiative recombination processes are a function of temperature, i.e., $\tau_{nr}(T)$. This means that $\tau_{TRPL,final}$ in our model should be dependent on temperature. In Fig. 7.18(a), the fitted results of the normalized TD-TRPL measurements are depicted using a natural logarithmic scale. They clearly show that the linear decay in the last stage of $P_{TRPL}(t)$ increases as temperature increases. This indicates that τ_{nr} is decreasing as the temperature increases. $\tau_{TRPL,final}$ is estimated from Fig. 7.18(a) and τ_{nr} is obtained from Eq. (7.54) and plotted in Fig. 7.18(b). The largest τ_{nr} is observed at 75 K. Since τ_{nr} is inversely proportional to the PL efficiency, we tried to compare τ_{nr} to the TRPL intensity by integrating the TRPL response from its

Fig. 7.19 (**a**) Measured relative EL efficiency curves at room temperature for four different blue LED chips and (**b**) the correlation between the injected current level to achieve the maximum EL efficiency (*left y-axis*, *open circles*) and the estimated τ_{nr} (*right y-axis*, *filled circles*) from the TRPL measurement for each sample
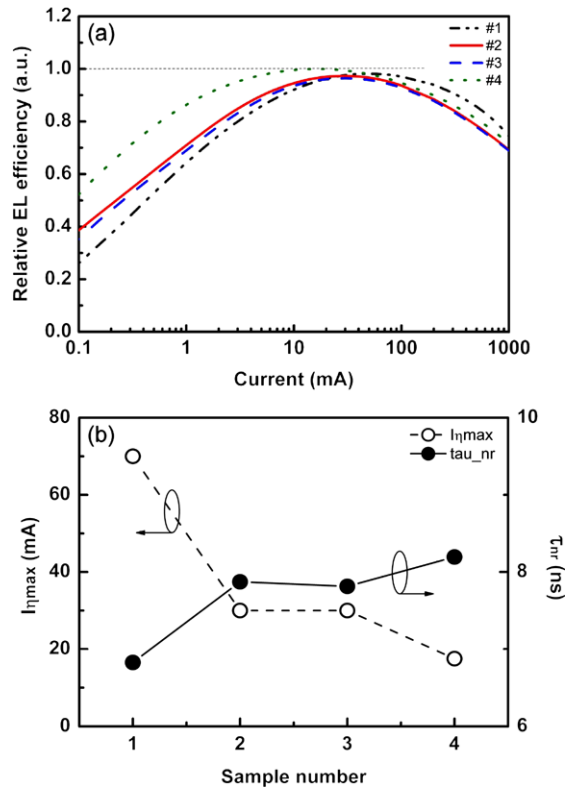
peak to -4 on the *y*-axis of Fig. 7.18(a). Relatively measured TRPL efficiency as a function of temperature is shown as open circles in Fig. 7.18(b) as well. Both curves in Fig. 7.18(b) show very similar tendency with temperature, which supports our analysis result that $\tau_{TRPL,final}$ is a measure of τ_{nr}.

Next, we show another application of our model. We also prepared four blue LED chips emitting with slightly different EL efficiencies from the same wafer. Thus, they have the same device size and active-layer structure but different defect levels, i.e., the nonradiative recombination rates. Figure 7.19(a) shows the relative EL efficiency as a function of injected current at room temperature. The current at the peak EL efficiency, $I_{\eta\,max}$, is quite different from each chip. This indicates that each sample has a different defect density. It is known that $I_{\eta\,max}$ becomes high as the nonradiative recombination rate increases in case of same physical dimensions. The estimated $I_{\eta\,max}$ and τ_{nr} for each sample are shown in Fig. 7.19(b). The figure indicates that the sample with a smaller τ_{nr} requires larger $I_{\eta\,max}$ of the peak EL efficiency peak. This shows quite consistent results between the proposed TRPL analysis and the EL characteristics. The proposed TRPL analysis method can be utilized to understand the carrier dynamics in semiconductors more easily.

7.4 Conclusion

We have summarized the definitions of various LED efficiencies, i.e., the wall-plug efficiency ($\eta_{\text{wall-plug}}$), the voltage efficiency (η_{VTG}), the injection efficiency (η_{INJ}), the radiative efficiency (η_{RAD}), the light extraction efficiency (η_{LEE}), the internal quantum efficiency (η_{IQE}), and the external quantum efficiency (η_{EQE}). Among these efficiencies, the IQE is the most important but its characteristics as well as measuring methods have not been clearly known yet. We have first introduced experimental results of the IQE droop depending on temperature and current. Then, the IQE droop mechanisms ever reported have been reviewed. Up to now, none of the popular droop model is able to explain all the experimental droop phenomena consistently. Auger and defect-related processes consider additional increase in the nonradiative recombination process inside the QW active region as well as sufficient carrier capture rates into the QW active region. On the other hand, carrier-transport-related models suspect additional increase in the nonradiative recombination outside the QW active region like the p-side and insufficient carrier capture rate into the QW active region. However, all these models focus on faster increases in the nonradiative recombination without taking into account the possible mechanisms of slower increase in the radiative recombination rate.

We have also investigated and compared the temperature dependences of the PL efficiency, the open-circuit voltage, and the EL efficiency as functions of operating temperature and excitation power. Through comparative studies, we have suggested that the inherent origin of the efficiency droop is the saturation of the radiative recombination rate in the InGaN QW at low current and subsequent increase in the nonradiative recombination rates at high current. The degree of saturation is determined by operating temperature and effective active volume. Although the saturation of the radiative recombination rate is common origin of the IQE droop, the shapes of the IQE versus current, i.e. the IQE droop curve, vary with the dominant nonradiative recombination process.

We have reviewed the IQE measurement methods both theoretically and experimentally. We have introduced a very simple fitting method based on the constant ABC model in the carrier rate equation. The method utilizes just two parameters of current at the IQE peak and the IQE peak value itself. It is possible to know the IQE and the carrier recombination current simultaneously without any knowledge of A, B, and C recombination coefficients. This method is very useful to extract the recombination coefficients with very small errors from the measured curve of current versus optical power.

Then, the IQE measurement methods have been reviewed by focusing on all-optical methods such as the temperature-dependent photoluminescence (TDPL) and the temperature-dependent time-resolved photoluminescence (TD-TRPL) methods. The assumptions used in the TDPL and TD-TRPL methods have been theoretically examined by using the carrier rate equation in semiconductors. Another method based on the intensity-dependent PL, i.e. IDPL, has also been presented and demonstrated experimentally. This method utilizes the laser excitation power as a varying parameter instead of the temperature in the conventional TDPL and TD-TRPL

methods and is able to evaluate the IQE at any temperature without the aid of any calculated parameters.

We have also developed a theory of carrier dynamics from the TRPL measurement just at room temperature and found a simple and effective method to obtain the carrier lifetimes and the IQEs. The meanings of the measured TRPL response curves have been verified rigorously by solving the carrier rate equation in the QW active layer. It has been shown that the radiative and nonradiative recombination processes compete with each other at the initial stage of the TRPL and is gradually dominated by the nonradiative recombination at the final stage. The proposed method can be regarded as a very unique and efficient tool for measuring the nonradiative recombination carrier lifetime without the tediously long measurements. The method also avoids the temperature dependencies of carrier lifetimes. We hope that this article is helpful for better understanding of the carrier dynamics in semiconductors and further improvements in optoelectronic device performances.

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Chapter 8
Electrical Properties, Reliability Issues, and ESD Robustness of InGaN-Based LEDs

M. Meneghini, G. Meneghesso, and E. Zanoni

This chapter describes the physical mechanisms that may induce the degradation of InGaN-based LEDs. Before going into a detailed description of the main degradation processes, we will give a brief overview on the electrical (current-voltage) characteristics of LEDs, to provide the basic information which is useful to understand the electrical characterization data reported in the subsequent part of the chapter. On the basis of literature data, we will analyze the main mechanisms responsible for current conduction, with focus on tunneling mechanisms, reverse-current conduction, and on the origin of the high ideality factors, always present in InGaN LEDs.

After this introductory section, we will discuss the main physical mechanisms that may limit the lifetime of GaN-based LEDs submitted to short and long-term operation. More specifically, we will focus on the following mechanisms: (i) the degradation of the blue semiconductor chip, due to the generation of non-radiative recombination centers, and to the degradation of the p-type semiconductor and contacts; (ii) the degradation mechanisms that affect the performance of the package and phosphors, which are typically activated at high temperature levels; and (iii) the failure of LEDs submitted to ElectroStatic Discharges (ESD).

8.1 Current-Voltage Characteristics

The electrical characteristics of LEDs based on nitrides are different from those of conventional diodes, such has *pn* or Schottky diodes. Several factors can influence the main parameters of the devices, and in particular the operating voltage, the reverse current, the ideality factor, and their dependence on temperature. In this

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section we briefly summarize the main features of the electrical characteristics of GaN-based LEDs: we focus on the origin of reverse-current conduction, tunneling-related components, and high ideality factors, since this information is of fundamental importance for understanding the degradation processes that limit the reliability of LEDs, which are described in the next section.

In Fig. 8.1 we report the typical current-voltage (I–V) characteristics of a multiple-quantum well (MQW) LED, measured at different temperature levels between 100 K and room temperature (RT). The curves are plotted both in linear (left) and semi-logarithmic (right) scale. With increasing temperature from 100 K to RT, the forward voltage of the LEDs shows a remarkable decrease, due to the following factors: (i) a decrease in the resistivity of the p-type and n-type layers; (ii) an increase in the injection of holes from the p-side of the diodes to the active layer; (iii) a decrease in the resistivity of the ohmic contacts. An accurate analysis of the I–V curves in semi-logarithmic scale reveals the presence of different conduction mechanisms.

For low voltage levels (here below 2.5 V), the semi-logarithmic I–V curves indicate that current has a nearly exponential dependence on voltage: the slope of the semi-logarithmic IV curves does not significantly depend on temperature, thus suggesting that in this voltage region the main mechanism responsible for current conduction is a field-dependent process—such as tunneling—rather than a thermally activated one. For GaN-based LEDs, forward-bias tunneling components are usually represented in the form

$$I \sim I_0 \exp\left(\frac{qV}{E}\right) \tag{8.1}$$

where E represents an energy parameter, which reflects the transparency of the energy barriers to tunneling carriers, and has a negligible temperature dependence [1, 2]. According to [3, 4], for a p^+-n junction, parameter E can be expressed as

$$E = \frac{4\hbar q \sqrt{N_D}}{\pi \sqrt{m^* \epsilon}} \tag{8.2}$$

where q is the electron charge, $\hbar$ is the Planck's constant divided by 2π, m^* is the effective mass of the electrons, and N_D represents the doping concentration. Typical reported values for E range between 72 and 220 meV for GaN based LEDs [1]. As deeply discussed in [1, 5, 6], several models have been proposed to explain the tunneling processes that occur in GaN-based LEDs: (i) tunneling of holes to the active region, (ii) tunneling of electrons to deep levels located within the barriers [5] or the p-GaN region [1], with subsequent recombination; (iii) tunneling recombination via multiple-step processes [1], (iv) diagonal tunneling [6]. Several defects may enhance carrier tunneling in GaN-based MQW LEDs, including point defects located within the quantum well region, clusters of defects close to dislocations, nitrogen vacancies or oxygen impurities [1], Mg-related defective structures, defects related to indium segregation [5].

Recent studies demonstrated that strong tunneling between the quantum wells and the barriers can be detected by means of spatially and spectrally-resolved electroluminescence measurements. Analysis on blue [7] and green [5] LEDs operated
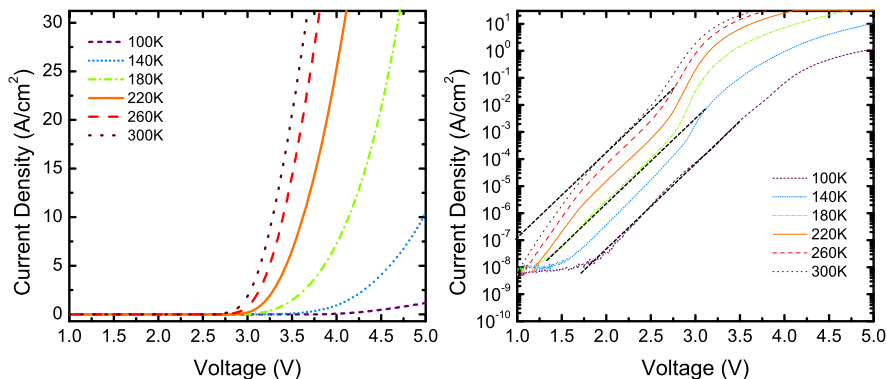
Fig. 8.1 Current-voltage characteristics of a multiple quantum well LED measured at different temperature levels. *Curves* are plotted in (*left*) linear scale and in (*right*) semi-logarithmic scale

at low forward bias levels reveal the presence of localized dots, with emission wavelength significantly longer than the peak wavelength of the devices. In Fig. 8.2 we report the false-color micro-electroluminescence images taken at two different wavelengths on a green LED, with a peak wavelength close to 540 nm: as can be noticed, this device emits a weak luminescence signal also in the yellow-red spectral region (see the image taken with the monochromator set to 600 nm). The presence of yellow localized luminescence dots was found to be strongly correlated to the existence of high tunneling components in the $I-V$ characteristics of the devices (see [5]): dot-like luminescence was ascribed to the injection/tunneling of electrons from the quantum wells to the barriers, with subsequent recombination via the defective states that are commonly considered to be responsible of yellow luminescence in GaN [8].

In Fig. 8.1, it is clearly noticeable that while current conduction is dominated by tunneling for $V < 2.5$ V, for higher voltage levels the slope of the semi-logarithmic $I-V$ curves shows a significant increase, indicating that current diffusion and recombination is starting to dominate. As a consequence, the relation between current and voltage can be expressed as

$$I = I_0\left(e^{qV/nkT} - 1\right) \tag{8.3}$$

where I_0 is the saturation current, q is the electron charge, k is the Boltzmann constant, and n is the ideality factor. In agreement with (8.3), in the region between 2.5 V and 3 V the slope of the $I-V$ curves is significantly temperature-dependent. For high current densities (in Fig. 8.1), the presence of the series resistance results in a significant decrease in the slope of the $I-V$ curves: series resistance originates from the non-ideal conductivity of the ohmic contacts and of the quasi-neutral regions, and may significantly impact the operating voltage of the devices. High resistive components may also result in the so-called current-crowding phenomenon: due to the non-ideal conductivity of the semiconductor layers, current crowds in proximity of the mesa edge (see a schematic representation of this mechanism in

Fig. 8.2 False-color
emission images showing the
intensity of the EL of green
LED, measured through a
monochromator set at
(**a**) 540 nm and (**b**) 600 nm
(monochromator spectral
width is 10 nm)
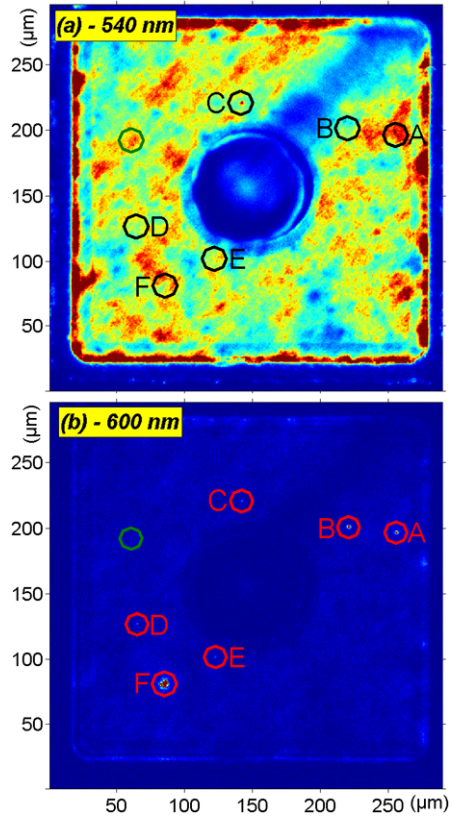
Fig. 8.3 Schematic
representation of the current
crowding phenomenon.
Dashed lines represent the
existence of preferential
current paths
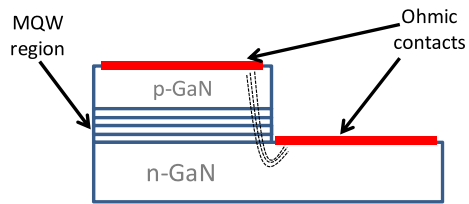
Fig. 8.3), for LEDs with horizontal structure. This mechanism has been investigated
by several research groups (see for instance [9–15]), who provided information on
the role of current crowding in limiting the optical performance of LEDs, and com-
plete analytical models for the description of this phenomenon. Distributed models
are commonly used for the description of current crowding in horizontal LEDs: as
widely discussed in [12], both the resistance of the *n*-type and *p*-type materials
must be considered for the definition of an accurate model for current crowding. In
Fig. 8.4 we report a typical schematic circuit used for the study of current crowding:
the vertical resistors are used to take into account the resistance of the ohmic contact
on *p*-GaN, and of the *p*-type material, while the horizontal resistors represent the
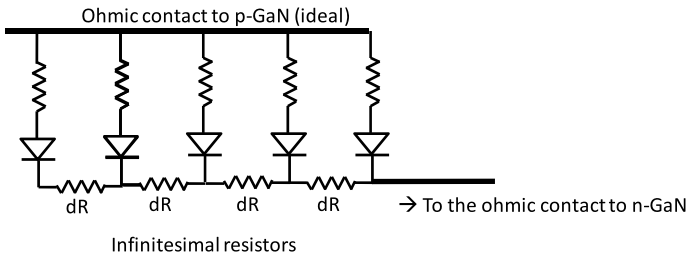
Fig. 8.4 Schematic representation of a model used for the study of current crowding in LEDs with horizontal structure (see [12] for a detailed description of this model)

ohmic losses in the n-type region. The uniformity of current distribution in a horizontal LED can be evaluated by calculating the *current spreading length* (see [12]), according to the formula

$$L_s = \sqrt{(\rho_c + \rho_p t_p) t_n / \rho_n} \tag{8.4}$$

where ρ_p, ρ_n and ρ_c represent the resistivities of the p-type material, of the n-type material, and of the ohmic contact to p-GaN, while t_p and t_n are the thicknesses of the p and n-type regions. In a first-order approximation, the *current spreading length* represents the distance (from the mesa edge) for a $1/e$ decrease in current density. As shown in (8.4), L_s depends only on the geometry of the samples (thickness of the p and n-side regions) and on the specific resistivity of the materials adopted for the realization of the LEDs. Current crowding can be significantly limited by decreasing the resistivity of the n-GaN layer, and by increasing the thickness of the n-type material.

Current crowding can have detrimental effects in terms of performance, thermal management and reliability. In fact, when current is focused on a limited region, this results in a reduction of the effective area of the LED. Furthermore, in the region characterized by crowding, current density is significantly higher than in other regions. This determines a decrease in LED efficiency since: (i) due to the efficiency droop, IQE decreases monotonically with current density; (ii) a stronger self-heating of the junction in proximity of the most effective conductive paths induces a further efficiency decrease.

A significant current crowding can be present even when the current spreading length has been carefully optimized: an example is shown in Fig. 8.5, which reports the false-color electroluminescence map measured on a $500 \times 500 \ \mu m^2$ LED chip with horizontal structure. As can be noticed, the device does not show any crowding related to the resistivity of the n or p-type material (i.e. current does not flow only in proximity of the mesa edge as schematically shown in Fig. 8.3). On the other hand, current focuses to the center of the LEDs: this mechanism can be partly explained by considering that the center of the LED is the region from which heat extraction is most difficult. During LED operation, junction temperature can be higher at the center of the LED, with respect to the borders. This may result in an increased

Fig. 8.5 False-color
emission microscopy image
showing the distribution of
light intensity on a
$500 \times 500 \ \mu m^2$ LED chip
with horizontal structure
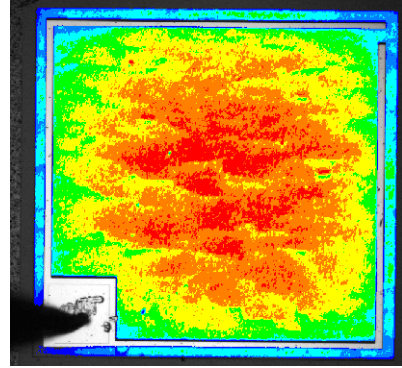
current density at the center, and therefore in a higher emission intensity, as shown
in Fig. 8.5.

8.2 The Ideality Factor of GaN-Based LEDs

The $I-V$ curves of diodes are commonly modeled by means of the first-order equation

$$I = I_0\big(e^{qV/nkT} - 1\big) \tag{8.5}$$

where I_0 is the saturation current, q is the electron charge, k is the Boltzmann constant, and n is the so-called ideality factor, which is a measure of how much the $I-V$ curve of the diode is close to the theoretical one. For conventional diodes (e.g. Schottky and pn-diodes based on silicon), this parameter ranges between 1 and 2. Based on the theory proposed by Sah-Noyce-Shockley [16], values of n close to 1 indicate that current conduction is dominated by diffusion processes, while on the other hand ideality factors equal to 2 indicate that current is strongly influenced by carrier generation and recombination. The ideality factor n can be easily estimated through the fitting of the semi-logarithmic $I-V$ curves of the diodes.

Several reports [17–23] indicate that the ideality factor of GaN-based diodes and LEDs can be significantly higher than 2. Many theories were proposed to explain such anomalously high ideality factors: to provide a more detailed view on this particular aspect, in the following we give a brief description of the most common theories.

In [4], Franssen et al. explained the high ideality factors by considering that in a MQW-based LED, electrons and holes do not simply diffuse towards the neutral regions, where they eventually recombine (as in a pn junction), but they are injected into the quantum wells where they undergo radiative recombination. In a first-order approximation, i.e. by assuming similar carrier concentrations in the neutral regions, recombination rate is supposed to be maximum at the center of the active layer, i.e. in the region where the concentrations of both electrons and holes are high. Under
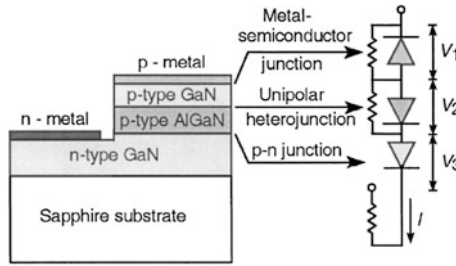
Fig. 8.6 Structure of one of the diodes studied in [23]. Reprinted with permission from Jay M. Shah, Y.-L. Li, Th. Gessmann, and E.F. Schubert, "Experimental analysis and theoretical model for anomalously high ideality factors $n > 2.0$ in AlGaN/GaN p-n junction diodes", J. Appl. Phys. 94, 2627 (2003) . Copyright [2003], American Institute of Physics

this hypothesis, described in [4], carriers must overcome an energy barrier which is half the intrinsic barrier of the junction, before undergoing radiative recombination. The authors of [4] suggest that an ideality factor close to 2 can be obtained by considering the presence of this (halved) barrier into the Shockley equation.

A different model was proposed by Shah et al. in [23]: in this study, the authors demonstrated that high ideality factors can be present even without the presence of a multi-quantum well structure. The devices analyzed in [23] were pn junction diodes with/without AlGaN/GaN heterostructure, with ideality factors in the range between 4 and 6.9. The presence of such high ideality factors was explained by considering that the structure of a diode based on GaN consists in the series connection of several heterojunctions, with non-linear characteristics (see Fig. 8.6 for a schematic representation of a typical diode): (i) the metal-semiconductor junctions of the ohmic contacts, which have the following (non-linear) I-V equation

$$I_{ms} \approx I_{s,ms} e^{\frac{qV_{ms}}{n_{ms}kT}} \tag{8.6}$$

and (ii) the AlGaN/GaN or InGaN/GaN heterojunctions, for which the following equation holds

$$I_h \approx I_{s,h} e^{\frac{qV_h}{n_h kT}} \tag{8.7}$$

In (8.6) and (8.7) $I_{s,ms}$ and $I_{s,h}$ (n_{ms} and n_h) represent the saturation current (ideality factor) of a generic metal-semiconductor junction and of a heterostructure, k is the Boltzmann constant, q is the electron charge, and T is the temperature.

These heterojunctions are placed in series to the main pn junction of the diode, with equation

$$I_{pn} \approx I_{s,pn} e^{\frac{qV_{pn}}{n_{pn}kT}} \tag{8.8}$$

According to [23], the current-voltage relation of a diode results from the series connection of the different heterojunctions, and can be expressed as

$$\ln(I) = \frac{q/kT}{\sum_i n_i} V + \frac{\sum_i n_i \ln(I_{si})}{\sum_i n_i} \tag{8.9}$$

Where the index i is used to take into account that several heterojunctions are contributing to the $I–V$ relation, according to equations (8.6)–(8.8).

Equation (8.9) clearly indicates that the ideality factor of a GaN-based diode is the sum of the ideality factors of each of the heterojunctions which constitute the diode ($n = \sum_i n_i$, see [23] for details). This result explains the high ideality factors recorded in GaN-based diodes and LEDs.

This interesting theory is partly in contrast with more recent data published on AlGaN-based quantum-well LEDs: in [17], the $I–V$ curves of ultraviolet LEDs with identical structures, but different EBL (electron blocking layer) thicknesses, were compared. Results indicated that the ideality factor strongly depends on the thickness of the EBL: LEDs without EBL showed an ideality factor of 7.2, while LEDs with an EBL thickness of 4 (10) nm were found to have ideality factor of 6.9 (5.5). This result indicates that the overall ideality factor cannot be directly determined as the sum of the contributions of the several heterojunctions which constitute an LED device. On the other hand, the authors of [17] indicate that the ideality factor of a LED can be strongly temperature dependent, while the slope of the semi-logarithmic $I–V$ curves does not depend on temperature, thus suggesting that tunneling dominates current conduction below the turn-on voltage of the diodes.

Results reported in the literature therefore suggest that the large ideality factors reported for LEDs cannot be explained by a simple, first-order, model: to accurately explain the $I–V$ characteristics of the LEDs, and the high values of the ideality factor, several factors must be considered, including the existence of significant tunneling components, the presence of several non-linear junctions, and the existence of efficient recombination paths within the active layer.

8.3 Current Conduction in Reverse-Bias

LEDs are not ideal blocking devices, and for this reason they can have a significant current conduction even when they are biased with a negative voltage. The existence of reverse current components has been analyzed by several authors [1, 2, 4, 5, 24–26]. In Fig. 8.7 we report typical current-voltage characteristics measured, at different temperature levels, on a GaN-based LED: data are plotted in log-log scale. As can be noticed, current has a power-law dependence on voltage ($I \sim V^n$), where the exponent n ranges between 5 and 6 [26, 27]. Other reports [2] indicated that reverse-current can have an exponential dependence on voltage, according to $I = I_0 e^{\frac{qV}{E}}$, where E is a characteristic energy parameter, ranging between 1 and 2. Previous reports [25] indicated that reverse-current is directly proportional to junction area, and this suggests that leakage is dominated by bulk components, rather than by the conduction through the mesa/surface. Interestingly, the shape of the reverse-bias $I–V$ curve does not significantly change with temperature. This result suggests that, under reverse bias, current conduction is dominated by a field-dependent mechanism, rather than by a thermally-activated one. Results reported in the literature therefore suggest that reverse-current conduction is dominated by tunneling.
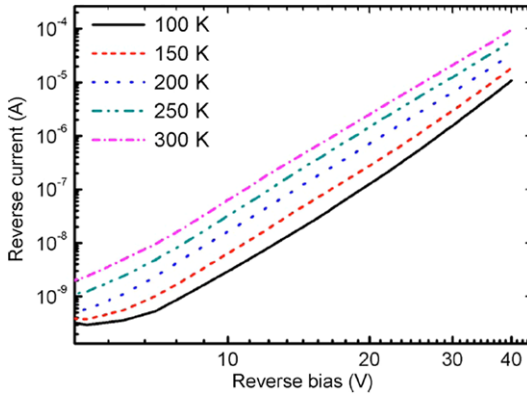
Fig. 8.7 Current-voltage characteristics of a GaN-based Light Emitted Diode measured at different temperature levels in the reverse-bias region. Reprinted with permission from M. Meneghini, N. Trivellin, M. Pavesi, M. Manfredi, U. Zehnder, B. Hahn, G. Meneghesso, and E. Zanoni, "Leakage current and reverse-bias luminescence in InGaN-based light-emitting diodes", Appl. Phys. Lett. 95, 173507 (2009). Copyright [2009], American Institute of Physics [27]

Fig. 8.8 Current-voltage characteristics measured on two LEDs with identical structure grown on templates with different Threading Dislocation Densities
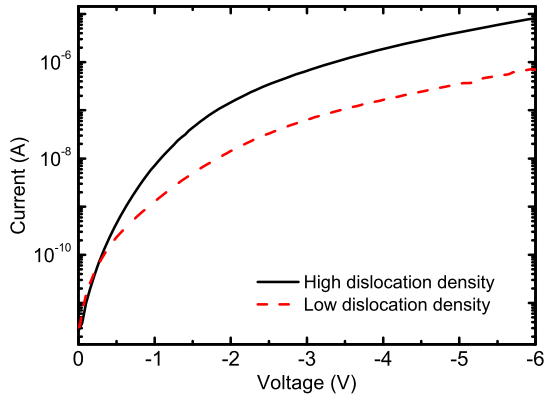
Leakage current strongly depends on the density of structural defects. In Fig. 8.8 we report the current-voltage characteristics measured on two LEDs with identical structure, grown on two templates with different Threading Dislocation Densities (TDD), namely 3×10^8 cm^{-2} and 8×10^9 cm^{-2}: the existence of high reverse-current components is strongly correlated to the density of threading dislocations, thus indicating that TDDs act as efficient channels for reverse-current conduction. Dislocations with screw component are supposed to give the stronger contribution to reverse-current conduction [2]: in many cases, dislocations can be connected to V-shaped defects, which are quite common in InGaN/GaN-based structures, and can act as preferential tunneling paths [28].
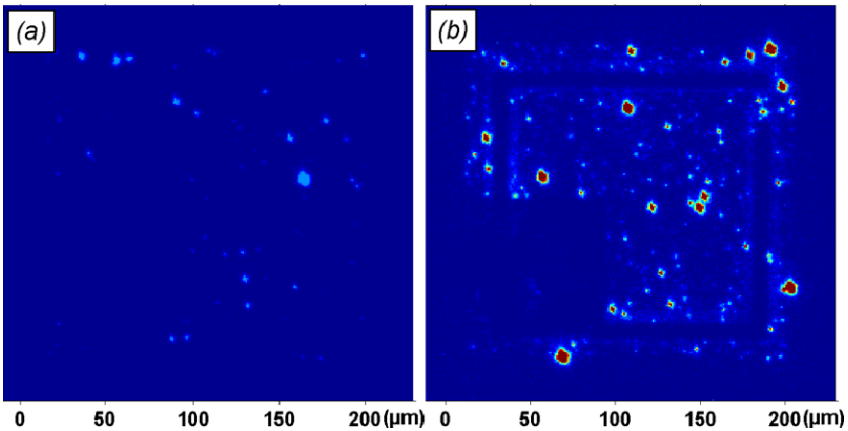
Fig. 8.9 False-color electroluminescence images measured in reverse-bias conditions on two LEDs with (**a**) low and (**b**) high Threading Dislocation Densities. Measurements were taken at the same voltage level of −11 V

A further confirmation on the role of structural defects in governing reverse-bias current conduction was obtained by means of electroluminescence measurements. As described in [5, 27, 80], LEDs can emit a weak electroluminescence signal even when they are under reverse-bias conditions. The intensity of reverse-bias luminescence (RBL) is directly proportional to the injected reverse-current level [27]: RBL occurs at localized spots (see the false color image in Fig. 8.9), which correspond to the presence of preferential paths for leakage current conduction. The density of luminescence dots recorded by EL measurements is strongly correlated to the density of threading dislocations (Fig. 8.9), and to the level of the reverse-current. The origin of reverse-bias luminescence can be understood through EL spectral investigation. In Fig. 8.10 we report the comparison between the EL spectra measured under forward and reverse-bias conditions on a green LED: as can be noticed, reverse-bias luminescence has a broad spectrum, which is slightly blue-shifted with respect to the EL peak recorded under forward-bias. Results reported in the literature suggest that RBL originates from the recombination of electrons and holes injected, by tunneling, to the quantum wells. The slight blue shift measured between forward and reverse-bias luminescence can be explained by considering that, under reverse-bias, the internal fields are partly compensated [29], and this leads to a change in the peak emission wavelength. Another, concurrent, mechanism can be the following: the localized emission paths, which are responsible for reverse-bias current conduction and luminescence, correspond to the presence of structural defects, such as V-pits. In correspondence of V-pits, the thickness of the quantum wells can be significantly smaller than in non-defective regions [30]. This leads to an higher tunneling probability (and this may explain the fact that reverse-current preferentially flows in proximity of defects), and to a higher transition energy [27].
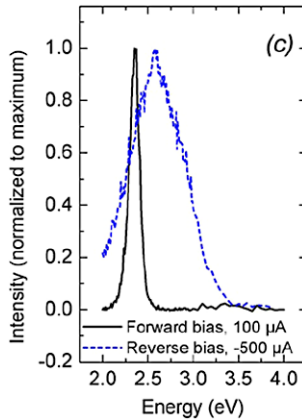
Fig. 8.10 Electroluminescence spectra taken in reverse and forward-bias conditions on a green LED. Reprinted with permission from M. Meneghini, N. Trivellin, M. Pavesi, M. Manfredi, U. Zehnder, B. Hahn, G. Meneghesso, and E. Zanoni, "Leakage current and reverse-bias luminescence in InGaN-based light-emitting diodes", Appl. Phys. Lett. 95, 173507 (2009). Copyright [2009], American Institute of Physics [27]

8.4 Degradation of LEDs

Over the last few years, important efforts have been done in order to improve the performance and the reliability of GaN-based LEDs. As a result, it is now possible to fabricate LEDs with a luminous flux of several hundreds or thousands of lumen. Thanks to these achievements, LEDs are expected to replace conventional light sources in the next future. To be used in the lighting field, LEDs must have a very high luminous output (in the range 100–1000 lm): as a consequence, devices can be operated at current densities around 100 A/cm^2, junction temperatures greater than 85 °C, power densities around 2–300 W/cm^2. Operation at high current densities or high temperature levels may determine an acceleration of the degradation processes: the reliability of LEDs must be therefore carefully optimized, in order to guarantee that the LEDs, mounted in their final application, are capable of the long-lifetime (typically 30000–50000 h) which is expected from the LED technology. This requires an accurate knowledge of the physical mechanisms responsible for the degradation of the LEDs.

A white Light-Emitting Diode is a complex system, which is typically constituted by several elements: the core of a white LED structure is a blue-emitting semiconductor chip (see Fig. 8.11), based on gallium nitride. In most of the cases, this chip has a dimension of approximately 1 mm^2, a typical operating current of 700 mA, and a voltage around 3 V. The conversion of blue light into white light is obtained by the use of a phosphor layer, which can be deposited directly on the chip (chip-level conversion, CLC), or placed some centimeters away from the light-emitting device (remote phosphor approach), in order to reduce the effects related to heating. The LED-phosphor system is then integrated within a package, which must maximize the light-extraction efficiency, and guarantee a good thermal dissipation. A lens is

Fig. 8.11 SEM micrograph of a blue LED chip mounted in its package. To take this picture, the lens was removed from the top of the package. This image is reproduced, with permission, from G. Meneghesso, M. Meneghini, E. Zanoni, "Recent results on the degradation of white LEDs for lighting", J. Phys. D: Appl. Phys. 43, 35, 354007 (2010), IOP Publishing, doi:10.1088/0022-3727/43/35/354007
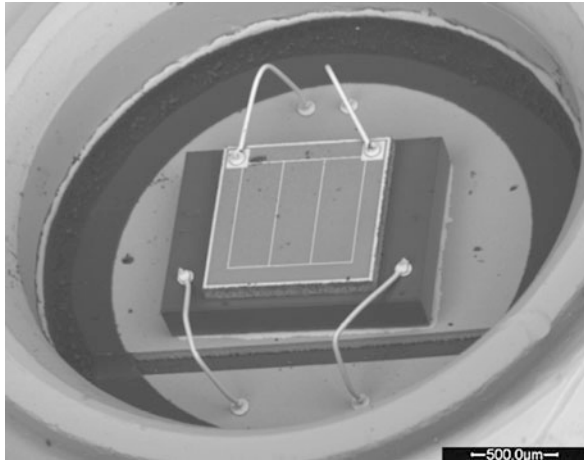
placed over the chip, in order to shape the output beam, and to improve the extraction efficiency. Bonding wires are used to bring current to the device, from metallic pads located on the package. All the components of an LED structure can degrade, when the device is operated at high current and/or temperature levels: the blue semiconductor chip, based on a InGaN/GaN Multi-Quantum Well (MQW) structure, may show an efficiency decrease due to the generation of lattice defects within the active layer, or to the degradation of the ohmic contact and semiconductor material at the p-side of the diode. On the other hand, the package/phosphor system may degrade due to the high temperatures reached by the device during operation, and this may significantly modify the output power emitted by the devices, and induce an important worsening in the chromatic properties (CCT, CRI) of the LEDs. Finally, bonding wires, metallization lines, and the whole LED junction can fail as a consequence of Electrostatic Discharges (ESD), or hot-plugging of the LEDs. Due to the intrinsic complexity of an LED system, studying how each individual component of a LED structure degrades over time can be complicated: to this aim, it is important to design appropriate test strategies, capable of analyzing specific degradation mechanisms, and of separately evaluating the role of the different driving forces (current, temperature) in determining the measured degradation process. In the following, we describe the main degradation mechanisms of GaN-based LEDs, and the experimental strategies that have been adopted in order to accurately study their origin.

8.5 Degradation of the Blue Semiconductor Chip Activated by Current

Several authors (see for example [26, 31–48, 71–74, 77]) have demonstrated that during constant current stress, LEDs can show a remarkable degradation of their

Fig. 8.12 Optical Power degradation measured on a GaN-based LED submitted to stress at 60 A/cm^2, with a case temperature of 50 °C
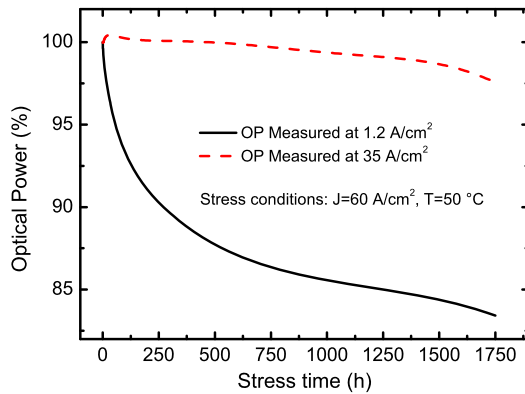
Fig. 8.13 Degradation of the efficiency vs current characteristics of a GaN based LED submitted to stress at 60 A/cm^2, with a case temperature of 50 °C
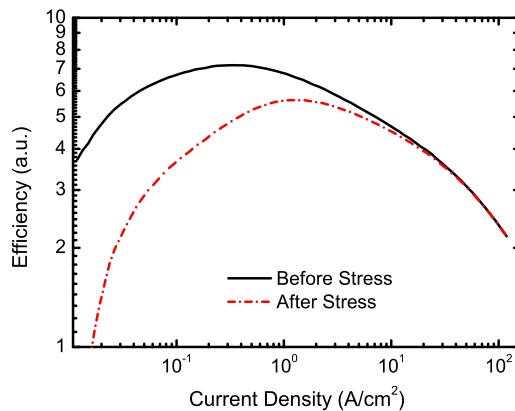
optical and electrical characteristics. In Fig. 8.12 we report typical degradation kinetics measured on a GaN-based LED chip submitted to constant current stress, with a current density of 60 A/cm^2, and a case temperature of 50 °C. During stress time, junction temperature (estimated by the method described in [82]) was smaller than 100 °C, in order to be able to evaluate stress mechanisms activated by the flow of current, without any strong thermally-induced degradation. For the execution of this stress experiment, device was mounted on a metallic TO package, without any phosphors and lens: in this way, it was possible to analyze the failure mechanisms related to the degradation of the blue semiconductor chip, without involving other elements of a conventional LED structure. As shown in Fig. 8.12, stress can induce a significant decrease in the optical power emitted by GaN-based LEDs. The decrease in optical power is more prominent at low measuring current levels, while on the other hand, at high measuring current levels no significant degradation is revealed. This is better explained in Fig. 8.13, which shows the efficiency vs current ($\eta - I$) characteristics measured before/after stress on a GaN-based LED. The $\eta - I$ curves can be ideally divided in two parts: in the high current region, the efficiency is mainly limited by the efficiency droop, while in the low current region, the efficiency of
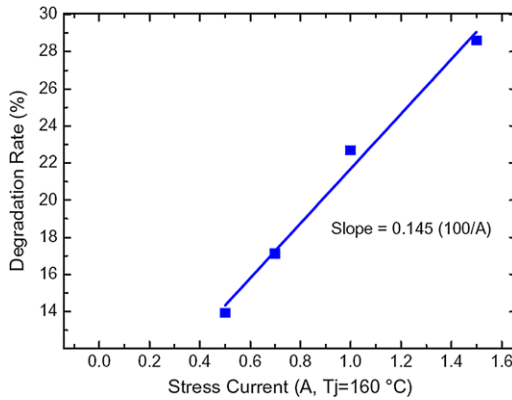
Fig. 8.14 Degradation rate (degradation after 1000 h) for high-power LEDs stressed at different current levels, same junction temperature (160 °C). *Solid line* represents a linear fit of the experimental data. This image is reproduced, with permission, from M. Meneghini, M. dal Lago, N. Trivellin, G. Mura, M. Vanzi, G. Meneghesso, and E. Zanoni, "Chip and package-related degradation of high power white LEDs", Microelectronics Reliability 52, 804–812, 2012

the LEDs is determined by the balance between the radiative and non-radiative recombination rates within the active layer of the devices. The fact that after stress the devices show a remarkable efficiency decrease in the low current region, is a clear signature that stress is inducing an increase in the non radiative recombination rate, due to the generation of lattice defects within the active region [35, 36, 41, 45, 49]. At low measuring current levels, defect-related centers can effectively act as alternative recombination paths, thus limiting device efficiency. On the other hand, at high current densities, defective recombination paths can be saturated by the high number of injected carriers, and device efficiency only depends on radiative recombination processes, and—for extremely high current levels—by the physical mechanisms responsible for droop.

The fact that stress induces an increase in the non-radiative recombination rate is usually explained by considering a defect generation process activated by the flow of electrons through the active layer [34]: even at low current densities, and at moderate temperatures, some of the carriers injected towards the active region may have enough energy to induce a damage in the lattice, thus limiting the performance of the devices. In this view, the degradation rate is strongly dependent on the stress current density, and current acts as a major driving force for the degradation process. Many studies (see for instance [50] and [42]) analyzed the role of current in determining the degradation of MQW-based diodes: results effectively indicated that the degradation rate of LEDs can have a linear dependence on the stress current density (see Fig. 8.14), thus confirming the important role of current as a major driving force for device degradation.

In order to achieve a better description of the physical mechanisms responsible for the degradation of the blue semiconductor chip, current-voltage (I–V), capacitance-voltage (C–V) and Deep-Level Transient Spectroscopy (DLTS) can be

Fig. 8.15 Current-voltage characteristics measured during stress of a LED
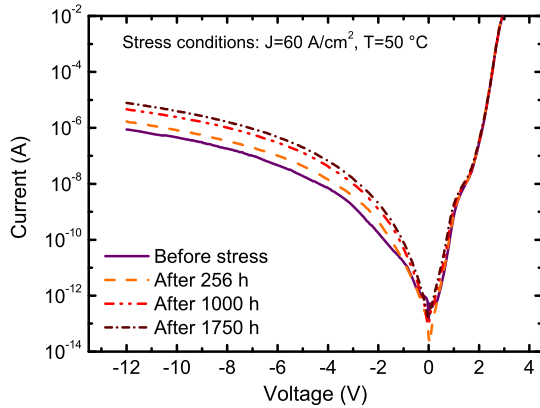
used in combination. Results of electrical ($I-V$) measurements usually indicate that stress induces an increase in the reverse-bias leakage components (Fig. 8.15): this result is consistent with the hypothesis that stress induces the generation of point defects within the active layer of the devices, and suggests that, by suitable analytical techniques, it is possible to extrapolate the properties of the defects involved in the degradation process. Capacitance-Voltage measurements represent a powerful tool for the analysis of the modification of the heterostructure of GaN-based LEDs arising as a consequence of stress. Stress may induce severe modifications in the $C-V$ curves of GaN-based LEDs (see typical results in Fig. 8.16), namely an increase in device capacitance, which occurs in a specific voltage region (here between -6 V and 2 V). In the hypothesis that the p-side of the diodes is more heavily doped with respect to the n-side, by using the well known formulas

$$N(W) = \frac{2}{q\varepsilon} \frac{1}{d(\frac{1}{C^2})/dV} \tag{8.10}$$

$$W = \frac{\varepsilon}{C} \tag{8.11}$$

it is possible to extrapolate the Apparent Charge Distribution (ACD) profile of the analyzed devices. In (8.9) and (8.11), $N(W)$ represents the charge profile as a function of the depth W, while q and ε are the electron charge and the permittivity of the semiconductor material.

Results of this analysis (see an example in Fig. 8.17) provide information on the distribution of free charge within an LED structure: an ACD profile of a MQW structure usually has one or more peaks, corresponding to regions where charge is accumulated (the quantum wells), one or more valleys, representing depleted regions (usually the barriers), and a region where the charge concentration is almost constant (the n-type material). From the comparison of the diagrams in Figs. 8.16 and 8.17, it can be easily understood that the semiconductor region interested by the increase in charge concentration detected after stress, is the region close to the active layer (here between -6 V and 2 V). Results therefore suggest that constant current stress may induce a significant increase in the charge close to the quantum
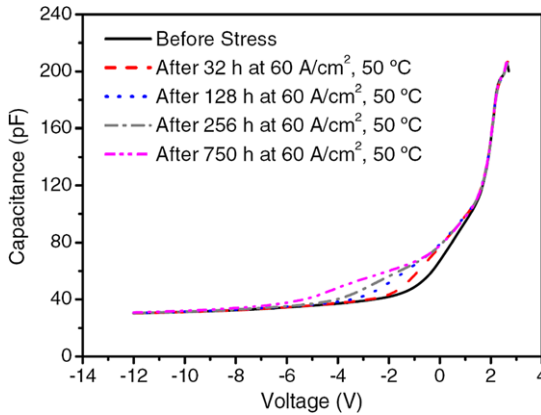
Fig. 8.16 Capacitance-voltage characteristics measured on a GaN based LED submitted to stress at 60 A/cm^2, with a case temperature of 50 °C. This image is reproduced, with permission, from G. Meneghesso, M. Meneghini, E. Zanoni, "Recent results on the degradation of white LEDs for lighting", J. Phys. D: Appl. Phys. 43, 35, 354007 (2010), IOP Publishing, doi:10.1088/0022-3727/43/35/354007

Fig. 8.17 Apparent charge distribution profile measured on a GaN-based LED starting from $C-V$ measurements
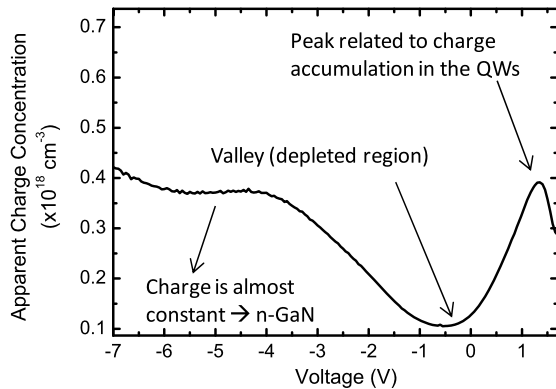
well region, due to the generation of electrically-active defects. This strong correlation between the increase in defect-related charge, and the optical degradation is confirmed by Fig. 8.18, which shows that the two parameters (integrated junction charge, and optical power) vary with similar kinetics. A powerful technique for the identification of the defects which are generated as a consequence of electrical stress within an LED structure is Deep Level Transient Spectroscopy (DLTS) [51]: this technique is based on the analysis of the capacitance transients generated by a sudden change in junction voltage from a filling voltage level (used to charge trap levels in a specific semiconductor region) to a reverse voltage level, at which the capacitance variation is recorded. DLTS provides—as a result—the extrapolation of the activation energy and of the capture cross section of the defective levels that are present in a semiconductor-based device. DLTS investigation is usually carried out

Fig. 8.18 Variation of the integral of the $C-V$ curves and of the optical power measured on a GaN based LED submitted to stress at 60 A/cm^2, with a case temperature of 50 °C
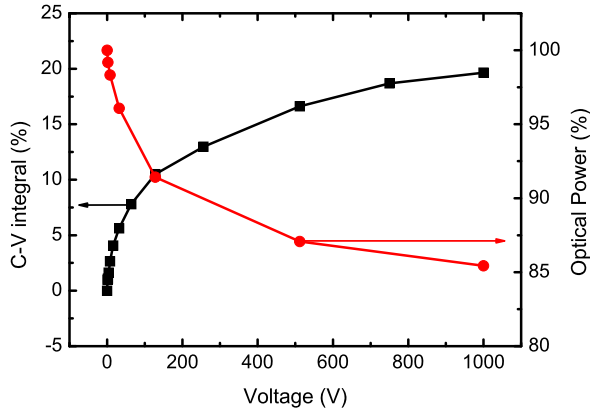
by selecting filling and reverse-voltage pulses that allow to explore a specific semiconductor region: for the analysis of MQW structures, voltage levels are chosen on the basis of 2D simulation. A typical example is reported in Fig. 8.19(a): 2D simulation reveals that, in order to explore the active region of these specific devices, voltage must be swept between -1 V and $+2$ V during the DLTS measurement. By using this voltage range, DLTS analysis reveals, as a consequence of stress, the increase in the peak amplitude (see Fig. 8.19(b)), which corresponds to a majority trap level, located within the active region of the devices (see [52] and [36] for details on this kind of investigation). To achieve more detailed information on the origin of this deep level, its Arrhenius plot must be compared with the Arrhenius plot of other deep levels reported in the literature for GaN: typical results of this analysis are summarized in Fig. 8.20, which shows the Arrhenius plots of several deep levels identified by DLTS in GaN-based materials and devices, together with the Arrhe-
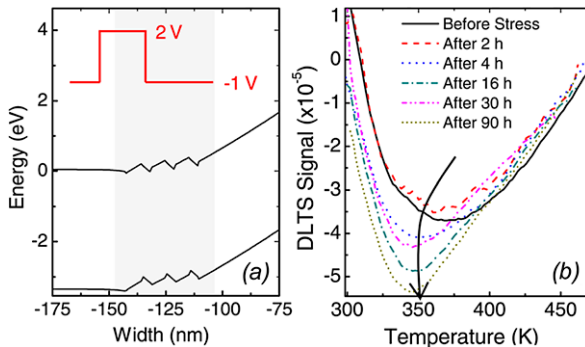
Fig. 8.19 (**a**) Simulated band diagram of a MQW LED structure; (**b**) change in the DLTS spectrum of a LED device submitted to stress at high current densities. Reprinted with permission from M. Meneghini, C. de Santi, N. Trivellin, K. Orita, S. Takigawa, T. Tanaka, D. Ueda, G. Meneghesso, and E. Zanoni, "Investigation of the deep level involved in InGaN laser degradation by deep level transient spectroscopy", Appl. Phys. Lett. 99,093506, 2011. Copyright [2011], American Institute of Physics [52]
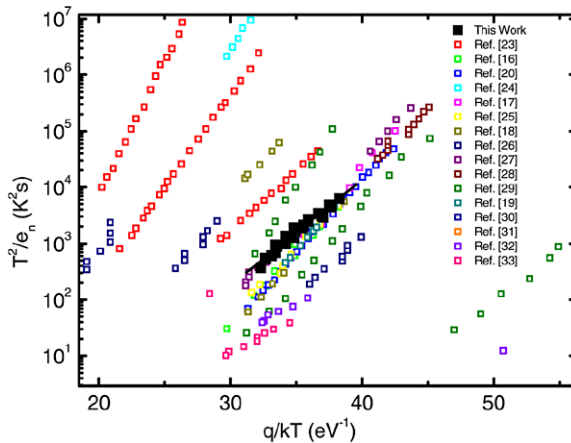
Fig. 8.20 Arrhenius plot of the deep level in Fig. 8.19(b), plotted together with several Arrhenius plots that have been reported in the literature for deep levels in GaN-based materials. The numbers of the references in the legend correspond to the references of [52]. Reprinted with permission from M. Meneghini, C. de Santi, N. Trivellin, K. Orita, S. Takigawa, T. Tanaka, D. Ueda, G. Meneghesso, and E. Zanoni, "Investigation of the deep level involved in InGaN laser degradation by deep level transient spectroscopy", Appl. Phys. Lett. 99, 093506, 2011. Copyright [2011], American Institute of Physics [52]

nius plot of the deep level in Fig. 8.19(b). By an accurate analysis of the deep levels reported in the literature, several types of point defects have been proposed as responsible for the degradation of GaN-based LEDs [52]: (i) nitrogen-antisite defects, whose properties have been described in [52–54], that have very high formation energy [55], and therefore are supposed not to be easily generated during stress time; (ii) defects related to foreign impurities, such as carbon, whose properties have been described in [56]; (iii) point defects related to gallium vacancies, such as those described in [57], that have a high probability of being generated near the dislocations. Further investigation are still in progress, with the aim of understanding if and how these deep level can influence the degradation kinetics of high power LED chips.

8.6 Degradation of the Blue Semiconductor Chip Activated by Temperature

In the previous section, we have described how the electrical and optical properties of a blue LED chip can significantly degrade as a consequence of the flow of current through the active region of the devices: most of the results described above, were obtained by stressing the devices at constant current, with a moderate junction temperature, smaller than 100 °C. In this section we demonstrate that the blue LED chip can significantly degrade also when it is submitted to stress at high temperature levels, usually greater than 140 °C.
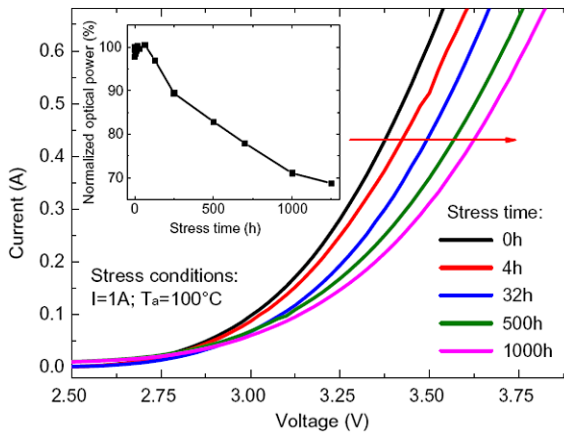
Fig. 8.21 Modifications in the electrical characteristics of a high-power LED submitted to stress at 1 A, with an ambient temperature of 100 °C (junction temperature equal to 160 °C). *Inset*: degradation of the optical power measured (At 700 mA) on the same LED device. Reprinted with permission from M. Meneghini, M. Dal Lago, N. Trivellin, G. Mura, M. Vanzi, G. Meneghesso, and E. Zanoni, "Chip and package-related degradation of high power white LEDs", Microelectronics Reliability 52, 804–812, 2012 [60]

In the inset of Fig. 8.21 we report the optical degradation of a high-power LED that was submitted to stress at 1 A, with an ambient temperature $T_a = 100$ °C, corresponding to a junction temperature of 160 °C. As can be noticed, stress induces a significant decrease in the optical power emitted by the device. High temperature degradation may correspond to important modifications in the electrical characteristics of the devices, namely in an increase in the series resistance, and/or in the increase in the operating voltage (see Fig. 8.21). An accurate analysis reveals that kinetics of the increase in the operating voltage and of the optical power degradation can be strongly correlated (Fig. 8.22). The high temperature degradation of blue LED chips can be explained according to the following interpretation [58, 59, 78, 81]: at high temperature levels the resistivity of the ohmic contact and of the semiconductor at the p-side of the diodes, can significantly increase, due to the interaction between hydrogen, which is normally present within the semiconductor and passivation layer, and the acceptor dopant (Mg). As a consequence, magnesium can be passivated by hydrogen, and this may result in a significant decrease in the effective acceptor concentration and, therefore, in a worsening of the conductivity of the ohmic contact and of the p-type GaN layer. This results in an increase in the operating voltage, and therefore in a remarkable decrease in device efficiency. In extreme cases, as a result of the increase in the resistivity of the p-type material, stressed devices can show an increase current and emission crowding: an exampleis reported in Fig. 8.23, reporting the false-color emission pattern of a blue semiconductor chip, measured after a 20 min stress at 250 °C. As can be noticed, stress induced a significant emission crowding, that can be attributed to the increase in the resistivity of the p-type material. This degradation mechanism occurs only at
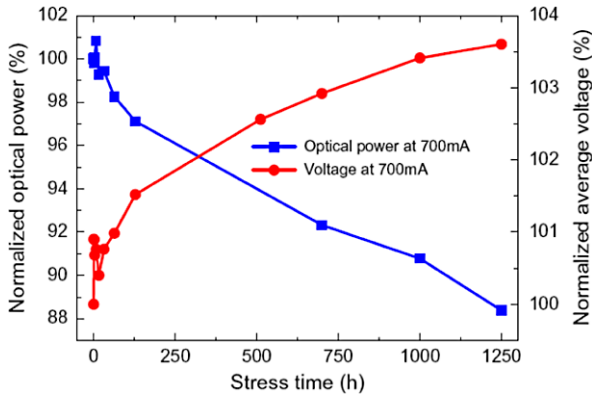
Fig. 8.22 Correlation between the optical degradation and the increase in operating voltage of a high-power LED submitted to stress at 1 A, with an ambient temperature of 100 °C (junction temperature equal to 160 °C). Reprinted with permission from M. Meneghini, M. Dal Lago, N. Trivellin, G. Mura, M. Vanzi, G. Meneghesso, and E. Zanoni, "Chip and package-related degradation of high power white LEDs", Microelectronics Reliability 52, 804–812, 2012 [60]
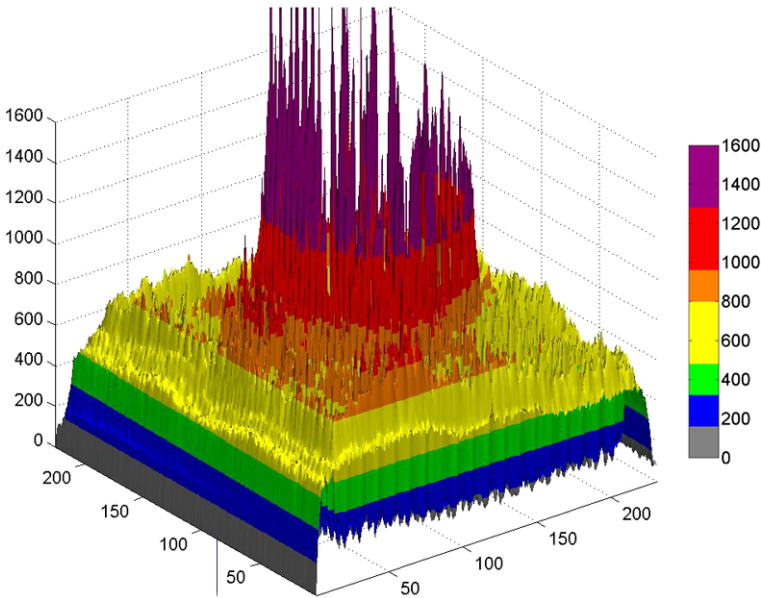
Fig. 8.23 False-color emission pattern measured on a LED submitted to high temperature stress (250 °C, 20 minutes)

relatively high temperature levels: for clarity, in Fig. 8.24 we report the results obtained by aging several high power LEDs at 1 A, with different temperature levels: results clearly show that an increase in the operating voltage is generated only when
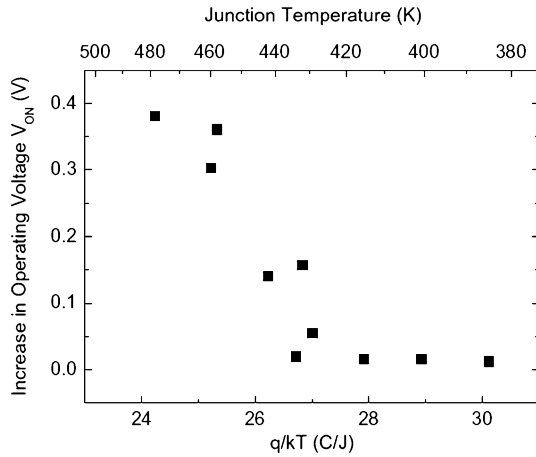
the LEDs are stressed with junction temperatures greater than 140 °C (413 K). Typical activation energies for this degradation process are in the range 1.1–1.3 eV, as demonstrated by recent reports [58, 60].

8.7 Degradation of the Package/Phosphor System

So far, we have discussed the degradation mechanisms of the blue semiconductor chip, activated by temperature and by current. The semiconductor chip is only one of the components of a power LED structure: the package and the phosphor layer play an important role in determining the overall efficiency and chromatic properties of the device. For this reason, it is of fundamental importance to understand how package and phosphors degrade when they are submitted to high temperature levels. Typically, these two components degrade at temperatures greater than 100 °C: the best methodology to study the thermally-activated degradation of high power LEDs is to carry out "purely-thermal" stress tests, in which the devices are exposed to relatively high temperature levels, with no applied current. In this way, all the degradation mechanisms can be ascribed to the high temperatures, without mixing the effects of temperature and current in the analysis.

Typically, exposure to stress temperatures greater than 100 °C may result in a remarkable decrease in the efficiency of the LEDs: in Fig. 8.25 we report typical degradation curves, measured on commercially available LEDs submitted to thermal storage. Most the analysis on package and phosphor degradation are usually carried out on LEDs with a very stable *p*-type material: as a result, the degradation mechanisms described in the previous section do not take place on these LEDs, and the electrical and optical characteristics of the blue chips are stable over stress time (for instance see [60] and references therein). Degradation can therefore be unambiguously attributed to the worsening of the package/phosphor system.

Fig. 8.25 Degradation of the optical power measured on a set of LEDs submitted to purely thermal stress at different temperature levels, without applied bias. Reprinted with permission from M. Meneghini, M. Dal Lago, N. Trivellin, G. Mura, M. Vanzi, G. Meneghesso, and E. Zanoni, "Chip and package-related degradation of high power white LEDs", Microelectronics Reliability 52, 804–812, 2012 [60]
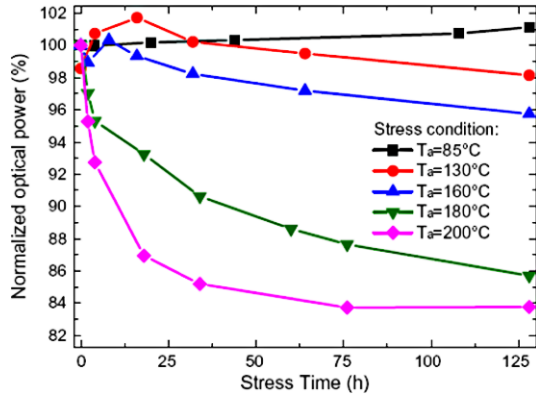
Fig. 8.26 (*Top*) Electroluminescence image taken before and after stress on a high power LED. (*Bottom*) Micrograph of a section of the same LED, showing the carbonization of the package. Reprinted with permission from M. Meneghini, L. Trevisanello, C. Sanna, G. Mura, M. Vanzi, G. Meneghesso, E. Zanoni, "High temperature electro-optical degradation of InGaN/GaN HBLEDs", Microelectronics Reliability 47, 1625–1629, 2007
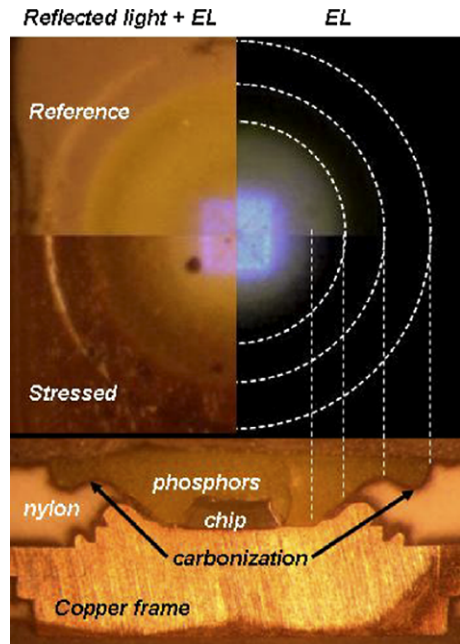
Package degradation can be particularly severe in the case of LEDs with a white, reflective, package. In this case the package has the important role of maximizing the extraction efficiency of the LEDs, by guaranteeing a high reflectivity. As a consequence of the high temperatures reached during stress, or (when devices are operated with a high current levels) due to the short wavelength of blue and white LEDs, the plastic material used for the realization of the package can show a remarkable browning. As an example, in Fig. 8.26 we report a micrograph of the section of a LED package which was submitted to stress at high temperature levels: in this case stress does not affect the quality of the phosphor material [75], but induces a severe darkening of the surface of the package [61], and this can significantly limit its abil-
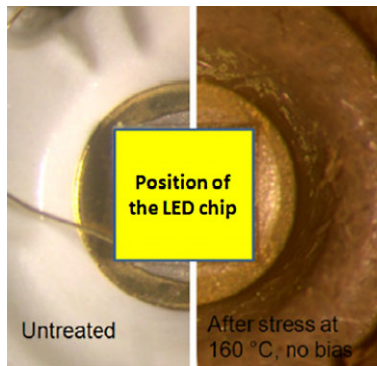
Fig. 8.27 Micrograph of the package of a power LED that was submitted to stress at high temperature. The image was taken after the chemical removal of the lens, and of the phosphors. Reprinted with permission from M. Meneghini, M. Dal Lago, N. Trivellin, G. Mura, M. Vanzi, G. Meneghesso, and E. Zanoni, "Chip and package-related degradation of high power white LEDs", Microelectronics Reliability 52, 804–812, 2012 [60]

ity of reflecting light out of the device. In Fig. 8.27 we show a front picture of the same kind of package, taken after the removal of the lens and of the phosphors. The browning of the surface of the package may determine a modification in the chromatic properties of the LEDs: as shown in Fig. 8.28, the angular distribution of light can be significantly affected by stress, especially for high emission angles. Degradation is also correlated to important modification in the spectral power distribution of the devices, as representatively shown in Fig. 8.29. Similar degradation processes are more prominent in devices with a white plastic package, rather than in LEDs with a ceramic package. However, since plastic package is widely used both for lighting, outdoor applications, backlighting and automotive, it is very important to study and optimize the degradation of LEDs with a plastic reflector, and to carefully optimize the lighting systems in order to avoid exposure to high temperature levels. As described above, the degradation of the package is strongly thermally activated, with activation energies in the range between 0.5 eV and 1.8 eV, depending on the specific analyzed technology (see for instance [60, 62, 79] and references therein).

Another part of a white LED that may be prone to thermally-activated degradation is the phosphor layer, used to convert the blue radiation of the LED chips into white light. In most of the cases, phosphors are dispersed in a matrix, that can become dark as a consequence of high temperature treatment [63]. An example is reported in Fig. 8.30, which clearly shows that, as a consequence of high temperature, high current stress, the power LED shows a remarkable darkening of the phosphor layer. Interestingly, degradation is mostly focused at the center of the LED, since this is the region which is more prone to self-heating. Recently, ceramic phosphor layers have been proposed as an interesting solution for the realization of highly reliable phosphors (see for instance [63, 64] and references therein): ceramic phosphors are good candidates for adoption in all those fields in which LEDs can be exposed
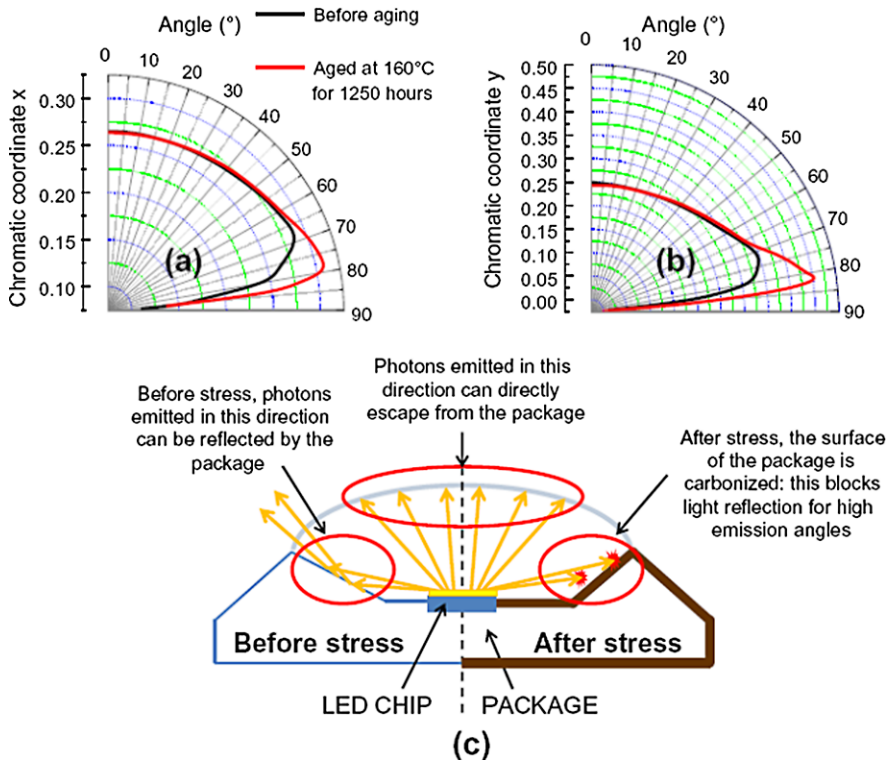
Fig. 8.28 (**a**) and (**b**) Variation of the chromatic coordinates x and y measured after high temperature stress on high power LEDs. (**c**) Schematic picture of the effect of the darkening of the package on the light extraction efficiency. Reprinted with permission from M. Meneghini, M. Dal Lago, N. Trivellin, G. Mura, M. Vanzi, G. Meneghesso, and E. Zanoni, "Chip and package-related degradation of high power white LEDs", Microelectronics Reliability 52, 804–812, 2012 [60]

to high power levels and high temperatures, including automotive headlamps, and indoor and outdoor lighting.

8.8 ESD-Failure of GaN-Based LEDs

Electrostatic Discharges (ESD) may be present during the whole life of a LED-based system. During the fabrication of the wafer, improper handling by a charged operator, or the contact with a charged machine, may result in the ESD failure of the LEDs. On the other hand, when the LEDs are mounted in the final system, they can receive voltage/current spikes through the electrical contacts of the whole luminaire, and this can, if there are specific technological weaknesses, damage the devices. For this reason, several research groups have investigated the origin of the ESD failures, and proposed methods for improving the robustness of the LEDs (see for instance [65–68, 76]).
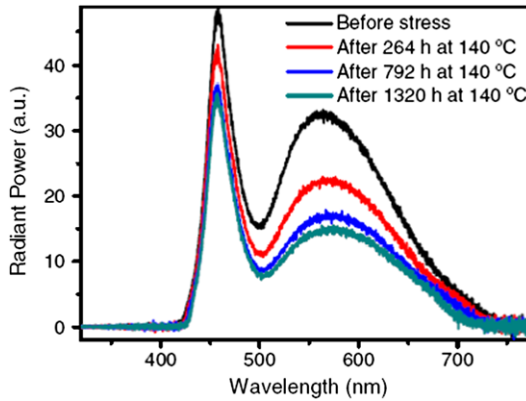
Fig. 8.29 Spectral power distribution measured at different stress time on a power LED submitted to high temperature stress. This image is reproduced, with permission, from G. Meneghesso, M. Meneghini, E. Zanoni, "Recent results on the degradation of white LEDs for lighting", J. Phys. D: Appl. Phys. 43, 35, 354007 (2010), IOP Publishing, doi:10.1088/0022-3727/43/35/354007

Fig. 8.30 Micrograph of a power LED taken before and after stress at 100 A/cm^2, 120 °C. This image is reproduced, with permission, from G. Meneghesso, M. Meneghini, E. Zanoni, "Recent results on the degradation of white LEDs for lighting", J. Phys. D: Appl. Phys. 43, 35, 354007 (2010), IOP Publishing, doi:10.1088/0022-3727/43/35/354007
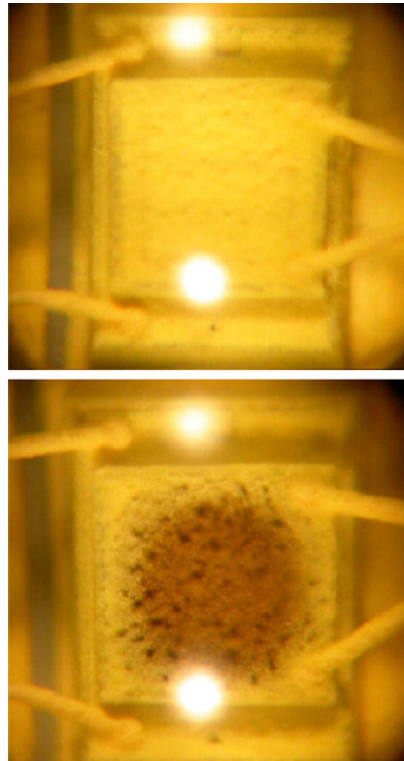
Fig. 8.31 Typical output of a TLP test carried out on a GaN-based LED with an area of $290 \times 290\ \mu m^2$. (**a**) Current-voltage characteristic measured during a TLP test, until the failure is reached. (**b**) Leakage current measured after each of the TLP pulses on the analyzed LED. Failure can be recognized as a sudden increase in device leakage current
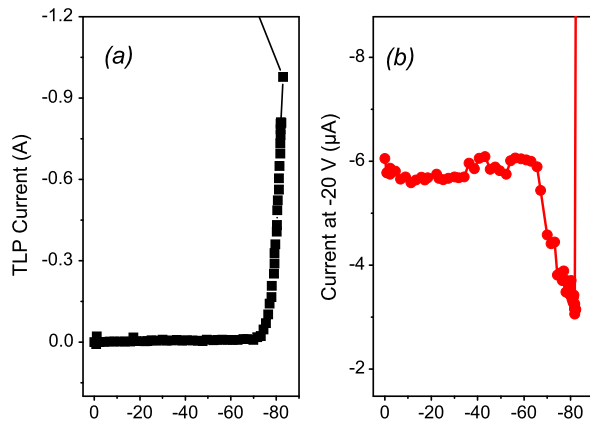
As a consequence of ESD events, LEDs can show either a catastrophic failure, or a latent damage. In the first case, after the ESD event, the impedance of the device shows a sudden decrease and, in most of the cases, devices behave as short circuits. In other cases, it may happen that the ESD event is not completely destructive: after the discharge, devices may still be able to be operated, but they can show an unstable behavior, due to the fact that an unstable leakage path is generated in parallel to the junction. As a consequence, luminous power may dim, or show a variable intensity over time, due to the intrinsic instabilities of the shunt-path generated after the ESD event. A similar behavior is the signature of a latent damage, that can significantly limit the stability of the LEDs.

Several models can be efficiently adopted to evaluate the ESD robustness of LEDs: the most important include the Human Body Model (HBM), the Machine Model (MM), the Charged Device Model (CDM) [69] and the Transmission Line Pulse (TLP) method [70]. Among these methods, the TLP analysis allows to achieve a detailed description of the damages introduced during ESD testing. The operating principle of a TLP system is the following: a transmission line of appropriate length is charged by means of a high-voltage generator, and then, by suitable switches, is discharged on the LED sample, thus obtaining very short pulses (duration around 100 ns). During the discharge, the voltage and the current that flow on the sample are measured by oscilloscope probes. After the electrical discharge, several device parameters—including leakage current, optical power, emission pattern—are measured in order to characterize the effects of a ESD pulse on the properties of the devices. This procedure is repeated for increasing voltage levels, until the failure is reached. The typical output of a ESD analysis is twofold: on one hand, a complete pulsed $I–V$ characterization of the characteristics of the devices up to the failure limit; on the other hand, a description of how some device parameters (for instance the leakage current) vary as a consequence of the application of voltage pulses with increasing amplitude. In Fig. 8.31 we report a representative output of a TLP test carried out on a GaN-based LED with an area of $290 \times 290\ \mu m^2$. This test was carried out by applying pulses with increasing (negative) amplitude to the devices. As can be noticed, during a TLP test the device could withstand a negative current of
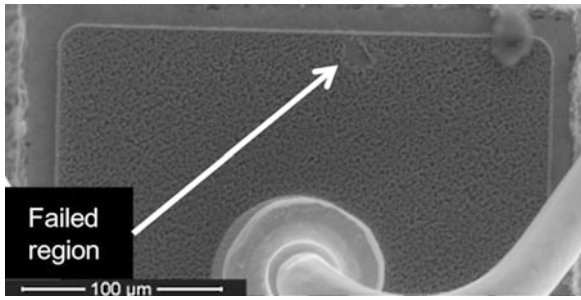
Fig. 8.32 SEM Micrograph taken on a LED after the execution of a ESD failure test. This image is reproduced, with permission, from G. Meneghesso, M. Meneghini, E. Zanoni, "Recent results on the degradation of white LEDs for lighting", J. Phys. D: Appl. Phys. 43, 35, 354007 (2010), IOP Publishing, doi:10.1088/0022-3727/43/35/354007

Fig. 8.33 Results of ESD test carried out under (**a**) forward and (**b**) reverse bias on LED chips. This image is reproduced, with permission, from G. Meneghesso, M. Meneghini, E. Zanoni, "Recent results on the degradation of white LEDs for lighting", J. Phys. D: Appl. Phys. 43, 35, 354007 (2010), IOP Publishing, doi:10.1088/0022-3727/43/35/354007
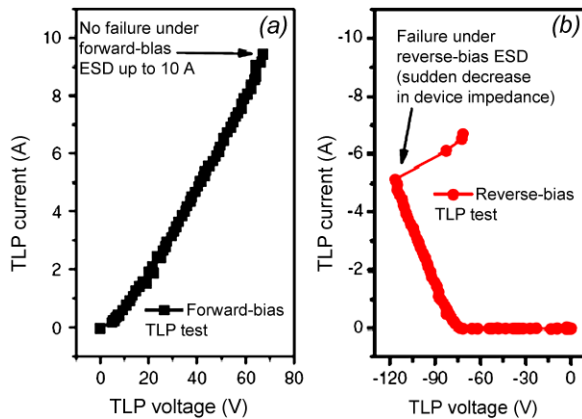
1 A (corresponding to a current density around -1.2 kA/cm^2) before reaching the failure, which can be clearly identified as a sudden decrease in device impedance. In most of the cases, device failure can be ascribed to the generation of a short-circuit path, that can be identified by SEM investigation (see an example in Fig. 8.32). The failure under reverse-bias ESD events isrelated to the fact that, under reverse-bias, conduction mostly occurs through preferential leaky paths, related to the presence of structural defects, such as dislocations and/or V-shaped defects (see Fig. 8.9). These defects have a nanometer-size, and for this reason, they can be crossed by very high current densities during the ESD event [62, 65]. When a critical current/energy level is reached on one of these defects, a localized failure may occur, and this determines the sudden shortening of the device. This kind of failure only occurs under reverse-bias ESD tests: when TLP robustness tests are carried out in forward bias, since the whole device are contributes to current conduction, devices can reach current densities in excess of 10 kA/cm^2 without showing any failure (see an example in Fig. 8.33(a)). Results presented in the literature indicate that, while the robustness to forward-bias ESD pulses does not represent a serious issue for state-of-the-art
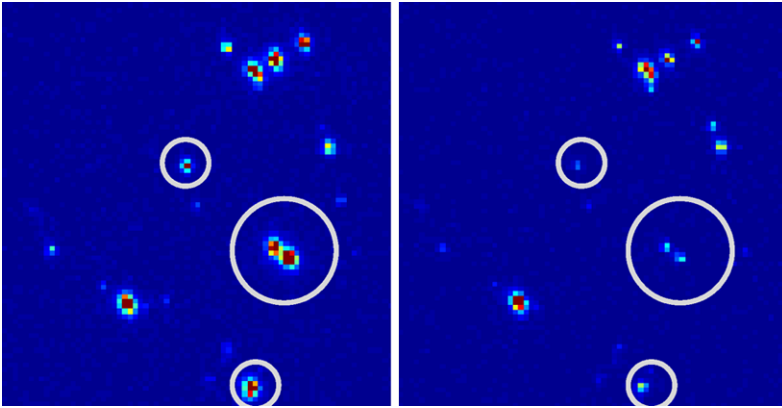
Fig. 8.34 Spatially-resolved electroluminescence profiles measured, under reverse-bias conditions, on a LED before and after a non-destructive ESD-like event

LEDs, the robustness of current LED technologies to reverse voltage pulses still has to be optimized, through a careful improvement of the epitaxial quality of the devices aiming at a significant reduction of the defects responsible for reverse current conduction.

Besides catastrophic (or hard) failures, GaN-based LEDs can show also a soft (or non-destructive) degradation when they are submitted to ESD-like events with negative voltage [68]. As shown in Fig. 8.31, during the execution of a TLP test, pulses with increasing amplitude are applied to the LED. Beyond a certain voltage (here for $V < -70$ V), the reverse-current of the ESD pulses significantly increases. As a result, the leakage current measured after each of the ESD-like pulses shows a remarkable decrease (see Fig. 8.31(b)). This effect has been attributed to the partial annihilation of the defective paths responsible for reverse-current conduction, that are related to the presence of structural defects (see Fig. 8.9 and its explanation). This mechanism is supposed to be related to the high energy released, during a negative-bias ESD pulse, in correspondence of the nanometer-size regions that carry most of the reverse-current that flows through the device. The annihilation of the reverse-leakage paths, can be detected by spatially-resolved electroluminescence measurements, carried out under reverse-bias [68]. An example is reported in Fig. 8.34, which shows the false-color electroluminescence profile measured before and after a series of non-destructive ESD-like events that have determined a remarkable decrease in leakage current. As can be noticed, after the execution of the non-destructive ESD tests, some of the luminescent dots detected before stress almost disappear. Since these dots indicate the presence of leaky regions, related to defects, results strongly support the hypothesis that ESD-like events induce the partial annihilation of the preferential paths responsible for leakage current conduction. This effect, which is still under investigation, indicates that ESD events may significantly affect the characteristics of GaN-based LEDs, even when they are not destructive. To improve the stability of the devices it is therefore necessary to minimize the

density of epitaxial defects, through the improvement of the growth conditions, the adoption of optimized interlayers, and the selection of suitable substrates.

8.9   Conclusions

Thanks to their high performance and potential reliability, LEDs based on gallium nitride are expected to widely penetrate the lighting market in the next years. However, the efficiency and the lifetime of these devices can be significantly limited by a number of factors, that have been extensively studied by several research groups. Within this chapter, we have discussed the main processes that limit the reliability of LEDs submitted to short and long-term stress experiments, by critically comparing the results with the data present in the literature.

Results described within this chapter provide information on the most common degradation modes and mechanisms, and on the role of the main driving forces (current and temperature) in determining device degradation. Furthermore, we have described specific methodologies, that can be adopted to study the individual degradation mechanisms that limit the lifetime of LEDs. The experimental data discussed within this chapter indicate that the robustness of the different components of an LED structure must be carefully optimized, in order to fabricate devices with intrinsically high reliability. On the other hand, since degradation mechanisms and kinetics are significantly dependent by the operating conditions, the designers of LED-based systems must carefully optimize the thermal management, the driving conditions, and the lamp layout in order to minimize the risk of early degradation.

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Chapter 9
Light Extraction Efficiency Part A. Ray Tracing for Light Extraction Efficiency (LEE) Modeling in Nitride LEDs

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Abstract We describe simulations of the light extraction efficiency (LEE) as a function of the major materials parameters and geometries in the three main LED structures used today, namely nitride LEDs on GaN substrates, on patterned sapphire substrates and flip chip nitride LEDs. We use ray tracing simulations of LEDs to show how the various extraction schemes operate. The simulations of this chapter show that both surface roughening and patterned sapphire substrates (PSS) lead to high efficiencies, although based on somewhat different mechanisms. The results appear here for the same device parameters (most of them with conservative values), which allow meaningful comparisons. Some industry results may be higher, either due to better materials quality (lower absorption in particular for ITO and metals) or more aggressive designs (smaller lossy contact areas). For LEDs on GaN substrates, the LEE is determined by substrate absorption. A desirable feature is the ability to scale the extraction efficiency for large chips in order to reduce costs. In almost all cases, the use of large surface LEDs is detrimental due to the long ray paths to reach the sidewalls except for those extraction schemes where light is continuously extracted or when substrates have very low absorption.

9.1 Introduction

The development and commercialization of high efficiency light emitters is a major technical challenge that will have a huge impact on efficient energy use, as worldwide 15–20 % of the electricity consumption is due to lighting. Today, lighting is performed mostly with incandescent light bulbs with efficiencies below 5 %, com-

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pact fluorescent lamps with efficiencies in the 20 %+ range, or fluorescent tubes with efficiencies in the 30 %+ range.

The direct conversion of electricity into light in semiconductor light emitting diodes (LEDs) could, in principle, occur with 100 % efficiency. The total efficiency of electricity to light conversion is given by the product of three terms:

$$Total\ Efficiency = IQE \times LEE \times \eta_{elec} \qquad (9.1)$$

where IQE is the internal quantum efficiency, describing the conversion of electrical current into photons inside the material, LEE is the light extraction efficiency of photons generated in the active region that are emitted to free space, and η_{elec} is the electrical efficiency, mainly related to ohmic losses in the electrical contacts and to the carrier injection efficiency into the active layers for radiative recombination, η_{inj}.

This chapter focuses on LEE, which is adversely impacted by the high indices of refraction of semiconductors ($n > 2$). The index mismatch between semiconductor and air leads to a large fraction of light being trapped within the material by total internal reflection (TIR), which occurs when light rays impinge on the semiconductor-air interface at an incident angle larger than the critical angle for refraction. The problem of total internal reflection is often addressed using chip shaping or textured interfaces, which cause light rays to impinge on the semiconductor-air interface at different angles as they make multiple escape attempts. Chip shaping and surface texturing features in LEDs have relatively large dimensions compared to the wavelength of light in the device, such that they can be described by geometrical or ray optics.

High light extraction and greater beam control is also achieved in LEDs through the use of optical mode quantization using microcavities of varied photonic dimensionalities [1, 2] and photonic crystals [7, 20, 25, 29]. Such solutions are still at the research level today and require more complex analysis incorporating wave optical physics. We mainly discuss here simulations of structures based on ray optics. This chapter aims to clarify the origin of improvements in LEE, identify the main loss mechanisms in given structures, and provide a method for designing LEDs with high LEE. We provide, in particular, simulations of the variation in LEE as a function of the major materials parameters and geometries associated with the three main structures used today, namely nitride LEDs on GaN substrates, on patterned sapphire substrates and flip chip nitride LEDs.

9.2 Background on Ray Optics

9.2.1 Snell-Descartes Law

The central issue in light extraction comes from the nature of light interaction at the surface between two dielectric materials. When a propagating light ray hits an interface between two different materials, part of it will be reflected, and part of it

Fig. 9.1 Schematic of the interaction of a light ray at a dielectric interface, displaying the critical angle θ_c beyond which rays are fully reflected by total internal reflection
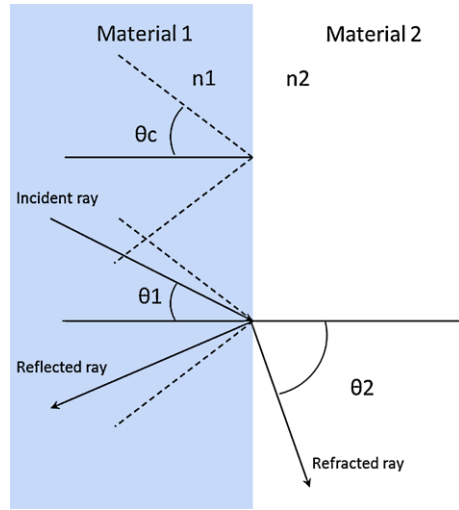
will be transmitted (or equivalently, refracted) into the new material. The angle of reflection will be the same as the angle of incidence, while the angle of refraction follows a more complicated law named the Snell-Descartes law, Eq. (9.2), where n_1 and n_2 are the refractive indices of the different materials and θ_1 and θ_2 are the angles of incidence and refraction, as shown in Fig. 9.1.

$$n_1 \sin(\theta_1) = n_2 \sin(\theta_2) \tag{9.2}$$

As shown in Eq. (9.2), if the refractive index n_1 of the original material is higher than that of the second material n_2, there can be some incidence angles that lead to unphysical solutions for the refracted angle when the term $\frac{n_1}{n_2}\sin(\theta_1)$ is greater than 1. This leads to an effect referred to as total internal reflection (TIR): beyond a certain critical angle, defining an "extraction cone" (also called a light cone), the light is entirely reflected at the interface, and no light is transmitted. Only light situated within the "extraction cone" may then be transmitted (Fig. 9.2).

9.2.2 Fresnel Coefficients

While the Snell-Descartes law describes the angles at which the light is refracted, it does not deal with the intensity of light in the reflected and refracted portions. This is described by the Fresnel coefficients and depends on the polarization of the light. For s-polarized incident light (the electric field is parallel to the interface) the reflection coefficient (for power) is:

$$R_s = \left| \frac{n_1 \cos(\theta_1) - n_2 \cos(\theta_2)}{n_1 \cos(\theta_1) + n_2 \cos(\theta_2)} \right|^2 \tag{9.3}$$
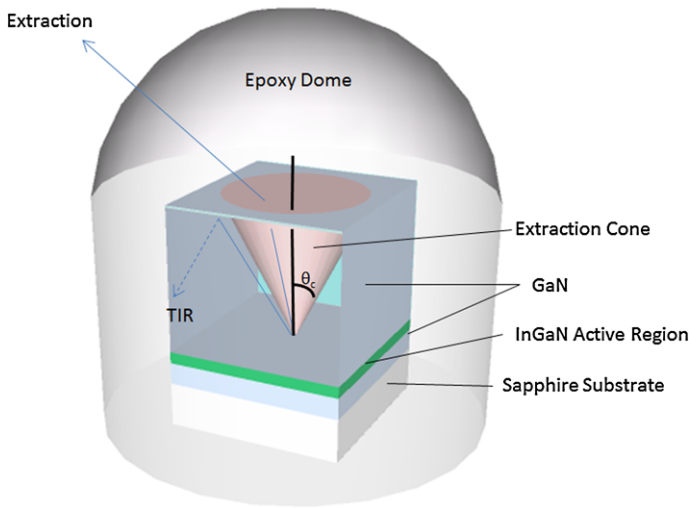
Fig. 9.2 Light emitted from the active region may be extracted via the epoxy extraction cone or internally reflected in the device. Most light that reaches the edge of the epoxy/air interface will be extracted due to the dome shape of the epoxy

For p polarized incident light (the magnetic field is parallel to the interface), it is:

$$R_p = \left| \frac{n_1 \cos(\theta_2) - n_2 \cos(\theta_1)}{n_1 \cos(\theta_2) + n_2 \cos(\theta_1)} \right|^2 \tag{9.4}$$

Therefore, the total reflection including both polarizations is $R_t = \frac{(R_s + R_p)}{2}$ and the transmission is $T_t = 1 - R_t$. Materials that absorb light can also be defined by their complex index of refraction; these formulas still hold and can be used to calculate the reflection on lossy metals, for instance. In that case, the angle θ_2 is not defined as a real number, and the complex value of $\cos(\theta_2)$ must be directly deduced from Snell-Descartes law.

9.2.3 Modeling with a Ray Optics Approach

The behavior and propagation of light is thoroughly described by Maxwell's equations, but their resolution require tremendous calculation power in finite, heterogeneous structures, and therefore cannot be used to calculate light propagation in complex and large geometries such as those used for LEDs. The laws of geometrical optics constitute a very good approximation in the limit of dimensions larger than that of the wavelength of the light travelling in the material (for GaN blue LEDs, the wavelength is typically 460 nm in air and 187 nm in GaN). Ray tracing is a Monte Carlo-based method that uses the laws of geometrical optics to statistically simulate the behavior of a great number of light rays in a given system. Throughout

Fig. 9.3 Transmitted power versus incidence angle at a GaN/air and GaN/epoxy interface
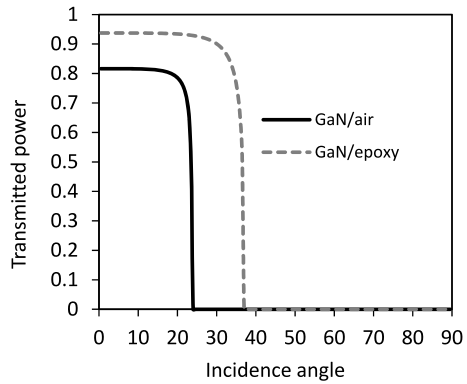
this chapter we will report results from the LightTools® ray tracing software. Ray tracing requires a detailed knowledge of the geometry and of the optical properties of all the different materials present in the system.

During a ray tracing simulation, many light rays are shot in random directions (or following a probability distribution corresponding to the characteristics of emission when emission is not random) from defined light emitting sources. The rays follow the rules of geometrical optics until they are either absorbed or exit the system. For example, during propagation in materials, rays will follow Beer-Lambert's absorption laws using the absorption coefficient of the given material or, equivalently, the complex refractive index of the material. Then upon arrival at an interface between two materials, rays of light can be reflected or refracted according to Fresnel's laws, which also require knowledge of the refractive indices of the materials in play. Surfaces can be defined as reflectors, absorbers, scatterers or smooth optical surfaces.

Once a large number of rays have been traced, the extraction efficiency and breakdown of the absorption mechanisms can be calculated from the fate of all the rays to give an overall picture of light behavior in the chip. Of course, because the ray tracing method relies on the rules of geometrical optics, it is unable to describe and calculate phenomena such as interference or diffraction, which can only be described by optical wave theory, but these effects are of minor importance for chips with large-scale features.

9.3 The Issue of LEE in GaN-Based LEDs

GaN-based LEDs tend to have low LEE, largely because of the high index of refraction of GaN ($n = 2.5$). This high value leads to a low critical angle ($23.6°$) and a small extraction cone. The total transmission versus angle for both interfaces is plotted in Fig. 9.3, using $n = 1$ for air and $n = 1.5$ for epoxy.

The integrated values over all solid angles of a half sphere is 6.5 % for the GaN/air interface. Extraction into dome-shaped silicone or epoxy encapsulant ($n \sim 1.5$) typically doubles the extraction value to $\sim$17.4 % by increasing the escape angle to $\sim$36.9°. Most light escapes the encapsulant material into free space

because a dome shape ensures that most rays impinge at the encapsulant-air interface at near-normal angles, and thus within the critical angle (Fig. 9.2). Further increasing the index of the encapsulant material could lead to unity extraction efficiency, but this requires higher indices than are typically available for most epoxies and resins [19].

In a simple first-pass light extraction analysis, it is assumed that the light that is not directly extracted is lost. This actually depends on the internal losses of the device structure and on its geometry; if there were no internal losses, one might expect that the light will eventually escape after many rebounds in the device. However, if the geometry of the chip is very regular, such as cubic, multiple reflections on the various interfaces may occur without change of angle, such that the rays propagate indefinitely under TIR (Fig. 9.4b). In this situation, even the tiniest loss mechanisms will destroy the optical power carried by such rays. Thus, even with a very transparent substrate, extraction is limited to ~30 % [11, 12] if no other extraction techniques are employed.

Some efficient schemes are based on roughened surfaces or interfaces that change the direction of light at each rebound [9, 28, 31, 32]. Another option is the use of chip shaping, such as pyramidal [13, 14] or cone [26, 30] shaping, which breaks the symmetry of the optical cavity and randomizes the light rays which can then escape after a few trips within the device (Fig. 9.4h–i). This can increase efficiency up to 60 % in transparent substrate LEDs.

9.4 Ray Propagation and Absorption in Layered Structures

It is useful to understand the modal distribution of emission in layered structures in order to better understand the issues for conventional LEDs grown on sapphire, which has a heterogeneous set of optical properties. Light is emitted into extracted photon modes or into a variety of guided modes. The emission distribution actually depends very little on the specific layer thicknesses or on the active layer structure [21, 22].

A ray tracing simulation can determine the intensities of the various modes in a semi-infinite planar structure. Assuming a two-layer structure of GaN ($n = 2.5$) on sapphire ($n = 1.7$) with light emission radiating from inside the GaN (as in Fig. 9.5) and no sources of light absorption, 8.6 % of light is directly extracted into the air cones, 23.3 % enters the sapphire mode and the remaining 68.4 % is guided in the GaN mode.

As rays propagate in various modes within the device, many absorption sources in the structure can contribute to reduction in LEE (Fig. 9.6) including:

1. Absorption in the substrate
2. Absorption (complete or partial) on the contact pads
3. Absorption upon reflection on an imperfect backside mirror
4. Absorption in the active region (doped regions, current spreading layer and quantum wells)

High Index Encapsulation

(a)

n=1

n=2.5

Extraction cone

TIR ray

(b)

n=1

n=2.5

Trapped ray

Extraction cone

(c)

Epoxy Dome

n=1.5

n=2.5

Dome-Shaped Encapsulant

Surface Texturing

(d)

10 µm

SEM Image of Roughened GaN

(e)

[1120]

[1702]   [1101]

2.0 µm 3.4 µm

10 µm

SEM Image of Patterned Sapphire

(f)

GaN Active Region

n=2.5

n=1.7

Patterned Sapphire Substrate

Ray randomization by Patterned Sapphire

Chip Shaping

(g)

Soraa GaN Substrate LED chip

(h)

1 mm

Lumileds AlGaInP Inverted Truncated Pyramid Chip

(i)

Quantum Wells

GaN Substrate

Silver Header

Light Extraction with Angled Sidewalls

Fig. 9.4 Summary of techniques used to enhance light extraction in LED chips. These include (**A**) high index encapsulation to increase the critical angle for extraction, (**B**) surface texturing such as roughening or patterning to randomize light, and (**C**) chip shaping to redirect rays at each bounce and increase the probability of extraction on the next attempt

Extracted (8.6%)

Quantum Wells

Guided (68.4%) in GaN

GaN

n=2.5

4 µm

Sapphire Substrate

Guided in Sapphire (23.3%)

n=1.7

130 µm
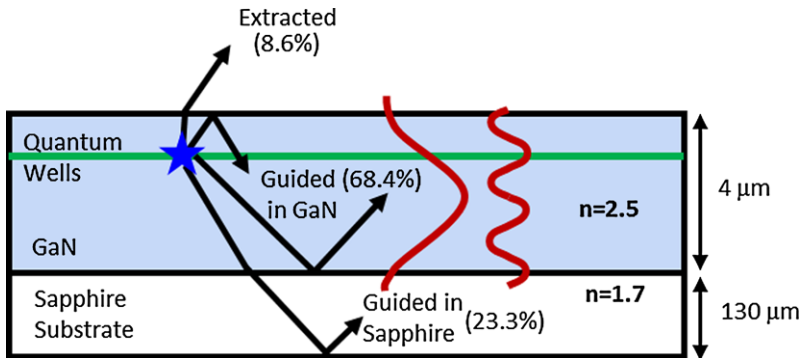
Fig. 9.5 Schematic of the various modes existing in a standard GaN on sapphire LED
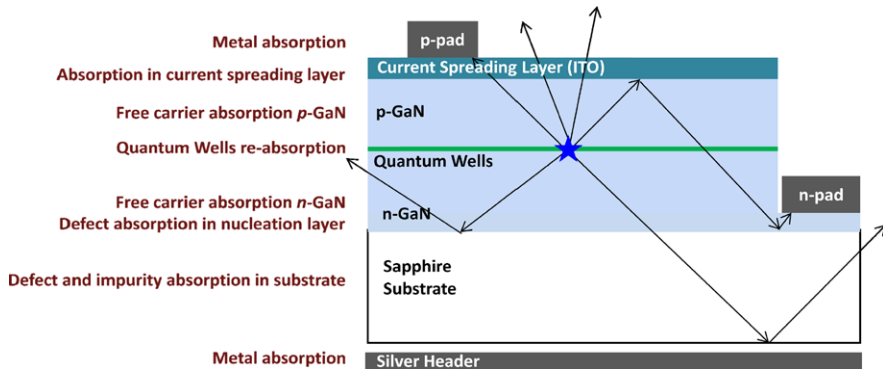
Fig. 9.6 Schematic of the various absorption phenomena within an LED

Depending upon the structure of the chip and on the extraction method used, the different sources of loss will not have the same importance and will not have to be optimized in the same way.

9.4.1 Localized vs. Distributed Loss Sources

Losses accumulate due to materials absorption when light propagates through the layers (losses are then distributed and connected to propagating rays) or can occur at selected areas of the device (localized losses), such as at contacts or mirrors. These will have different impacts on LEE, depending on the extraction mechanisms. One of the most common methods for light extraction in LEDs uses surface roughness or substrate patterning to break the guided modes and randomize light in the structure. Light then bounces in the LED until it ends in an extraction cone at a surface of the chip and is effectively extracted. After each unsuccessful extraction attempt, a light ray must travel through the LED, and thus propagate through a certain path length of absorbing material. In such cases, these distributed losses must be kept to a minimum to ensure high extraction. Hence, the goal will be to minimize the roundtrips of light in the device by enhancing the extraction probability at each incidence at the semiconductor-air (or encapsulant) interface. For localized losses, the goal is to diminish the interaction of rays with lossy regions, e.g. by minimizing contact area, insulating lossy mirrors from the device material by a low index dielectric gap, or using current apertures away from the contact regions. Such strategies will be discussed in the section on real-world LED structures.

9.4.2 Material Absorption

Optical losses due to propagation in absorbing materials are described by the Beer-Lambert law. To the first order they are directly proportional to the absorption coef-

ficient and the thickness of the materials. This type of loss will occur in the substrate material and more importantly in the active region, which is often more lossy and interacts more with the rays. It is therefore important to keep these regions as thin as possible while enabling satisfactory electrical properties, especially for very lossy materials such as the current spreading layer or p-doped GaN layer. Tradeoffs for the substrate thickness can occur when the chip geometry is such that sidewall extraction is important. This will be explained in more detail for a GaN chip on a bulk GaN substrate.

9.4.3 Quantum Well Absorption and Photon Recycling

Absorption of light in the quantum wells will generate electron-hole pairs, which may be recycled by radiative recombination to form a new photon. In this case, the effective loss every time a ray of light goes through the quantum wells is only $(1 - IQE) \times$ (absorption probability), which can, however, lead to a large decrease in EQE when repeated multiple times [27, 28]. A favorable effect of photon recycling is that the photons absorbed and reemitted from the quantum well will effectively loose directional information and will be reemitted in random directions. We will neglect this phenomenon because: (i) the randomization effect is small compared to that induced by the roughened surface or substrate; (ii) absorption is rather low for the typical few tens of nm total QW thickness due to the Stokes shift between emission and absorption peaks. This is even more pronounced for c-plane LEDs where the Quantum Confined Stark Effect (QCSE) leads to a very large Stokes shift (for measurements of the Stokes shifted emission and absorption, see e.g. [6]). This is compounded with the very high IQE so that the net contribution to losses per pass, $(1 - IQE) \times$ (absorption probability) would be quite small.

9.4.4 Metal Losses: Mirrors and Contacts

The metallic contacts in the LED are a potential source of loss by absorption. The typical absorption length of these metals is in the order of 20 nm, much thinner than the thickness of the contacts (typically 300 nm), so all the incoming light is either reflected or absorbed as the contacts can be considered optically thick (i.e. infinite).

Metal reflectivity can be accurately described for all angles of incidence using Fresnel's laws. The metal is then characterized by its real and imaginary indices, and the reflectivity is given by Eqs. (9.3) and (9.4) where the complex value of the metal index for n_2 is used. The result of the calculation for silver and titanium in air at a wavelength of 450 nm is given in Fig. 9.7 using $1.69 + 2.27i$ for titanium's refractive index and $0.15 + 2.46i$ for silver's refractive index.

As can be observed, the reflectivity significantly varies with the incidence angle, especially for titanium, so the angular dependant reflectivity is used in the simulations through the complete Fresnel law for the Ti contacts rather than the normal
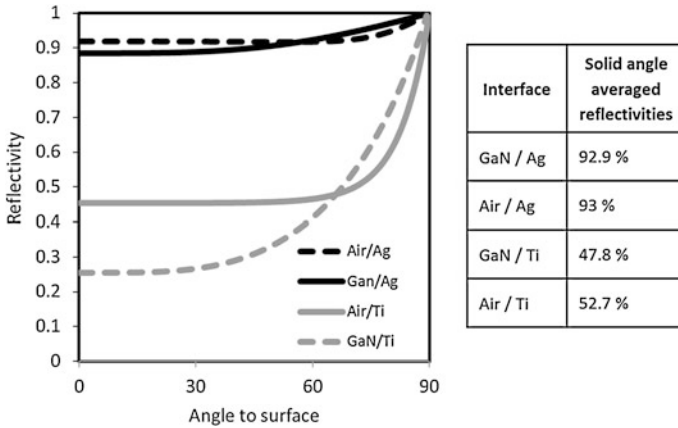
Fig. 9.7 Angular dependence of reflectivity and average reflectivity for silver and titanium at either GaN or air interfaces

or average reflectivity. One should also note that these metals are usually used in combination with other materials to ensure better electrical and optical properties, so this is a first order approximation.

The contact regions are a major source of loss, which will be shown in the precise modeling of LED structures. This is due, in particular, to the emission of light directly below the contacts, a fraction of which is promptly absorbed when it strikes the contacts. Strategies to diminish their impact include reducing the surface area of the top contacts, using current blocking layers below the contacts [5], or using current apertures away from the contacts.

9.5 Extraction Strategies and Considerations

9.5.1 Index Matching and Guided Modes

In addition to roughening and chip shaping techniques that modify the geometry of the substrate, packaging schemes that utilize high index, transparent encapsulant materials may also be used to increase light extraction [19]. In this scenario, instead of attempting to break the guided modes in the material, the rays are simply given a larger extraction cone. If the encapsulant is large compared to the chip and dome-shaped, then the rays propagating to the edge of the encapsulant would likely strike the surface at angles near perpendicular, and thus be extracted to free space (the normal incidence reflected intensity for an epoxy ($n = 1.5$)/air is only 4 %).

Semiconductor materials in LEDs have high refractive indices, exceeding $n = 2$. The rays incident on infinite-area flat surface between the semiconductor and the encapsulant will only have a complete extraction cone when the encapsulant index equals the semiconductor index. In a real cuboid-shaped chip with finite surface
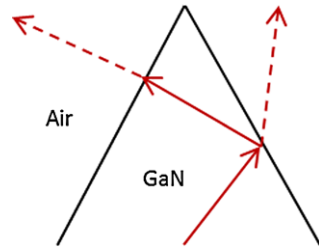
area, however, full extraction can be achieved when the encapsulant index is somewhat lower than that of the semiconductor. This is because the extraction cones from different faces of the chip will begin to overlap, enabling all of the rays to escape. For a semiconductor with an index of $n = 2.5$, like GaN, overlap of the cones occurs around $n = 2$. As shall be demonstrated later, in chips with roughening or advanced geometries, extraction saturates at as low as $n = 1.8$.

9.5.2 The Effect of Surface or Interface Texturing

A very common method used to break the guided modes in LED structures is to introduce surface roughness on one of the interfaces to randomize the propagation direction of light rays as they strike the surface (for a review, see e.g. [35]; see also [10, 17, 18, 33]). This enables all the optical modes in the LED to be radiative after some impacts on the interface, i.e. to have an in-plane wave vector smaller than the wave vector in air. Anisotropic wet etching of the N-face of GaN that forms hexagonal pyramids at the surface is a common way of achieving this surface roughness [9]. Interface roughness can be produced by using a patterned sapphire substrate before growth [32].

While the main effect of surface roughness is randomization of light rays, another very important effect is often overlooked: the single path escape probability is increased [15]. We define single pass escape probability as the probability for a random ray of light propagating from the high index material to be transmitted to the lower index material rather than reflected. This value will be modified by surface roughness since light, at the second escape attempt, will not be intersecting the material interface with the same angle, giving the ray a greater chance to be in the extraction cone. Figure 9.8 shows a ray bouncing twice against the surface, increasing its probability of being extracted by roughly a factor of two.

The two important effects of surface roughness are, therefore, the randomizing efficiency and the single pass escape probability enhancement. We discuss in the appendix the randomization effect of surface roughness and sapphire substrate patterning, and give a brief description of the enhancement of single pass extraction by surface roughness.

The first pass extraction probability was calculated by ray tracing for a surface of randomly roughened pyramids, such as would result from a wet chemical etching.

Fig. 9.9 GaN roughened
surface transmission in air
($n = 1$) and in an epoxy
encapsulant ($n = 1.5$)
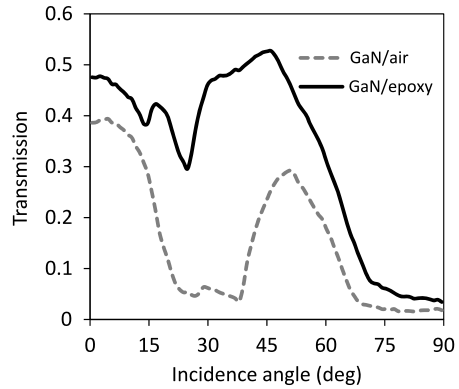
As is shown in Appendix C, randomly distributed pyramids or regularly patterned
pyramids give similar results for ray tracing purposes, so regular patterns are used
in the rest of the simulations for simplicity.

The extraction probability for a roughened surface is completely changed com-
pared to a flat surface: less light is transmitted at angles below the critical angle, but
light transmission is increased in the range of 40°–70°, outside of the flat surface
extraction cone. Once averaged over all solid angles, it leads to a 13.4 % average
first attempt transmission, twice better than for a flat surface (6.5 %).

Detailed simulations (see Appendix A) show that due to randomization of re-
flected rays, the transmission probabilities of 2nd and further attempts are nearly
constant around 13 %. So even if the rays are not extracted on the first try, they have
a large probability of being extracted later. This is very different from a flat surface;
if the ray is not in the extraction cone, it will never be extracted, so in that case, the
transmission probability of further attempts drops to zero.

Increasing the index of the surrounding medium has a similar effect for rough-
ened surfaces as for flat surfaces; for a given ray hitting the interface, the reflection
angle will not be changed, but the fraction of the extracted light will be improved
as more incident rays fall in the increased extraction cone. Comparing the angular
transmission of the hexagonal pattern in air and epoxy (Fig. 9.9), one can see that
the peak of extraction is around 50° for both, but is more widely spread for the
epoxy case. The total transmission goes from 13.4 % to 26.9 % (compared to 6.5 %
for flat GaN/air and 17.5 % for flat GaN/epoxy interfaces). However, we can note
that in both cases the transmission at high incidence angles is low.

9.5.3 Limit to the Ray Tracing Method

When dealing with devices where the feature size of the texture is close to or smaller
than the wavelength of light considered, the ray optics method is no longer suit-
able. The only wave optics results on texturing to our knowledge are found in [7]
and [8]. We find that the effects of roughness described here and in Appendix A

(namely, randomization and increased single pass transmission) are similar in both ray optical and wave optical treatments. The single pass extraction found by David is larger than for our large features (31 % for the single pass extraction instead of 13.4 % here), which should improve the competition between extraction and internal losses within the chip. The same trend at high incidence angles on the interface is observed, however, in both the ray optics and the wave optics approaches: the transmission efficiency tends to zero, meaning that the roughness does not improve light extraction. Thus, the trends reported here *between* structures are valid, at least semi-quantitatively.

9.5.4 Sidewall Extraction and Chip Shaping

While surface roughening strongly improves light extraction, it can also induce some fabrication difficulties. For example, photoelectrochemical etching mainly works on the *N* face of GaN, which is only accessible in flip-chip geometries by removal of the substrate. Some LED geometries work around this problem and rely on different extraction mechanisms that do not require surface roughening or patterned substrates.

High extraction in LEDs may rely strongly on the extra extraction cones on the sides of the chip [11, 12]. Extraction from these cones is referred to as sidewall extraction. Two features are crucial for sidewall extraction to contribute to optimized efficiency. First, the light must be able to reach the sidewalls without being significantly absorbed, and second, the geometry must be adapted to not have any indefinitely propagating guided modes.

For the first requirement, LEDs with GaN substrates are the most naturally adapted, since light can propagate easily in the substrate which is naturally index matched to the active region. However, GaN substrate LEDs are limited by a maximum size that allows light to reach the side without undergoing substantial absorption losses in the substrate material.

The second requirement can be fulfilled by shaping the chip in a way that non-extracted light impinging on an interface is reflected in the structure at a different angle, directing it into an extraction cone. This can be achieved by either in plane or out of plane chip shaping. In plane shaping means that the sidewalls remain vertical, and that the top view of the LED is a non-rectangular shape, such as a triangle or parallelogram (Fig. 9.4g). The triangular shape meets the above requirements, and has the benefit of allowing cleaving along the natural planes of the crystal. Additionally, this geometry allows close packing of a larger number of die on the wafer [4].

A ray tracing simulation of different chip shaping schemes shows that out of plane chip shaping such as the truncated inverted pyramid (TIP) (Fig. 9.4h) are most effective at breaking the guided modes (Fig. 9.10). The in plane shaping (triangle) provides a benefit over the perfect cuboid, but is not as effective at breaking all guided modes.
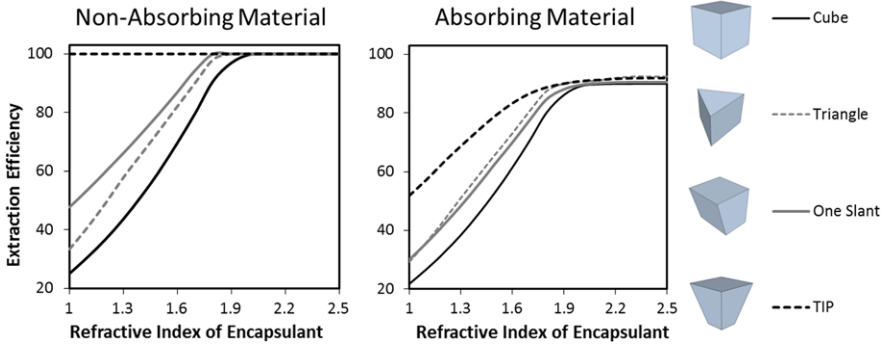
Fig. 9.10 Four different chip geometries ($n = 2.5$) with smooth surfaces and central isotropic point sources were compared using ray tracing to determine how efficiently they break the guided modes as a function of increasing ambient refractive index. Trends are shown for two cases: for non-absorbing materials (where LEE goes to 100 %) and for absorbing materials where the LEE reaches a maximum ∼90 % LEE at high ambient refractive indices. The absorption value for the latter was set to yield a typical amount of loss that is representative of a full LED structure (including contacts, etc.)

Introducing just a single angled sidewall to a non-roughened GaN-based device can increase extraction dramatically (∼15 %) by breaking the initial symmetry of the chip. In an idealized model of a point source emission at the center of a non-absorbing cube with refractive index $n = 2.5$, only two sidewalls need to be angled to achieve 100 % light extraction into air, provided that those sidewalls are adjacent and not opposite. Further, the angle can be as small as 3 or 4 degrees to achieve this.

It is also useful to note that even for the geometry with no chip shaping, the system does not need to be index matched to the encapsulant to get 100 % extraction. The cube shaped non-absorbing semiconductor ($n = 2.5$) in Fig. 9.10 reaches 100 % extraction with an encapsulant index of just $n \sim 2$.

In a real chip with lossy materials, it is advantageous to angle all four sidewalls and to have a greater degree of angle in order to minimize the number of bounces, and thus the path length in the material. An example of this situation will be described further in a later section.

9.6 Simulation of Real LEDs

In order to quantify all these loss mechanisms and to compare different extraction strategies, several representative chips of the state of the art in industry were compared using ray tracing simulations. While this does not aim to be an exhaustive comparison of all chips being manufactured today, it highlights the crucial parameters for different chip geometries. Different parameters are swept independently to understand their effect on LEE. Material properties were chosen based on the most commonly accepted values in the literature, but would differ from one growth
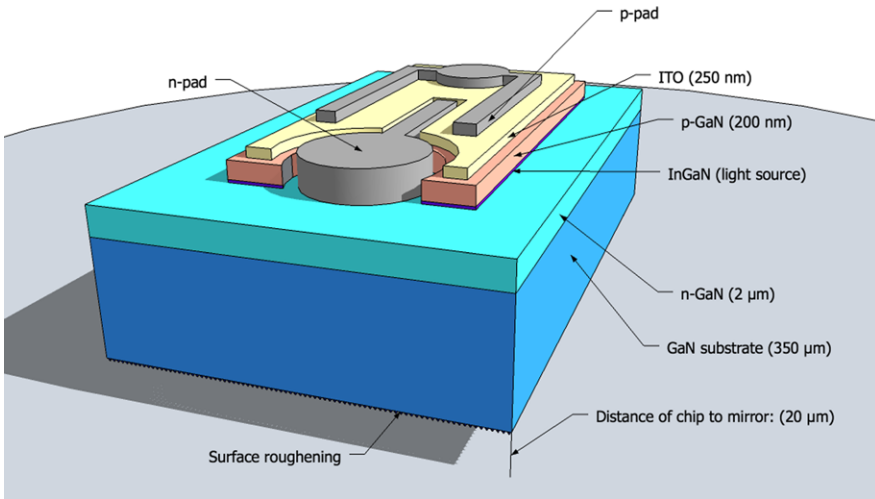
Fig. 9.11   Schematic of GaN substrate chip geometry as used in simulations (not to scale)
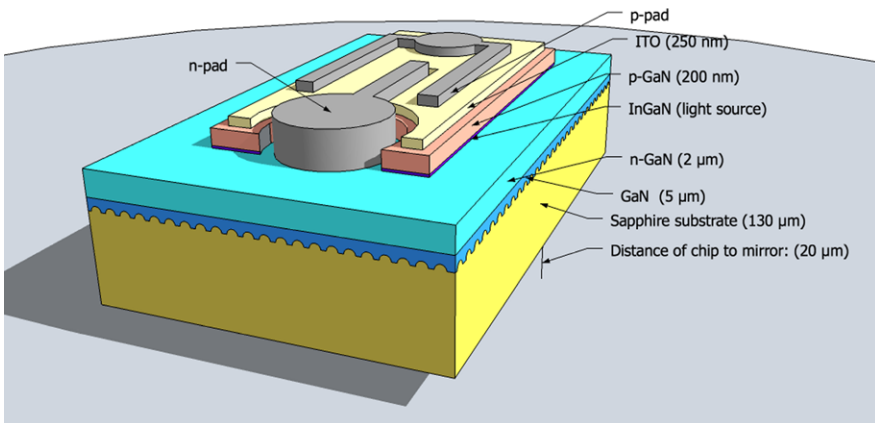
Fig. 9.12   Schematic of PSS chip geometry used for simulations (not to scale)

method to another. Therefore, the calculated values of absorption in the layers, while providing much insight on the extraction mechanisms of the chip, should not be considered as universal values.

All simulated chips are GaN-based chips with the same active region, described in Figs. 9.11–9.13. The first two chips are similar but the first one is grown on a GaN substrate with a roughened backside (Fig. 9.11) while the second is grown on a patterned sapphire substrate (Fig. 9.12). Both of them have a back mirror separated from the chip by an air or dielectric spacer. The third is a simplified version of a flip chip configuration that outlines the important effects, in particular that of the lossy back mirror acting as the *p*-side contact of the chip (see below Fig. 9.21). All chips

Fig. 9.13 Schematic of the top geometry of GaN and PSS reference chips used for the simulations and table of the layer thicknesses in the active region

Layer	thickness
ITO	250 nm
p-GaN	200 nm
n-GaN	2 μm
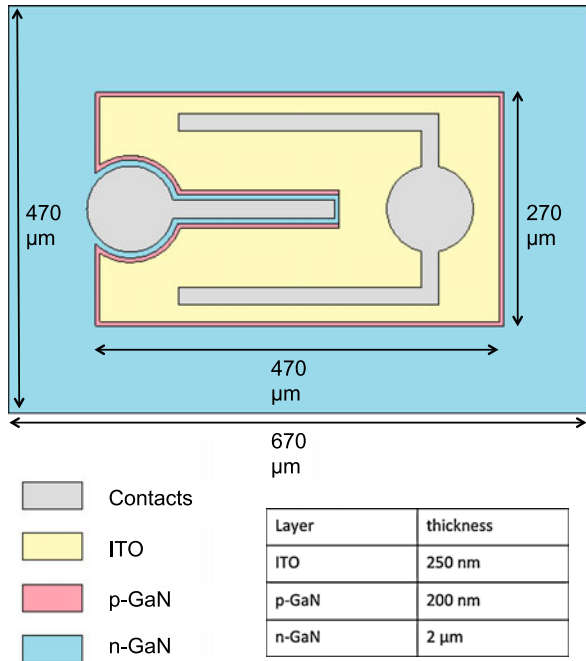
Table 9.1 Relevant intrinsic materials properties for ray tracing simulations for reference chips: absorption coefficient and refractive index

	Refractive index	Absorption coefficient
Bulk GaN	2.5	1 cm^{-1}
Sapphire	1.7	0.1 cm^{-1}
p-GaN	2.5	100 cm^{-1}
n-GaN	2.5	7 cm^{-1}
ITO	2.1	500 cm^{-1}
Epoxy	1.5	0
Silver header	–	92 % isotropic reflectivity
Contacts	$1.69 + 2.27i$	Reflectivity given by Fresnel law (Ti on GaN, 47.8 % average)

are simulated in an infinite environment of refractive index 1.5 to model of air or encapsulation. All rays entering the encapsulant are considered as extracted, as is the case to a good approximation for low loss, dome shaped, large encapsulants where light back-reflected at the encapsulant air interface represents only a very small portion, and has only a very small probability to hit a highly absorptive region of the LED. The exact geometries of the simulated chips are given in Fig. 9.13, and the material properties used are described in Table 9.1. In first simulations of the

Table 9.2 Extraction efficiency of the reference GaN on GaN chip simulation in different encapsulation environments and the breakdown of loss mechanisms by material layer

Encapsulant	Epoxy	Epoxy	Air
Current blocking under P contact	No	Yes	No
Percent extracted	72.6 %	75.1 %	51.9 %
Loss on P contact	4.4 %	2.2 %	5.1 %
Loss on N contact	1.6 %	1.6 %	3.7 %
Loss in N doped GaN	0.6 %	0.6 %	1.5 %
Loss in ITO	2.5 %	2.5 %	4.4 %
Loss in undoped GaN substrate	12.1 %	12.1 %	28.8 %
Loss on backside mirror	5.5 %	5.5 %	3.8 %

structures, we used the geometries and parameters of Fig. 9.13 and Table 9.1. We later report structures with varying geometries and parameters.

9.6.1 GaN on GaN Chip Simulations

The GaN substrate chip presented in Fig. 9.11 is simulated with the material properties described previously. Three scenarios are chosen: one when the encapsulant is air, and two where the encapsulant is epoxy. The source of light is situated at the interface of the p-GaN and the n-GaN layers. In one case though, the emission is removed from under the p contact, so as to simulate a current blocking layer. The different losses are extracted from the simulation results and are tabulated in Table 9.2. Various simulations were then performed while varying only one parameter to evaluate the sensitivity of the LEE to the value of that parameter for the GaN substrate chip LED, and for the PSS substrate chip.

9.6.1.1 Effects of Propagation in the GaN Substrate

The losses are dominated by the absorption in the chip's GaN substrate, the contacts, and the backside mirror. The modeling indicates that rays hit the top surface once and a half on average. This is consistent with the loss observed in the substrate, which corresponds to approximately one millimeter of distance travelled in the substrate per ray before extraction. This illustrates the fact that the light propagates in the entire chip and is quickly extracted from one of the sidewalls or from the top surface with very few reflections, and that the material quality of the GaN is of crucial importance for the LEE in these chips. This is shown in Fig. 9.14 where the absorption coefficient of the GaN substrate has been varied while keeping all other parameters constant.

Fig. 9.14 Influence of
absorption coefficient of the
undoped GaN substrate on
the LEE for a GaN substrate
chip (the *dot* indicates the
reference chip for 1 cm^{-1})
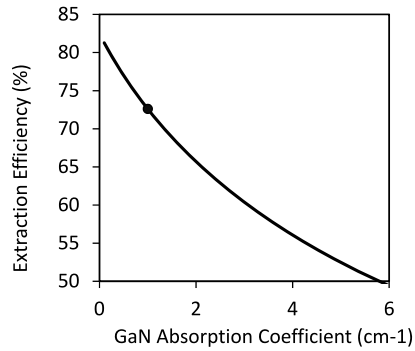
Fig. 9.15 Influence of
substrate thickness on LEE
for PSS or GaN substrate
LED chips (the *dot* indicates
the reference chip for 130 μm
and 300 μm respectively)
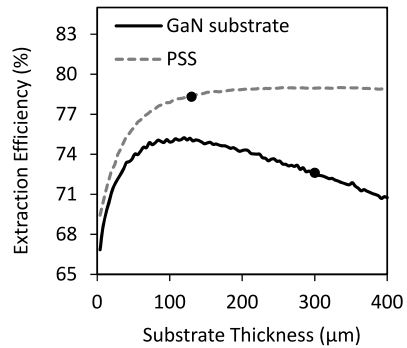
The transparency of bulk GaN substrates has seen improvements as defect concentrations in the growth have been reduced. As a result, absorption coefficients in the literature range over two orders of magnitude, as high as 125 cm^{-1} [3]. Modern substrates from industry suppliers typically have absorption coefficients between 1 and 2 cm^{-1} and are grown by hydride vapor phase epitaxy. Ammonothermal growth is another technique that is attracting research attention for high efficiency, high transparency bulk GaN growth.

Because light propagates extensively in the substrate, its thickness has an optimal value: a thinner substrate will induce less absorption by the GaN but will also make it harder for rays to reach the sidewalls and will therefore rely more on top surface extraction, which is inherently more lossy due to the contacts and the active region. On the contrary, because the PSS has a less absorbing substrate and benefits from sidewall extraction, its LEE does not decrease with increased thickness (Fig. 9.15).

9.6.1.2 Roughening Interaction with Backside Mirror

The roughening pattern in the GaN-on-GaN chip design is situated on the backside of the chip, which limits its benefits to its randomization effects: As previously discussed, roughening patterns at an interface help to both randomize rays and to
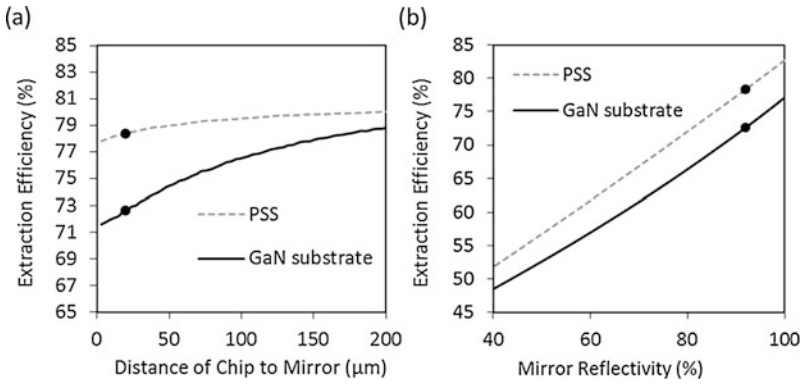
Fig. 9.16 Influence of (**a**) the backside mirror distance to the chip (the chip and mirror are separated by a transparent dielectric or air), and (**b**) the backside mirror reflectivity on LEE (the *dot* indicates the reference chip at 20 μm and 92 % respectively)

increase the single path extraction. In the case of the GaN-on-GaN chip with backside roughening, the rays extracted from the backside will bounce off the mirror and back into the LED, canceling the benefit of improved single path extraction on that side. Increasing the distance between the chip and the mirror improves this, such as shown in Fig. 9.16 because it will enable a larger fraction of rays extracted from the backside to escape the structure without reentering in the LED. This design causes other problems, however, such as poorer thermal management due to a larger thermal resistance of the dielectric or air gap. Various strategies have been explored to solve this problem, but have not proven practical [24]. Finally, the backside mirror reflectance is also of high importance, justifying the use of silver that has a very high reflectivity as the backside mirror.

9.6.1.3 Angled Sidewalls

The GaN substrate LED is an interesting design because the naturally index matched substrate enables light to escape the thin, lossy active region (as opposed to the patterned sapphire substrate, which will be discussed next). The challenge is that the GaN substrate must be of superior quality (low absorption) to ensure low enough losses, since light propagates extensively in it, and requires an extraction strategy such as surface roughening or angled sidewalls to break the guided modes.

Angled sidewalls are effective options for increasing light extraction in GaN chips because of both the high refractive index and high absorption of GaN substrates. This section demonstrates the extraction improvements achieved by shaping the chip into a truncated pyramid structure created by angling all four sidewalls [13]. This shape breaks all of the guided modes propagating to the sidewalls, while reducing the path length of the light in the lossy substrate. The large improvements based on such techniques were first demonstrated in the GaAlInP materials system for red-orange LEDs where the initial GaAs absorbing substrate is substituted by

Fig. 9.17 A truncated inverted pyramid chip shaping design is applied to the roughened and non-roughened GaN-on-GaN chip simulation. The LEE is viewed as a function of sidewall angle. Simulations were performed using an epoxy ambient environment
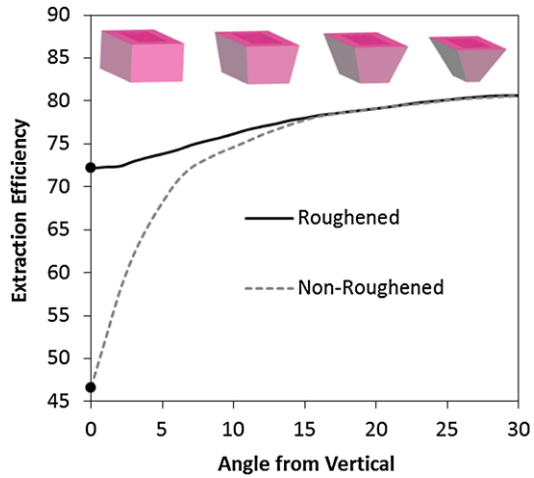
a shaped, transparent GaP substrate leading to extraction efficiencies up to 60 % [13, 14] (Fig. 9.4h).

In Fig. 9.17, extraction efficiencies are shown for increasing sidewall angle for a GaN-on-GaN chip in an epoxy environment. It is apparent that non-roughened GaN substrates experience a greater increase in extraction by angled sidewalls than roughened GaN substrates. When angles of ∼15 degrees or greater are employed, the distinction between roughened and non-roughened chips vanishes. Furthermore, 15 degrees is an attainable angle in device processing.

In this simulation, all four sidewalls of the chip were angled to increase extraction, but it should be noted that ∼50 % of the improvement in extraction is obtained by angling just the first sidewall (at 15 degree angle from vertical). Furthermore, ∼96 % of the extraction improvement is achieved by angling just two of the four sidewalls, provided that they are adjacent sidewalls, which will break all guided modes in the chip.

The effects of combining angling and roughening methods are discussed in Appendix B.

9.6.1.4 Impact of Current Spreading Layers

Current spreading in the LED is a necessity because p-GaN is such a poor conductor. Without current spreading, there is significant current crowding under the p contact to the p type material, which lowers IQE. This is done in "epi up" structures (p side of active layer is always grown last because p layers introduce defects and roughness detrimental to active layer growth) by using transparent current spreading layers (thin metal e.g. Ni/Au, transparent conducting oxides e.g. ITO or ZnO), or by having a p contact covering the p type material in flip chip device structures.

Table 9.3 Extraction efficiency of the PSS chip simulation in different encapsulation environments, and the breakdown of loss mechanisms by materials

Encapsulant	Epoxy	Epoxy	Air
Current blocking under P contact	No	Yes	No
Propagated	78.3 %	82.3 %	66.4 %
Losses:			
P contact	6.8 %	2.6 %	7.9 %
N contact	1.5 %	1.7 %	4.5 %
P GaN	1.3 %	1.3 %	2.1 %
N GaN	1.3 %	1.3 %	2.9 %
ITO	5.6 %	5.7 %	9.5 %
Undoped GaN	0.2 %	0.2 %	0.5 %
Sapphire	0.3 %	0.3 %	1.7 %
Mirror	4.5 %	4.5 %	4.3 %

In such cases one can assume in the simulations that the light is uniformly generated throughout the active region. If no spreading layer were used, current crowding would occur under the p contact leading to a localization of emission under the contact, which would reduce light extraction efficiency, in addition to diminishing IQE.

When a current spreading layer is used, the p contact absorption can be mitigated through the use of a current blocking layer under the p contact, forcing hole current and emission away from the p contact [5]. The effect of this is shown for the GaN and patterned sapphire substrate chips: the result of a simulation where light emission has been suppressed under the p contact region is given alongside the other results in Tables 9.2 and 9.3. Quite naturally, it displays that the improvement in extraction efficiency comes from lower absorption in the p contacts.

9.6.2 Patterned Sapphire Substrate (PSS)

Sapphire substrates are the most common substrates for industry GaN LED growth because they are low cost, very transparent (0.1 cm^{-1}) and can also lead to high IQE devices. Patterns can be formed on top of the sapphire in order to create the same effect as a rough surface, i.e. randomizing light ray trajectories. Ray tracing results of a PSS chip, described in Fig. 9.12, are given in Table 9.3.

9.6.2.1 Propagation of Light in the Substrate and Active Region for PSS Chips

The low refractive index of PSS leads to much of the light staying confined in the active region layer: it hits the top surface of the chip on average three times before

Fig. 9.18 ITO absorption impact on LEE for GaN substrate and PSS chips (the dot indicates the reference chip at 500 cm^{-1})
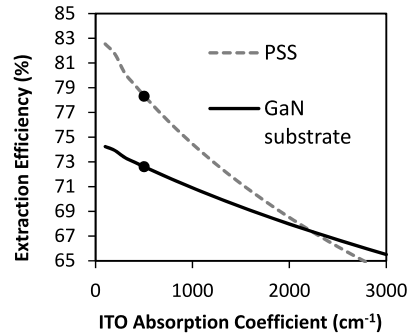
being extracted, which is twice as often as for the GaN substrate chip and thus leads to a much higher absorption in the doped regions. The light that does make it into the substrate is almost all directly extracted into the surrounding epoxy because the indices of sapphire and epoxy are close. This requires only that the sapphire is thick enough that light can easily reach the sidewalls and does not bounce back into the active region. In fact, the simulations show that light rays only travel on average a distance on the order of 150 μm in the substrate, compared to 1 mm for a GaN substrate. Because the sapphire is also very transparent, there is very little loss in the substrate.

The two exit paths for light are therefore the top surface and the sapphire substrate sidewalls. In this regime there is therefore no need for any modification of the substrate shape to enhance extraction. The most important region to optimize in a PSS chip is the active region. As can be seen from the single parameter sweeps, the PSS chip is shown to be much more sensitive to the parameters of the active region (such as ITO absorption coefficient, Fig. 9.18) than the GaN chip. Optimization of these layers can be achieved in two ways: ensuring that the active region is as transparent as possible (which depends mostly on material quality, contact geometry and current spreading), and making sure the light spends as little time as possible in it (which is the motivation for patterning the sapphire surface on which the GaN is grown). The LEE is also much less sensitive to the distance between the substrate and the mirror (Fig. 9.16a) because light extracted into the sapphire will very likely be extracted by the sidewalls or the top side if it is not extracted by the backside, since the index of sapphire and epoxy are almost matched, which is not the case for GaN.

9.6.2.2 Effect of the Patterning the Sapphire Substrate

The patterned sapphire surface is what is responsible for randomizing the direction of the light rays trapped in the active region and ensuring that light can be extracted quickly from the active region. Increasing the pattern density increases the randomization efficiency, and the light therefore propagates less in the active region, improving extraction efficiency (Fig. 9.19). Patterns that enable more light to

Fig. 9.19 Influence of pattern density on LEE for a PSS chip (the *dot* indicates the reference chip at 34.9 %)
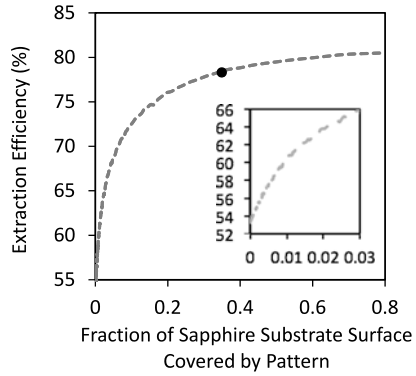
refract into the sapphire are also beneficial, as subsequently, this light will be easily extracted from the substrate to air. If no patterning is done, only rays emitted within the extraction cone of the top air surface and bottom epoxy surface may be extracted quickly. The rest must propagate all the way to the sidewall in the lossy active region, and will only be extracted if they lie in one of the sidewall extraction cones. This results in high losses and underlines the importance of the substrate patterning.

9.6.3 Limitations in GaN Chip Dimensions

In summary, the PSS chip and the GaN substrate chip both rely on extraction from the sidewalls of the substrates, although the two mechanisms differ slightly. GaN substrates are of the same material as the active region (and are therefore index matched) so that light penetrates easily into them, but suffer from losses due to a higher absorption coefficient of the substrate. PSS chips on the other hand have an almost lossless substrate, but are not index matched to the active region. Light thus takes a longer time in the lossy active materials before getting into the substrate, but can then propagate with very low loss and quick extraction. The crucial levers for high LEE are therefore substrate absorption for a GaN chip, and active region absorption for a PSS chip.

This results in a fundamental limitation for the dimensions of GaN chips that rely strongly on sidewall extraction. The light must be able to propagate to the sidewalls without undergoing large losses, so their size is limited by the absorption coefficient of the GaN. PSS substrate chips, however, can be made quite large because the absorption coefficient of sapphire is low, as long as the aspect ratio of the substrate is kept constant. Light can then propagate enough horizontally to reach the sidewalls, as shown in Fig. 9.20, where both types of chips are increased in size. In one scenario the substrates are kept the original thickness, in the other they are also scaled up to allow more sidewall extraction. The GaN chip suffers strongly in both schemes but the PSS extraction remains satisfactory when the aspect ratio is kept constant. For simplicity, contacts were omitted and the active region covered the chip entirely in

Fig. 9.20 Influence of chip size on LEE for simplified GaN-on-GaN and PSS chip simulations that have no top contacts and for which the active region covers the entire top area of the chip. The simulations are performed using an ambient epoxy environment
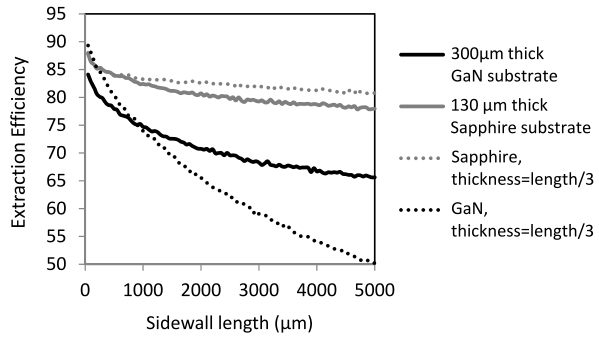
these simulations. It is also evident from the large chip limits that the GaN substrate chips are the most dependent on sidewall extraction.

9.6.4 Flip Chip LEDs

While standard LEDs have historically been top *p* layer emitters that use current spreading layers and back side substrate texturing, the flip chip design is a more recent commercialized design with several benefits.

In flip chip designs, the backside mirror covers the whole *p*-side of the chip and serves as the *p*-contact, enabling excellent current spreading. The sapphire substrate is removed and the exposed *n* doped GaN surface is roughened [9]. The resulting extraction mechanism is for the light to bounce several times between the backside mirror/contact and the GaN surface until it is extracted. The two benefits associated with surface roughness are therefore fully utilized, since the high single path extraction probability leads to better extraction at each attempt. Every extraction attempt comes with its associated losses, such as on the mirror or in the active region of the device. A simplified structure without top electrode shown in Fig. 9.21 simulated through ray tracing gives the following loss sources (an *n* side electrode would only introduce a small loss, see Table 9.2).
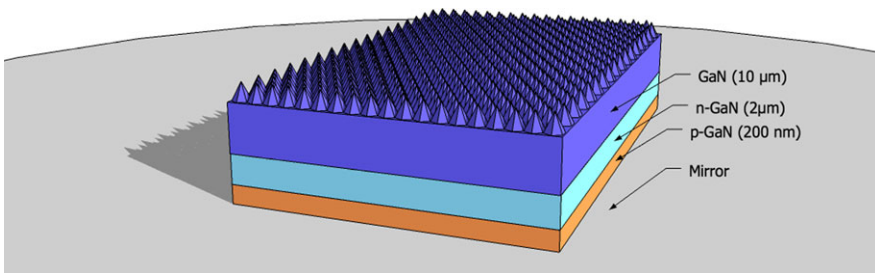
Fig. 9.21 Schematic of a flip chip used in simulations (not to scale, chip is approximately infinite in lateral dimensions)

Table 9.4 Extraction efficiency for simple flip chip simulation and breakdown of losses in material layers

Encapsulant	Epoxy	Air
Propagated	78.7 %	58.0 %
Mirror losses	16.8 %	34.2 %
p-Gan	1.9 %	3.2 %
n-Gan	1.4 %	2.4 %
GaN	1.2 %	2.0 %

Fig. 9.22 Influence of mirror reflectivity on LEE for a flip chip LED in epoxy (the *dot* indicates the reference chip at 92 %)
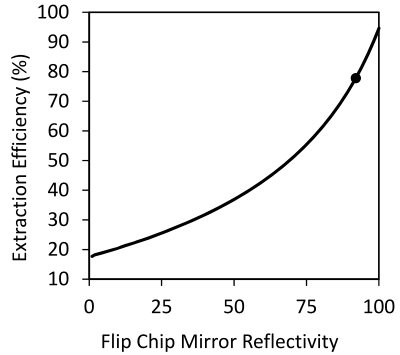
The mirror's non-perfect reflectivity is responsible for the bulk of the losses, and is the crucial element for this configuration, since a ray of light will, on average, need approximately six bounces before being extracted. Even a slight absorption in the mirror will lead to large losses, as seen in Fig. 9.22. Losses in the active region each time a photon passes through will play a similar role, but will only dominate as a loss mechanism in the event that the loss per pass is higher than the loss per interaction with the mirror. In the case of the modern materials properties and chip geometry studied here, the mirror is a much greater source of loss than the active region. The flip chip structure is the one most sensitive to mirror losses, and high performance can only be obtained, besides using highest possible reflecting contacts, by reducing the surface in contact with the mirror.

9.6.5 Index Effects

It is also of interest to compare the effect of the encapsulant index on all of the LED case studies discussed in this chapter. As shown in Fig. 9.23, the GaN-based chips, which experience greater internal reflection due to their high refractive index ($n = 2.5$), will see a much greater enhancement in their light extraction as the refractive index of the transparent encapsulant is increased.

Fig. 9.23 Extraction
efficiencies with increasing
ambient refractive index for
four chip simulations (the *dot*
indicates the reference chip
at 1.5)
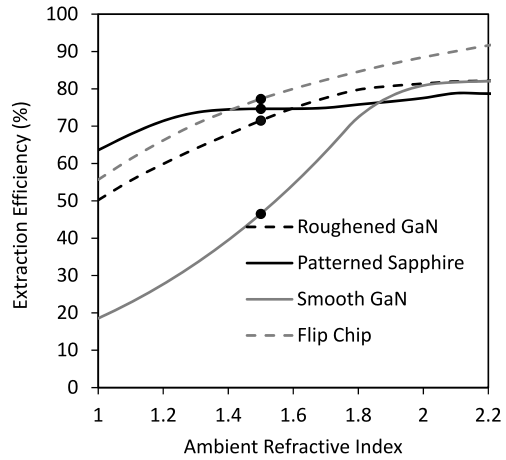
High index encapsulation may allow GaN substrate chips to compete with PSS chips for high light extraction efficiency. Use of native GaN substrates for nitride LEDs is well known to improve other aspects of LED optical performance due to the low dislocation densities and high thermal conductivity relative to sapphire substrates. Thus, the use of encapsulation and other techniques may allow GaN substrates to achieve high extraction efficiency while capturing the benefits of GaN substrates for high IQE. Additionally, at high refractive indices, the distinction between roughened and non-roughened GaN vanishes, so there would be no need to roughen the device if sufficiently high index, transparent encapsulants are used.

The extraction efficiencies in these various chips all exceed 80 % at a relatively low index of $n = 1.8$–1.9. Refractive indices of this magnitude are readily available in commercial glasses and some epoxies and polymers. In practice, however, most of these materials are somewhat absorbing, so there will need to be a balance of the index, absorption coefficient, and thickness of the encapsulant to maximize light extraction.

9.7 Conclusions

This chapter on ray tracing simulations of LEDs shows how the various extraction schemes operate. Besides the simple use of a high index, dome-shaped encapsulant, the use of index matched transparent substrates, such as GaN substrates, is very efficient, provided the substrate is very transparent, which reduces absorption resulting from long propagation path lengths in the substrate.

The most advanced techniques today yield extraction efficiencies in the 80 % range, and rely mostly on texturing of the LED surfaces (surface roughening or surface photonic crystals) or interfaces (patterned sapphire substrate or embedded photonic crystal). Such values are indeed obtained in industry [14, 16, 23].

Surface roughening and PSS operate differently: the former relies on both ray randomization and increased single pass extraction while the latter only relies on ray randomization. Surface roughening is not well adapted to p-side up LEDs on sapphire substrates because it is difficult to etch the p materials, and it requires a current spreading layer on top of the p-GaN, usually ITO or a semi transparent metal because of the poor p conductivity. Attempts have been made to structure the ITO spreading layer, by roughening or etching a PhC but the index difference between the GaN and ITO limits the efficiency of that solution because a large fraction of light will be guided in the GaN layer and will only weakly interact with the surface texturing [10]. One can both etch the ITO and the p layer, but this comes at the expense of some degradation of the p contact [10]. The high performance surface roughened LEDs are therefore obtained either using a p-side up LED on GaN substrate with back roughened n-type surface or in flip chip geometries using a substrate lift off.

Although based on somewhat different mechanisms, the simulations of this chapter show that both surface roughening and PSS lead to high efficiencies (Tables 9.2, 9.3, 9.4). The results appear here for the same device parameters (most of them with conservative values), which allow meaningful comparisons. Some industry results may be higher, either due to better materials quality (lower absorption in particular for ITO and metals) or more aggressive designs (smaller lossy contact areas).

A desirable feature is the ability to scale the extraction efficiency for large chips in order to reduce costs. In almost all cases, the use of large surface LEDs is detrimental due to the long ray paths to reach the sidewalls (see Fig. 9.20) except for those extraction schemes where light is continuously extracted or when substrates have very low absorption.

Appendix A: Discussion of the Origins of the Effects of Surface Roughness and of Sapphire Patterning

A.1 The Effect of GaN Surface Roughening: Randomization

We discussed above the increased single pass extraction due to surface roughness. A second essential effect is the randomization of light. Even if a ray is not extracted on first try, it will be reflected at a random angle, and may escape on the next attempt. We can quantify this using ray tracing by considering the fate of a random collection of rays and their angular distribution for each incident angle. Figure 9.24 shows the
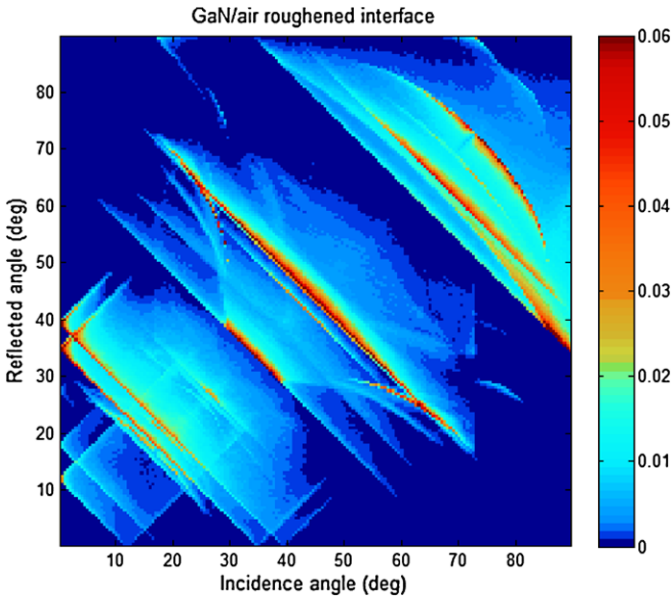
Fig. 9.24 Angular distribution of reflected rays as a function of incidence angle for a roughened surface (hexagonal non-random pyramids, with a top angle of 28°, for a GaN/air interface) (Color figure online)
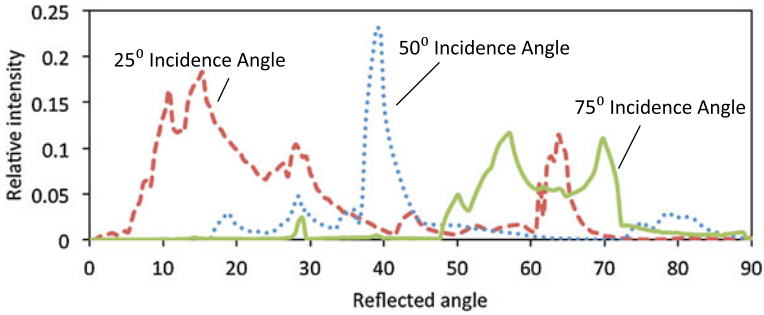
Fig. 9.25 Example of reflected light angular distribution pattern on a roughened surface (hexagonal non-random pyramids, with a top angle of 28°) for 25°, 50° and 75° incidence angles

angular distribution of reflected rays as a function of incidence angle and Fig. 9.25 gives some examples of reflected rays distributions for a given incidence angle.

It is shown that for a given angle of incidence from an isotropic source, the reflected light pattern is complicated and very different from a specular reflection pattern. The angular distribution of reflected light is obtained by summing over all incidence angles (Fig. 9.26).

These reflected rays will hit the header and will be specularly reflected back to the roughened surface for a second attempt. As we know their angular distribution,

(a)

(b)
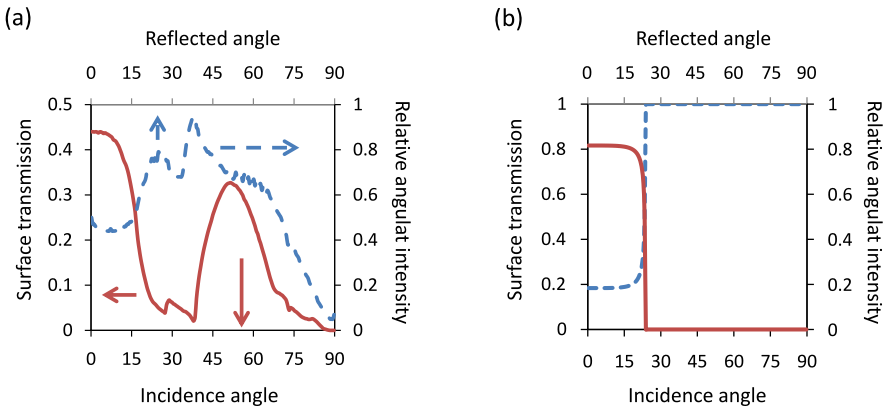
Fig. 9.26 (**a**) Surface transmission as a function of incident angle (*solid*) and angular distribution of the reflected rays integrated on all incidence angles (*dashed*) from a rough surface. (**b**) Similar results for a flat GaN/air surface

Fig. 9.27 Successive *n*th attempt extraction probability for a roughened (*diamonds*) and flat surface (*circles*)
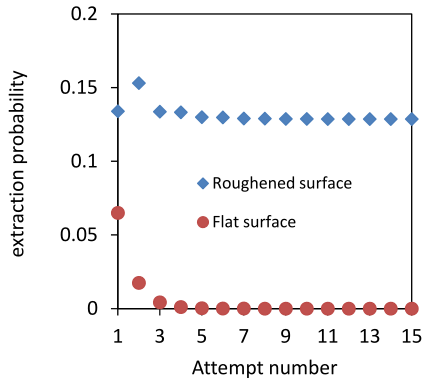
the second attempt extraction can be computed using the transmission coefficient as a function of the incidence angle of Fig. 9.26a. It is 15.3 %, slightly more than the first attempt probability (13.4 %). We can iterate this process and compute the *n*th attempt extraction probability (Fig. 9.27). Due to randomization of the rays, nearly the same fraction of the light is extracted after each attempt (around 12.9 % after a few bounces). These values were computed for a GaN/air interface; adding a high index encapsulating epoxy would significantly increase them.

In contrast, a flat surface with no randomization effect exhibits specular reflection and the second attempt rays are mostly outside of the transmission window of the surface. Thus, the extraction probability drops to zero after two to three successive attempts (Fig. 9.27).
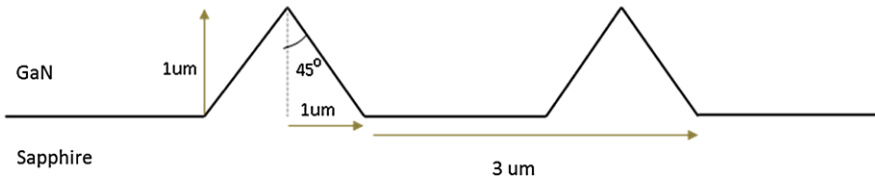
Fig. 9.28 Cone patterning of the GaN/sapphire interface
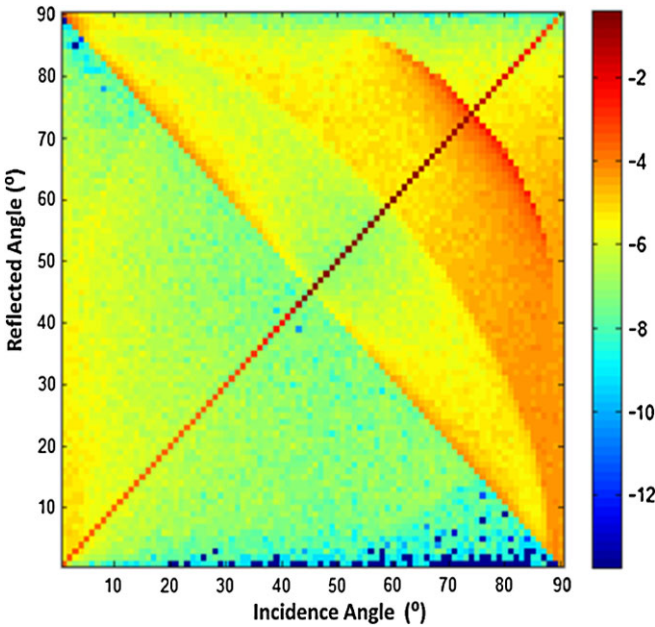
Fig. 9.29 Angular distribution of reflected rays as a function of incidence angle (intensity is on a log scale) (Color figure online)

A.2 The Effect of Patterned Sapphire Substrate (PSS)

We can also study PSS interface the same way. We will consider a cone shaped pattern in a square packing arrangement. The exact geometry is given in Fig. 9.28. This results in a filling factor of 34.9 %.

We can quantify the intensity of specular reflection, randomized reflection and transmission as a function of incidence angle by averaging over the reflected and transmitted solid angles (Fig. 9.30).

The angular reflection pattern of the PSS is given in Fig. 9.29. We can see that, compared to the GaN surface, the pattern is overall simpler because of the cylindrical symmetry of the cone compared to the hexagonal pyramid. We can also see that there is high intensity along the main diagonal; this is specular light that is re-

Averaged values :
-Specular reflection 30.0%
-Random reflection 42.1%
-Transmission 27.9%
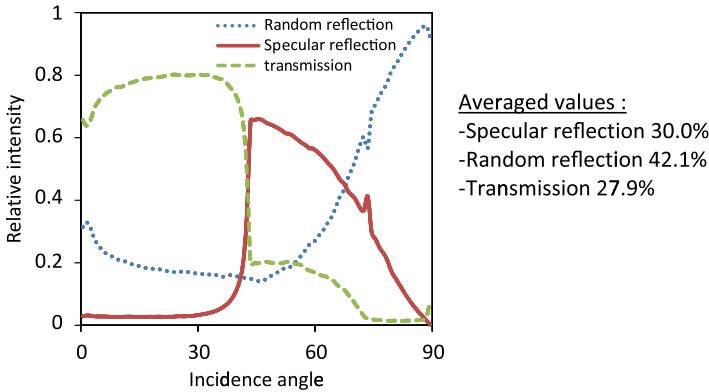
Fig. 9.30 Fate of the incident rays for a PSS surface

flected exactly at the incidence angle. It comes from the fact that contrary to GaN roughening, the filling factor is not 100 %; some of the interface is still flat.

Figure 9.30 is the result of the superposition of two effects: the specular reflection by the flat part of the GaN-sapphire interface and the randomization by the cones. Below the critical angle (42.6°), the cones transmit part of the light and reflect the rest at random angles. The flat surface transmits all of the light that strikes it (modulo the Fresnel transmission coefficient). For higher angles, the flat surface reflects all the rays striking it, and because the filling factor is ∼35 %, this explains the 65 % specular reflection for incidence angle around 43° (Fig. 9.31a). However, for angles larger than 45°, some of the specularly reflected light can immediately hit another cone and be randomized (Fig. 9.31b). This explains the decrease in specular reflection. This also corresponds to the "shading" of the flat surface by the cones (Fig. 9.31c–d).

The purpose of patterning the sapphire differs subtly from that of roughening the surface of GaN. The GaN roughening is used both for its randomization effect and improvement of first pass extraction. However, in the case of GaN/sapphire patterned interface, the situation is different: extracting the rays from GaN to the sapphire substrate is not the final purpose of the texture. Like for the GaN on GaN chip with roughened backside, it is the randomization effect that is the most beneficial.
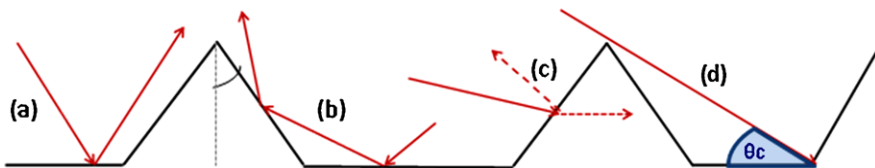
Fig. 9.31 (**a**) 44° specularly reflected ray, (**b**) high angle ray undergoing specular and random reflection, (**c**) high angle ray that is directly randomized by a pattern that "shades" the flat surface, and (**d**) a ray impinging at the critical angle beyond which there is no specular reflection

Fig. 9.32 First attempt
extraction rays: (**a**) reflected
at low angles in GaN and
(**b**) transmitted at low angles
in sapphire
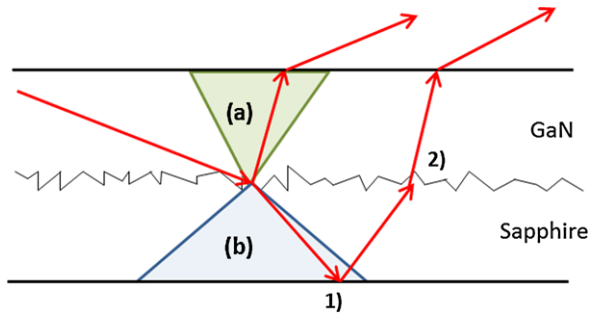
Indeed, getting the light into the sapphire has two opposite effects: due to the sapphire's lower index and absorption, sidewall extraction is facilitated, but on the other hand, light in the substrate will most likely hit the lossy mirror. So, contrary to flip chip roughening, we cannot simply compute the *n*th pass extraction in sapphire to determine the efficacy of the structure. We have to study in more detail the extraction process.

As seen in Fig. 9.32, the PSS interface will extract a downwards traveling ray in two ways: (a) it can reflect the ray into the GaN extraction cone or (b) it can refract the ray into the sapphire extraction cone, which leads to the ray being extracted after it bounces back on the LED rear reflector (Fig. 9.32(1)). For (a), rays reflected into GaN must be below 23.6° (GaN/air TIR angle), and for (b), transmitted rays must be in the sapphire/air extraction cone (36.0°).

The fraction of incident rays reflected into the GaN light cone will be referred to as the "GaN efficacy". The smaller number of rays that are transmitted into the sapphire light cone can be extracted, too. For a simplified first pass calculation, we will not take into account the effect of the second randomization of rays going back through the sapphire/GaN interface (Fig. 9.32(2)). The "sapphire efficacy" is then defined as the fraction of rays inside the sapphire light cone (Fig. 9.32b). The total first pass efficacy is then defined as the sum of the GaN efficacy and of the sapphire efficacy. It estimates the fraction of the incident rays that will be extracted on first attempt via scattering by the patterned interface into the air cone directly into GaN, or through the sapphire and back through the GaN into the air cone.

One should note that this model does not take into account the reflectivity of the GaN/air interface inside the light cone (some of the low angle incident rays will still be reflected according to Fresnel laws), the losses on the bottom mirror (taken as zero), or the effect of the sidewall extraction. However, all these effects are mostly independent of the pattern choice, so the total first pass efficacy is a good relative indicator to compare different patterns.

To find the optimal pattern to achieve maximal first pass efficacy, we first find the optimal shape, comparing a flat surface, half-sphere and cone. The cone pattern is still according to Fig. 9.28, and the half sphere has the same filling factor as the cone (34.9 %). The total efficacy, for different incidence angles is given in Fig. 9.33.

We can compute the average efficacy over all incidence angles for the three surfaces. We find 8.6 % for the flat surface, 13.0 % for the half-sphere and 14.8 % for

Fig. 9.33 Comparison of total efficacy (sapphire efficacy + GaN efficacy) for different pattern geometries (cone, half-sphere with 34.9 % filling factor, and flat surface)
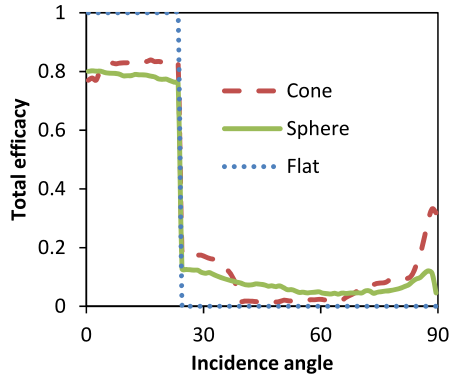
the cone pattern. The higher performance of the cone can be explained by the high efficacy at high angles.

In more detail, we can see that this high angle efficacy comes from GaN efficacy: the cone pattern reflects some of the light incident at high angles into GaN extraction cone (Fig. 9.34a). Qualitatively, we can understand it from simple ray tracing. (Fig. 9.34b).

Once the cone pattern was identified as the optimal pattern, simulations were performed to determine the cone apex angle that maximizes the efficacy. This was done by varying the height of the cones (while keeping the fill factor and all other parameters the same) from 0.15 μm to 7 μm. These cone heights correspond to cone apex angles from 85.7° to 17.7°. For each geometry, the average low angle transmission and reflection were computed. The results are given in Fig. 9.35.

We can see that the fraction of light transmitted to the sapphire extraction cone does not change much with the cone shape. On the contrary, the reflection in GaN is very sensitive to the cone angle. It reaches a maximum for a cone apex of 66°.
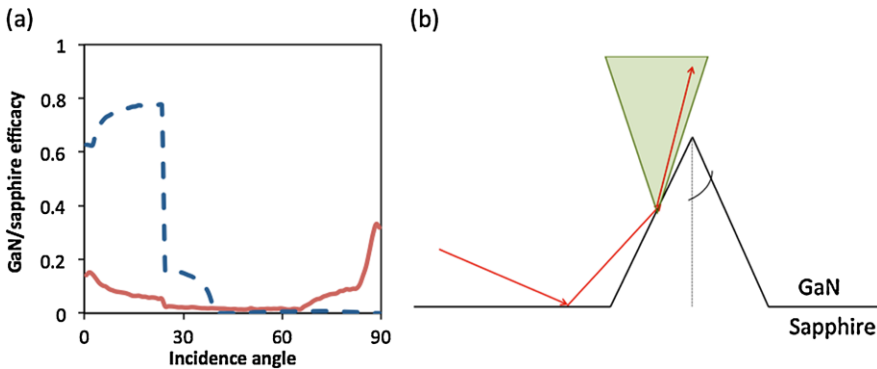
Fig. 9.34 (**a**) Breakdown of the efficacy of the cone pattern between the sapphire efficacy (*dashed line*) and GaN efficacy (*solid line*), (**b**) Simple 2D model of the high angle GaN efficacy

Fig. 9.35 Effect of cone apex angle on PSS single pass efficacy
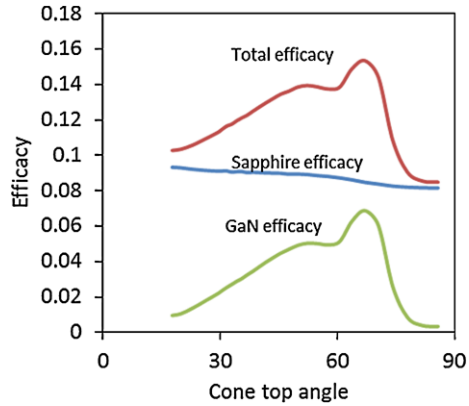
Fig. 9.36 Comparing the GaN efficacy for different cone geometries
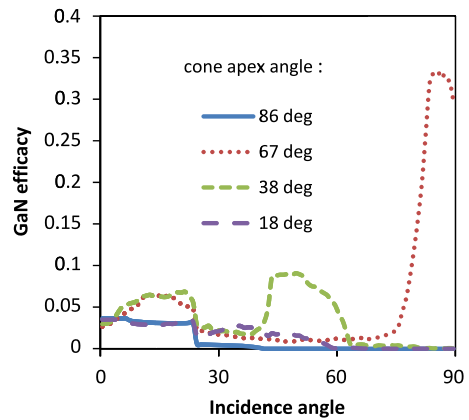
We can understand why if we compare the GaN efficacy as a function of incidence angle for four geometries (Fig. 9.36).

For very flat cones (high apex angle), the result is close to that of a flat surface, with little reflection at low angles and diminishing to nearly zero for angles beyond the GaN/air critical angle (23.6°). An angle of 67° is approximately optimal, with good reflection at high angles (as explained in Fig. 9.34b). For more peaked cones, we lose this high angle extraction, but this is partly balanced by an increase in reflection for angles around 40°–60°. Then for an even more peaked cone, this reflection diminishes.

Finally, the optimal filling factor for these 66° cones was determined by varying the distances between cones (2 μm to 20 μm) in the simulations (Fig. 9.37). As the patterns are in square packing, this results in filling factors from 3.1 % to 79 %. We can see that the part of light reflected in GaN increases with the filling factor. The light extracted in sapphire is almost constant (8.2–8.5 %).

These results can qualitatively predict the LEE, for example increasing the filling factor leads to better extraction [34]. However, both the full chip modeling and the experimental results show that the incremental improvement in LEE with increasing

Fig. 9.37 Effect of varying the filling factor for a given cone shape on the pattern first pass efficacy. To obtain an even better filling factor, a hexagonal packing can be used, resulting in about 91 % filling factor. For this filling factor, the resulting efficacy is 18.1 % compared to just 8.4 % for a flat GaN/sapphire interface
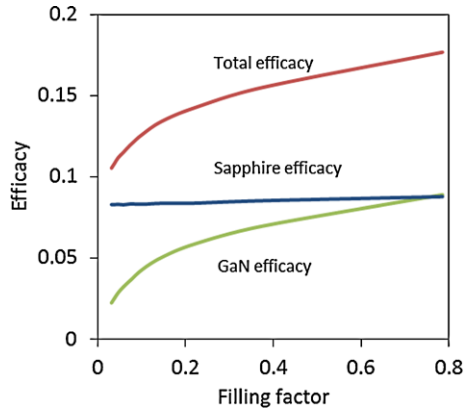
filling factor tends to diminish for high filling factors. This is because only the first pass efficacy was computed. In reality, if the light is not extracted on first attempt, it can reach the patterned interface again, where it would be more likely to be extracted after fewer attempts for a high filling factor.

Appendix B: Comparison of Roughening and Angled Sidewalls for GaN Substrate LEDs

A comparison was done to demonstrate the hierarchy of techniques of roughening, out of plane angled sidewalls (truncated inverted pyramid structure) and high refractive index encapsulation. An angle of 15 degrees was used for the sidewalls (Fig. 9.38). The most effective extraction techniques have high extraction values at

Fig. 9.38 Comparison of roughening and chip shaping techniques for light extraction
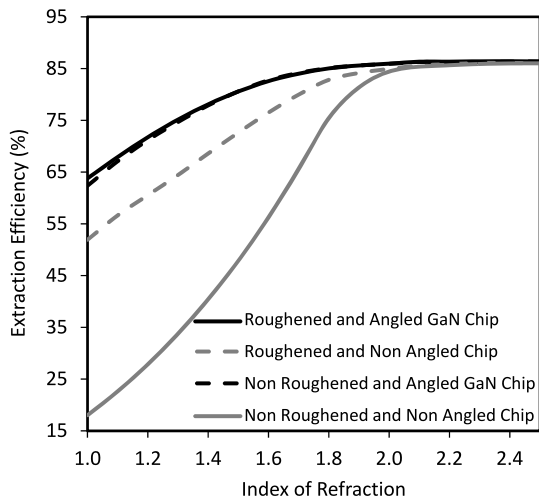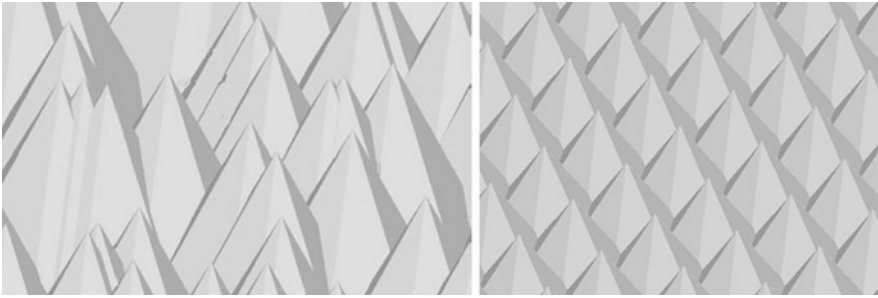
Fig. 9.39 Random simulated pyramids (*left*) and evenly spaced pyramids (*right*)

Fig. 9.40 Transmission function of GaN/air interface with random or periodic rough patterns
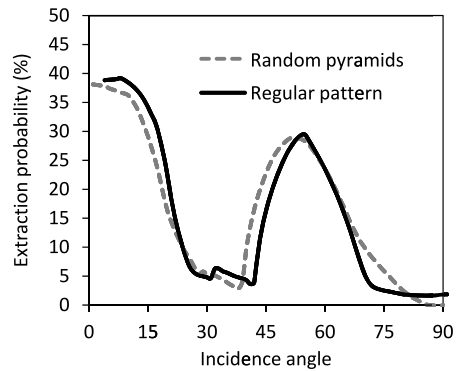
low index that saturate earlier with increasing index. The combination of roughening and the sidewall angling gives the highest extraction for low index encapsulants and the benefit of increasing encapsulant index saturates at $\sim n = 1.8$.

The advantage of combining both roughening and sidewall angling over sidewall angling alone, however, is small, differing at most by less than 2 %. Roughening without sidewall angling follows after that, indicating that at least for the chip dimensions modeled here, sidewall angling into a truncated pyramid structure is a more effective extraction method. It should be noted, however, that for larger chip sizes, backside roughening may gain an advantage over chip shaping. This is because the improvement from angled sidewalls diminishes with increasing chip area as the rays must travel further through the device to reach a wall where they may be extracted or redirected.

Appendix C: Simulations of Periodic "Rough" Surfaces vs. Random Rough Surfaces

The first-pass extraction probability can be quantified by analyzing the intensity of light emission through a roughened surface as a function of incidence angle. In

real systems, the surface texturing is not of constant size or regularly distributed on the surface, as for GaN roughening from a chemical etching. Generating random patterns in ray tracing considerably increases the complexity of the simulations, so a simpler model was considered, which uses a periodic pattern rather than a random one (Fig. 9.39). For comparison, the transmission function was computed in the two cases of random and periodic surfaces (Fig. 9.40). Little difference was found between the two, not surprisingly as ray tracing is a geometric optics approach, so the size of the pyramids does not affect the results. Randomizing the position of the pyramids can only modify the trajectory of those already extracted rays that hit a second pyramid, which is very rare. Therefore, the periodic pattern is used to simplify the simulations without losing accuracy compared to the actual random patterns.

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Chapter 10
Light Extraction Efficiency Part B. Light Extraction of High Efficient LEDs

Ja-Yeon Kim, Tak Jeong, Sang Hern Lee, Hwa Sub Oh, Hyung Jo Park, Sang-Mook Kim, and Jong Hyeob Baek

Abstract This chapter deals with methods for fabricating high efficient light-emitting diodes (LEDs) with higher light extraction efficiency (LEE). Some LED prototypes are then reviewed to investigate how their performance was enhanced by utilizing a variety of chip processes. The most efficient devices were found to be produced with unique fabrication processes, having at least one patterned sapphire substrate, chip shaping, vertical configuration, and/or flip-chip packaging, among others.

Keywords Light extraction efficiency · Vertical chip · Flip-chip · PSS · Chip shaping · Texturing · DBR

10.1 Enhanced Light Extraction for GaN LEDs Using Surface Shaping

In order to improve the external quantum efficiency (EQE) in LED structures, it is essential to enhance both the light extraction efficiency (LEE) as well as the internal quantum efficiency (IQE) [1, 2]. In particular, improving the LEE is considered to be crucial for fabricating high-efficiency LEDs. To date, methods such as chip-shaping [3–9], surface texturing [10–13], flip-chip packaging [14–16], use of a patterned sapphire substrate (PSS) [17–24], and use of photonic crystals [25–29] have been proposed in attempts to enhance the light extraction.

The descriptions of some of these proposals are outlined as follows. Kao et al. [6] reported that ∼22° etching undercut sidewalls in a nitride-based LED (LED II) could be achieved using controllable inductively coupled plasma etching, which was comparable to results obtained using a conventional LED (LED I). Note that an undercut angle could be definitely observed when the ∼22° oblique sidewall profile was compared to the initial unetched mesa (Fig. 10.1(a)). In the figure, the angle is very close to the critical angle (∼23°) between GaN and air. It was also found that

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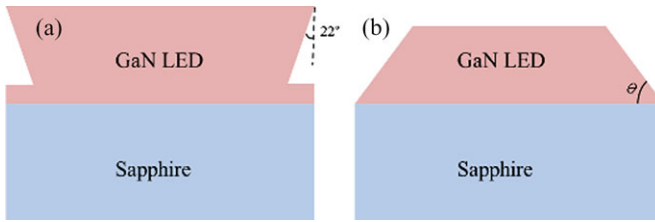
Fig. 10.1 (**a**) Schematic of the fabricated LED device having a 22° undercut angle and (**b**) SDI-LED structure

the current-voltage (I–V) curves shown at around 3.5 V at 20 mA for an LED with undercut sidewalls were almost identical to the LED with normal sidewalls. This observation indicates that undercutting the sidewalls should not result in any degradation in the electrical properties of nitride-based LEDs. In the electroluminescence (EL) measurement, the current was continuously injected into the device at room temperature.

Lee et al. [7] also reported LEDs with modified mesa sidewalls, referred to as sidewall-deflector-integrated (SDI)-LEDs, in which smooth mesa sidewalls were prepared at an oblique angle (Fig. 10.1b) all over the GaN epilayer; in this case, the air-GaN-sapphire waveguide structure terminates at the angle between the GaN LED and sapphire substrate.

Kim et al. [8, 9] investigated the characteristics of a triangular (TRA) LED and compared it to a standard quadrangular (QDA) LED having the same surface area and contact area. The range of the incident angle for the total internal reflection (TIR) for photons kept inside the QDA LED is given by

$$23.5° < \Theta_{i\,(\text{QDA})} < 66.5°. \tag{10.1}$$

For an equilateral TRA LED, the range of incident angles for TIR inside the LED can be estimated by

$$23.5° < \Theta_{i\,(\text{TRA})} < 36.5°. \tag{10.2}$$

As such, this simple calculation shows that the range of the incident angle for TIR of the TRA LED is much smaller than that of the QDA LED. Therefore, the LEE of the TRA LED can be greatly enhanced when using the increased probability of the emitting photon from the LED chip compared to that of the QDA LED. These considerations indicate that the light extraction from the packaged TRA LED is substantially enhanced compared to that of the packaged QDA LED.

This theoretical analysis can be proven by comparing properties of the TRA LED to the standard QDA LED. Here, the optical output power of the TRA LED is higher than that of the QDA LED over the entire input current range. The total radiant fluxes of the QDA and TRA LEDs measured in the integrating sphere were 8.9 mW and 13.2 mW at an input currents of 20 mA and 38 mW and 47.2 mW at 100 mW, respectively [8]. The enhancement of the total radiant flux of the TRA LED compared to that of the QDA LED was 48 % at 20 mA and 24 % at 100 mA. These
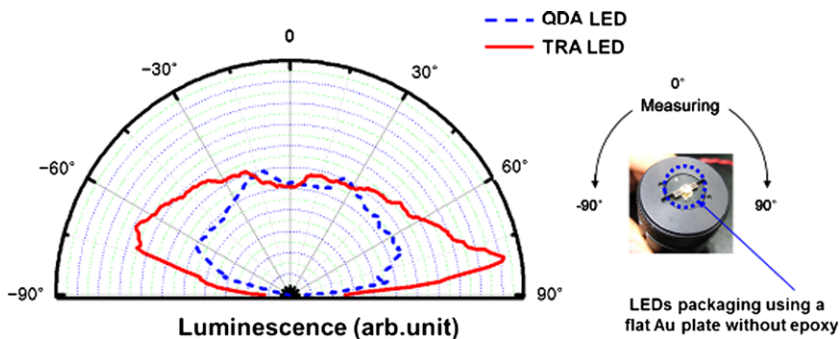
Fig. 10.2 Angle-dependent scan of the far-field beam distribution of the TRA and QDA LED at a current injection of 20 mA (*left*). Photo shows the angle dependent measurement of the emitted light (*right*). Reprinted with permission from Ref. [8]. Copyright [2007], IEEE
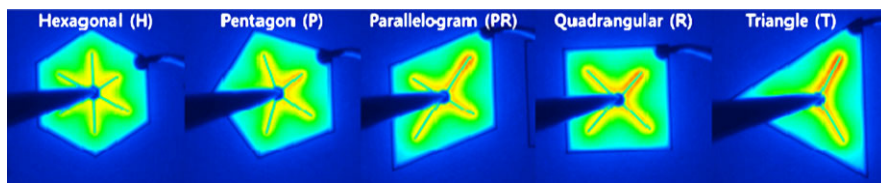
Fig. 10.3   EL intensity of polygonal LEDs at an injection current of 120 mA

improvements in the optical power enhancement of the TRA LED can be attributed to a significant reduction in TIR.

Figure 10.2 shows the angular dependence of the light far-field beam distribution of TRA and QDA LEDs at an injection current of 20 mA. In the figure, it can be clearly seen that the emission intensities of QDA and TRA LEDs after packaging were almost identical in the vertical direction, but the emission intensity of the TRA LED in the horizontal direction was greater than that of the QDA LED. This enhancement of the light emission from the TRA LED in the horizontal direction can be attributed to the significant reduction of TIR at the sidewalls of the TRA LED. According to this study, the light extraction efficiency (LEE) can be greatly enhanced in the TRA LED, with increasing the probability of photon escape, by significantly reducing TIR at the sidewalls of the TRA LED, compared to the QDA LED.

Kim et al. [9] compared QDA LEDs to various polygonal LEDs in an attempt to determine an optimum structure for improving the extraction and current injection efficiencies. Figure 10.3 presents the EL intensity mapping images of LEDs at an input current injection of 120 mA. The hexagonal LED shows good current spreading characteristics, compared to the pentagon, parallelogram, quadrangular, and triangular LED shapes. To measure their electrical and optical properties, LEDs were
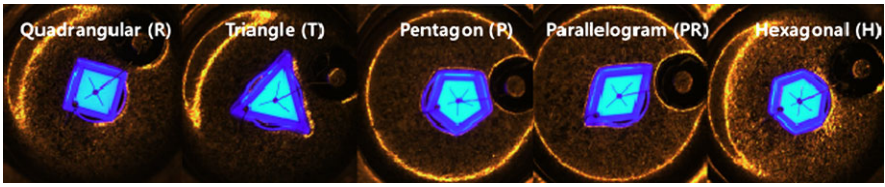
Fig. 10.4 Emission images of scribed polygonal LEDs with TO-can package at an injection current of 120 mA
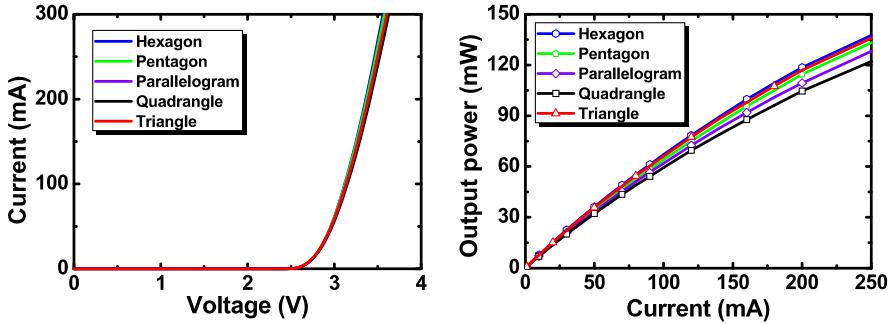
Fig. 10.5 (**a**) $I-V$ and (**b**) $L-I$ curves of polygonal LEDs over the entire input current range

packaged onto a TO-can after scribing with a diamond tip; the total radiant flux was then measured using an integrating sphere measurement system (Fig. 10.4). This study revealed similar $I-V$ characteristics, though the polygonal LED showed an increase in the total radiant flux, compared to that of the quadrangular LED. Specifically, the hexagonal LED had the highest value, resulting from an increase in its extraction and injection efficiency (Fig. 10.5). These results indicate that the fabrication of LEDs with polygonal shapes is a very promising way to develop high-efficiency LEDs.

10.2 Textured Surfaces for High Extraction

Roughening the top surface of an LED is one method for improving its light extraction [10–13], as the roughened top surface reduces the internal light reflection and scatters the light outward. Huh et al. [10] reported the fabrication of an InGaN-based LED with a top p-GaN surface microroughened using metal clusters as a wet etching mask. The light-output power of LED chips having microroughening was increased compared to LED chips with no roughening; by employing this structure, the power conversion efficiency was increased. Compared to conventional LED chips, the light-output power for the LED chip with a microroughened top surface increased by 52.4 % for the front side and 30 % for the backside.

Fig. 10.6 Scanning electron microscopy (SEM) images of top *p*-GaN surfaces of LEDs etched using KE solution
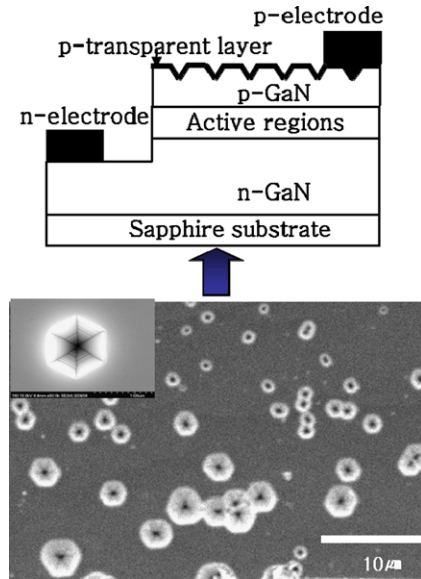
In another study, Na et al. [11] reported the selective wet etching of a *p*-GaN layer by using a solution of KOH in ethylene glycol (KE), in an attempt to enhance the optical and electrical performances of GaN-based LEDs. The selectively etched *p*-GaN surface showed hexagonal-shaped etch pits (Fig. 10.6), and the light-output power of the LEDs was improved by 29.4 % compared to that of non-etched LEDs.

10.3 Patterned Sapphire Substrate

LEDs grown on a patterned sapphire substrate (PSS) have also shown higher performances than those grown on an *a*-plane sapphire substrate. Figure 10.7 shows typical SEM images of PSS(a) and the GaN epilayer grown on the PSS(b). Several groups have demonstrated the direct epitaxial growth onto various PSS shapes, in attempts to enhance both the epitaxial quality and LEE. For instance, Tadatomo et al. [17, 18] adopted models of parallel grooves along the (11–20) sapphire to fabricate nitride LEDs, Bell et al. [19] used the trenches of sapphire (11–20) direction to growth Mg-doped AlGaN epilayers, and a hexagon pattern parallel with the axis of the sapphire was reported by Yamada et al. [20]. In addition, Chang et al. [21] described the use of parallel stripes along the sapphire (100) direction for blue LED growth [17–24], and Wuu and Wang et al. [22, 23] proposed an approach for growing a high-quality GaN film using a PSS; this technique eliminated dislocations and increased the probability of light extraction.

Another type of PSS was prepared using periodic hole patterns, in which the etching depths ranged from 0.5 μm to 1.5 μm [23]. For the PSS LED, the GaN epilayer grew laterally from the top of the sapphire substrate and covered the holes.
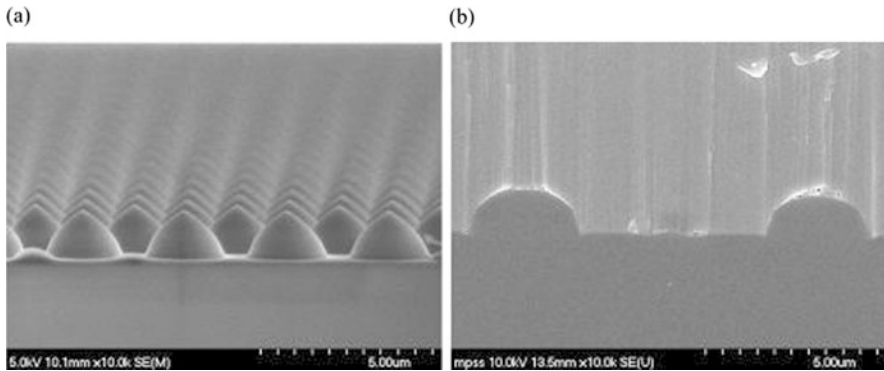
Fig. 10.7 (**a**) Tilted SEM images of PSS and (**b**) cross-sectional SEM images of GaN epilayer grown on a PSS

The enhanced output power (∼21 %) can be attributed to the combination of improved LEE and the crystalline quality by the reduction of threading dislocations (TDs) density with the PSS structure. Lee et al. [24] reported a chemical wet-etched PSS, used to eliminate the sapphire surface damage induced by dry etching. The crystallography-etched pattern of a (0001)-oriented sapphire substrate has a flat-surface {0001} C-plane that is triangle-shaped in the center. Surrounding this plane are three facets: an {1–102} R-plane with an angle of 57° against the [0001] C-axis; this inclined facet with a steep slope is useful for light extraction. It was also found that the sapphire etching rate depends on the crystal orientation and decreases in the order of C-plane > R-plane > M-plane > A-plane. They reported that the output power of the PSS LED increased by approximately 15 % relative to conventional LEDs at an injection current of 20 mA. This improvement was attributed to not only the geometrical shape of the {1–102} crystallography-etched facets that efficiently redirected the trapped light to escape, but also to the reduction of the dislocation density by adopting the PSS growth scheme.

Recently, nanosized PSSs (NPSSs) were fabricated using nanotechnologies such as laser holographic lithography (Fig. 10.8), e-beam lithography, and nanoimprint technology, to enhance the LEE of LEDs. Relaxation of the residual strain in GaN layers on the PSS was found to reduce the resultant piezoelectric field in the quantum wells (QWs) [30].

10.4 High-Power Vertical-Type LEDs

Sapphires are one of the most commonly used materials for substrates in GaN-based LEDs. However, the relatively poor thermal conductivity of the sapphire substrate has been recognized as a major limitation to its further application in high-brightness LEDs. In addition, sapphire substrates impose constraints on attempts to improve the quality of GaN films due to the lattice and thermal-expansion coefficient mismatch between these two materials. To resolve these problems, an LED structure
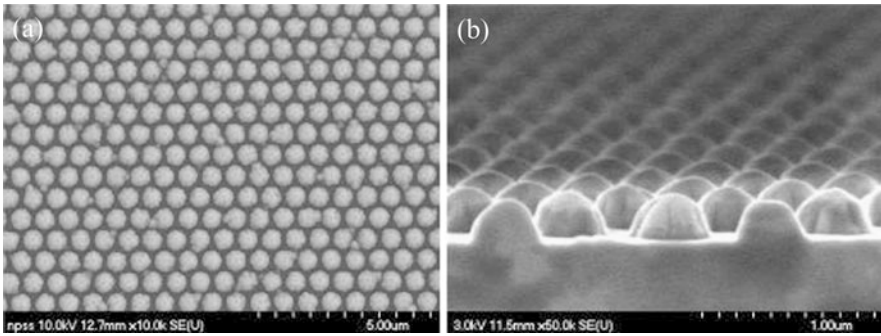
Fig. 10.8 (**a**) Top and (**b**) tilted SEM images of NPSSs fabricated using nanoimprint technology
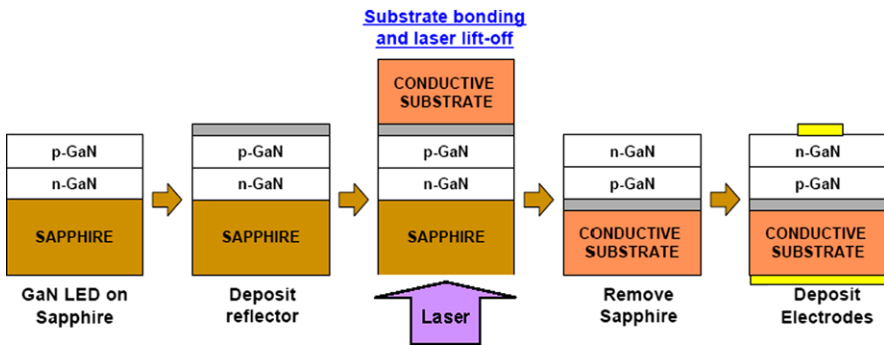
Fig. 10.9 Schematic of vertical-type GaN LED device fabrication process

with GaN epilayers can be transferred to another substrate with good thermal conductivity using wafer bonding and laser lift-off (LLO) techniques [31–33]. LLO processes were first demonstrated by Kelly et al. [34]; these types of LED structures have been extensively investigated [35–39] since. Compared to the conventional lateral structures, these vertical-type LEDs (VLEDs) have advantages such as better current injection, good heat dissipation (due to wafer bonding or electroplating with highly conductive materials), and enhanced reliability against electrostatic discharge. For example, Huang et al. [40] used high-power GaN-based VLEDs to improve light extraction and heat dissipation. They showed that VLEDs having a textured surface produced a 65 % higher extraction efficiency compared to conventional GaN-based LEDs at an injection current of 20 mA. Indeed, compared to conventional LEDs, VLEDs are known to have better performance, in particular, at high current injection due to their excellent heat dissipation capability. Figure 10.9 presents the schematic for vertical-type GaN LED device fabrication processes using wafer-bonding technology. First, metal reflectors aredeposited on the top side of the epitaxial wafer using an electron-beam evaporator or sputter system. Then, rapid thermal annealing (RTA) is performed to make an Ohmic contact. The process is normally performed within a temperature range of 400 °C to 650 °C in air or N_2

Fig. 10.10 Schematic cross-sectional image of the hexagonal pyramidal LED structure
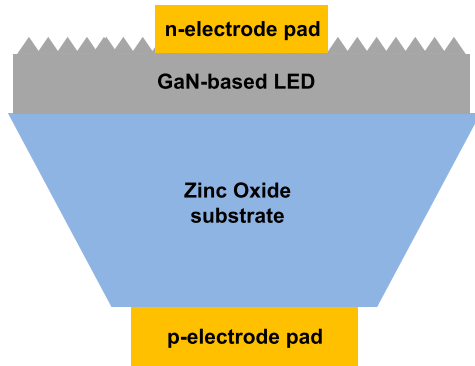
ambient conditions for 1 min to 5 min in order to minimize the thermal damage. Then, a bonding metal is deposited onto the reflector for the wafer bonding process. To facilitate a good bonding scheme for GaN, different approaches have been reported in literature [41–44]. In order to prevent diffusion of the bonding metal into the epitaxial layer, multiple barrier layers consisting of Ni, Ti, W, and Pt can be applied onto the reflectors prior to depositing the bonding metal. After the bonding process, the sapphire substrate is detached using a 248 nm-KrF excimer laser during the LLO process, and the undoped GaN layer is etched off using an inductively coupled plasma (ICP) etching system to expose the *n*-GaN layer. Finally, Cr/Au, or Ti/Al layers are deposited in sequence on the exposed *n*-GaN layer to serve as the *n*-type electrode.

10.5 Wafer Bonding and Electroplating Techniques

To fabricate GaN-based VLEDs, the formation of a receptor, diffusion barrier, and soldering (seeding) layer is critical. Recently, a great deal of effort has been made to fabricate a receptor using Si, Cu, Ni, etc., for wafer-bonding and electroplating techniques. For example, Peng et al. [45] reported the fabrication of VLEDs on a 50 mm diameter Si wafer using wafer-bonding and LLO technologies. The light output power of the VLEDs was 20 % higher than for conventional sapphire-based LEDs, and the VLED chip had a relatively smaller viewing angle. In addition, the performance of the vertical InGaN LEDs did not degrade during a 1000 h aging test. Uang et al. [46] used a thick electroplated Ni layer as the conductive supporter to realize large-area VLEDs using a similar LLO technique. They showed that the VLEDs (compared to conventional LEDs having the same size as the proposed VLEDs) produced a significantly improved operating voltage drop and light output power. Figure 10.10 shows the hexagonal pyramidal LEDs fabricated through a process including the wafer bonding of ZnO to GaN, LLO, and photochemical etching. ZnO contacts allow for the creation of a vertical current path; selective etching was then used to form truncated hexagonal ZnO pyramids. At a current of 20 mA,
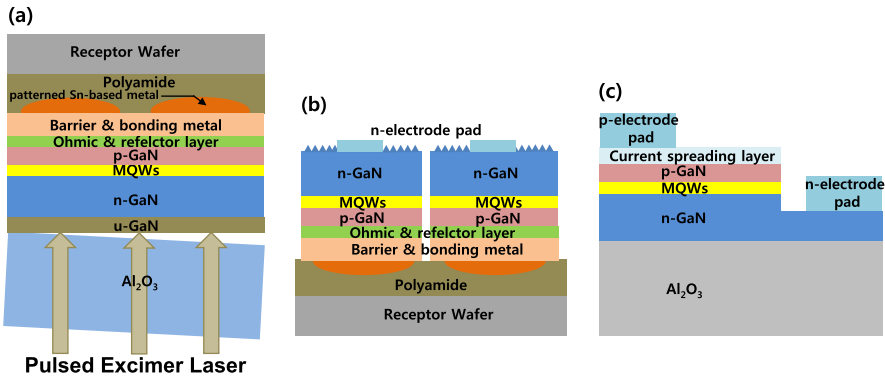
Fig. 10.11 Schematics of VM-LED processes using the Sn-based metal substrate technology: (**a**) Sn-based metal substrate bonding and laser lift-off process, (**b**) device isolation and formation of *n*-contact process, and (**c**) schematic of a conventional LED

the output power of the hexagonal pyramidal structure LED was 23.6 mW, with an EQE of 43.6 % [47].

Kuo et al. [48] also proposed a metal substrate technology, using Sn-based solder balls and a patterned LLO technique to fabricate vertical-structured metal substrate GaN-based light-emitting diodes (VM-LEDs). Figure 10.11 schematically illustrates the layer structure and key fabrication processes of VM-LEDs using the proposed metal substrate engineering process. As compared to regular LEDs, the proposed VM-LEDs improve the optical power by about 145.36 %, with a voltage drop of 1.05 V at 350 mA.

10.6   Chemical Lift-Off

It has been reported that the high-energy laser treatment of the LLO process leads to local heating of the GaN layer and causes its destruction [39]. An alternative method to detach the GaN epilayer from the sapphire substrate is the chemical lift-off (CLO) technique [49, 50]. Figure 10.12(a) presents VLEDs fabricated by CLO in which a CrN buffer layer was placed between the epilayer and the sapphire substrate, and this CrN buffer was selectively etched laterally resulting in lift-off of the epilayer grown over it [49]. Park et al. [50] also reported an electrochemical etching method based on the use of oxalic acid. However, although the CLO method has several advantages over LLO such as low stress and low cost, there is still a need to improve the epitaxial quality of the layer grown.

10.7   *n*-Type Contacts

The GaN substrate has two faces with different crystal polarities, the Ga-face and N-face polarities (Fig. 10.13), which greatly influence the electrical properties at the
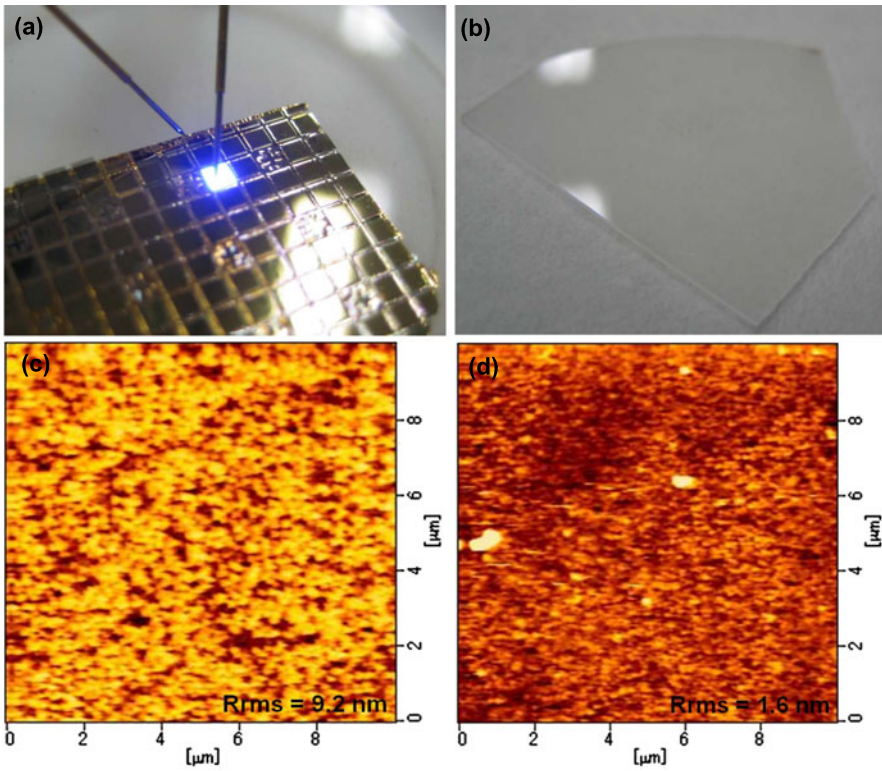
Fig. 10.12 (**a**) CLO GaN vertical LED, (**b**) CLO sapphire substrate, (**c**) atomic force microscopy surface images of chemical lift-off GaN, and (**d**) that of sapphire substrate

Fig. 10.13 Atomic structure of N-polarity GaN crystal
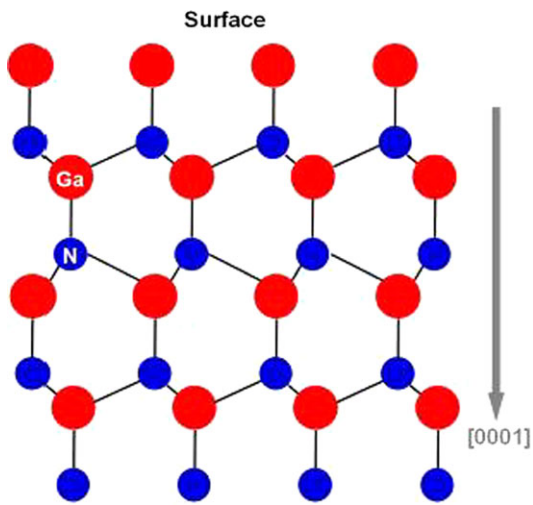
metal/GaN interface [51]. During VLED fabrication, the exposed n-type GaN layer is formed to have an N-face polarity after LLO. Note that n-Ohmic contacts for the Ga-face GaN could easily be formed using either Ti/Al or Cr/Au schemes; however, the N-face surface of GaN has displayed somewhat different Ohmic contact properties compared to the Ga-face surface. For instance, Kwak et al. [52] investigated the dependence of the Ohmic contact property on both the Ga- and N-face surfaces of an n-type GaN layer. They showed that it was difficult for the Ti/Al contacts to the N-face n-GaN to have Ohmic properties. Jang et al. [53] then introduced a thin Pd layer at the interface between a typical Ti/Al scheme and an N-face n-GaN layer. The Pd interlayer resulted in a reduction of the specific contact resistance by about one order of magnitude as compared to the Ti/Al scheme when annealed at temperatures in excess of 400 °C. After annealing at 450 °C, the contact resistivity reached 9.08×10^{-6} $\Omega \cdot cm^2$ and remained stable up to an annealing temperature 600 °C.

In addition, Jeon et al. [54] reported the fabrication of TiN/Al contacts for N-face n-GaN layers for high-performance VLEDs. The electrical characteristics of the TiN/Al contacts before annealing were similar to those of Ti/Al contacts; however, annealing the samples at 300 °C caused a degradation of their electrical properties. Jeong et al. [55] then demonstrated high performance VLEDs having non-alloyed reflective Cr/Al/Pt/Au n-type electrodes; the Cr/Al/Pt/Au contacts for n-GaN exhibited good Ohmic properties.

10.8 Randomly Roughened Structure for VLEDs

One method for increasing the LEE in GaN-based VLEDs is to form a roughened structure on the surface of VLEDs (Fig. 10.14a). By using the LLO technique, followed by a wet etching process to roughen the VLED surface, an n-type side-up GaN-based LED with hexagonal cone-like features was achieved [13]. Another method to improve the LEE is to form a corrugated pyramid-shaped (CPS) structure on the VLED surface [56]. A technique to further enhance the number of photons escaped from VLED slab of forming randomly roughened CPS structure was then developed (Fig. 10.14b). The CPS structure increased the surface areas having angular randomization allowing more scattering of photons.

10.9 Photonic Crystal Structure

A photonic crystal in general is the 1-, 2-, or 3-dimensional periodic arrangement of materials with different refractive indices, in which the characteristic length scale is of the order of the wavelength [57]. Using photonic crystals in VLEDs is a possible next step to improve performance, as they can be used to increase the LEE and also enhance the IQE by means of the Purcell effect [58]. Coupling a photon into guided modes is allowed, though these modes are subsequently out-coupled by the diffractive properties of the photonic crystal. Several research groups have
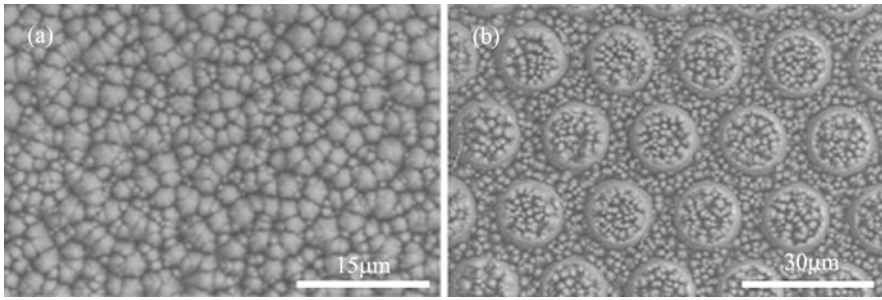
Fig. 10.14 SEM micrographs of VLED surface covered with (**a**) randomly roughened surface (**b**) corrugated pyramid-shaped (CPS) surface
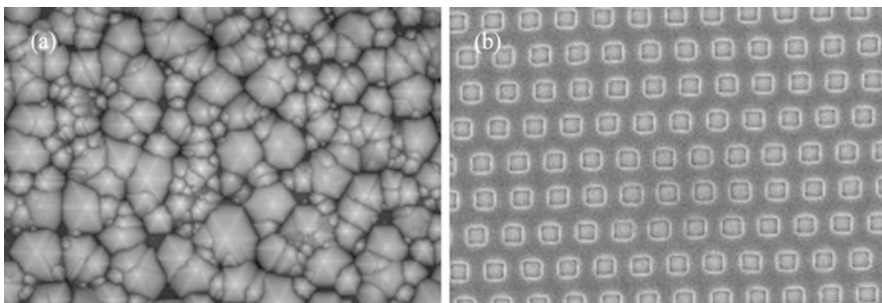
Fig. 10.15 Schematic of VLEDs with a (**a**) roughened structure and (**b**) photonic crystal structure

demonstrated GaN-based VLEDs having photonic-crystal diffractive layers on the surfaces, which display an improvement of light extraction and control of the radiation pattern [26, 58, 59]. In photonic crystal LEDs, various parameters contribute to the overall LED performance, such as the thickness of the epitaxial layers, as well as the type, depth, filling factor, and pitch of the photonic crystal. Within the parameter range investigated, the best photonic-crystal LED shows a very high. In randomly roughened LEDs, all waveguide modes inside the VLED have an average probability of being extracted into an outer medium; the extraction solely depends on the reabsorption occurring along this path length (Fig. 10.15a). In a photonic crystal LED, the discrete diffraction of the modes according to Bragg's Law means that not all waveguide modes can be extracted into an outer medium with high efficiency (Fig. 10.15b). However, the diffraction of certain modes is very efficient and leads to a fast extraction of that mode, thereby reducing the effect of absorption, and the design of a photonic crystal LED can be optimized for improved directionality [58, 59].

In VLED structures having a several microns thickness, a large fraction of the photon is not extracted even with an optimized photonic crystal, such that the overall light output of a roughened LED is superior-even in a forward direction. A higher light output in the forward direction can also be obtained using ultra-thin (>1 μm)
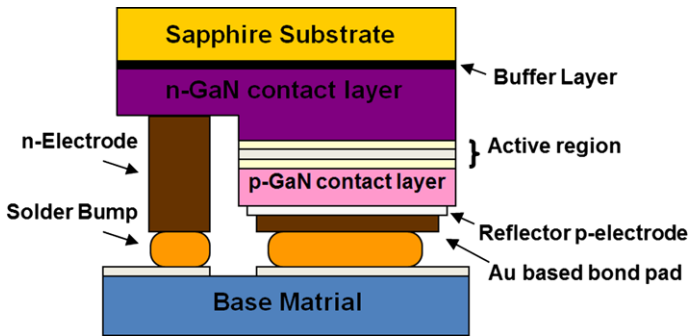
Fig. 10.16 Structure of flip-chip LEDs having a reflector electrode

LED structures; however, such ultra-thin structures require additional challenges in terms of homogeneous current injection [60].

10.10 High Power Flip-Chip LEDs

The basic structure of a GaN flip-chip LED (FCLED) is shown in Fig. 10.16. In light extraction, flip-chip type LEDs have advantages over top-emitting LEDs because the thick reflective mirror on the p-GaN layer is quite effective in enhancing the LEE, thus enabling the structure to be used extensively in high power and high efficiency LEDs [61–63].

In order to develop high-performance FCLEDs, a key technique might be to realize high-quality p-Ohmic contacts that have characteristics of high reflectivity, low contact resistance, and good thermal stability. To this end, Ag is considered the most attractive candidate for p-type reflectors for fabricating GaN-based LEDs having flip-chip or vertical-structure configurations because of its high reflectance in the visible wavelength range and reasonable Ohmic behavior [64, 65]. However, Ag shows poor adhesion to p-GaN due to agglomeration during the annealing process [66–69], leading to deterioration in its electrical and optical properties [68, 69]. Therefore, thin transparent layers such as nickel-oxide and transparent conducting oxides are typically inserted between Ag and p-GaN in order to improve the adhesion of Ag to p-GaN and to decrease agglomeration [64, 65, 69–75]. However, these thin interlayers can absorb light generated from the InGaN/GaN multiple quantum well (MQW), resulting in a decrease in the optical power of LEDs [76, 77].

To resolve these problems, Kim et al. [78] reported that an Ag-based alloy reflector for the FCLED should first be made. In other studies, AgCu was also applied to alloy p-type reflectors for GaN FC LEDs [79], and an AgAl reflector showed a specific constant resistance of 5.3×10^{-4} $\Omega \cdot cm^2$ and reflectance of 86.7 % at a 450 nm wavelength after annealing at 500 °C (Fig. 10.17(b)).

Figure 10.18 shows the light output-current (L–I) characteristics of FCLEDs having Ag and AgAl layers that were annealed at 300 °C and 500 °C for 1 min
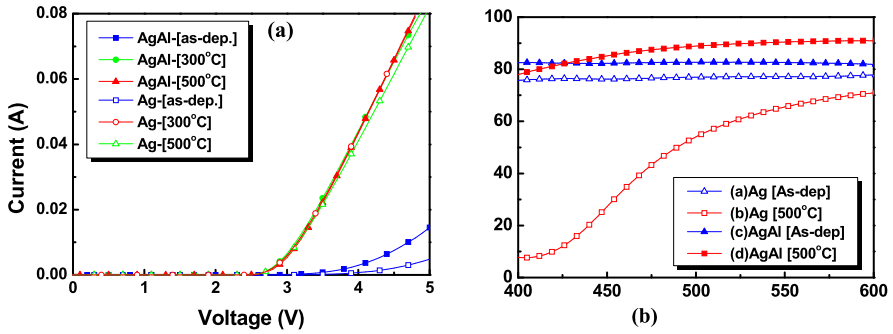
Fig. 10.17 (**a**) Typical $I-V$ characteristics of LEDs with Ag and AgAl layers on p-GaN annealed at different temperatures. (**b**) Reflectance of Ag and AgAl layer before and after annealing at 500 °C for 1 min in air. Reprinted with permission from Ref. [78]. Copyright [2006], American Institute of Physics
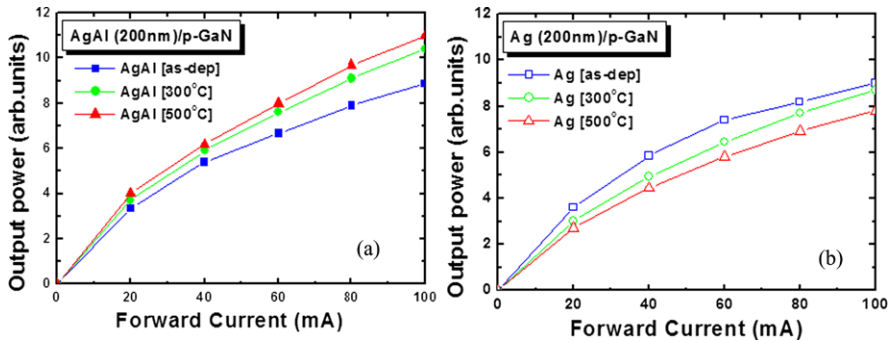
Fig. 10.18 $L-I$ characteristics of LEDs having an (**a**) AgAl layer before and after annealing at 500 °C for 1 min in air and (**b**) Ag layer before and after annealing at 500 °C for 1 min in air. Reprinted with permission from Ref. [78]. Copyright [2006], American Institute of Physics

in air. Figure 10.18(b) shows that the optical output power of a LED with the Ag layer decreases with increasing annealing temperature. This result can be attributed to the decrease in reflectance, which is largely due to the formation of holes and agglomeration after the thermal annealing of the Ag layer. Figure 10.18(a), however, shows that the light output power of LEDs having AgAl layers is enhanced with increasing annealing temperature due to the higher thermal stability of the AgAl layer compared to the Ag layer [78].

Kim et al. [79] reported the specific contact resistance of 7.5×10^{-5} $\Omega \cdot cm^2$ of AgCu after annealing at 400 °C. Furthermore, compared to the Ag contacts, the AgCu contacts remain reasonably stable at higher temperatures of 500 °C and 600 °C, and their contact resistivity is comparable to that of Ag contacts. The AgCu alloy reflector also has a high reflectance of 89.5 % at a 400 nm wavelength after annealing (Fig. 10.19). Note that despite the absence of interfacial voids, the AgCu sample shows a lower reflectance compared to the as-deposited sample, possibly
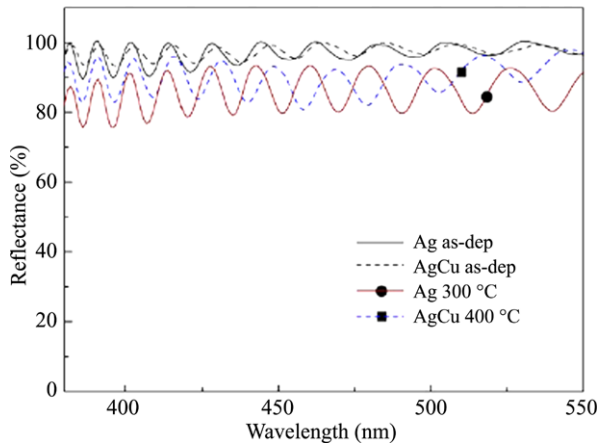
attributed to the presence of an interfacial Cu-oxide phase. Compared to the LED having an Ag reflector, the LED with the AgCu reflector shows better $L-I-V$ characteristics. For example, the forward bias voltage and output power measured at 20 mA were 3.56 V and 1.21 mW for the LED with the Ag reflector, and 3.26 V and 1.31 mW for the LED with the AgCu reflector, respectively.

Conductive omnidirectional reflectors (ODRs) consisting of GaN, an ITO nanorod low-refractive-index layer, and an Ag layer have been demonstrated in GaInN LEDs [73]. GaInN LEDs with GaN/low-n ITO/Ag ODR showed a lower forward voltage and a 31.6 % higher LEE than LEDs having an Ag reflector. This increase in efficiency could be attributed to the enhanced reflectivity of an ODR employing a low-n ITO layer, which has a refractive index of 1.34 at 461 nm.

The LEE in LEDs is limited by TIR due to the high index of the GaN relative to the surrounding medium, air, or epoxy. Therefore, LEE enhancement is typically attributed to an escape cone that is produced by the change in propagation direction of light rays. Subsequently, a pyramid reflector consisting of an array of three-dimensional (3D) SiO_2 pyramids and an Ag layer was developed [74], and a diffused reflector with Ag on nanotextured ITO was proposed (Fig. 10.20) [75]. Experimental results showed that the pyramid and reflector improved the LED extraction efficiency by 13.9 % and output power by 153.9 % compared to a conventional GaInN LED with a planar Ag reflector.

Lee et al. [14] reported the LEE enhancement obtained by adopting a hole-shape patterned ITO Ohmic contact layer (HITO) and AgIn reflector in GaN-based FCLEDs (Fig. 10.21).

Figure 10.22 shows the $L-I-V$ curves of GaN-based FCLEDs with hole-shaped patterned ITO (HITO) and AgIn reflector with microsized hole-shaped patterned ITO layers. In this configuration, the optical output power of FCLEDs was improved by approximately 30 %, which can be attributed to the good adhesion and low resistance, high reflection, and transmission of the patterned ITO/Ag layer.

Shen et al. [15] investigated an FCLED having a double-sided PSS (Fig. 10.23). For comparison, a conventional FCLED with no PSS patterning and an FCLED

(a) Reflector on textured ITO

Substrate
n-type GaN
MQW
p-type GaN
Textured ITO
Reflector

(b) Reflector on planar ITO

Substrate
n- type GaN
MQW
p-type GaN
ITO
Reflector
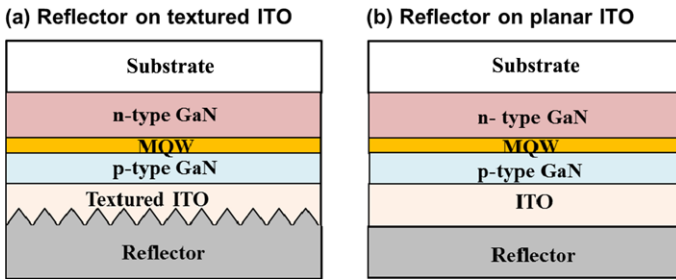
Fig. 10.20 Schematic cross-sectional views of FCLEDs with (**a**) Ag reflector on nanotextured ITO layer, and (**b**) Ag reflector on planar ITO layer
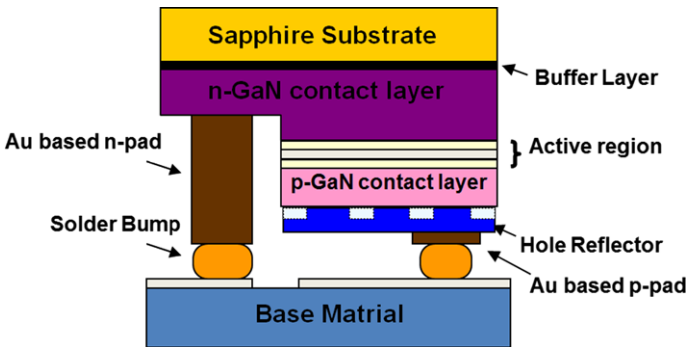
Fig. 10.21 Structure of GaN filp-chip LEDs with AgIn reflector on ITO hole-shape pattern

Fig. 10.22
Light-current-voltage
($L-I-V$) curves of
GaN-based filp-chip LEDs
with hole-shaped patterned
ITO Ohmic contact layer and
AgIn reflector as a function of
hole radius and spacing
between holes
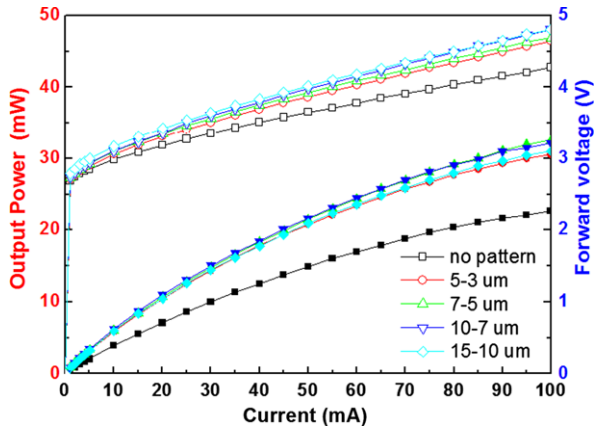
prepared on a PSS were prepared at the same time using exactly the same epitaxial layers [15].

Under a 350 mA current injection, the forward voltages were 3.24 V, 3.26 V, and 3.25 V for the conventional FCLED, FCLED prepared on a PSS, and an FCLED
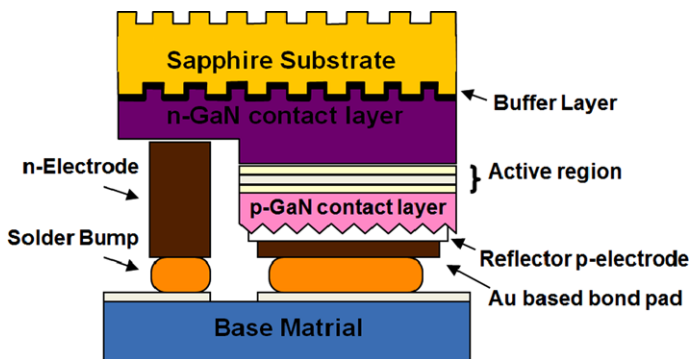
Fig. 10.23 Schematic of fabricated FCLED with double-side PSS

Fig. 10.24
(**a**) Cross-sectional schematic
of InGaN/GaN thin-film
flip-chip LED structure.
(**b**) Plane view
photomicrograph of the
device during operation
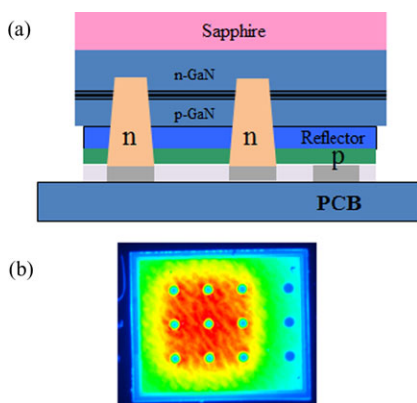
having a double-sided PSS, respectively. Compared to the conventional FCLED with no PSS patterning, the 24 % EL intensity enhancement observed for the FCLED on the PSS should be attributed to the decreased dislocation density in the epitaxial layer and to enhanced light scattering at the GaN-sapphire interface [15]. However, it is expected that the LED output intensity can be further improved by texturing the sapphire bottom surface, thereby enabling photons to undergo multiple scattering and easily escape from the roughened sapphire surface.

Shchekin et al. [16] compared the performances of standard FCLEDs to vertical-injection thin-film (VTF) and thin-film flip-chip (TFFC) LEDs by performing a split-wafer experiment, in which the three types of devices were processed from the same epi-structure. Figure 10.24 presents a cross-sectional schematic of the InGaN/GaN TFFC-LED structure. The radiance of the blue TFFC-LED was 168 mW/mm^2 sr, and is 1.2 times and 2.3 times that of the VTF- and FCLEDs, respectively. Standard deviations for all measured radiances in the group were less than 5 %. The optical power output of the encapsulated TFFC-LED was 750 mW, which is 51 % and 15 % higher than for the encapsulated FC- and VTF-LEDs, respectively. The superior performances of the VTF- and TFFC-LEDs compared to

the FCLED are due to the enhanced extraction efficiency of the roughened GaN thin film. In LEDs, the roughened surface disrupts the guiding of light in the high refractive index layers and enhances the out-coupling of light, resulting in a dramatically increased EQE. The higher radiance and output power of the TFFC-LEDs compared to VTF-LEDs are the result of adopting a superior flip-chip device structure that eliminates occlusion caused by the top-contact and wire bond pads. However, even though patterning of the top metal contact in VTFLEDs can be further optimized, there is a trade-off between the extraction efficiency and high current operation. The top-contact area has to be minimized to maximize the LEE, whereas the higher the electrical input power applied, the greater the contact area required to minimize current crowding effects [16].

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Chapter 11
Packaging.
Phosphors and White LED Packaging

Rong-Jun Xie and Naoto Hirosaki

Abstract Both phosphors and the LED packaging play a crucial role in optical properties of white LEDs. The selection criteria of the host lattice, activator ions as well as phosphors utilized in white LEDs are firstly discussed, followed by the overview of the technologically important phosphors, including yttrium aluminum garnets, alkaline earth orthosilicates, alkaline earth sulfides/thiogallets, and (oxo)nitridosilicates (i.e., nitrides). Typically, (oxo)nitridosilicate phosphors are highlighted introduced in this chapter as they have significantly redshifted photoluminescence spectra, abundant emission colors and high reliability. Then, the LED packaging configurations, such as "phosphor-in-cup", "remote-phosphor" and quantum dot (QD) white LEDs are discussed in terms of their luminous efficacy, color rendering index and color temperature. Finally, some issues on phosphors (materials searching, color tuning, particle size and geometry) and white LEDs (packaging configurations and reliability assessments) are addressed.

Keywords Luminescence · Phosphors · Garnet · Nitride · Orthosilicate · Remote-phosphor · LED packaging · Luminous efficacy · Color rendering · Color temperature

11.1 Introduction

Although there are lots of issues needed to be addressed in white LEDs, such as chip processing, light extraction efficiency, heat sink structures, resin materials, reliability, life test, and etc., this chapter focuses on phosphors materials and their applications in white LED packaging.

White light-emitting diodes are attracting considerable attention from both academic and industrial communities, because they are generally accepted as the next-generation energy-saving and green solid state lighting sources. A world-record

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of 130 lm/W at 1 A drive current for white LEDs with the color temperature of 4700 K was announced recently by Niichia [1], and Cree also claimed a R&D record of 254 lm/W at a drive current of 350 mA [2]. Nevertheless, such performance levels greatly surpass that of fluorescent tubes (∼80–100 lm/W), and far exceed that of incandescent (∼17 lm/W) and halogen (∼25 lm/W) lamps. Thanks to the great progress in optical qualities (luminous efficacy, color temperature, color rendition, lifetime, etc.), white LEDs are now penetrating into the lighting market very rapidly, which are steadily replacing incandescent lamps and fluorescent tubes for general lighting, cold cathode fluorescent lamps (CCFL) for liquid crystal display TVs for backlighting, or substitute Xeon lamps for automotive headlights. The use of white LEDs will significantly save huge electrical energy, and reduce carbon emissions [3]. For example, the worldwide energy saving would reach ∼1000 TW·h/year if all traditional lighting sources were replaced by white LEDs. This corresponds to a budget savings of ∼US$100 billion/year, and a carbon emission reduction of 200 M/year [4]. For improving the energy savings and reducing the greenhouse effect, lots of governments around the world have banned or will ban the use of incandescent bulbs for general lighting, encouraging peoples to use more energy efficient lighting alternatives, such as LED lamps.

There are basically three options to create white lighting using LED chips, and each of them has their own advantages and disadvantages. The first option is called multi-chip white LED, which combines individual R (red), G (green) and B (blue) monochromatic LED chips. The multi-chip white LEDs, however, have the problem of that three different LEDs age over time and temperature at different rates, causing the white light to take on colored hues. Moreover, the luminous efficacy is not high because the green LEDs are not bright enough (this is called green window). To solve those problems, a green phosphor instead of a green LED is combined with RB LEDs. The second option is to pump a yellow phosphor (usually the Ce^{3+}-doped yttrium aluminum garnet, YAG:Ce) with a blue LED chip. This bichromatic (blue LED + YAG) white LED is the simplest one, and produce very efficient light with a medium color rendering index (CRI) of ∼70–80. To achieve a higher CRI, green/red (RG) or green/red/yellow (RGY) phosphor blends are used instead of a single yellow phosphor. The third option is to down-convert an ultraviolet (UV) or near-UV LED by using a red/green/blue (RGB) phosphor blend. Among the last two approaches, phosphors are often utilized as the down-conversion luminescent material, which absorb the emitted light from the LED chip and convert it into visible light (green, yellow, or red). LEDs using phosphor materials are also named phosphor-converted white LEDs (pc-wLEDs).

In the following, the development and photoluminescence properties of phosphor materials used in white LEDs will be overviewed. Furthermore, the status of white LED lamps utilizing phosphors will be discussed, and the optical properties of these lamps will be demonstrated.

11.2 Phosphor Materials for White LEDs

11.2.1 Selection Criteria of Phosphors

In pc-wLEDs, phosphor materials play a crucial role in the color quality of LED lamps, such as luminous efficacy, color rendering index, color temperature, lifetime, and etc. Therefore, to realize highly efficient and reliable white LEDs, it is of quite importance to choose appropriate phosphors. Although the history of luminescent materials dates back to one century ago, and there are numerous phosphors that have been developed for cathode-ray tube displays and fluorescent lighting tubes, only a limited number of phosphors are known to be suitable for use as wavelength conversion materials in wLEDs [5]. wLED phosphors are excited by lower energy irradiations, i.e., 440–470 nm of blue light, compared to those phosphors for displays and traditional lightings. This makes most traditional phosphors impossible for use in wLEDs, and novel phosphors are to be developed to meet new requirements.

The selection rules of phosphor materials for wLEDs, in terms of spectral features, quantum efficiency, stability as well as particle characteristics, are summarized as below [6, 7].

(i) An *excitation spectrum* is determined by monitoring fluorescence emission at the wavelength of maximum intensity while the phosphor is excited through a group of consecutive wavelengths. To convert the emitted light of LED chips more efficiently, the excitation spectrum should be broad enough to cover the emission spectrum, and to endure variations of the emission spectrum of LED chips. This requires the phosphor to absorb strongly the light of LEDs, and to have a flat excitation spectrum near the maximum emission wavelength of LEDs.

(ii) An *emission spectrum* records a set of fluorescence intensity of light given off by a phosphor when it is excited by a certain wavelength of light. The emission spectrum of a phosphor is required to be useful (i.e., aiming at a certain color) and broad sufficiently to realize high color rendering white light, when it is combined with that of a LED chip. On the other hand, the emission spectrum is required to be as narrow as possible (typically for green or red phosphors) to achieve wide color gamut LED backlights and high efficiency white LEDs. Furthermore, to minimize the re-absorption of light, the overlap between the excitation and emission spectra should be as small as possible.

(iii) The *quantum efficiency* of a phosphor is subdivided in *internal quantum efficiency* and *external quantum efficiency*. The internal quantum efficiency is defined as the ratio of the number of emitted photons to the number of absorbed photons, and the external quantum efficiency is defined as the ratio between the number of emitted photons and the number of incident photons. Higher quantum efficiency of phosphors leads to higher luminous efficiency of white LEDs. It is required to reduce the concentration of lattice defects that absorb the incident photons and to decrease the probability of the non-radiative energy

transfer. In addition, the light scattering by phosphor particles is also required to be minimized.

(iv) The *stability* of a phosphor requires that the shape of luminescence spectra, luminescence intensity, and quantum efficiency do not change over the entire operating temperature range and the lifespan of wLEDs. The thermal quenching of the phosphor should be as small as possible. In addition, the phosphor is required to be stable against chemical attacks (including moisture, O_2, CO_2, and etc.) and irradiation attacks (including UV or laser light). Moreover, there should be no saturation effects in phosphors when the excitation flux increases, typically in high-power wLEDs.

(v) The *particle size and morphology* of phosphor particles affects the quantum efficiency of phosphors as well as the luminous efficiency of wLEDs greatly. Generally, to reduce the light scattering and improve the quantum efficiency of the phosphor, the particle size is required to be as uniform as possible, and the distribution as narrow as possible. Spherical particle morphology is highly desired. In addition, phosphors should be easily mixed with and well dispersed in silicones and epoxy resins, which again requires the homogeneous particle size, narrow size distribution and spherical morphology.

(vi) The *production cost* of the phosphor should be as low as possible. The phosphor is required to be synthesized cost-effectively.

11.2.2 Selection of Host Crystals and Activators

Phosphors are inorganic luminescent materials that usually consist of a host lattice and a small concentration of optically-active metal ions (dopant). To obtain the phosphor with desired luminescence spectra (shape, emission color) and high quantum efficiency, both host lattice and activators should be carefully selected.

For the *host crystal*, the following requirements will be fulfilled [6].

- *It has a very large crystal-field strength* that splits the 5d energy levels to a great extent and thus reduces the electronic transition energy (see Fig. 11.1). The crystal field is the electric field of the surrounding ions at the site of activator using the approximation of point charges. It increases with increasing the coordination number and the charge of neighboring ions.

- *It has strong covalent chemical bonding* that lowers the excited state of the activations ions through the nephelauxetic effect (see Fig. 11.1).

- *It has a wide band gap* that enables it to be optically transparent. In addition, this also enables the excited state of the activator ions to be located at a right place in between the conduction band and the valence band of the host crystal.

- *It is chemically and thermally stable*. Air-, moisture-, or temperature-sensitive materials are not suitable for use as host crystals unless they are well protected by the surface coating or other treatments.

- *It is easily to be synthesized* to reduce the production cost.

Fig. 11.1 Schematic of lowering the 5d energy levels of Eu^{2+} by the nephelauxetic effect and the crystal-field splitting
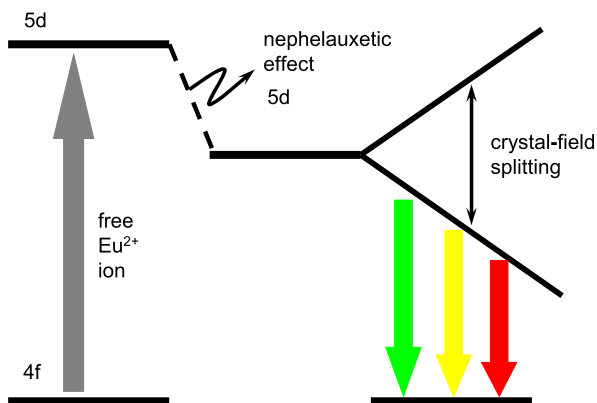
Shifting to the *activator ion*, the following suggestions should be considered while they are selected [6].

- *It has a small coordination number when bonded to the nearest ligands*. This will lead to a large crystal-field splitting of the dopant ion.
- *It is covalently bonded to the nearest ligands at relatively short distance*, so that the crystal-field splitting and nephelauxetic effect will be enhanced.
- *It has the spin-allowed electronic transitions* to realize intense luminescence. The $4f \rightarrow 5d$ electronic transitions of Eu^{2+} or Ce^{3+} are spin-allowed, yielding intense emission of Eu^{2+} or Ce^{3+}. So these ions are popularly used as activators in rare earth phosphor. Whereas the d-d electronic transitions of Mn^{2+} in some cases are spin-forbidden, resulting in weak Mn^{2+} emission due to the low oscillator strength.
- *It shows the desired emission color and spectrum*. Some ions having the f-f electronic transitions (e.g., Tb^{3+}, Eu^{3+}) usually show their characteristic narrow emission spectra and colors which are not sensitive to the local environment. Their emission wavelengths are not tunable, so that they are not commonly used in LED phosphors.

With these in mind, numerous compounds have been investigated as phosphor host lattices, and a number of phosphors have thus been developed and suggested for use in wLEDs. In the next section, an overview of some technologically important phosphors will be given.

11.2.3 Type of Phosphors

Phosphors can be divided into many groups, depending on their chemical composition, crystal structure, emission color, as well as the dopant type. Here, the phosphors are classified into garnets, orthosilicates, sulfides, nitrides, and so on [6]. In

Fig. 11.2 Excitation and emission (*shadow*) spectra of $Y_3Al_5O_{12}:Ce^{3+}$, showing the crystal field splitting ε_{CFS}
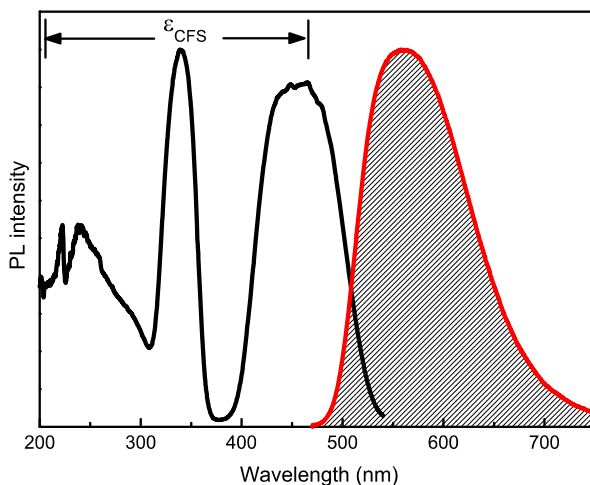
this part, the discussion will be concentrated on the phosphors that are mostly popular in the market, including garnets, alkaline earth orthosilicates, alkaline earth sulfides and (oxo)nitridosilicates.

11.2.3.1 Garnets

Garnet phosphors refer to a family of Ce^{3+}-doped luminescent materials that have a general chemical formula of $X_3Y_2Z_3O_{12}:Ce^{3+}$ (X = Ca, Mg, Y, Lu; Y = Al, and Z = Si, Ge) [8, 9]. The three different cations X, Y and Z are coordinated to the nearest oxygen with the coordination number of 8, 6 and 4, respectively, which forms a very stiff cubic structure with the space group of Ia–3d.

The yellow-emitting Ce^{3+}-activated yttrium aluminum garnet ($Y_3Al_5O_{12}:Ce^{3+}$, simply YAG:Ce) is the first phosphor that was used to combine with a blue LED to generate white light by Nakamura in 1995 [10]. Although it was reported as a phosphor for flying spot cathode-ray tubes by Blasse in 1967 [11], YAG:Ce is now the most popular yellow phosphor in wLEDs because of its excitation spectrum matching perfectly with the blue LED, very intense yellow emission color, high quantum efficiency, composition-tunable emission wavelength, high thermal quenching temperature, and ease of synthesis.

Figure 11.2 shows the typical excitation and emission spectra of YAG:Ce. The excitation spectrum is significantly split into several bands due to a very large Ce^{3+} crystal field splitting ε_{CFS} of ∼26,050 cm^{-1} [12], so that the lowest-energy excitation band covering the spectral range of 380–550 nm has the peak wavelength (450–470 nm) equivalent to the emission maxima of blue LEDs. The emission spectrum shows an asymmetric broad band peaking at 560 nm as a result of the electronic transitions from the ^5D excited states to the $^2F_{5/2}$ and $^2F_{7/2}$ doublet ground states. The full-width at half maximum (FWHM) of YAG:Ce is as large as ∼120 nm, en-

abling to achieve an acceptable color rendering index ($R_a \sim 70$–80) of wLEDs for
general illumination.
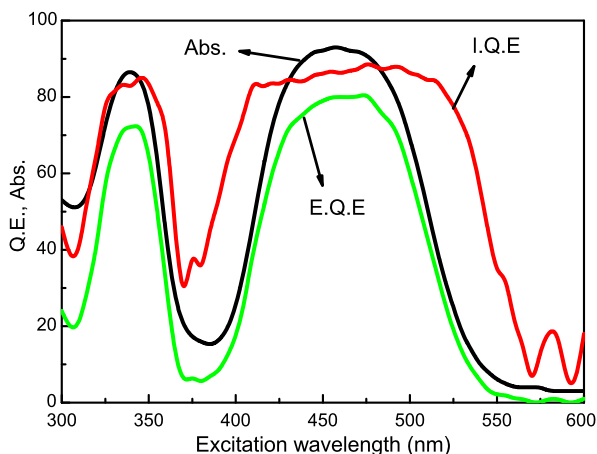
As shown in Fig. 11.3, the yellow-emitting YAG:Ce phosphor (1 mol% Ce) has
a very good luminescence efficiency, the internal quantum efficiency and external
quantum efficiency being 86 % and 80 % under the 460 nm excitation, respectively.
The absorption at 460 nm is 93 %, only 7 % incident light being reflected.

The Ce concentration affects the luminescence properties (i.e., efficiency, emis-
sion color, absorption) and thermal quenching temperature greatly. Silver et al. [13]
reported that the luminous efficacy of YAG:Ce decreased continuously as the Ce
concentration increased, as shown in Fig. 11.4. Although the diluted phosphor has
the luminous efficacy approaching the theoretic value, it is not recommended for
the practical use because of its high reflectance at the emission wavelength of blue
LEDs. A redshift in emission spectrum is often observed when the Ce concentra-
tion increases, which is due to the re-absorption and the enhanced Stokes shift.
Bachmann [14] conducted a comprehensive investigation on thermal quenching of
YAG:Ce. He addressed that the thermal quenching temperature is intrinsically high
($\gg$680 K), and YAG:Ce with the lower Ce concentration had the higher thermal
quenching temperature. Ueda et al. [15] did a temperature dependence of photo-
conductivity measurement, and reported that the threshold of photoionization from
the 4f level of Ce^{3+} to the conduction band in YAG:Ce was 3.8 eV, and the main
quenching process was caused by the thermally stimulated ionization process from
the $5d_1$ to the conduction band.

The $Y_3Al_5O_{12}$:Ce composition is usually modified to achieve green or orange-
yellow emitting colors [14, 16–19]. Substitutions on the Y dodecahedral site with
a smaller Lu ion or on the Al octahedral site with a larger Ga ion will lead to the
blue shift. On the contrary, considerable red-shifts can be realized by substituting Y
by larger La, Gd, Tb ions, Al by a Mg–Si pair, or Al–O bonds by Si–N ones. The
modified YAG:Ce phosphors compensate the loss of thermal stability and visual ef-
ficiency by achieving lower color temperature (red-shift) or higher color rendering
index (blue-shift). A yellow $Tb_3Al_5O_{12}$:Ce phosphor was used by Osram, which has

Fig. 11.4 Luminous efficiency of the YAG:Ce phosphor as a function of the Ce concentration (excitation wavelength = 470 nm) (Reproduced from Silver, J. and Withnall, R. 2008. Color conversion phosphors for LEDs in Luminescent Materials and Applications (Edited by A. Kitai). New York: John Wiley & Sons. Ltd. With permission)
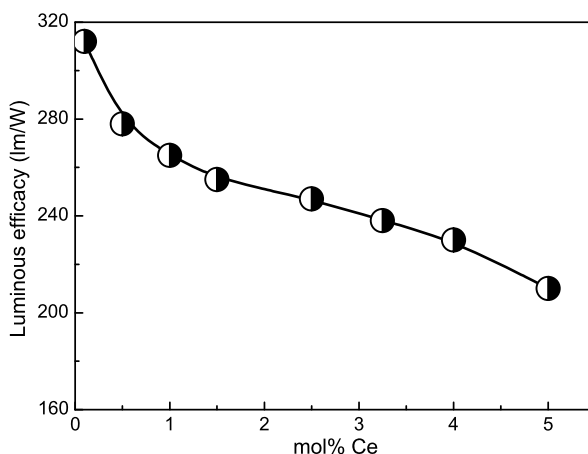
Table 11.1 Peak excitation wavelength (λ_{exc}), peak emission wavelength (λ_{em}), and thermal quenching temperature (T_{tq}) of some garnet phosphors

	λ_{exc} (nm)	λ_{em} (nm)	T_{tq}	Reference
$Ca_3Sc_2Si_3O_{12}$	455 nm	505/560	High	[21]
$Lu_2CaMg_2(Si,Ge)_3O_{12}$ (6 mol%)	470	605	Medium	[22]
$Mg_3Gd_2Ge_3O_{12}$ (2 mol%)	464	600	–	[23]

no intelligence property (IP) problems with Niichia's YAG-based phosphors. However, Jang et al. [20] indicated that the luminous efficacy of wLEDs using TAG:Ce was lower than that of YAG-based wLEDs, which was due to the significant sedimentation of TAG:Ce particles in the epoxy resin caused by the higher density of TAG:Ce (6.085 g/cm^3 vs. 4.657 g/cm^3 of YAG:Ce).

Other Ce^{3+}-activated garnet phosphors are listed in Table 11.1. These garnet phosphors show redshifted emission spectra compared to YAG:Ce, which enables them to create warm white light when combined with blue LEDs.

11.2.3.2 Alkaline Earth Orthosilicates

Alkaline earth metal orthosilicates, such as Sr_2SiO_4 and Sr_3SiO_5, are interesting host lattices for wLED phosphors. The highly polarisable Si–O bonds in these orthosilicates yield strongly covalent binding between alkaline earth metals and oxygen, thus resulting in a large nephelauxetci effect. This strong nephelauxetic effect considerably reduces the transition energy of Eu^{2+}, leading to red-shifted excitation spectrum of alkaline earth metal orthosilicate phosphors and thus a significant overlap with the emission spectrum of blue LEDs.

Barry [24] did a comprehensive survey on the photoluminescence of Eu^{2+}-doped Sr_2SiO_4–Ba_2SiO_4 solid solutions. A continuous color shift from green

Fig. 11.5

Photoluminescence spectra of
the green- and
yellow-emitting
$(Sr,Ba)_2SiO_4$:Eu phosphor
(*1, 3*—excitation and
emission spectra for the green
phosphor, and
2, 4—excitation and emission
spectra for the yellow
phosphor). The emission
spectrum of a blue LED
(λ_{em} = 460 nm) is also
plotted
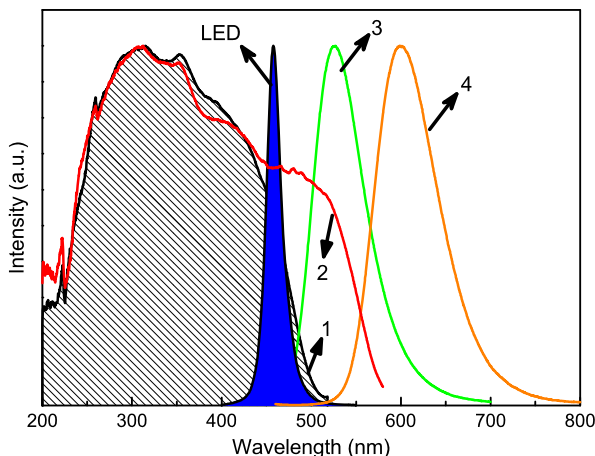
(λ_{em} = 505 nm, Ba_2SiO_4:Eu) to yellow (λ_{em} = 575 nm, Sr_2SiO_4:Eu) was observed,
indicative of the composition-tunable emission wavelength that is very important
for practical applications. Typical photoluminescence spectra of green and yellow
$(Sr,Ba)_2SiO_4$:Eu phosphors are given in Fig. 11.5. The emission spectrum shows a
broad band, which is due to the 4f → 5d electronic transition of Eu^{2+}. This Eu^{2+}
emission band is slimmer than the Ce^{3+} one because of the single electronic tran-
sition of Eu^{2+}. The excitation spectrum extends from UV to the visible spectral
range (∼500 nm), but the excitation maxima does not perfectly match with the peak
emission wavelength of blue LEDs. It indicates that $(Sr,Ba)_2SiO_4$-based phosphors
are more suitable for use in combination with short-wavelength blue LEDs (i.e.,
λ_{em} = 430–450 nm). On the other hand, the blue-light absorption can be enhanced
by the compositional tailoring (see curve 2 in Fig. 11.5). In addition, the low ab-
sorption of the blue light is compensated by higher visual efficiency of orthosilicate
phosphors. As seen in Fig. 11.6, under the 450 nm excitation the internal quantum
efficiency and external quantum efficiency of a green orthosilicate phosphor are 81
and 65 %, respectively. The absorption at 450 nm is around 80 %, much lower than
that of YAG:Ce (93 %).

Compared to YAG:Ce, Sr_2SiO_4:Eu-based phosphors show stronger temperature-
dependent efficiency and a larger shift of chromatic coordinates, typically the end
members of the Ba_2SiO_4–Sr_2SiO_4 system (see Fig. 11.7). Besides this, another dis-
advantage is the lower CRI of the wLEDs utilizing a single Sr_2SiO_4:Eu-based yel-
low phosphor. These shortcomings prevent it from their use in high-power LEDs.

Both of the excitation spectrum and the emission spectrum are considerably red-
shifted in Eu^{2+}-doped Sr_3SiO_5 phosphors, leading to orange-yellow or orange emis-
sion colors (see Fig. 11.8) [25–27]. The peak emission wavelength varies in the
range of 570–610 nm, depending on the Sr/Ba ratio [27]. Therefore, the Sr_3SiO_5:Eu
based phosphors are one of the best choices to compensate for the red component
in wLEDs to realize warm white when it is blended with a green phosphor. In ad-
dition, a much broader emission band (FWHM = 121.8 nm) can be obtained in the

Fig. 11.6 Plot of the absorption (Abs.), internal quantum efficiency (I.Q.E) and external quantum efficiency (E.Q.E) of the $(Sr,Ba)_2SiO_4$:Eu green phosphors vs. excitation wavelength
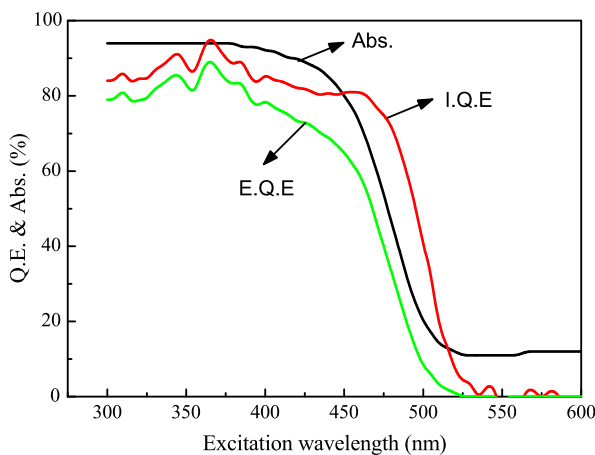
Fig. 11.7 Thermal quenching of Sr_2SiO_4:Eu-based green phosphors
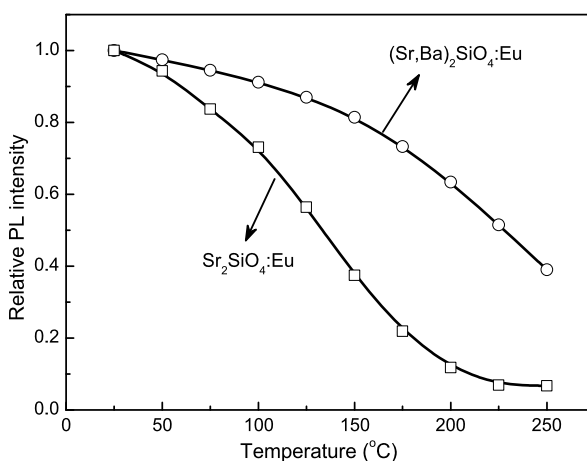
Ce^{3+}-doped Sr_3SiO_5 yellow phosphor, thus enabling to achieve a color rendering index comparable to YAG:Ce [28].

The orange-yellow/orange $(Sr,Ba)_3SiO_5$:Eu phosphors have higher thermal quenching temperatures than the green/yellow $(Sr,Ba)_2SiO_4$:Eu ones. Moreover, $(Sr,Ba)_3SiO_5$:Eu absorbs the blue light ($\lambda_{em} = 450–470$ nm) more efficiently than $(Sr,Ba)_2SiO_4$:Eu. For example, the absorption at 450 nm is about 91 % for $(Sr,Ba)_3SiO_5$:Eu, and a higher external quantum efficiency of 70 % is thus achieved (see Fig. 11.9).

Other interesting orthosilicate phosphors reported so far include Eu^{2+}-doped $MMg_2Si_2O_7$ (M = Ca, Sr, Ba) [29], $CaAl_2Si_2O_8$ [30], $M_3MgSi_2O_8$ (M = Ca, Sr, Ba) [31], Li_2SrSiO_4 [32], and etc. These phosphors show intense blue-to-yellow emission colors under the excitation of the UV light. However, most of these orthosilicates (except Li_2SrSiO_4) show a very low absorption of the blue light, making them impossible to be combined with blue LEDs.

Fig. 11.8 Excitation and emission spectra of the $(Sr,Ba)_3SiO_5:Eu$ orange-yellow phosphor
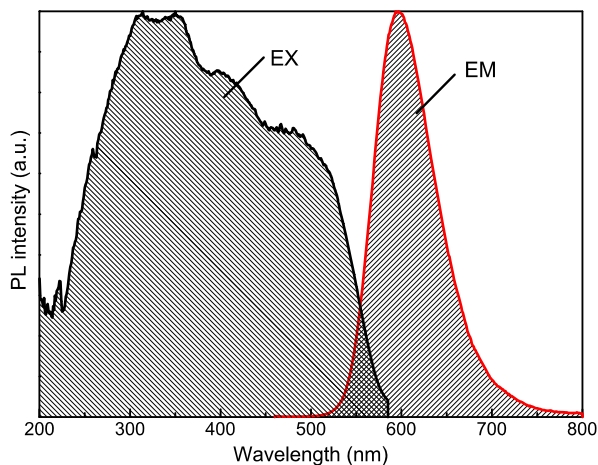
Fig. 11.9 Plot of absorption (Abs.), internal quantum efficiency (I.Q.E.) and external quantum efficiency (E.Q.E) of $Sr_3SiO_5:Eu$ vs. excitation wavelength
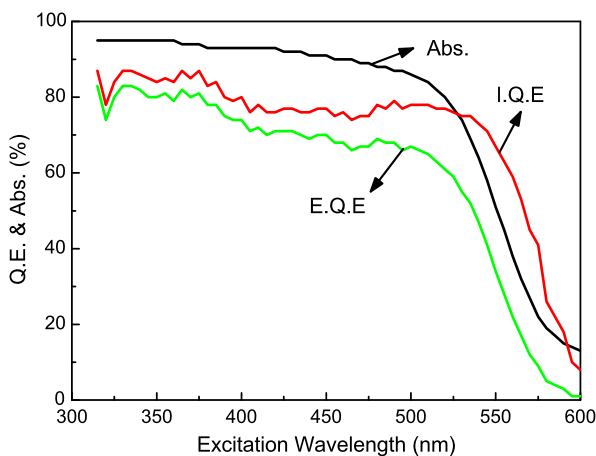
11.2.3.3 Alkaline Earth Sulfides/Thiogallates

Inorganic sulfides are also covalent compounds that become interesting host lattices for phosphors. The most notable materials are Eu^{2+}-doped alkaline earth metal sulfides (MS:Eu, M = Ca, Sr, Ba) [32, 33] and alkaline earth metal thiogallates (MGa_2S_4:Eu, M = Ca, Sr, Ba) [34–37].

(Sr,Ca)S:Eu phosphors show very broad excitation bands covering the UV-to-green spectral range, and absorb the blue and green light very efficiently. The phosphors emit red colors in the spectral range of 600–655 nm, depending on the Sr/Ca ratio. These features indicate that the (Sr,Ca)S:Eu phosphors are suitable for white LEDs. However, the strong moisture-sensitivity and large thermal quenching of these red phosphors make them impossible for practical applications. The surface coating of Al_2O_3 and SiO_2 is an effective way to protect (Sr,Ca)S:Eu phosphors against moisture and thermal attacks.

Fig. 11.10 Excitation (*left*) and emission (*right*) spectra of Ca–α-sialon:Eu and Li–α-sialon:Eu (*hatched*) phosphors. The emission spectrum of a blue LED ($\lambda_{em} = 450$ nm) is also plotted
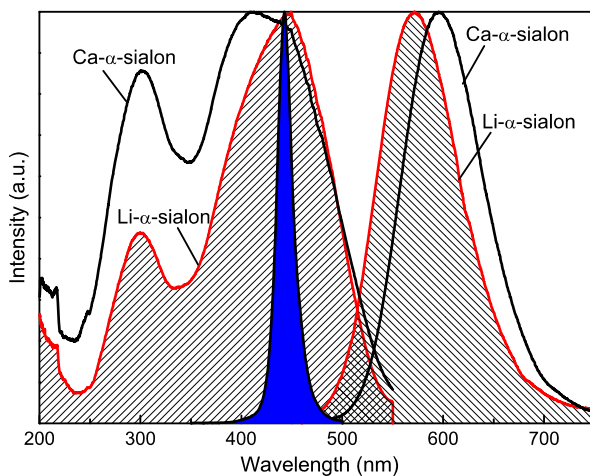
The $SrGa_2S_4$:Eu phosphor has a broad excitation band maximum at 470 nm, and it emits a green light with the peak wavelength of 535 nm. The emission color is further red-shifted when Sr is partially substituted by Ca or Ba [36]. This phosphor has already been considered as a green conversion luminescent material in high color rendering wLEDs due to its promising luminescent properties [36]. On the other hand, the poor thermal and chemical stabilities again will not enable it to be used in wLEDs widely.

11.2.3.4 (Oxo)nitridosilicates

(Oxo)nitridosilicate phosphors or nitride phosphors are newly developed lumines-cent materials that are very suitable for wLED applications [6, 38]. (Oxo)nitridosili-cates have stiff structures built up on highly dense three dimensional SiN_4 tetra-hedral networks, and have covalent chemical binding characters and strong crys-tal field strength [6, 39, 40]. Therefore, (oxo)nitridosilicate phosphors show red-shifted photoluminescence spectra, high thermal/chemical stability, high quan-tum efficiency, abundant emission colors, and etc. Extensive investigations on (oxo)nitridosilicate phosphors have been done in recent years, but the number of phosphors reaching the sufficient level of practical applications is limited. In this section, only three (oxo)nitridosilicate phosphors will be discussed, and others are summarized in Table 11.2.

α-sialon has a chemical formula of $M_x Si_{12-m-n} Al_{m+n} O_n N_{16-n}$ ($x = m/v$, v is the valence of metal M), which is limited solid solutions in the system α-Si_3N_4–Al_2O_3–AlN–$MO_{v/2}$ [41]. The α-sialon:Eu is a yellow-emitting phosphor [42–45]. As seen in Fig. 11.10, the excitation band centered at 400–450 nm matches well with the emission spectrum of blue LEDs. The excitation spectrum changes in the shape with accommodating different metal ions, the low-energy band being enhanced by substituting Li for Ca [46]. The emission spectrum shows a broad band peaked at

Table 11.2 Photoluminescence properties of (oxo)nitridosilicate phosphors

Phosphors	λ_{exc}	λ_{em}	I.Q.E/E.Q.E	Ref.
$LaAl(Si_{6-z}Al_z)N_{10-z}O_z$ (2.75 %Ce)	380	496	–/48 %*	[61]
$LaSi_3N_5$ (10 %Ce)	360	440	–/67 %	[62]
AlN (0.28 %Eu)	290	465	76/46 %**	[63]
β-sialon (0.3 %Eu)	450	535	49/33 %	[51]
$BaSi_2O_2N_2$ (2 %Eu)	458	492	–/41 %	[64]
$SrSi_2O_2N_2$ (2 %Eu)	456	544	–/69 %	[64]
$Ba_3Si_6O_{12}N_2$	–	525	–	[65]
$Sr_5Al_{5+x}Si_{21-x}N_{35-x}O_{2+x}$ (10 %Eu)	400	510	–	[66]
$Sr_{14}Si_{68-s}Al_{6+s}O_sN_{106-s}$ ($s \sim 7$) (6 %Eu)	376	508	–	[67]
$Sr_3Si_{13}Al_3O_2N_{21}$ (8 %Eu)	365	515	75/67 %	[68]
Ca–α-sialon (7 %Eu)	450	580	61/49 %	[6]
$La_3Si_6N_{11}$ (10 %Ce)	460	595	–/42 %	[69]
$CaAlSiN_3$ (10 %Ce)	483	570	71/50 %	[70]
$CaAlSiN_3$ (1.0 %Eu)	450	660	83/70 %	[6]
$Sr_2Si_5N_8$ (2 %Eu)	450	627	87/71 %	[71]
$SrAlSi_4N_7$ (5 %Eu)	450	639	76/53 %	[72]

Fig. 11.11
Temperature-dependent
luminescence intensity of
Ca–α-sialon:Eu ($m = 2$,
$n = 1$, 7 mol% Eu) and the
red emission enhanced
$(Y,Gd)_3Al_5O_{12}$:Ce phosphors
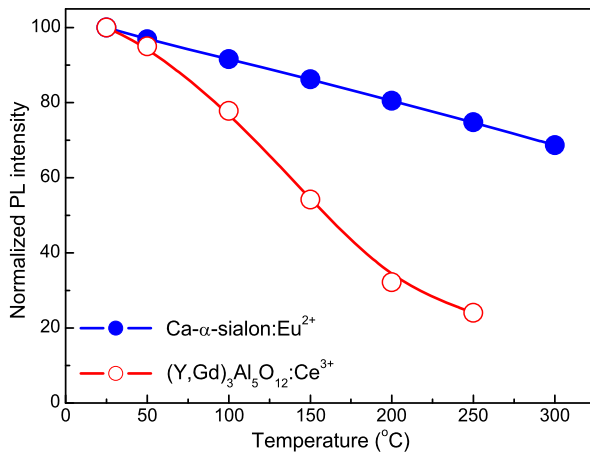
585 nm for Ca–α-sialon and 570 nm for Li–α-sialon [47]. The full width at half maximum for both sialons is about 94 nm, which is narrower than that of YAG:Ce. α-sialon:Eu is the first nitridosilicate phosphor that is suggested for use in warm wLEDs [44, 48].

Compared to YAG:Ce, α-sialon:Eu has a much higher thermal quenching temperature due to its extremely rigid structure. As shown in Fig. 11.11, the luminescence of $(Y,Gd)_3Al_5O_{12}$:Ce is significantly quenched at temperatures above 100 °C,

Fig. 11.12
Temperature-dependent
emission spectrum of
Ca–α-sialon:Eu ($m = 2$,
$n = 1$, 7 mol% Eu)
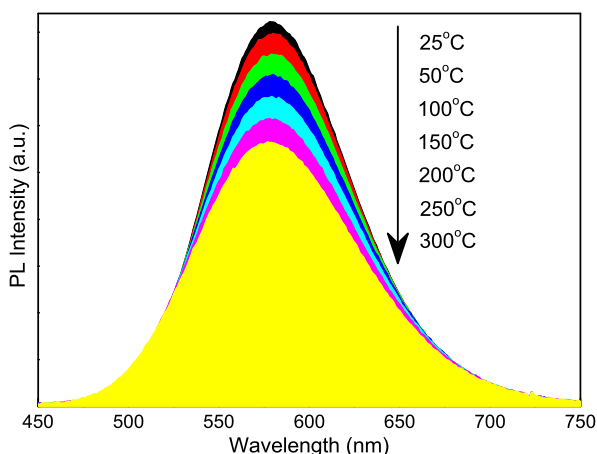
Fig. 11.13 Plot of absorption
(Abs.), internal quantum
efficiency (I.Q.E) and
external quantum efficiency
(E.Q.E) of Li–α-sialon:Eu
($m = 1$, $n = 1.8$, 7 mol% Eu)
vs. excitation wavelength
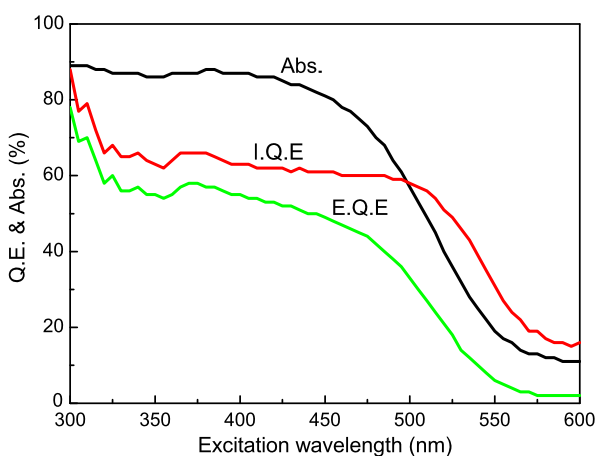
whereas that of α-sialon:Eu slightly decreases. In addition, with increasing temperature the shape of the emission spectrum of α-sialon:Eu remains unchanged (see Fig. 11.12), indicating that its chromaticity coordinates are quite stable against temperature. This is of importance when it is applied in high-power wLEDs.

Although it is superior to YAG:Ce in thermal stability, enhanced red emission, and color point stability, α-sialon:Eu has a lower quantum efficiency than YAG:Ce. The external quantum efficiency is 49 % when Li–α-sialon:Eu is excited at 450 nm (see Fig. 11.13). It is increased up to 60 % by modifying the processing conditions. The lower efficiency is ascribed to the low absorption and lattice defects.

Thanks to the wide range of the solid solution forming area, the emission color of α-sialon:Eu can be tuned in a wide range by changing the m and n values and the type of accommodating metal ions [6]. The composition-tunable emission of α-sialon:Eu enables it to fabricate wLEDs with varying color temperatures.

Fig. 11.14 Excitation and emission (*hatched*) spectra of β-sialon:Eu. The emission spectrum of a blue LED ($\lambda_{em} = 450$ nm) is also plotted
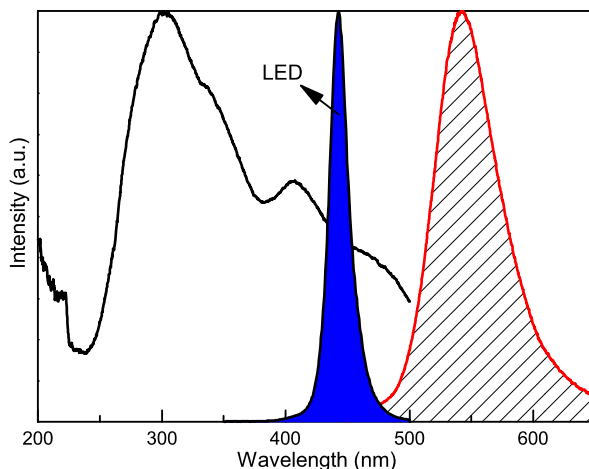
β-sialon ($Si_{6-z}Al_zO_zN_{8-z}$) is a limited solid solution between β-Si_3N_4 and Al_2O_3 with $0 \leq z < 4.2$ [49, 50]. The Eu^{2+}-doped β-sialon shows an intense emission band centered at 530–550 nm, depending on the z value and the Eu concentration [51, 52]. The emission band is abnormally narrow and the FWHM value is only 55 nm. The narrow band results in a higher color purity of β-sialon:Eu in comparison with green-emitting $Y_3(Al,Ga)_5O_{12}$:Ce and $(Sr,Ba)_2SiO_4$:Eu phosphors. On the other hand, β-sialon:Eu does not absorb the blue light strongly (see Fig. 11.14), leading to the external quantum efficiency less than 50 %.

The z value and the Eu concentration in β-sialon are optimized as 1.0 and 0.3 mol%, respectively [51, 52]. The absorption is inevitably low because of the low Eu concentration. It is reported that Eu ions are accommodated in the large channel parallel to the c-axis, and are coordinated by six (N,O) anions at equal distances [53, 54].

Under the 450 nm excitation, the internal quantum efficiency and external quantum efficiency of β-sialon:Eu are 49 and 33 %, respectively. The external quantum efficiency is greatly enhanced ($\sim$50 %) by the compositional tailoring and the processing control. Like α-sialon:Eu, β-sialon:Eu is also very stable against thermal and chemical attacks. The luminous efficiency of β-sialon:Eu basically does not change over time and temperature, making it more suitable for use in wLED backlights [55]. Due to its narrow emission band, β-sialon:Eu is not the best choice for use in general lighting.

Eu^{2+}-doped nitridosilicates $CaAlSiN_3$, $M_2Si_5N_8$ (M = Ca, Sr, Ba) and $CaSiN_2$ are red-emitting phosphors that have the similar excitation and emission spectra [56–59]. As presented in Fig. 11.15, $CaAlSiN_3$:Eu considerably extends its excitation spectrum from UV to the red spectral region. The emission band is centered at 660 nm for $CaAlSiN_3$:Eu (0.8 mol%), and can be blue-shifted to 620 nm when Sr substitutes for Ca [60]. Besides these excellent photoluminescence properties, $CaAlSiN_3$:Eu shows higher thermal quenching temperature than the $Ca_2Si_5N_8$:Eu

Fig. 11.15 Excitation and emission spectra of (Ca,Sr)AlSiN$_3$:Eu. Emission spectrum from right to left: CaAlSiN$_3$, (Ca$_{0.6}$Sr$_{0.4}$)AlSiN$_3$, (Ca$_{0.4}$Sr$_{0.6}$)AlSiN$_3$, and (Ca$_{0.1}$Sr$_{0.9}$)AlSiN$_3$
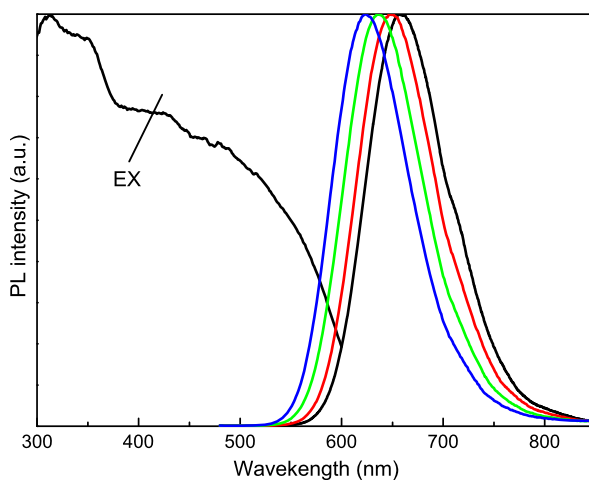
counterpart. In addition, CaAlSiN$_3$:Eu is more stable than the Sr$_2$Si$_5$N$_8$:Eu red phosphor.

The luminous efficiency CaAlSiN$_3$:Eu is much higher than most (oxo)nitridosilicate phosphors. The absorption and external quantum efficiency of CaAlSiN$_3$ (1.0 %Eu) are 86 and 70 % under the blue light excitation (λ_{em} = 450 nm), respectively. On the other hand, CaAlSiN$_3$:Eu has an excitation spectrum tail largely covering the green spectral region, therefore it will also absorb the emission from the combined green phosphor and reduce the total green light of wLEDs. To solve this problem, a larger quantity of the green phosphor should be used.

Photoluminescence properties of other interesting (oxo)nitridosilicate phosphors are summarized in Table 11.2.

11.3 Color Issues and Luminous Efficacy of White LEDs

For all types of white LEDs, color rendering and luminous efficacy are two most important criteria for general lighting, which determine the optical quality and the application purpose of wLEDs. There is a trade-off between these two important criteria because color rendering and luminous efficacy are best achieved by different shapes and broadness of the emission spectra of wLEDs.

The color rendering of wLEDs will be discussed in detail in Chap. 13.

11.3.1 Color Rendering

Color rendering of a light source is determined by comparing the appearance of varying object colors illuminated by the given light source with that illuminated by reference illuminant. The reference illuminant is the Planckian radiation for test

sources with a correlated color temperature (CCT) of <5000 K, or a phase of daylight for test sources with CCT > 5000 K. In the standardized method, the *Color Rendering Index* (CRI), fourteen Munsell samples of various different colors were carefully chosen [73, 74]. The color differences, denoted as ΔE_i, of these color samples under a given illumination and under the reference illumination are calculated on the 1964 $W^*U^*V^*$ uniform color space. The *Special Color Rendering Index*, R_i, for each sample is then given by [75]

$$R_i = 100 - 4.6\Delta E_i \quad (i = 1, \ldots, 14) \tag{11.1}$$

The R_i value is an indication of color rendering for each particular color. The *General Color Rendering Index, R_a*, is often used to evaluate the color rendering of a light source. R_a is calculated by averaging the values of R_i for the first eight color samples that have the medium color saturation.

$$R_a = 1/8 \times (R_1 + R_2 + \cdots + R_8) \tag{11.2}$$

The low R_a value means that some colors (e.g., red and green) may appear unnatural when illuminated by the light. So, high values of CRI or R_a are always pursued for general lighting. Basically, $R_a > 70$ is considered acceptable for illumination purposes, values greater than 80 are regarded as high quality and suitable for interior lighting, and $R_a > 95$ is suitable for visual inspection purposes. High color rendering is realized by broadband spectra covering the whole visible region. Therefore, the spectral distribution of wLEDs should be designed or modified to achieve the R_a value required for the application in mind.

11.3.2 Luminous Efficacy

Luminous efficacy of a light source, or *luminous efficacy of radiation*, is the efficiency of energy conversion from electrical power to optical power, or from radiant flux to luminous flux. The *luminous efficacy of radiation*, in lm/W (lumen per watt), describes the brightness of the radiation perceived by the average human eye, and is calculated by the eye sensitivity over the entire spectral distribution of light through the following equation [76]:

$$LER \ (\text{lm/W}) = 683 \ \text{lm/W} \times \frac{\int_{380}^{780} V(\lambda) S(\lambda) d\lambda}{\int_0^\infty S(\lambda) d\lambda} \tag{11.3}$$

$V(\lambda)$ is the spectral luminous efficiency of photopic vision (eye sensitivity curve) defined by CIE which is defined in the visible range of 380 to 780 nm, as shown in Fig. 11.16. $S(\lambda)$ is the *spectral power distribution* (SPD), which describes the power per unit area per unit wavelength of wLEDs. The theoretical limit of the luminous efficacy of monochromatic green radiation (555 nm) is 683 lm/W, whereas it is only 26 lm/W for a blue monochromatic light at 450 nm, since the eye sensitivity peaks at 555 nm. For creating white light, blue and red components in the emission spectrum of wLEDs are essential, so the luminous efficiency of wLEDs is always considerably lower than the theoretical limit.

Fig. 11.16 Eye sensitivity
spectral curve for photopic
vision $V(\lambda)$
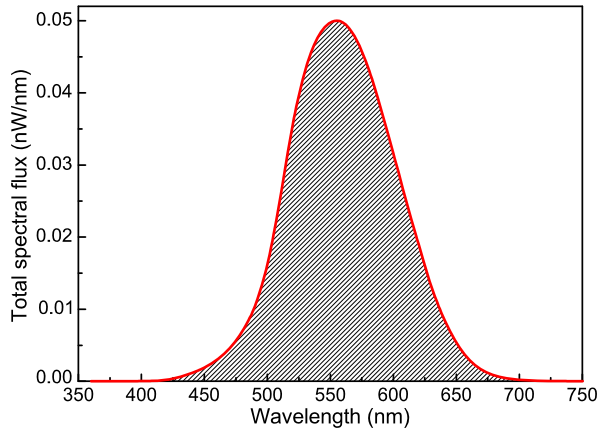
11.3.3 Chromaticity Coordinates and Color Temperature

White light generated by wLEDs is required to show all the colors of illuminated objects appropriately, so that the white color should be reasonable and acceptable. To quantitatively describe the color of the white light, chromaticity coordinates and color temperature are utilized.

For a given light the spectrum $S(\lambda)$ can be weighted by the *XYZ color matching functions* as seen in Fig. 11.17. The resultant three weighted integral values X, Y, Z, also called triastimulus values, are obtained by using the following equations [77].

$$X = \int \overline{x}(\lambda)S(\lambda)d\lambda \tag{11.4a}$$

$$Y = \int \overline{y}(\lambda)S(\lambda)d\lambda \tag{11.4b}$$

$$Z = \int \overline{z}(\lambda)S(\lambda)d\lambda \tag{11.4c}$$

The *chromaticity coordinate x, y* of the given light is thus calculated by

$$x = \frac{X}{X + Y + Z} \tag{11.5a}$$

$$y = \frac{Y}{X + Y + Z} \tag{11.5b}$$

Any color of light can be expressed by the chromaticity coordinate (x, y) on the CIE 1931 chromaticity diagram, as shown in Fig. 11.18. In the horseshoe-shaped diagram, the boundaries are called *spectrum locus*, representing the monochromatic light. *Planckian locus* or *blackbody locus* is plotted near the center of the diagram, which represents the chromaticity coordinates of a blackbody having various temperatures.

The colors on the Planckian locus can be specified by *color temperature* in Kelvin [78]. The colors around the Planckian locus from the color temperature of

Fig. 11.17 CIE 1931 XYZ color matching functions
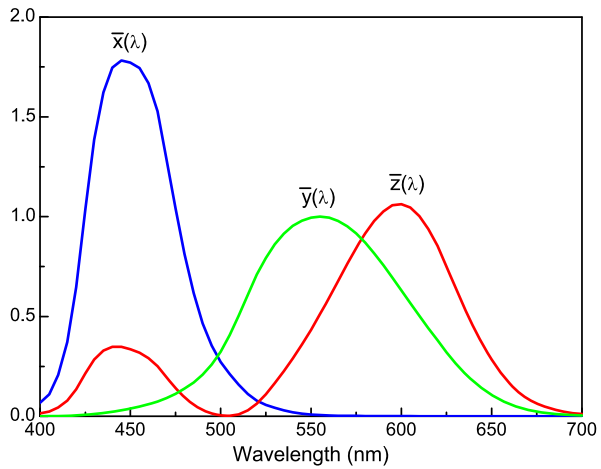
Fig. 11.18 1931 CIE chromaticity diagram
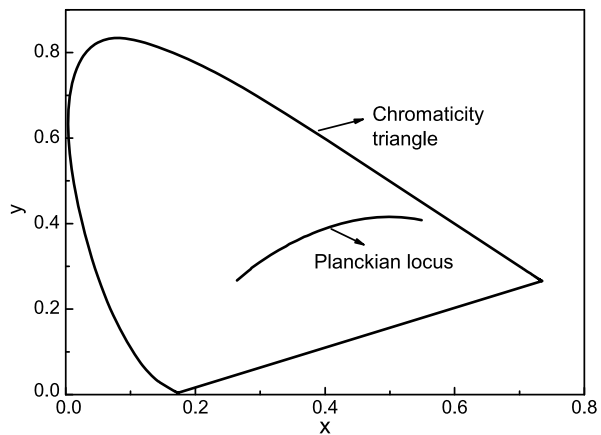
~2500 to 20000 K are recognized as *white*. For most of traditional lamps their colors fall in the region between 2850 and 6500 K. The color of white light can be purposely shifted along the Planckian locus (i.e., from warm to cool) for desired mood. However, the color shifted away from the Planckian locus is not regarded as white, and hence is hardly accepted for general lighting. Strictly speaking, color temperature cannot be used for the colors whose chromaticity coordinates are beyond the Planckian locus. In this case, correlated color temperature (CCT) is used instead. The CCT for any white light can be determined by drawing a perpendicular line from the measured 1931 CIE chromaticity coordinates to the Planckian locus to obtain the (x, y) intercept. Then the CCT of the white light would correspond to the surface temperature of a blackbody radiator with the same (x, y) chromaticity coordinates.

The chromaticity coordinate and color temperature of four standard white points are shown as below.

Tungsten lamp: 2865 K, $x = 0.448$, $y = 0.408$
Direct sunlight: 4870 K, $x = 0.348$, $y = 0.352$
Overcast sunlight: 6770 K, $x = 0.310$, $y = 0.316$
D65 (daylight): 6504 K, $x = 0.313$, $y = 0.329$

11.4 White LED Packaging

As mentioned previously, white light can be generated by multi-chip or phosphor-converted white LEDs. The multi-chip white LEDs have advantages of high luminous efficacy and high color rendering index, but they are costly. In addition, the choice of monochromatic LED chips is strictly limited. On the other hand, pc-wLEDs are cheaper and allow a much more flexible design in spectra as phosphors are easily selected and combined. Furthermore, they are getting more efficient and reliable, making them to be the mainstream products of the market.

Besides the above two packaging methods, white light generated by using nanotechnology such as quantum dots (QDs) and quantum wells (QWs) is also extensively investigated. It is well known that the size-tunable emission wavelength is usually observed in low-dimensional structures (i.e. QDs or QWs), which is ascribed to the quantum confined Stark effect which is induced by the presence of large piezoelectric field [79]. The generation of white light by stacking the GaN/AlGaN QDs with different sizes, as well as the incorporation of InGaN/GaN QWs with different thicknesses in the active region has been intensively investigated [80–84].

The manufacturing of wLEDs consists of several steps, but not limited to: die attach, wire bonding, wavelength-conversion materials application, encapsulation, curing, optic lens attach, singulation, testing, binning and taping. The wLED products can be divided into various groups depending on the *chip type* (i.e., ultraviolet, near-ultraviolet, or blue), *wavelength-conversion materials (phosphors or QDs) and their locations*, the *input power* (low or high-power), and the *construction*. From the viewpoint of conversion materials and configurations, we will group the wLEDs into "phosphor-in-cup" white LEDs, "remote-phosphor" white LEDs and QDs white LEDs, and discuss them in the following sections.

11.4.1 Phosphor-in-Cup White LEDs

Usually phosphor(s) is dispersed in the transparent epoxy resin, and then mounted on the LED chip. In the case of phosphor(s) dispersed epoxy resin is uniformly distributed in the reflector cup (see Fig. 11.19), such a LED configuration is called "*phosphor-in-cup*" [85]. In an extreme case, a "*conformal*" configuration is recognized when the phosphor(s) is not homogeneously distributed in the epoxy resin but conformally coated on the LED chip (phosphor replicating the contour of the chip) [85].

Fig. 11.19 Schematic of white LEDs with the phosphor-in-cup configuration
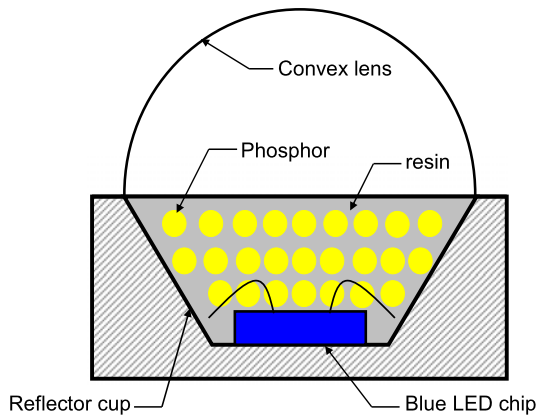
Table 11.3 Color rendering indices of white LEDs using $Y_3Al_5O_{12}$:Ce

R_a	R_1	R_2	R_3	R_4	R_5	R_6	R_7	R_8	R_9
71	67	74	80	71	68	65	82	57	−39

The most popular way to fabricate "phosphor-in-cup" wLED is using a blue LED chip in conjunction with a yellow-emitting phosphor. Such a LED type is also known as "bichromatic" or "one-phosphor-converted (1-pc)" wLEDs. The commonly used yellow phosphor is cerium-doped yttrium aluminum garnet ($Y_3Al_5O_{12}$:Ce). As this method is patented by the Niichia company, to avoid the IP problems, alternative yellow phosphors, such as orthosilicates and (oxo)nitridosilicates, have been developed.

Figure 11.20 shows a typical emission spectrum of wLEDs using a single YAG:Ce phosphor. It consists of two bands: the narrow blue band is the emission spectrum of the LED chip, and the broad yellow band is from the YAG phosphor. For the $Y_3Al_5O_{12}$:Ce phosphor, the wLED has a correlated color temperature of 5200 K, the color rendering index R_a of 71, and the chromaticity coordinates of $x = 0.340$, $y = 0.354$. Krames et al. [4] reported a similar YAG-based wLED with CCT = 5600 K. The first 9 special color rendering indices of the wLED are shown in Table 11.3. To obtain warm white light, the emission wavelength of $Y_3Al_5O_{12}$:Ce needs to be red-shifted by partial substitution of Gd for Y. So the use of the modified $(Y,Gd)_3Al_5O_{12}$:Ce phosphor leads to a lower color temperature of 2960 K. At the same time, the color rendering index is reduced to 63 due to the larger gap in between the blue band and the yellow band. As the thermal quenching of $(Y,Gd)_3Al_5O_{12}$:Ce is higher than that of $Y_3Al_5O_{12}$:Ce, the $(Y,Gd)_3Al_5O_{12}$:Ce-based LED achieves warm white by sacrificing reliability or lifetime.

Warm white can also be realized by using orange-yellow phosphors, such as Sr_3SiO_5:Ce^{3+} [28], Eu^{2+}-doped Ca–α-sialon [38, 44, 48], and Ce^{3+}-doped CaAlSiN$_3$ [70], and La$_3$Si$_6$N$_{11}$ [69]. As these (oxo)nitridosilicate or orthosilicate phosphors have higher thermal quenching temperature than $(Y,Gd)_3Al_5O_{12}$:Ce, the

Fig. 11.20 Emission spectra of white LEDs using $Y_3Al_5O_{12}$:Ce and $(Y,Gd)_3Al_5O_{12}$:Ce, respectively. The bias current is 20 mA
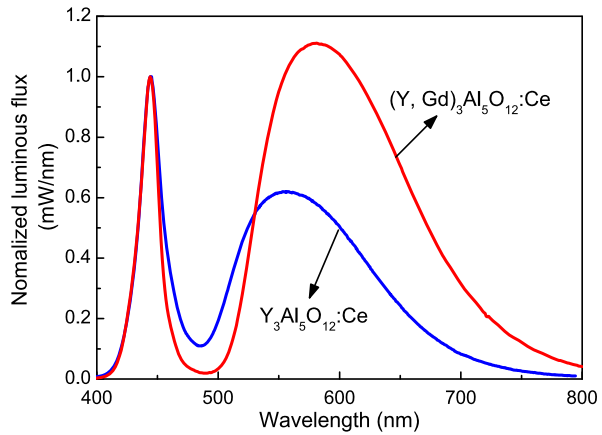
Table 11.4 Color rendering indices of white LEDs using $CaAlSiN_3$:Ce

R_a	R_1	R_2	R_3	R_4	R_5	R_6	R_7	R_8	R_9
70	71	78	77	65	67	63	79	63	8

warm white LEDs using them have stabilized luminous efficacy and chromaticity coordinates when the temperature changes. Sakuma et al. [48] addressed that the chromaticity coordinates of Ca–α-sialon:Eu-based wLEDs were only 1/3–1/4 of those of YAG-based ones when the temperature increases. In Fig. 11.21 is shown the white emission spectrum for blue (∼450 nm) pumped LED based on orange-yellow $CaAlSiN_3$:Ce. The wLED has the color temperature of 3700 K, the color rendering index of 70, and the luminous efficacy of 51 lm/W, respectively [70]. As given in Table 11.4, the color rendering index R_9 is significantly improved (compared to YAG-based wLED), since the red spectral part of the emission spectrum is enhanced.

Although the use of a single yellow phosphor in combination with a blue LED results in high luminous efficacy, the color rendering index is less than 80. To enhance the color rendering, it is necessary to utilize red and green phosphors, i.e., the multi-phosphor approach (see Fig. 11.22). In this approach, green and red phosphors, or green, yellow and red phosphors are blended with the epoxy resin, and then mounted on the LED chip. In addition to realizing high color rendering index, the color temperature of white light could also be varied by controlling the phosphor blend ratio.

In Fig. 11.23 are shown the luminescence spectra of warm white LEDs using multi-phosphors. High color rendering index is then expected (see Table 11.5). Typically, the deep red color rendering for the multi-phosphor-converted white LEDs is excellent [86, 87]. The emission spectrum of the 4-pc white LED nearly mimics the spectrum of the CIE standard illuminant A, leading to a super-high color rendering index of 98 [86]. The color temperature is 2840 and 2900 K for the 3-pc and 4-pc

Fig. 11.21 Emission spectra of white LEDs using CaAlSiN$_3$:Ce. The bias current is 20 mA
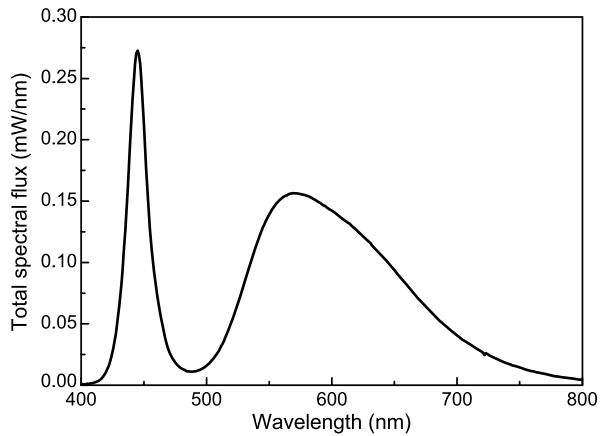
Fig. 11.22 Schematic of white LEDs with the phosphor-in-cup configuration
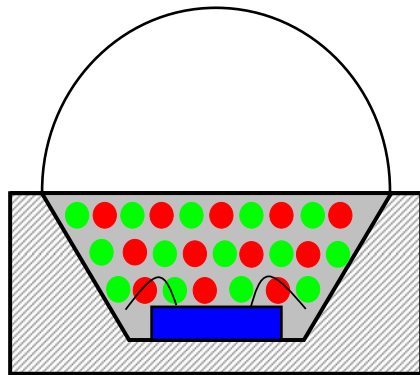
Table 11.5 Color rendering indices of 3-pc and 4-pc warm white LEDs

	R_a	R_1	R_2	R_3	R_4	R_5	R_6	R_7	R_8	R_9
3-pc	88	97	90	73	92	91	80	89	91	96
4-pc	98	99	99	98	97	98	99	96	94	88

white LEDs, respectively. Compared to the 1-pc white LEDs, the multi-phosphor-converted white LEDs have lower luminous efficacy. For example, the luminous efficacy of the 4-pc warm white LED is 30 lm/W, but it far exceeds the incandescent lamps ($\sim$15 lm/W). The lower luminous efficacy for the multi-phosphor-converted white LEDs is due to increased re-absorption and optical losses, and lower lumen equivalent.

The "hallo effect" or bleed-through effect occurs in the architecture using the blue LED chip because the light from the blue LED is directional while the light from the phosphor radiates over a $2p$ solid angle. In addition, much higher color rendering index ($R_a > 95$) is usually not realized by using the blue LED chip be-

Fig. 11.23 Emission spectra
of the 3-pc warm white LED
using β-sialon:Eu,
Ca–α-sialon:Eu and
CaAlSiN$_3$:Eu, as well as the
4-pc warm white LED using
BaSi$_2$O$_2$N$_2$:Eu, β-sialon:Eu,
Ca–α-sialon:Eu and
CaAlSiN$_3$:Eu. The bias
current is 20 mA
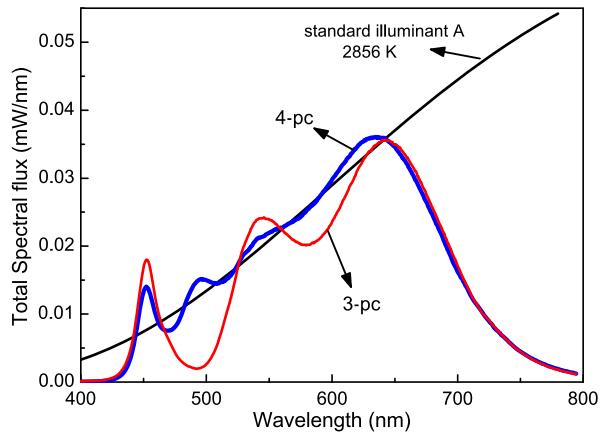
cause there is a large gap in between the blue and green emission spectra (i.e., the
number of green phosphors with $\lambda_{em} < 525$ nm is extremely limited). To avoid the
hallo effect and further improve the color rendering index, it is required to use the
UV or near UV LED chips [61].

Table 11.6 lists some examples of white LEDs by using difference phosphor
combinations.

11.4.2 Remote-Phosphor White LEDs

The "phosphor-in-cup" configuration has the advantages of compactness and ease of
mass production. On the other hand, the disadvantages of this configuration include
the photodegradation of the epoxy resin (yellowing) and low phosphor extraction
efficiency. The yellowing of the epoxy resin is caused by the circulation of the emit-
ted light from the phosphor and the unconverted blue light between the phosphor-
distributed epoxy and the LED reflector. Moreover, the heat generated by the LED
junction accelerates the degradation of the epoxy. The efficiency loss of phosphor is
attributable to the reflection of the blue light back into the LED chip and the large
fraction of re-emitting light directed into the LED chip. It has been reported that
~60 % of the re-emitted light backscatters to the LED chip, which significantly
decreases the luminous flux. Therefore, separating of the phosphor-dispersed resin
from the LED chip by a certain distance is one of the approaches to alleviate the
yellowing of the resin and the reduction of the absorption of the re-emitting light
by the LED chip. This packaging mode is named the "remote-phosphor" configura-
tion [85], as shown in Fig. 11.24.

Kim et al. [100] addressed that the conversion efficiency of the blue-pumped
YAG:Ce yellow phosphor was enhanced by 15.4 % by using the remote-phosphor
configuration and diffuse reflector cup. Zhu et al. [101] suggested that for remote-
phosphor white LEDs with multi-phosphor layers the highest light output was real-
ized by placing the red phosphor as the second layer. And he also pointed out that

Table 11.6 Examples of phosphor-converted white LEDs

λ_{LED} (nm)	Phosphor(s)	η (lm/W)	CRI	CCT (K)	Ref.
385	$La_{0.827}Al_{11.9}O_{19.09}$:Eu,Mn	–	–	3100–3600	[88]
395	$Sr_6B_5O_{20}$:Eu^{2+} + $(Sr,Ca)_5(PO_4)_3Cl$:Eu^{2+}, Mn^{2+}	–	87.3	5686	[89]
405	α-$Ca_2P_2O_7$:Eu,Mn + Ba_2SiO_4:Eu	9	78	10465	[90]
405	Sr_3SiO_5:Ce,Li	25.0	69	–	[28]
405	JEM:Ce + β-sialon + α-sialon + $CaAlSiN_3$:Eu	19–20	95–96	2830–4300	[61]
450	Ca–α-sialon:Eu	25.9	61	2750	[48]
450	Li–α-sialon:Eu	46–55	60–65	3000–5200	[47]
450	$(La,Ca)_3Si_6N_{11}$:Ce	34.8	65	2911	[69]
450	$CaAlSiN_3$:Ce	50	70	3722	[70]
450	$Sr_{1.95-x}Eu_{0.05}Bi_xSiO_4$	10–25	65–68	8000–10000	[91]
450	$Sr_3Si_{13}Al_3O_2N_{21}$:Eu + $(Sr,Ca)_2SiO_4$:Eu	56–62	82–88	3230–6450	[68]
450	Ca–α-sialon:Yb + $Sr_2Si_5N_8$:Eu	17–23	82–83	2744–6508	[92]
450	$Sr_2Si_5N_8$:Eu + YAG	94	72	3791	[93]
450	$Sr_2Si_5N_8$ + $SrSi_2O_2N_2$:Eu	25	>90	3200	[94]
450	$Ca_3Sc_2Si_3O_{12}$:Ce + $(Sr,Ca)AlSiN_3$:Eu	–	90	–	[95]
450	$(Sr,Ca)_3(Al,Si)O_4(F,O)$:Ce + K_2TiF_6:Mn^{4+}	82	90	3088	[96]
450	β-sialon + α-sialon + $CaAlSiN_3$:Eu	24–28	81–88	2840–6580	[87]
450	$BaSi_2O_2N_2$ + β-sialon + α-sialon + $CaAlSiN_3$:Eu	25	95–98	–	[86]
455	$SrSi_2O_2N_2$:Eu + $CaSiN_2$:Ce	30	90.5	5206	[97]
460	Sr_3SiO_5:Ce,Li	31.7	81	–	[28]
460	$La_{1-x}Ce_xSr_2AlO_5$	10–17	80–82	4340–5122	[98]
460	YAG:Ce + Sr_2SiO_4:Eu	40.9	76	5716	[99]
460	$Tb_3Al_3O_{12}$:Ce	34.1	79	5026	[99]
460	Sr_3SiO_5:Eu	41.3	57	3387	[99]
460	YAG	38.9	75	7756	[99]

the performance of the white LEDs was controlled by several factor, such as mixture or stacked layers, specific layer order, phosphor density, phosphor external quantum efficiency, overall spectral power distribution, phosphor photoluminescence spectra. A lighting system based on the white LEDs using a remote phosphor film (YAG:Ce) was investigated by Huang et al. [102]. Compared with the lighting system using the traditional white LED with the phosphor-in-cup configuration, the novel system yielded higher lumen efficiency (73 lm/W vs. 64.7 lm/W), less angular color deviation, and more uniform luminous distribution (Uniformity = 82 % vs. 12 %).

In addition to the use of the mixture of phosphor powders and the organic resins, some other forms of the remote luminescence layer are also applied, such as transparent YAG:Ce ceramics, glass-ceramic phosphors and phosphor(s)/glass compos-

Fig. 11.24 Schematic of the remote-phosphor white LED configuration
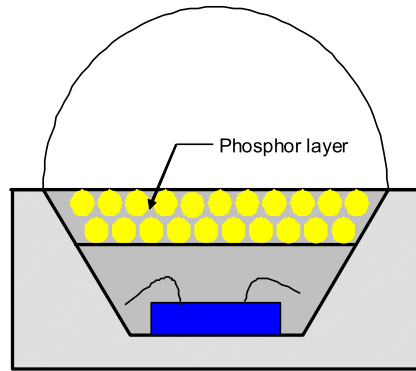
Fig. 11.25 Emission intensity change in the heat-resistant test of the YAG glass-ceramic phosphor and commercial white LEDs (Reprinted from Fujita, S., Yoshihara, S., Sakamoto, A., Yamamoto, S. and Tanabe, S. 2005. YAG glass-ceramic phosphor for white LED (I): background and development. Proc. of SPIE. 5941: 594111-1. With permission)
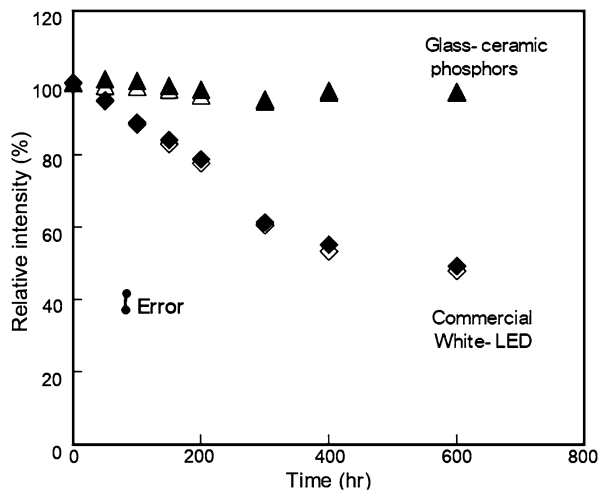
ites. Nishiura et al. [103] prepared a transparent yellow YAG:Ce ceramic which has the emission peak at 530 nm and transmittance of 70–87 % at 800 nm. The white light was achieved by packing the ceramic with a blue LED ($\lambda_{em} = 465$ nm), and the emission color could be adjusted by controlling the thickness of the transparent ceramic. The luminous efficacy of 73.5 lm/W was obtained for the ceramic with the thickness of 0.632 nm. Tanabe et al. [104] suggested that the YAG glass-ceramic phosphors (Ce^{3+}-doped glasses in SiO_2–Al_2O_3–Y_2O_3 and SiO_2–Al_2O_3–Y_2O_3–Gd_2O_3 systems) could be used in high-power white LEDs. The quantum efficiency of the glass-ceramic phosphor is around 30 %. Moreover, the glass-ceramic phosphor has the high thermal conductivity of 2.2 $W\,m^{-1}\,K^{-1}$ (compared to 0.19 $W\,m^{-1}\,K^{-1}$ of the resin) as well as a high bending strength of 125 MPa (compared to 93 MPa of the resin). Fujita et al. [105] reported that the luminous efficacy of the white LED using the glass-ceramic phosphor achieved 80 % of that using the phosphor powder. Furthermore, the white LED using the glass-ceramic phosphor exhibited higher thermal resistance than the commercial white LEDs us-
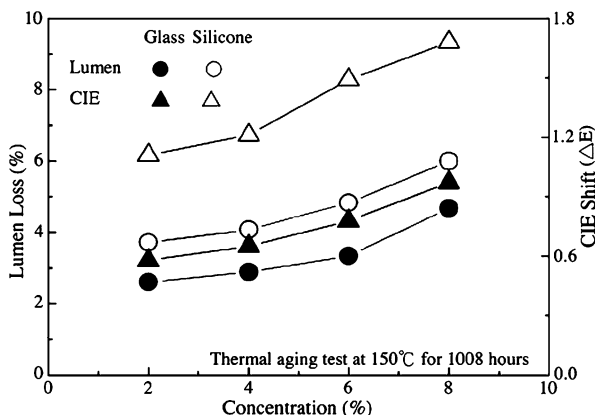
Fig. 11.26 Lumen loss and CIE shift as a function of the YAG:Ce concentration in the thermal aging test for white LEDs using the YAG-doped glass and the YAG-distributed silicone resin (Reprinted from Wang, J., Tsai, C.C., Cheng, W.C., Chen, M.H., Chung, C.H. and Cheng, W.H. 2011. High thermal stability of phosphor-converted white light-emitting diodes employing Ce:YAG-doped glass, IEEE J. Selected Topics in Quan. Electron., 17(3): 741–746. With permission)

ing the YAG phosphor powder (see Fig. 11.25). Cheng et al. [106] prepared the YAG-doped low temperature glass (with the glass transition temperature of 400 °C) by distributing the YAG:Ce phosphor powder in a SiO_2–Na_2O–Al_2O_3–Li_2O glass, and firing it at 650 °C. The YAG-doped glass has the luminous efficacy of 51 % with the YAG:Ce concentration of 4 wt%. Wang et al. [107] addressed that white LEDs employing the YAG:Ce-doped glass demonstrated better performances than the conventional ones using the YAG:Ce-distributed silicone resin (see Fig. 11.26), in terms of lumen loss, chromaticity shift, transmittance loss as well as peak emission intensity undergoing three industry-standard reliability tests (i.e., 150 °C thermal aging for 1008 h, 85 % RH/85 °C damp heat for 1008 h, and −40 to 125 °C thermal shock for 200 cycles).

11.4.3 Quantum Dots White LEDs

Semiconductor nanocrystals, i.e., quantum dots (QDs), have been also proposed to be alternative color conversion materials in wLEDs, because they exhibit high photon conversion efficiency (∼80 %), high photostability and wide absorption spectrum. Their optical properties can be conveniently tuned by controlling size and shape, and using different material systems [108]. Moreover, the narrow emission linewidths (FWHM ∼30 nm) enable the custom-design of a desired emission spectrum by using proper combinations of differently sized QDs.

CdSe QDs are one of the extensively investigated nanocrystals with high quantum efficiencies, and their emission colors can be readily adjusted in the range of
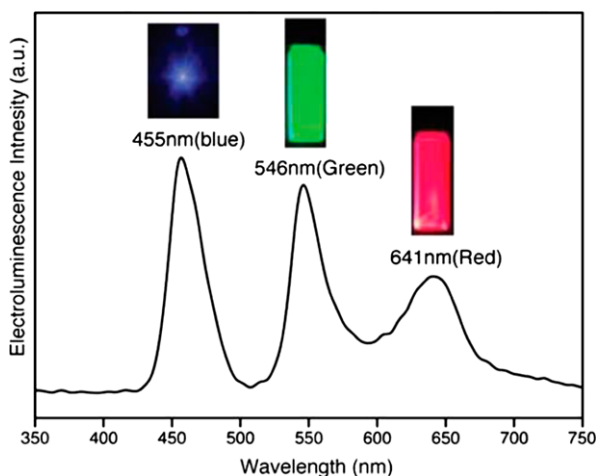
500–650 nm by manipulating the particle size [109–112]. The use of CdSe QDs in
wLEDs has been reported in the literature recently. Chen et al. [113] synthesized
core-shell CdSe–ZnSe QDs which exhibited a high quantum yield of ∼40 % and
size-tunable emission colors (510–620 nm). A three band white LED with the color
rendering index R_a of 91 was attempted by combining CdSe–ZnSe QDs with the
InGaN blue chip (see Fig. 11.27). Jang et al. [110] prepared a high color render-
ing white LED by using $Sr_3SiO_5:Ce^{3+},Li^+$ (yellow) and CdSe QDs (red), the color
rendering index R_a and the luminous efficacy being 90.1 and 14 lm/W respectively.
Lita et al. [114] synthesized a CdSe-silica nanocomposite phosphor, and packed it
with the blue GaN chip to obtain stable and efficient wLEDs.

As the Cd-based compounds are not green materials, Cd-free QDs with high
quantum yields are thus required. Zhang et al. [115] reported white LEDs using
heavy metal-free quantum dots. White light was obtained by mixing the red emis-
sion of ZnCuInS/ZnS core/shell QDs and the blue-green emission of poly(N,N″-
bis(4-butylphenyl)-N,N″-bis(phenyl)benzidine) with blue LED chips. A high color
rendering index of 92 was achieved as compared to a 5310 K blackbody reference.

White QD-LEDs are very suitable for use in the display applications because of
their high color rendering properties. Jang et al. [112] prepared highly luminescent
core-shell CdSe/ZnS/CdSZnS (green) and CdSe/CdS/ZnS/CdSZnS (red) QDs with
external quantum efficiency of 72 and 34 %, respectively. A white QD-LED back-
light with the color gamut of 100 % of the NTSC standard and a luminous efficacy
of 41 lm/W was achieved, which was then applied in a 46 inch LCD TV panel (see
Fig. 11.28). By using the white QD-LEDs, Nanosays also announced a LCD back-
light product which produced a triangular color gamut exceeding 100 % of Adobe
RGB and provided an improvement of 20 % in power efficiency compared to exist-
ing backlight solutions [116].
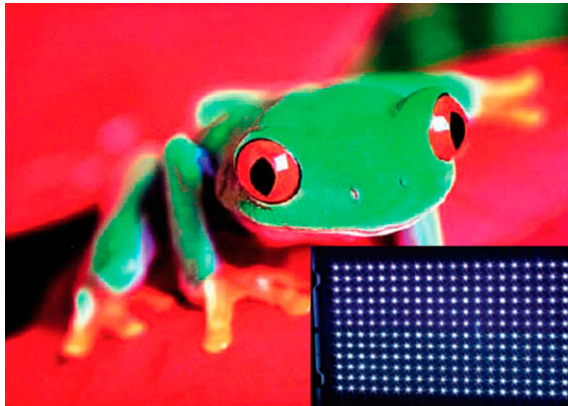
11.5 Summary and Perspective

Recent trends and developments in LED phosphor materials and white LED packaging are overviewed in this chapter. Although a large quantity of phosphors have been investigated as down-conversion luminescent materials for white LEDs, only a very limited number of them can be used practically. To judge whether the phosphors are valuable for wLEDs or not depends mainly on their photoluminescence spectra, quantum efficiency, thermal quenching/degradation, chemical stability, and etc. According to these requirements, the selection rules of host crystals and activator ions have been proposed.

The rare-earth phosphors that are commercially available now include yttrium aluminum garnets, alkaline earth orthosilicates, and (oxo)nitrido(alumino)silicates. Among these, (oxo)nitrido(alumino)silicates demonstrate their superior suitability in producing highly reliable and efficient wLEDs with high color rendition and tunable color temperatures. Luminescent quantum dots with very slim emission spectra and high efficiencies have been shown as promising alternative color-conversion materials in lighting and displays, but the production cost, stability/reliability and the limited host crystals are the major problems for QDs.

Novel phosphor materials with excellent photoluminescence properties are continuously pursued by material scientists and chemists, with focus on understanding the crystal structure, photoluminescence spectra, and the relationship between them. In addition to the searching for new materials, the modifications of the existing phosphors are also the interesting topics, such as the enhancement of photoluminescence properties, tuning of the emission colors, particle size and morphology control, optimization of processing conditions, and so on.

Considering their use in the LED packaging, phosphors are also required to have the refractive index close to that of epoxy resin to enhance the luminous efficacy of wLEDs. However, the refractive index is not available for most phosphors. In addition, the phosphors including (oxo)nitridosilicates have the thermal quenching/degradation problems when they are combined with LED chips. It is thus necessary to understand the quenching/degradation mechanisms through the reliability

test (for example, high temperature and high humidity test, thermal attack test, and etc.) and finally to solve the problems.

The optical quality of wLEDs (i.e., luminous efficacy, chromaticity, lifetime, and etc.) is significantly dependent on chip processing, packaging configuration, heat sink material/structure, phosphor materials, and the epoxy resin. The "remote-phosphor" wLEDs reduce the re-absorbed light by the LED chip as well as the thermal attacks on phosphors and the epoxy resin, and show higher luminous efficacy than traditional "phosphor-in-cup" wLEDs. By using heat sink materials and structures, the thermal effect on the packaging materials, such as the yellowing of the epoxy resin, will be significantly reduced. In addition, to achieve high performance of wLEDs, the phosphor concentration, geometry, and particle size/morphology should be carefully considered [117–120]. The low concentration phosphor mixture employed in LED packaging usually leads to flux enhancement and stable color point, due to the reduced re-absorption of the phosphor mixture and the absorption saturation. Further, the LED chip/green/red phosphors geometry is superior to the chip/phosphors (mixed randomly) geometry in reducing the re-absorption of green light by the red phosphor.

Much attention is now paid to the reliability assessments of wLEDs. Degradations induced by external stresses (i.e., temperature and moisture) dramatically affect the long lumen maintenance of wLEDs, leading to the lumen loss of LED chips, the transparency reduction and refractive index change of the resin, and the compositional change and lumen loss of phosphors as well [121]. Therefore, to truly understand the factors determining the reliability and use them to set an appropriate warranty period can reduce the risks of a higher than expected rate of claim against the warranty and the end users' dissatisfaction with the products. The reliability of wLEDs is thus becoming an urgent topic that needs to be strengthened.

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Chapter 12
High Voltage LEDs

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Abstract In this chapter, the basic mechanisms, opto-electrical characteristics, variety designs, and development process of the high voltage light emitting diode (HVLED) and alternating current light emitting diode (ACLED) were extensively discussed. Some important issues, such as the efficiency drooping, flickering illumination, harmonic distortion limitation, and lifetime result of the HV/ACLED were also studied. Furthermore, packaging methods, safety considerations, and measurement techniques of the HV/ACLED were mentioned to assist readers in applying the HV/ACLED safely and properly.

12.1 Introduction of Assembled HVLED Modules and Single-Chip HVLEDs

With the phasing out of incandescent light bulbs, the development of high efficiency, low pollution alternative light sources has become a common objective of the world's leading manufacturers and research institutes. In particular, light emitting diodes (LEDs), with their high efficiency, lengthy lifetime, compact size, absence of mercury, etc., are widely considered to be the most promising candidates for becoming the mainstream light sources of the future. Nonetheless, for end-users, the decision on the kind of new lighting products to purchase hinges not only on factors such as efficiency and environmental friendliness but also on prices and compatibility with existing equipment. In view of this, researchers, while continuing their works on improving LED luminous efficacy, should at the same time ponder how end-users can more conveniently utilize LED lighting products in their everyday lives.

Due to the inherent limitations imposed by the physical mechanism underpinning the operation of a LED, the typical operating voltage of a single LED die, depending on the materials utilized in the fabrication process, falls between 1.8–3.3 V. The mains voltages worldwide range from 100–220 V AC, and the use of a single LED

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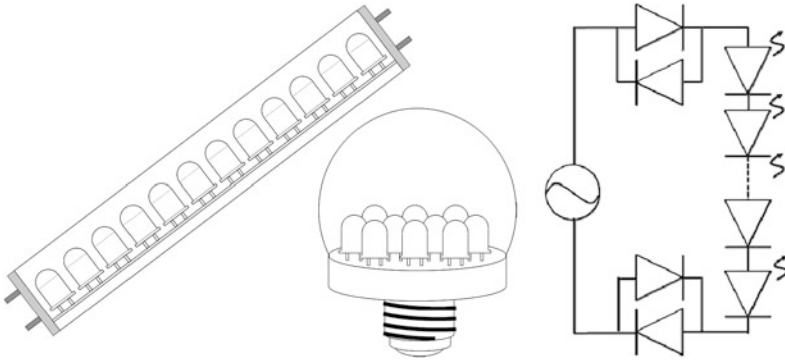
Fig. 12.1 Schematic diagrams for a HVLED lighting modules

die for lighting purposes at such mains voltages will require the coupling of the single LED die with rectifier and transformer elements. However, the addition of rectifier and transformer elements will inevitably increase the cost and scale of the LED lighting module. Additionally, the shorter lifetimes of the rectifier and transformer elements compared to those of the LEDs, as well as the waste heat generated during the operation, will further lower the efficiency and shorten the lifetime of the entire LED lighting module. In view of these shortcomings, researchers and manufacturers alike have attempted to develop LED lighting modules operating at mains voltages and with less reliance on electronic devices.

Figure 12.1 are schematic diagrams of fluorescent lamp type and light bulb type high voltage LED (HVLED), or also so-called alternating current LED (ACLED), lighting modules obtained by assembling several encapsulated LED dies on a circuit board. By connecting the LED dies in series via a circuit board, one enables their use in a high voltage AC environment with only rectifier elements. In comparison with traditional high power LED dies which easily consume hundreds of milliampere of drive current, assembled HVLED modules not only eliminate the use of transformers for operations but also reduce transmission line losses due to their low drive currents, thereby further enhancing the luminous efficacy of the LED modules. On the other hand, assembled HVLED modules also suffer from several shortcomings; for example, they are difficult to reduce in size and their light emission mode is similar to that of flat light sources and thus limits the scope and flexibility for application.

To resolve the aforementioned issues, the concept of single-chip HVLED die (henceforth referred to as HVLED) was proposed. Compared to assembled HVLED modules, HVLEDs not only retain the advantage of LED operation at high voltage, but also are compact and possess a concentrated emission area, rendering them eminently suitable for use as light source elements in such standard lighting lamps as the A19, PAR38, and MR16. In this chapter, we shall discuss in details the development processes as well as operating characteristics of HVLEDs, and compare the differences in the opto-electrical properties of HVLED and traditional low voltage DCLED dies during practical use.
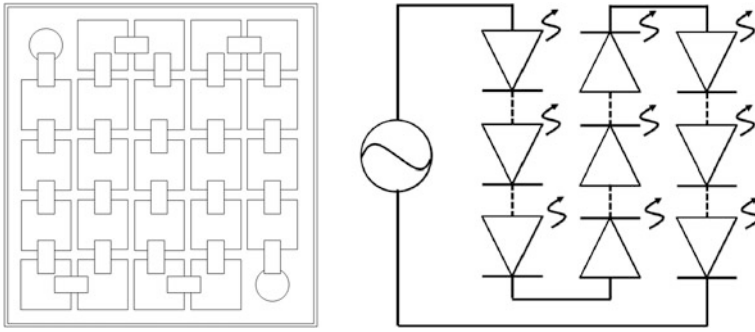
Fig. 12.2 The top view of a HVLED structure

12.2 Fabrication, Development, and Characteristics of HVLEDs

HVLEDs are fabricated with semiconductor manufacturing processes, through which several micro-chips are formed on the same substrate and electrically connected with metal wires. The fabrication technology is very similar to that utilized in the fabrication of traditional LEDs. However, the distinct drive modes, especially under AC driving conditions, necessarily mean that HVLEDs will exhibit distinct characteristics from traditional LEDs and thus require significantly different design considerations. In this section, we shall describe the fabrication flow of HVLEDs as well as the unique opto-electrical properties of the LED devices under AC, and discuss the development processes and differences of the disparate types of HVLEDs.

12.2.1 Structure and Fabrication of HVLED Micro-Chip

Figure 12.2 shows the structural diagram for a HVLED die. The micro-chips were fabricated on the same substrate and connected via metal wires, while the high voltage drive current is injected into the micro-chip array via the two wire pads at the opposite ends. From the side view of the micro-chip structure in Fig. 12.3, one can see that the structure of a single micro-chip differs from that of a traditional LED only in size, while other features such as the transparent conductive layer, surface roughness, and patterned sapphire substrate, all employed extensively to enhance the efficiency of traditional LEDs, are readily transferable to HVLEDs. The key differences in the fabrication of HVLED and traditional LED dies are the use of the insulating substrate, the etching of the insulation trench, as well as the fabrication of the metal wire.

The core concept of HVLED is to fabricate and connect several micro-chips in series on the same substrate. Therefore, the use of the insulating substrate to ensure electrical insulation between the micro-chips is a fundamental condition for the normal operation of HVLEDs. For LEDs grown with GaN materials, the sapphire substrates used already possess excellent insulating properties, and one can achieve
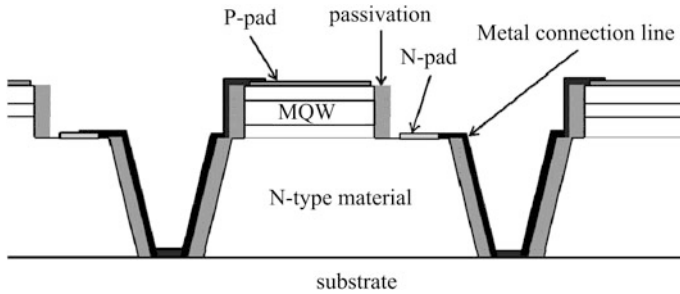
Fig. 12.3 The side view of the structure of a GaN micro-chip grown on a sapphire substrate
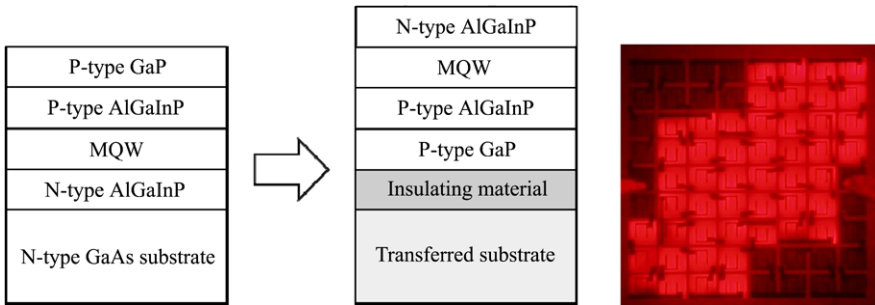
Fig. 12.4 Structural diagram and actual operating conditions of a device where the AlGaInP substrate was replaced with an insulating substrate

excellent electrical insulation simply by etching trenches among the micro-chips until the substrate is exposed. However, most red LEDs grown with AlGaInP materials employ electrically conducting GaAs as the substrate, and one must necessarily resolve the issue of the conductivity of the substrate before the mass production of high voltage AlGaInP LEDs is even possible. The feasible approaches include the use of insulating GaAs materials as the substrate, the transfer of the AlGaInP epitaxial structure to the insulating substrate, or the use of insulating materials to effect the transfer of the AlGaInP epitaxial structure to any arbitrary substrate, etc., all of which will serve the purpose of ensuring electrical insulation among the micro-chips. Figure 12.4 shows the side-view structural diagram and actual operating conditions of an AlGaInP HVLED device where the original substrate was replaced with an insulating substrate.

In addition, from Fig. 12.5, one can see that although the insulation trenches amongst the micro-chips are critical for the normal operation of the HVLED, they also reduce the overall light-emission area of the HVLED die. Although increasingly thin insulation trenches will theoretically increase the light emission area of the HVLED die, the complexity of the fabrication process will also increase accordingly. The sidewalls of the insulation trenches need to be coated with a layer of dielectric materials to avoid a short circuit between the PN materials where the metal wires are deposited across the surface. However, when the insulation trenches

Fig. 12.5 SEM photographs of a HVLED die
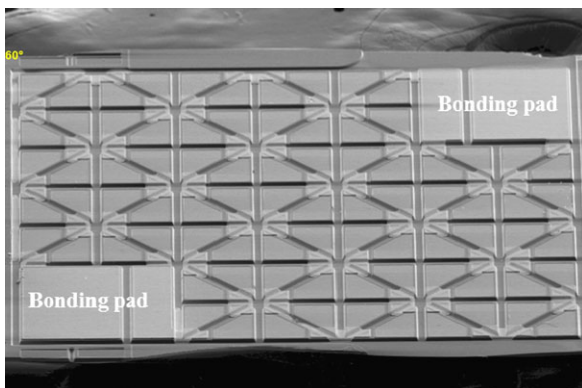
are overly thin, incomplete coating is likely to occur, regardless of whether one employs chemical vapor deposition or evaporation for the fabrication of the dielectric films, leading to the short circuit of the PN materials and thus failure of the micro-chip. Likewise, insulation trenches that are overly thin are not easily accessible to the metal vapor, resulting in an overly thin or even discontinuous metal wire film, and subsequently an increase in the series resistance of the HVLED die or even open circuit failure. In order to enhance the process yield of the dielectric coating and conducting metal film, one feasible approach is to employ an inverted trape-zoidal structure for the fabrication of the insulation trench, with the opening facing upward.

The sloped sidewall structure shown in Fig. 12.6 can improve the manufacturing yield of the micro-chips and the HVLED, and the non-rectangular geometrical structure will also help enhance the light extraction efficiency of the micro-chip. In addition, to prevent the metal shielding from reducing the luminous efficacy of the HVLED, the metal wires amongst the micro-chips must also simultaneously possess low impedance and low light obscuration. The fabrication of narrow yet thick metal wire is one feasible approach for achieving the aforementioned objective. Another viable possibility is the use of transparent metal oxides (such as ITO or ZnO) as the wire material. Both approaches will help to enhance the luminous efficacy of the HVLED.
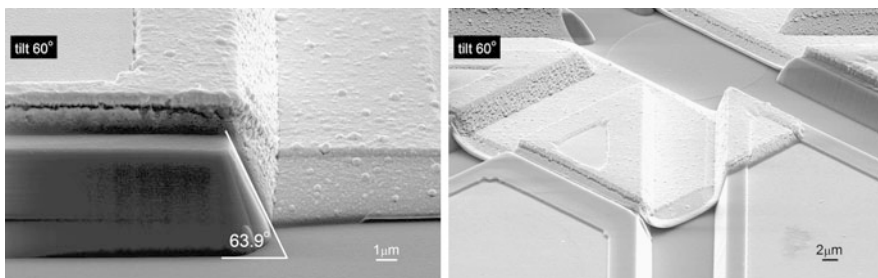
Fig. 12.6 Micro-chip structure with sloped sidewalls and inverted trapezoidal opening
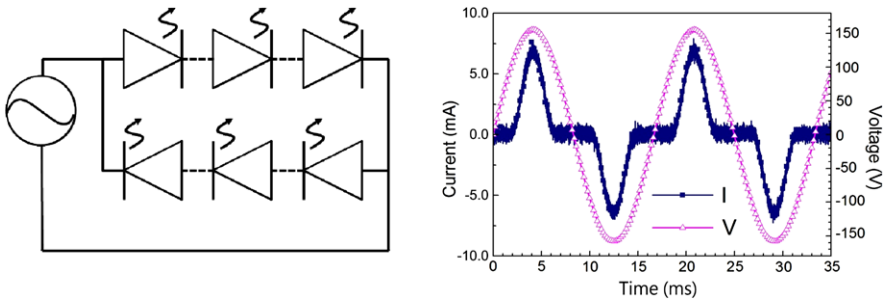
Fig. 12.7 Equivalent circuit and the time-dependent relationship between the voltage and current for a HVLED

12.2.2 Basic Characteristics and Methods of Measurement for HVLEDs Under AC Operation

In addition to the chip structure and a number of the fabrication steps, the key differences between traditional DCLEDs and HVLEDs are the low voltage DC and high voltage AC drive modes employed respectively. The different drive modes result in drastically different opto-electrical properties of the devices, meaning that the measurement techniques must also be modified accordingly. A traditional DCLED is driven by a constant direct current and one can simply integrate over time at any instance after the device characteristics have stabilized to obtain accurate measurements of its operating power and luminous intensity. However, this is not the case for AC-driven HVLEDs. Figure 12.7 shows the equivalent circuit of a HVLED device, as well as the time-dependent relationship between the voltage and current. When a HVLED is driven by a sine wave type AC voltage, it will only start emitting light when the input voltage is higher than the threshold voltage. Therefore, when driven under AC, HVLEDs often exhibit a flashing phenomenon also known as pulsed emission.

From Fig. 12.7, one can see that for this HVLED operating under AC, the frequency of the light output will be identical to the frequency of the current, both twice the frequency of the input AC voltage. In addition, under AC operation, the current and light output intensity of a HVLED at any instance will vary continually with the time-dependent input voltage, and it is difficult to extrapolate the full characteristics of the HVLED from data accrued from an arbitrary instance in time. The current consensus is that a better measurement approach is to set the integration time of the measurement to an integer multiple of the period of the input AC voltage. Take a HVLED operating under a 60 Hz AC voltage as an example; the integration time for the measurement can be set to 50 ms, which covers three full AC cycles, whereas the integration time for the measurement for a HVLED operating under a 50 Hz AC voltage can be set to 20 ms.

Fig. 12.8 Time-dependent
forward current of the
HVLED driven by DC and
AC
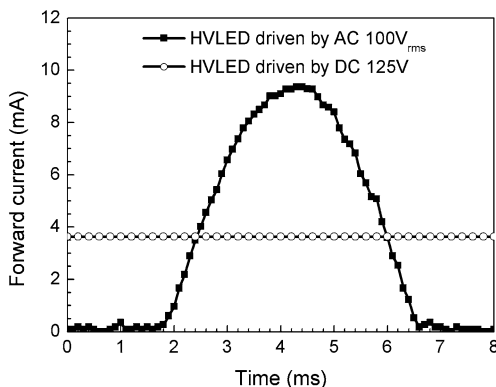
12.2.3 Light Emission Characteristics of HVLEDs Under AC Operation

Being driven by AC voltage not only transforms the light emission of HVLEDs from the steady mode under DC to the flashing mode under AC, but also significantly impacts the luminous efficacy. In Fig. 12.8, we compared the current characteristics of the same HVLED operating under DC and AC. When the operating power was 0.45 W in both cases, the operating voltage of the HVLED under DC and AC were 125 V and 100 V_{rms} respectively. Since the threshold voltage of the HVLED was about 90 V, an input voltage between 90–140 V (100 V $\times \sqrt{2}$) was required to induce a response current under AC operation and thus light emission. This duration accounted for about 60 % of the AC voltage cycle. In addition, under DC operation, the current in the HVLED was constant at 3.6 mA. Under AC operation, the current in the HVLED ranged from 0 to 9.5 mA and varied with changes in the input voltage. If one compares only the duration in which a response current could be observed in the HVLED under AC operation, for more than 70 % of the time, the response current in the HVLED under AC operation is higher (or much higher than) than that under DC operation.

Figure 12.9 shows the relationship between the injection current density and the LED luminous efficacy. Due to the efficiency drooping effect, the internal quantum efficiency of the LED will decrease with increasing current density passing through the LED, thereby lowering the overall luminous intensity and efficacy of the LED. Since the current or current density in a HVLED driven by AC is higher during most of the light emission cycle, a HVLED operating under AC exhibits lower luminous intensity or efficacy than when operating under DC, provided that the operating power and light emission area are identical. From the above discussions, one can see that although operating a HVLED under AC sidesteps the drawbacks of the rectifier and transformer elements, the luminous efficacy is limited by the physical mechanism and will necessarily be lower than that of a traditional DCLED with the same light emission area and operating power. To rectify the aforementioned issue,

Fig. 12.9 Relationship between the injection current density and luminous efficacy of a LED
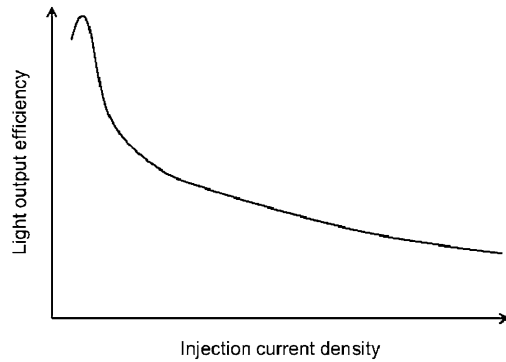
Fig. 12.10 Equivalent circuit for an anti-parallel HVLED die
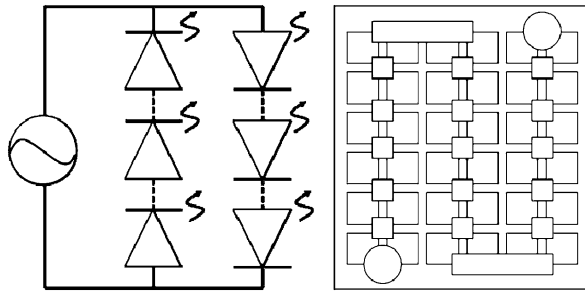
different designs for HVLEDs have been proposed. We shall introduce, in chronological order of their development, the designs and characteristics of HVLEDs that are representative of their respective generations.

12.2.4 Design and Characteristics of HVLED Devices

12.2.4.1 Anti-parallel HVLED

The equivalent circuit of the first single-chip HVLED is shown in Fig. 12.10. The core concept is to connect two chains of micro-chips fabricated on the same insulating substrate in an anti-parallel configuration, enabling at least one chain of micro-chips to be under forward bias and thus light emitting, regardless of the bias of the input AC voltage. However, clearly, in such an anti-parallel type HVLED (AP-HVLED), only half of the micro-chips will be luminescent at any instance under AC operation, with the remaining half of the micro-chips not illuminated since they are under reverse bias. Therefore, when calculating the ratio of the light emission area of a HVLED at any instance, one obtains chip area utilization efficiency (CAUE) of only 50 % for the AP-HVLED. Although the two chains of micro-chips in the AP-HVLED design will be alternatively luminescent, the CAUE is a paltry 50 %. As such, under AC operation, the current density passing through the micro-chips within the AP-HVLED is still about two times that in a traditional DCLED
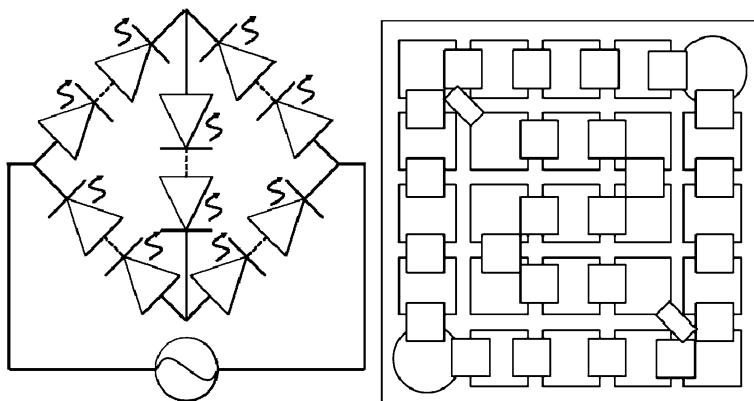
Fig. 12.11 Equivalent circuit of a Wheatstone bridge type HVLED die

with identical chip area and effective operating power. In other words, if one wishes to operate an AP-HVLED under AC with similar luminous efficacy to that of a traditional DCLED with the same effective operating power, even without taking into account the efficiency drooping effect on the luminous efficacy, the chip area of the AP-HVLED must still be at least two times that of the traditional DCLED.

Although the AP-HVLED successfully realizes the concept of a single-chip HVLED, the low CAUE necessarily implies that its operating current density will be higher than that of a traditional DCLED with the same area and operating power. While boosting the area of the AP-HVLED die is a feasible approach for solving the aforementioned problem, an excessively large chip area will result in increased production costs. Therefore, developing a HVLED die with a higher CAUE became a primary goal of researchers.

12.2.4.2 Wheatstone Bridge HVLED

The equivalent circuit of a Wheatstone bridge type HVLED (WB-HVLED) with greater than 50 % CAUE is shown in Fig. 12.11. In the standard design of a WB-HVLED, the micro-chips on the insulating substrate are arranged into an array that assumes the form of a Wheatstone bridge and placed in four rectifier branches and an output branch. Regardless of the bias direction of the input AC, the micro-chips in two of the rectifier branches and the output branch of the WB-HVLED will be under forward bias and thus luminescent, thereby enabling the WB-HVLED to exhibit a higher CAUE than the AP-HVLED does. For example, suppose the number of micro-chips (with identical area) in the rectifier branches and output branch of the WB-HVLED is 5 and 10 respectively. Under AC operation, there will be 20 micro-chips illuminated at any instance in the WB-HVLED, accounting for 67 % of the 30 micro-chips. If one decreases the number of micro-chips in the rectifier branches while simultaneously increasing the number of micro-chips in the output

Fig. 12.12 Side-view SEM images of the destruction to the micro-grain structure caused by the GaO$_x$ oxide
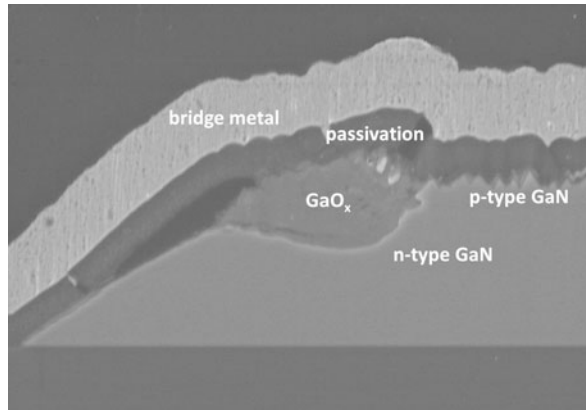
branch, one can further enhance the CAUE of the WB-HVLED without altering the total number of uni-directional light emitting micro-chips. For example, when the number of micro-chips in the rectifier and output branches of the WB-HVLED is 3 and 14 respectively, the number of uni-directional light emitting micro-chips remains the same at 20. Therefore, the total number of micro-chips is reduced to 26, enabling the CAUE of this type of WB-HVLED to soar to 77 %. However, there are limitations on the use of such a mechanism to enhance the CAUE of the WB-HVLED. Figure 12.12 shows GaN micro-chips placed under excessively high reverse bias for a prolonged duration. Lumps of GaO$_x$ oxide can arise in the n-type GaN, leading to the destruction of the epitaxial structure of the micro-grain and thus device failure [1]. Although this phenomenon can be alleviated by fabricating a denser coating layer to block out water vapor, when the number of micro-chips in the rectifier branches is too low, the micro-chips in the rectifier branches may still suffer from reverse breakdown due to the excessively high reverse bias electric field, leading to short circuit failure of the WB-HVLED. In view of the above two factors, a more appropriate ratio of the number of micro-chips in the rectifier and the output branches of a GaN WB-HVLED is 1:3, with a CAUE of about 70 %.

For HVLEDs, a chip design with a higher CAUE implies a higher luminous efficacy (for identical total chip area and operating power) or lower manufacturing costs (for identical light emission area under uni-directional bias). WB-HVLEDs that incorporate the concept of the Wheatstone bridge exhibit an increase in the CAUE to 70 %, which represents an increase of 40 % on the 50 % CAUE exhibited in AP-HVLEDs, but the improvement still falls far short of the nearly 100 % CAUE exhibited by traditional LEDs. Further attempts to enhance the CAUE of the WB-HVLEDs have floundered due to the fact that the LED micro-chips in the rectifier branches are unable to withstand too large a reverse bias. Extended operation at high reverse bias will lead to the formation of GaO$_x$ oxide in the rectifier micro-chips and thus device failures. As such, it is not feasible to drastically enhance the CAUE without compromising on the reliability and longevity of the WB-HVLEDs. One possible solution is to employ selective area regrowing techniques in combination with the epitaxial structures of the LED and Schottky Barrier Diode (SBD) for the

Fig. 12.13 A WB-HVLED utilizing 4 SBDs as the rectifier elements
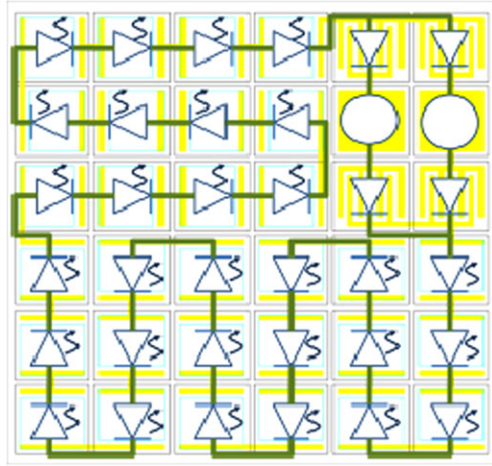
fabrication of WB-HVLEDs with SBD rectifier elements. The equivalent circuit is shown in Fig. 12.13. Since the capacity of the SBD to withstand reverse bias is far superior to those of the LED micro-chips, the use of merely four SBDs as rectifier elements in a WB-HVLED will enable normal operation under 110–220 V AC and an increase in the CAUE to more than 90 %. However, at this point in time (early 2012), technologies enabling the epitaxial stacking of SBD and LED structures is still under research. Further, simultaneously optimizing the characteristics of the two devices is fraught with complications, and numerous issues will need to be clarified and resolved before the use of SBD as a rectifier element in a WB-HVLED and the subsequent mass production can be realized.

12.2.4.3 Hybrid HVLED

In view of the technical bottlenecks encountered in the continuing attempts to improve the CAUE of single-chip high HVLEDs, LED manufacturers and researchers begun attempts to redefine the mode of operation of the single-chip HVLED. In assembled HVLED modules, numerous encapsulated LEDs are typically connected in series to withstand high voltages, and then coupled with an external bridge component to enable the LED arrays to be illuminated regardless of the bias of the input AC. The same concept also applies to single-chip HVLEDs. The use of the external bridge component does away with the need for rectifier elements within the single-chip HVLED, and one needs only to configure the number of micro-chips capable of withstanding the uni-directional AC voltage in the array. Figure 12.14 shows the equivalent circuit of such a hybrid type HVLED (H-HVLED) chip coupled with external bridge components and operating under AC. Since all the micro-chips in the H-HVLED can be illuminated under bi-directional AC bias, its CAUE can be enhanced to the 100 % exhibited by traditional DCLEDs, if one does not take into
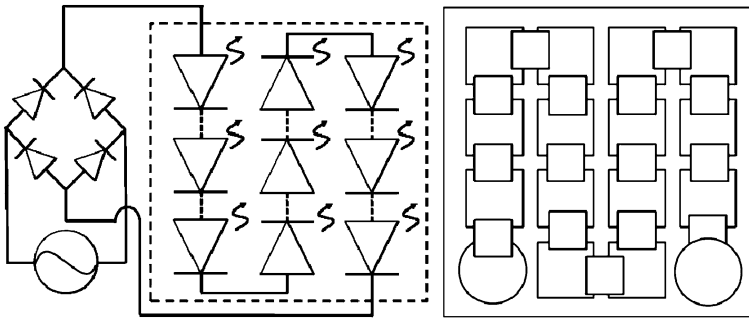
Fig. 12.14 Schematic diagram of the entire module consisting of a single-chip H-HVLED coupled with external bridge components

DCLED

Anti-parallel
High Voltage LED

Wheatstone bridge
High Voltage LED

Hybrid
High voltage LED
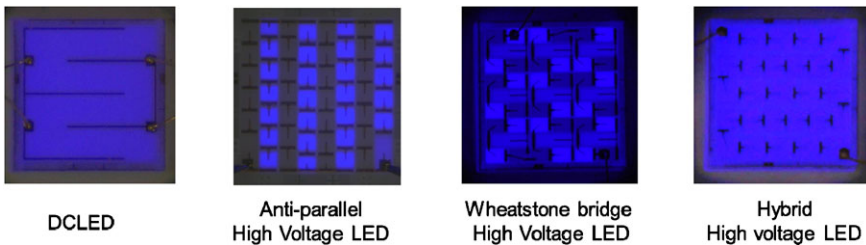
Fig. 12.15 Photos of a traditional DCLED and three different HVLED dies under DC operation

consideration the insulation trenches. Although the external bridge components increase the size and cost of a lighting module employing the H-HVLED as the light source to levels slightly exceeding those of modules employing the AP-HVLED or WB-HVLED, when one considers the die by itself, the H-HVLED actually exhibits superior CAUE to both the AP-HVLED and WB-HVLED, and thus a higher luminous efficacy (for identical chip area) and lower manufacturing costs (for identical uni-directional light emission area). This is the main reason why the H-HVLED has become widely adopted by numerous LED die manufacturers at this point in time (early 2012).

12.2.5 Characteristics of Various HVLEDs and Traditional DCLED

Figure 12.15 shows the external views of a traditional DCLED as well as three different HVLED dies. All four dies were fabricated with the same GaN wafer and the same die area. We compared the opto-electrical properties of the four LEDs at the same operating power. In particular, the AP-HVLED, WB-HVLED and H-HVLED consisted of 54, 45, and 27 micro-chips respectively, and in all three

Fig. 12.16 Relationship between the input power and light output power for the four different LEDs under DC operation
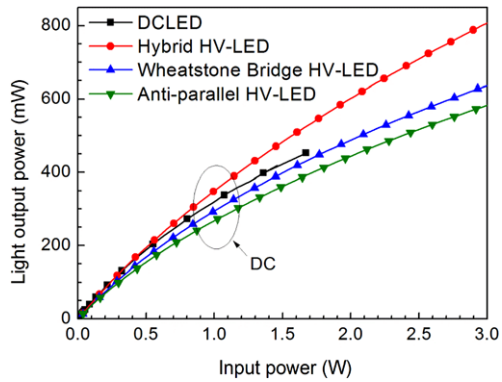
designs, 27 micro-chips were illuminated under uni-directional bias. The areas of the micro-chips in the AP-HVLED and H-HVLED were identical, whereas the areas of micro-chips in the output branch of the WB-HVLED was twice that in the rectifier branches.

From Fig. 12.16, one can see that when all four LEDs were operated under DC, the order of the luminous intensity and efficacy of the dies was H-HVLED > DCLED > WB-HVLED > AP-HVLED. At the same operating power, an LED die with a smaller light emission area will exhibit higher operating power and current density, and the efficiency drooping effect will result in the internal quantum effectiveness of the LED die and the overall luminous efficacy becoming lower than that of the LED die with a larger light emission area. Since the CAUEs of the WB-HVLED and AP-HVLED were significantly lower than those of the H-HVLED and DCLED, the luminous efficacy of the former two must be less than those of the latter two. A comparison of the H-HVLED and DCLED revealed that although the CAUE of the H-HVLED was slightly lower than that of the DCLED after taking into consideration the losses due to the insulating trenches, the current diffusion uniformity of the H-HVLED was actually superior to that of the single-chip DCLED with greater area, since the H-HVLED was composed of numerous micro-chips with small area. Therefore, in comparison to the DCLED, in which the current tends to be concentrated near the electrodes, the H-HVLED exhibited better current diffusion properties. As such, the efficiency drooping effect was less pronounced. Further, the side wall structure of the micro-chips further increased the light extraction area and efficiency of the H-HVLED, enabling the H-HVLED to exhibit a higher luminous efficacy under DC operation than the DCLED did.

Figure 12.17 shows the comparison of the light emitting efficacy of the AP-HVLED, WB-HVLED, and H-HVLED under AC operation with that of the DCLED under DC operation. By comparing with the characteristics measured under DC operation, one can see that the luminous efficacy of three HVLEDs under AC operation all suffered significant declines. As discussed in Sects. 12.2.3 and 12.2.4, when the operating wattage or voltage is identical, the response current in a HVLED will be significantly higher than the response current under DC operation most of the time, resulting in the HVLED exhibiting lower luminous efficacy when operating under

Fig. 12.17 Comparison of the input power and light output power of the four LEDs under different operating conditions
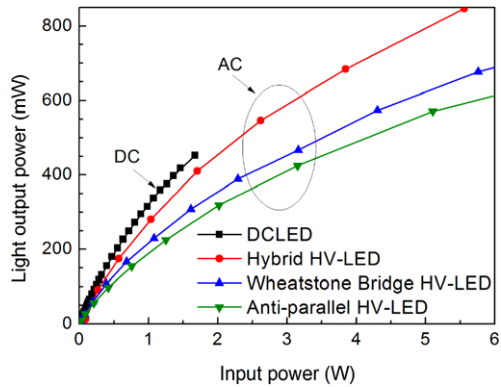
AC than when operating under DC because of the pronounced efficiency drooping effect. Thus, although the H-HVLED still exhibited higher luminous efficacy under AC operation compared to the WB-HVLED and AP-HVLED, it lagged slightly behind the performance of the DCLED.

In sum, our findings imply that, under DC operation, the H-HVLED exhibited higher luminous efficacy, lower circuit losses, and higher module efficiency than the DCLED did. However, under AC operation, the H-HVLED, due to the impact of efficiency drooping effect, exhibited luminous efficacy that was slightly lower than that of the DCLED under DC operation. Since the WB-HVLED and AP-HVLED had CAUEs that were both significantly lower than those of the H-HVLED and DCLED, their luminous efficiencies lagged rather far behind those of the H-HVLED and DCLED both under DC and AC operations. Under the premise that the chip area was identical, all three of the HVLEDs, namely H-HVLED, WB-HVLED, and AP-HVLED, exhibited inferior luminous efficacy under AC operation to the traditional DCLED. Nonetheless, the elimination of some of the electronic devices offers tangible competitive advantages for the HVLED modules in terms of cost, space, and lifetime. Should technological breakthroughs allow one to reduce or even completely avert the negative impacts of the efficiency drooping effect on the internal quantum efficiency of the LED through modifying the epitaxial structure, the luminous efficacy of HVLEDs operating under AC can be further brought closer to the level of DCLEDs, thereby allowing one to make the former even more competitive in comparison.

12.3 Important Issues on the Applications of HVLEDs

The applications for HVLEDs can be classified into two categories, namely high DC voltage drive mode and high AC voltage drive mode. The former is characterized by stable optical output characteristics, and particular attention needs to be paid to the flammability and pressure withstanding capability of the encapsulation materials. For the latter, the optical output characteristics of the devices will vary according

Fig. 12.18 Variation in the
intensity of the HVLED
output light and fluorescent
light
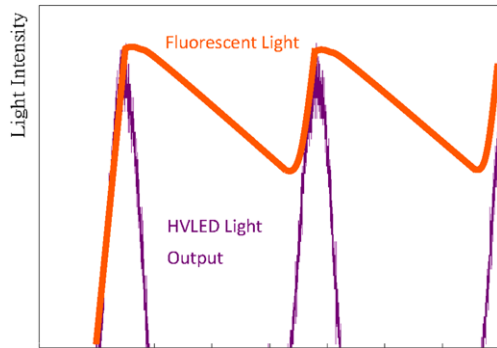
to changes in the AC input. Therefore, on top of the aforementioned encapsulation material properties, one must also further explore issues such as light flickering, harmonic distortion and optical-electrical measurements. In addition, since the opto-electrical characteristics vary significantly from those of traditional DCLEDs, the evaluation and estimation of the lifetime will also be one of the most important issues.

12.3.1 Characteristics of and Possible Solutions for Light Flickering

HVLEDs typically operate at a frequency of 60 Hz. Due to the AC driving mode of the devices, the brightness as perceived by the human eye will vary dynamically at the rate of 120 Hz. The perception threshold of the human eye to flickering is about 100 Hz. Therefore, the 120 Hz variation in the HVLED light source can result in the perception of flickering for some people. In order to overcome the potential problems associated with the flickering phenomenon of the HVLED light source as well as to enhance the luminous performance, thereby enabling HVLEDs to gain wider acceptance in the lighting application markets, numerous studies are currently in progress to identify possible solutions, one of which is the development of high thermal stability, long afterglow phosphor materials for encapsulating HVLEDs and thus eradicating the flickering phenomenon. When a HVLED is driven by an AC voltage, the latter assumes the form of an alternating sine wave. The HVLED light source will exhibit cycles of bright and dark light emission states, and the development of the new long afterglow phosphors will alter the luminescence decay time after excitation of the fluorescent powder, enabling the decay time to 50 % brightness to exceed the period of the bright and dark cycles. This will reduce the differences in the luminous intensity of the bright and dark states and thus vastly stabilize the light source output of the HVLEDs, as shown in Fig. 12.18.

An alternative approach is to employ an external combinational circuit to sidestep the light flickering of the HVLEDs. Both the industry and academia have successfully developed simple integrated circuits for overcoming the flickering, and these

Table 12.1 IEC/EN 61000-3-2, class C, harmonic current limits (input power ≥25 W)

Harmonic order (n)	Maximum permissible harmonic current expressed as a percentage of the input current at the fundamental frequency
2	2
3	$30 \times$ power factor
5	10
7	7
9	5
$11 \leq n \leq 39$	3

circuits, on top of alleviating the flickering, can also boost the power factor of the HVLED light source system. However, the issues of effectively reducing the size of the combinational circuit and its subsequent integration into the light source system, as well as considerations on the operating lifetime of the applications, are all challenges to be overcome.

12.3.2 Total Harmonic Distortion Limits

In addition to the impacts of the input operating voltage on the optical properties of the HVLEDs, one must also consider the issue of Total Harmonic Distortion (THD) for applications in lighting. The limits on the harmonic distortion for lighting fixtures consuming 25 W and above are clearly spelled out under the International Standards on Lighting Equipment IEC/EN 61000-3-2 (Harmonic Standards) Class C. The limits for lighting fixtures consuming less than 25 W are not specified [2]; however, under the U.S. ANSI C82.77 Standards (Harmonic Emission Limits—Related Power Quality Requirements for Lighting Equipment), it is expressly stipulated that the total harmonic distortion of light sources slated for illumination purposes must be less than 32 % [3].

Although current HVLED devices typically consume less than 25 W of operating power by themselves, they will necessarily need to be integrated into light source modules in order to break into the field for lighting applications, in view of current LED device light radiation flux. In that case, the operating power will likely exceed 25 W. In other words, the total harmonic distortion for HVLEDs must also be in compliance with the IEC/EN 61000-3-2 Class C specifications and the ANSI C82.77 standards, as shown in Tables 12.1 and 12.2.

For compliance with the IEC/EN 61000-3-2 Class C and ANSI C82.77 Total Harmonic Distortion specifications, one can alter the number of micro-chips on a single chip HVLED to lower the threshold voltage and control the conduction angle of the current, but this method will also test the high current density tolerance of the HVLED die itself. A more plausible solution is perhaps to employ simple combinations of electronic devices and highly efficient driver circuit designs to improve the

Table 12.2 ANSI C82.77 harmonic emission limits (commercial indoor hard wired ballasts or luminaries)

Input power (P)	Minimum PF	Maximum line current THD (fundamental)
All	0.9	32 % and the requirements of ANNEX I

total harmonic distortion of the light source system and at the same time improve the power factor of the entire lighting system.

12.3.3 The Effect of Floating Driving Voltage

The floating range of most mains voltages is about ±10 %, and light sources that directly utilize AC will also exhibit cycles of brightness and darkness in their light outputs during operation in accordance with the floating voltage. The impacts on the human vision caused by the variations in the brightness of the light source are known as "stroboscopic effects", and they are characterized by a tendency to visually perceive fast-moving objects as slow-moving or stationary ones, which constitute obvious safety hazards. The unit for evaluating the light flux is known as the strobe depth, where a strobe depth of less than 5 % may be regarded as the absence of stroboscopic effects.

The threshold of the human eye for perceiving strobe depth is 100 Hz. In HVLED applications, when the power supply exhibit fluctuations of ±10 % for frequency of 100 Hz or less, the HVLEDs will exhibit strobe effects synchronized with the floating frequency, with a strobe depth of up to 50 %. In such cases, prolonged use of said applications will result in discomforts to the human body.

Therefore, in order to overcome the strobe effects of the light source brought about by the variations in mains voltages, the development of a simple driving circuit to stabilize the driving voltage of the light source and hence reduce the strobe depth of the HVLED light output will also be a critical research thrust.

12.3.4 Safety Considerations of HVLED Encapsulation Structural Design

There is essentially little difference in the encapsulation mode of HVLEDs and that of DCLEDs. In both cases, the operating power of the LED will determine the design for the thermal dissipation structure, and the encapsulation modes can roughly be categorized into Surface Mounted Device (SMD), Chip-on-Board (COB), etc. In addition to continually developing and improving upon such traditional encapsulation approaches, LED manufacturers have begun contemplating other possible uses of LED components, and gradually devised a LED encapsulation device that
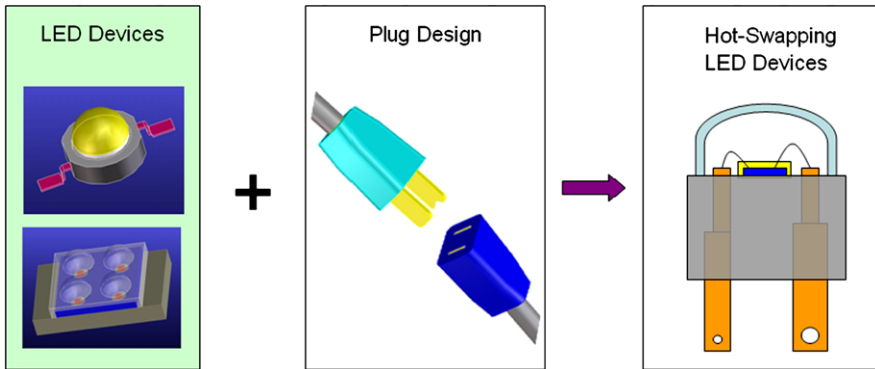
Fig. 12.19 Possible uses of HVLED encapsulation device

can be plugged and unplugged and thus installed and replaced easily, as shown in Fig. 12.19. In other words, no welding is required, and one needs only plug the pins into a connector assembly. Especially for HVLED applications, the direct use of AC mains has made it more expedient for users to upgrade and replace HVLED light sources, thereby dramatically improving the ease of operation.

There are more restrictions on the structural design of HVLED encapsulation, the most important of which is the pin specifications of the encapsulated device. HVLED encapsulation devices still employ the traditional two-pin approach for electrical connection, and special attention needs to be paid to the pin spacing design.

In the official ANSI/UL 8750 (Standard for Light Emitting Diode (LEDs) Equipment for Use in Lighting Product) safety specifications [4] released by the product safety testing and certification organization Underwriters Laboratories (UL) for LED lighting products and related devices, it was specifically pointed out that the two inputs must comply with the insulation requirements for meeting the highest operating voltage conditions of the entire circuit. Otherwise, the designer must do an excellent job of separating the pins. It is further stipulated that the minimum insulation spacing should be no less than 6.44 mm, in order to prevent electrical hazards arising from the two inputs coming into contacts.

The commonly used insulating materials in LED encapsulation, such as the LED bracket coating materials, encapsulation materials, substrate support materials, etc. must all be moisture-proof materials whose structural strength will not be altered under real world operating temperatures. It is also stipulated in ANSI/UL 8750 that the evaluation of these insulation materials should be performed in accordance with the specifications of the ANSI/UL 746C, using criteria such as the mechanical strength, fire resistance, dielectric strength, electrical insulation, thermal resistance before and after aging, degree of coating, anti-moisture capacity in a non-dry environment, and other characteristics that will affect the risk of fire and electric shock. When it comes to the choice of the encapsulation materials, such as the plastic colloid for the bracket, the encapsulating epoxy, silicone, aluminum substrates, etc., the fire-resistance rating cannot be less than that specified by the ANSI/UL 94V-1 safety

Table 12.3 Dielectric voltage withstand potential (UL8750 Table 8.3)

Applied potential	Circuit location
2 V + 1000	Between primary circuits or secondary circuits operating at greater than 70 V peak and accessible dead conductive parts
	Between the primary and secondary of a transformer
	Between PWB traces or other parts operating at different potentials
500	Between a secondary circuit operating at no more than 70 V peak and accessible dead conductive parts

limits. Since HVLED encapsulation components operate in a high voltage system, special consideration must also be taken for safety concerns such as the combustivity of the material, anti-electrical sparks characteristics, anti-moisture resistance, etc. when it comes to the choice of the aforementioned polymer materials.

The stipulated moisture and insulation tests for HVLED encapsulation devices involve placing said devices (or modules) under normal operations in a test environment with a relative humidity of 91 % to 95 %, with the temperature maintained in the room temperature range, for up to 48 hours. Thereafter, the insulation devices are subjected to a 500 V DC voltage, followed one minute later with an insulation resistance measurement. Under normal conditions, the insulation resistance must be greater than 2 MΩ, to ensure the anti-moisture and insulation capacity of the encapsulation materials.

The test conditions for the Dielectric Voltage Withstand Test referred to in ANSI/UL 8750 are shown in Table 12.3. A potential of 1000 V plus twice the maximum rated AC operating voltage is applied between a line and a conductor that carries no charge and can be placed in contact (i.e. the input end of the HVLED and a neighboring conductor) for one minute, during which the HVLED encapsulation devices must not suffer dielectric breakdown. The test voltage needs to be applied to the power supply, LED modules, and control circuit modules that are directly connected to the circuit branch. Double or reinforced insulation test conditions are also described under the IEC/EN 60598–1:2008 (General Requirements and Tests for Lightings) specifications, whereby a voltage of 2000 V plus four times the maximum rated AC operating voltage is applied for advanced electrical insulation test [4].

12.3.5 Measurement Techniques for the Optical, Electrical, and Thermal Properties of HVLED Modules

During the operation of a HVLED under AC, its opto-electrical characteristics as well as temperature will exhibit cyclical changes with the input AC frequency. Therefore, one cannot simply use the "input of a small instantaneous current for eliciting and measuring the response voltage" approach for DCLEDs to deduce the T_j of HVLED.
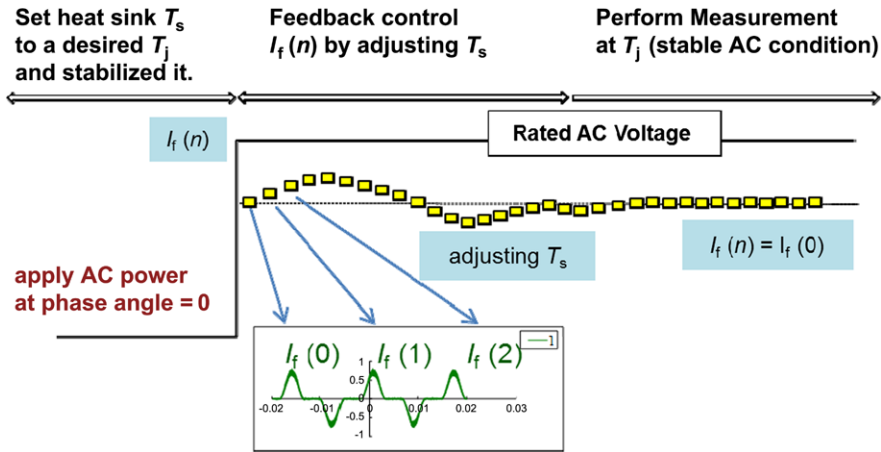
Fig. 12.20 A viable HVLED measurement approach

In 2008, the NIST (National Institute of Standards and Technology) and the Taiwan Industrial Technology Research Institute (ITRI) jointly proposed a reference HVLED opto-electrical measurement approach based on extending the measurement time and which borrows from and modifies the aforementioned traditional measurement approach for high power DCLEDs. The scheme is outlined step-by-step in Fig. 12.20 and described in details as follows: (1) Install the HVLED to be tested onto a metal core printed circuit board (MCPCB), where the heat sink with temperature control function is located. (2) Turn on the temperature controls to maintain the measurement point temperature (T_s, typically a specified point on the MCPCB) of the HVLED (yet to be turned on and tested) at the specified $T_{j(0)}$. (3) Turn on the AC power for measurement with zero phase, and simultaneously commence measurements. (4) Adjust the temperature T_s such that during any half-cycle, the root-mean-square current value If (n) is equal to the If (0) of the first half-cycle, i.e. $T_{j(n)} = T_{j(0)}$. (5) Perform opto-electrical measurements of the HVLED at $T_{j(n)} = T_{j(0)}$. It should be noted that the measurement time should be integer multiples of the driving cycle of the HVLED to be tested, so as to ensure the accuracy of the data.

The aforementioned measurement approach can be integrated with existing measurement equipment for lighting, e.g. integrating sphere for discharge lamp, goniophotometer, array or scanning spectrometer. Utilizing this approach, one can also precisely calculate the thermal resistance of a HVLED by computing the difference between the default T_j and the actual measured T_s.

12.3.6 Operating Lifetime

The output intensity of the LED light source will gradually decline over time, and one can typically calculate the lumen maintenance rate of a LED light source by

comparing its initial light output intensity and its light output intensity after use. In the LM-80-08 testing standards developed by the North American Illuminating Engineering Society (IESNA), the standard measurement procedures [5] for the lumen maintenance rate of a LED light source, including the measurement equipment, environmental settings, definition of testing conditions, etc. are all clearly defined. Further, the actual operating time of the LED light source will be symbolically denoted as a percentage of the initial output light intensity, e.g. L_{70} (hours) indicates that after said hours have elapsed, the lumen of the light source is maintained at 70 % of its initial value, while L_{50} (hours) indicates that after said hours have elapsed, the lumen of the light source is maintained at 50 % of its initial value. The U.S. ENERGY STAR standards stipulate that for indoor lightings, the lumen maintenance rate after 25,000 hours must be no lower than 70 %. For outdoor lightings, the lumen maintenance rate after 35,000 hours must be no lower than 70 %. The lumen maintenance rate is further used as a criterion for gauging the operating lifetime of a LED light source. The above criterion is based on calculating the decay of the light intensity from the variations in the light intensity of the DCLED, whose optical output is more stable. The operation of a HVLED differs from the stable current drive mode of a traditional DCLED in that it utilizes a constant AC voltage. Therefore, it is recommended that the decay of the light output intensity under AC driving is used as the basis for calculating the lumen maintenance rate.

In addition, due to prolonged operations, the operating power of a HVLED device or module may vary. Therefore, another possibility is to utilize the decay in the luminous efficacy as the basis for gauging the operating lifetime of a HVLED.

In sum, the relevant lifetime testing, failure criteria, and lifetime testing standards for HVLED are still undergoing discussions. Before lifetime testing procedures are standardized, there still remain several related or corresponding opto-electrical characteristics worthy of further research, discussions, and analysis.

12.4 Summary

HVLED utilizes semiconductor fabrication techniques to grow and connect in series several micro-chips on the same substrate, enabling a single-chip LED to operate directly under AC. By avoiding (or partially circumventing) the use of rectifier and transformer elements, HVLED modules perform lower manufacturing costs and smaller module volumes than traditional DCLED modules. As opposed to traditional DCLEDs driven by a constant current, HVLEDs are designed to operate under AC voltage. Therefore, issues such as the luminous intensity varying with the input voltage, limitations due to total harmonic distortion, etc. can arise. All the aforementioned issues have been discussed in great details in this chapter, along with the feasible solutions.

In addition, due to the fact that the characteristics of HVLEDs under AC operation are very different from those exhibited by traditional DCLEDs, the relevant safety and measurement standards will also need to be modified in accordance with

the actual operating conditions of HVLEDs. In this chapter, the relevant standards and recommendations have also been discussed, enabling the users to more safely and properly operate HVLEDs and measure the characteristics exhibited under AC operation.

The purpose of HVLEDs is to provide a single-chip solution capable of operation under AC. Although the efficiency drooping effect means that for identical operating power and chip area, the luminous efficacy of a HVLED operating under AC will be lower than that of a traditional DCLED, the compactness, low cost, and possibly longer operating lifetime of the overall module are all vast improvements on traditional DCLEDs with regards to application.

References

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2. IEC/EN 6 1000-3-2, Harmonic Standards, Class C: Lighting Equipment
3. ANSI C82. 77:2002, Harmonic Emission Limits—Related Power Quality Requirements for Lighting Equipment
4. IEC/EN 60598-1:2008 Luminaries. Part 1: General Requirements and Test
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Chapter 13
Color Quality of White LEDs

Yoshi Ohno

Abstract This chapter provides an overview of the fundamentals of chromaticity and color rendering, the two important aspects of color quality of light sources for general illumination. There is a special focus on the use of solid state light sources. The section on chromaticity discusses chromaticity coordinates and diagrams, correlated color temperature (CCT), Duv, and specifications for color differences. The section on color rendering discusses object color evaluation, the color rendering index (CRI) and the shortcomings thereof, and color quality beyond CRI, introducing the color quality scale (CQS), a proposed alternative metric. The chapter also discusses luminous efficacy of radiation and the color characteristics used for single-color LEDs. Finally, future considerations on color quality for white LED developments are given.

13.1 Introduction

It is critical for LED sources for general illumination to have acceptable color quality suitable for intended applications. There are two aspects to the color quality of lighting sources: chromaticity and color rendering. These two aspects are interrelated. Both of these characteristics are critical for general lighting sources.

The chromaticity of a light source is the color of light emitted from the light source, and is normally expressed in chromaticity coordinates, (x, y) or (u', v') [1], and is often plotted in two-dimensional chromaticity diagrams using these coordinates. The chromaticities of illumination sources are normally chosen to be around the Planckian locus or the daylight locus, which are generally considered as the center curves for white light. Shifts of chromaticity along these center curves result in warmer or cooler color appearance, and are specified by correlated color temperature (CCT) [1]. Shifts of chromaticity across these center curves make the lights appear yellowish, greenish, or pinkish and are specified by the term, Duv [2]. While

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a wide range of CCT is suitable depending on application, the range of acceptable Duv is much narrower on the color space. Duv is important especially for indoor applications. There have been standards for chromaticity of fluorescent lamps for many years [3]. A standard for chromaticity specifications for solid-state lighting products [2] was recently developed and widely used. However, there are still some questions about whether the Planckian or the daylight locus represents optimal white light for lighting, e.g., [4]. Chromaticity coordinates and CCT are fundamental terms defined by International Commission on Illumination (CIE) and their official definitions are available in [5]. The details these chromaticity terms and formulae are presented in Sect. 13.2.

Another important aspect of color quality is color rendering. Even if the light has perfect white chromaticity, color rendering can be very poor. Color rendering is a property of a light source that quantifies how natural the colors of objects look under the given illumination. If color rendering is poor, the light source will not be useful for general lighting. The term color rendering is defined by CIE [5] with a rather narrow meaning, but this term is often used with a broader meaning including the aspects of color fidelity and color preference. The current international standard to evaluate color rendering is the Color Rendering Index (CRI) [6] defined by CIE. The CRI has been used very widely for a number of years. The formulae used in CRI are old, and recently problems related to the CRI when used for LED sources have been identified, e.g., [7] and are being addressed by CIE (TC1-69 Color rendition by white light sources). Several new metrics to replace or supplement CRI have been recently proposed [8–12], and many other old research results and metric proposals are available, e.g., [13–15]. The details and issues related to color rendering are presented and discussed in Sect. 13.3.

Color rendering depends on the spectrum of the light source, which also affects the energy efficiency of light source. Color rendering is generally more superior if the spectrum is broadly distributed over the visible region. However, such broadband spectra are generally less energy-efficient. The theoretical energy efficiency of a spectrum is defined by CIE [5] as *luminous efficacy of radiation* (LER). The formula to calculate the LER is introduced, and some examples of white LED spectra are analyzed for LER and discussed in Sect. 13.4. Section 13.5 introduces the definitions of the color-related terms used for single-color LEDs including dominant wavelength, centroid wavelength, and peak wavelength.

In Sect. 13.6, the color rendering of typical phosphor white LEDs are examined and the improvements of color rendering are discussed. To analyze the color rendering characteristics of white LEDs and also the problems of the CRI, a color simulation program was developed [16]. Various white LED spectra, multi-chip type and phosphor type, were modeled and analyzed in comparison to conventional lamps. Some results of the past studies [17] are presented, and the problems and necessary improvements of the CRI are discussed in Sect. 13.3.3.

13.2 Chromaticity

13.2.1 Chromaticity Coordinates and Diagrams

The CIE 1931 (x, y) and CIE 1976 (u', v') chromaticity coordinates and corresponding diagrams are defined by the CIE [1]. Both are widely used to specify chromaticity of light sources for illumination. The (x, y) diagram is very spatially nonuniform in terms of color differences. Figures 13.1 and 13.2 show the plots of MacAdam Ellipses (just noticeable color differences) [18] on these diagrams, with the ellipses' sizes magnified by ten. For the same difference values in x or y, the perceived color differences of a green LED would be much smaller than blue LEDs. The (u', v') diagram is more spatially uniform and is preferred, especially to specify chromaticity of colored sources including single color LEDs.
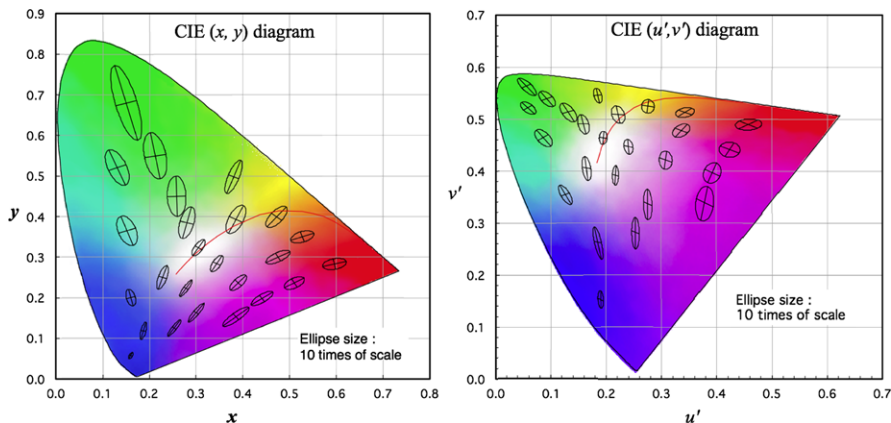
Fig. 13.1 CIE 1931 (x, y) chromaticity diagram and CIE 1976 (u', v') chromaticity diagram with MacAdam Ellipses plotted

Fig. 13.2 CIE 1931 color matching functions
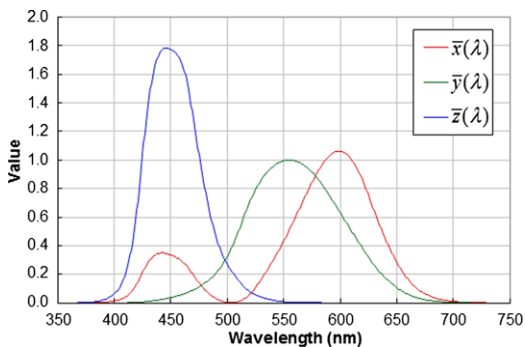
The chromaticity coordinates of a light source are calculated from its spectral power distribution $\phi(\lambda)$. The tristimulus values X, Y, Z are first calculated as

$$X = k \int_\lambda \phi(\lambda)\bar{x}(\lambda)\mathrm{d}\lambda$$

$$Y = k \int_\lambda \phi(\lambda)\bar{y}(\lambda)\mathrm{d}\lambda \qquad (13.1)$$

$$Z = k \int_\lambda \phi(\lambda)\bar{z}(\lambda)\mathrm{d}\lambda$$

where $\bar{x}(\lambda)$, $\bar{y}(\lambda)$, $\bar{z}(\lambda)$ are the CIE 1931 color matching functions (Fig. 13.2).

The values of these functions at 5 nm interval and 1 nm interval are available in Refs. [1, 19]. The chromaticity coordinates (x, y) and (u', v') are calculated by

$$x = \frac{X}{X + Y + Z} \qquad y = \frac{Y}{X + Y + Z} \qquad (13.2)$$

$$u' = \frac{4X}{X + 15Y + 3Z} \qquad v' = \frac{9Y}{X + 15Y + 3Z} \qquad (13.3)$$

The (u', v') coordinate can also be calculated from (x, y);

$$u' = \frac{4x}{(-2x + 12y + 3)} \qquad v' = \frac{9y}{(-2x + 12y + 3)} \qquad (13.4)$$

Likewise, (x, y) coordinate can be converted from (u', v') by

$$x = \frac{9u'}{(6u' - 16v' + 12)} \qquad y = \frac{2v'}{(3u' - 8v' + 6)} \qquad (13.5)$$

The CIE 1960 (u, v) coordinates (now obsolete) is introduced here only for the purpose of calculating CCT and Duv, and is obtained by,

$$u = u', \qquad v = \frac{2}{3}v' \qquad (13.6)$$

The (u, v) coordinates should not be used for calculations other than CCT and Duv. The curve formed by the plots of the chromaticity coordinates of monochromatic radiation at various wavelengths is called the *spectrum locus* (this forms the horseshoe shape of the diagram), and the curve formed by the plots of Planckian radiation at various blackbody temperatures is called the *Planckian locus*, as depicted in Fig. 13.3 on the (x, y) diagram. The straight line connecting both ends of the spectrum locus is called the *purple line*.

The Planckian locus is generally considered as the center curve of white light for low CCT sources (CCT < 5000 K). For higher CCT sources (CCT ≥ 5000 K), the different phases of daylight (called the *daylight locus*), also plotted in Fig. 13.3, is generally considered as the center curve of white light for lighting. The formula to calculate the daylight locus is available in [1]. Several specific points on this locus, at

Fig. 13.3 The spectrum locus and the Planckian locus on the (x, y) diagram
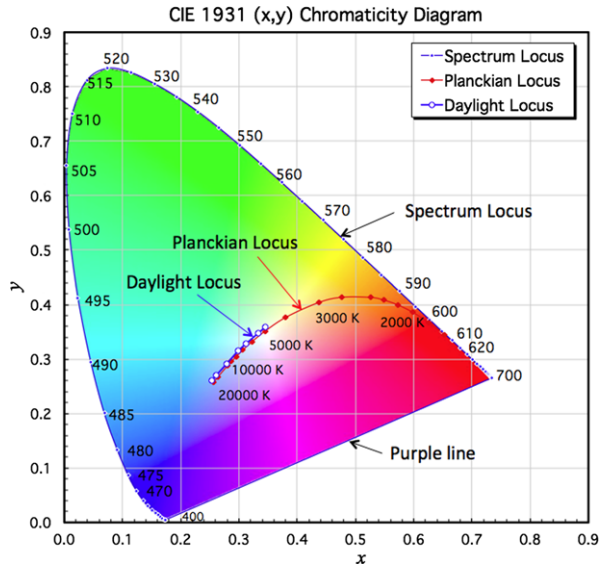
5000 K, 5500 K, 6500 K, and 7500 K are defined as standard illuminants D50, D55, D65, and D75 [1, 20]. The boundary between the two curves is not definite, and the 5000 K is used, e.g. in [6], for standardization purpose only. Also, if such a curve for center white exists, there should be a smooth transition between the two curves. The traditional chromaticity specifications of fluorescent lamps [3] somewhat follow such a smooth transition between the Planckian and the daylight locus, though not exactly.

The CIE also specifies the color matching functions $\bar{x}_{10}(\lambda)$, $\bar{y}_{10}(\lambda)$, $\bar{z}_{10}(\lambda)$, for the 10 degree field of view [1], and the corresponding chromaticity coordinates are
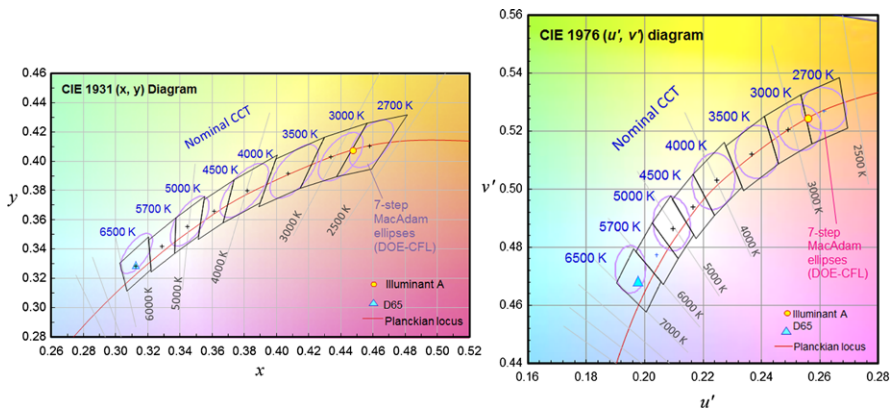
Fig. 13.4 The chromaticity specifications for solid-state lighting products by ANSI C78.377 [2] on the CIE (x, y) diagram (*left*) and on the (u', v') diagram (*right*)

also available but these are mainly used for object color specification and generally not used for specification of light sources.

Light sources for general lighting must have acceptable white light chromaticity. ANSI C78.377 [2] specifies the ranges of chromaticity for solid state lighting products as illustrated in Fig. 13.4, and is often used as the basis for color binning of white LED products. There are eight nominal CCTs in this standard, and also the Flexible CCT specification is provided as an option, in which nominal CCTs at every 100 K can be selected, and the formulae to calculate each quadrangle are provided.

13.2.2 CCT and Duv

Correlated Color Temperature (CCT, Symbol: T_{cp}) is defined as the temperature of the Planckian radiator having the chromaticity that is closest to the chromaticity of the given light source, on the CIE $(u', 2/3v')$ coordinate, as illustrated in Fig. 13.5. A more official definition is available elsewhere [1, 5].

The $(u', 2/3v')$ coordinate is the same as now-obsolete CIE 1960 (u, v) coordinate. The Planck's equation generally used in photometry and radiometry is given in the form:

$$L_{e,\lambda}(\lambda, T) = \frac{c_1 n^{-2} \lambda^{-5}}{\pi} \left[\exp\left(\frac{c_2}{n\lambda T} \right) - 1 \right]^{-1} \qquad (13.7)$$

where L is the spectral radiance, T is the temperature [K], n is the refractive index of the medium. For historical reasons, $n = 1$ (exactly 1) has been used and must be used for calculation of CCT. The parameters c_1, c_2 are the radiation constants; but only c_2 is relevant to the value of CCT. The value of c_2 used in the calculation of

Fig. 13.5 Illustration of the definition of CCT on the 1960 (u, v) diagram (now obsolete)
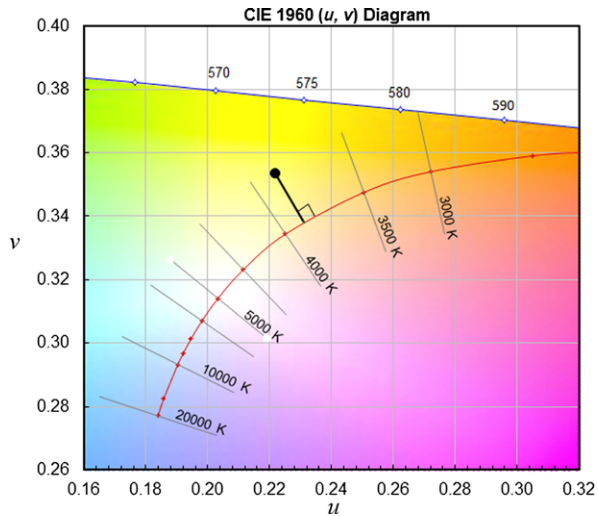
Fig. 13.6 The scales of Duv shown on the 1960 (u, v) diagram
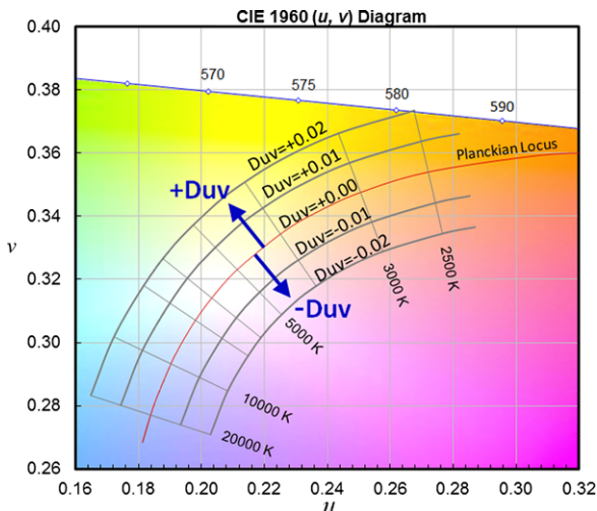
CCT must follow the value adopted in the current International Temperature Scale (ITS-90), namely $c_2 = 1.4388 \times 10^{-2}$ m·K. See Appendix E of CIE 15:2004 [1] for the details of these constants used for calculation of CCT.

Duv (symbol: D_{uv}) is defined as the closest distance from the Planckian locus on the CIE $(u', 2/3v')$ coordinates, with $+$ sign for above and $-$ sign for below the Planckian locus [2]. The distance is measured on the CIE 1960 (u, v) diagram. A positive value of Duv indicates a shift in yellowish/greenish direction, and a negative value indicates a shift in pinkish direction. Figure 13.6 shows the scales of Duv depicted on the (u, v) diagram.

While the ANSI specification [2] is widely accepted, it is still a question whether the Planckian locus and Daylight locus truly represent most preferred white light for lighting. Some lighting engineers experience that light with slight shift of Duv to negative side provides better color quality perception. The same effect is experienced with the NIST spectrally tunable lighting facility [34]. This is a subject that needs thorough study in the future.

Duv can normally be calculated in the process of calculating CCT, which requires determination of the closest point on the Planckian locus to the (u, v) chromaticity coordinates of the light source being tested.

If it needs to be calculated separately, and if the CCT of the light source is known, Duv can be obtained by calculating the spectral distribution of Planckian radiation at the CCT of the source using the Planck's equation (Eq. 13.7), then calculate $(u_0', 2/3v_0')$ of that Planckian spectral distribution, then Duv is given by

$$D_{uv} = \left[(u' - u_0')^2 + \left(\frac{2}{3}v' - \frac{2}{3}v_0' \right)^2 \right]^{\frac{1}{2}} \cdot \mathrm{SIGN}(v' - v_0') \qquad (13.8)$$

where $\mathrm{SIGN}(z) = 1$ for $z \geq 0$ and $\mathrm{SIGN}(z) = -1$ for $z < 0$.

13.2.3 Color Differences for Light Source

Color difference of light sources need to be specified in many cases, for example, for angular color uniformity, shift of color during lifetime (color maintenance), and individual color variation of products. The color difference of light stimuli, in general, is expressed by the distance on CIE (u', v') diagram, which is the uniform color space for light stimuli officially recommended by the CIE. The term "color difference" should not be confused by that for object colors, which are measured as the Euclidean distance in a three-dimensional object color space such as CIE $L^*a^*b^*$ (also called CIELAB). Color difference of object color is briefly discussed in Sect. 13.3.1.

The color differences of light sources for illumination are often expressed by the size of MacAdam ellipses [18] as introduced in Sect. 13.2.1. For example, the ellipses included in Fig. 13.4 are seven-times the size of MacAdam ellipses, so-called 7-step MacAdam ellipses, used for specification of compact fluorescent lamps in USA [21]. The chromaticity specification for linear fluorescent lamps in USA [3] uses 4-step MacAdam ellipses. Similar expressions occasionally used are SDCM (standard deviation of color matching) or JND (just noticeable difference), meaning the color difference corresponding to 1-step MacAdam ellipses. However, these are not internationally defined standard metrics. While these served many years for fluorescent lamps, which have only six nominal CCTs (in the specifications in the USA [3]), this method does not work for new lighting technologies, which are more versatile. Note that the shape and size of the 7-step MacAdam ellipses in Fig. 13.4 are all slightly different. Their shape and size change depending on the chromaticity coordinates and they have to be interpolated between the original 25 chromaticity points measured by MacAdam (Fig. 13.1). Such interpolation formulae are not defined in any international standard. This is a problem for LED light sources that do not follow the traditional nominal CCTs of fluorescent lamps. For example, the 4500 K and 5700 K nominal CCTs or any of the flexible CCTs in the ANSI specification [2] do not have corresponding MacAdam ellipses defined. Even if interpolation of ellipses can be specified, it would be complicated for practical use.

As described in Sect. 13.2.1, the CIE (u', v') diagram is the uniform color space for light sources. Figure 13.7 shows the white region of the CIE (u', v') diagram, on which 7-step MacAdam ellipses (solid line) [21] and circles with radius of 0.008 (dashed line) are plotted on top of each other. As shown, these ellipses and the circles are closely overlapped, which means that circles can be used to replace MacAdam ellipses in this white region on (u', v') diagram. Previous research has already indicated this [22]. 7-step MacAdam ellipses are equivalent to the circles with radius 0.008. The n step roughly corresponds to $0.001 \times n$ in radius on the (u', v') diagram. There are small differences between the circles and the ellipses, but it is considered insignificant in practical applications. Especially, the experimental uncertainty of MacAdam ellipses measured 70 years ago with a small number of experimental subjects is not well known. Rather than MacAdam ellipses, it is recommended that simply the distance in (u', v'), e.g., expressed as $\Delta u'v'$, be used to

Fig. 13.7 7-step MacAdam
ellipses and circles (radius
0.008) on the CIE (u', v')
diagram
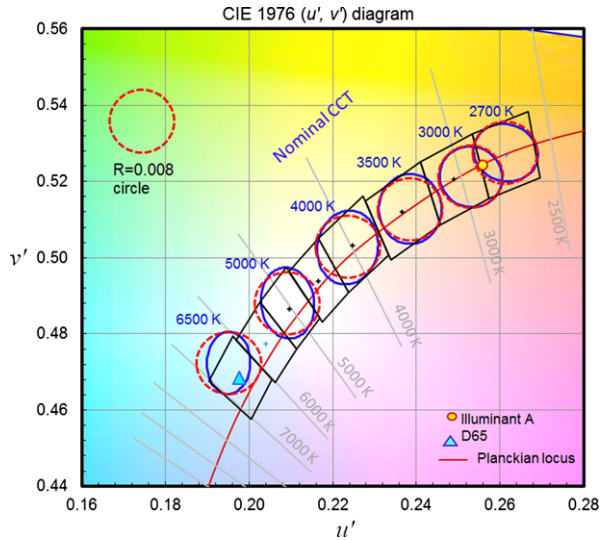
specify the color uniformity, color maintenance, and color variation of SSL prod-
ucts.

13.3 Color Rendering Characteristics

13.3.1 Object Color Evaluation

Color rendering is a phenomenon in which colors of objects appear differently under
illumination by different spectra. Thus, in discussing color rendering, it is impor-
tant to understand how object colors and their differences are evaluated, so a short
introduction is given in this section. The essential component of evaluating color
rendering is to measure or calculate color differences of objects under given illumi-
nation, and this approach is used in CRI and other new proposed metrics. Unlike
light source color, which has two dimensions, object color requires another dimen-
sion, lightness—a scale from black, gray, to white. For example, there is no brown
or black in the (x, y) diagram. Brown color has the same chromaticity as red but
has a different lightness. Therefore, a three-dimensional color space is needed for
object color.

There are two standardized object color spaces defined by CIE, which are $L^*a^*b^*$
(CIELAB) [1, 23], and $L^*u^*v^*$ (CIELUV) [1, 24]. CIELAB is commonly used in
lighting research as well as imaging applications, and is illustrated in Fig. 13.8.
The horizontal two axes, a^* and b^*, determine *hue* and *chroma*, and the vertical
axis L^* gives *lightness*. The center is gray, and the further a point is shifted outside
horizontally, the more colorful the object is. The three coordinates L^*, a^*, b^* of an

Fig. 13.8 Illustration of CIELAB object color space
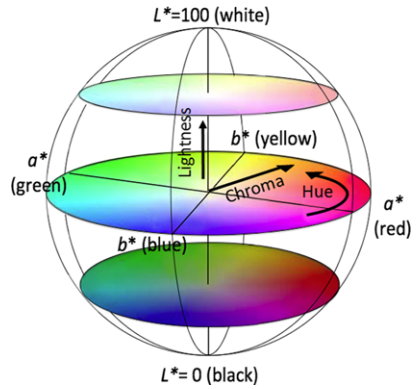
Fig. 13.9 The CIELAB (a^*, b^*) plots of 1269 Munsell color samples under D65
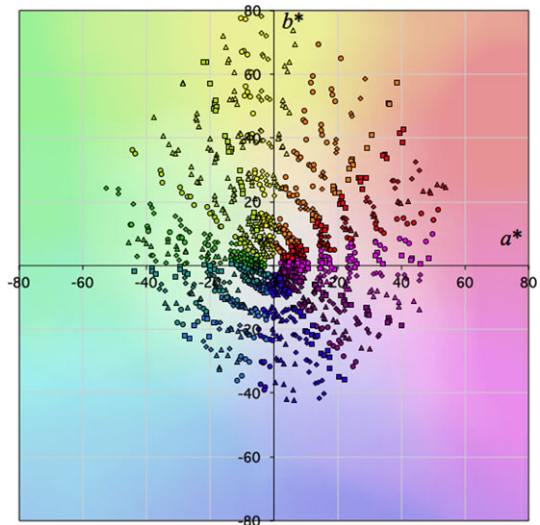
object surface are calculated by

$$L^* = 116(Y/Y_n)^{1/3} - 16$$
$$a^* = 500\left[(X/X_n)^{1/3} - (Y/Y_n)^{1/3}\right]$$
$$b^* = 200\left[(Y/Y_n)^{1/3} - (Z/Z_n)^{1/3}\right] \qquad (13.9)$$
$$\text{(when } X/X_n, Y/Y_n, Z/Z_n > 0.008856)$$

where X, Y, Z are the tristimulus values of the object surface, and X_n, Y_n, Z_n are the tristimulus values of the white reference (perfect diffuser) under a given source of illumination. Note that there are further formulae for samples with very low reflectance, but these are practically not needed. To specify object colors, one
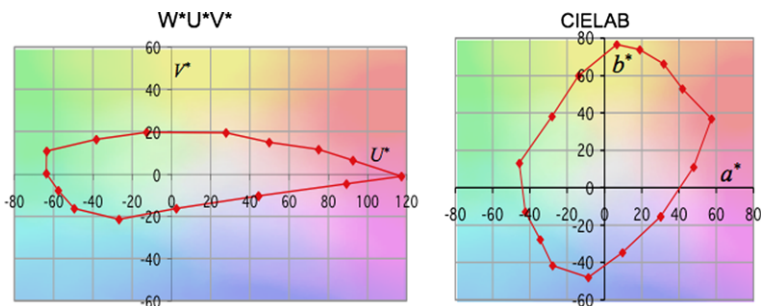
Fig. 13.10 Two-dimensional plots of 15 saturated color samples under Illuminant A in $W^*U^*V^*$ and CIELAB object color space

of standard illuminants such as CIE D65 [20] is used to calculate the tristimulus values. The color difference between two objects are calculated by

$$\Delta E_{ab}^* = \left[\left(\Delta L^*\right)^2 + \left(\Delta a^*\right)^2 + \left(\Delta b^*\right)^2 \right]^{1/2} \qquad (13.10)$$

For evaluation of color rendering of light sources, object colors are calculated with the reference source and given light source, and the difference ΔE_{ab}^* between them are evaluated. Since it is not convenient to draw 3-dimensional diagram, only (a^*, b^*) is often plotted as a two-dimensional graph. As an example, Fig. 13.9 shows (a^*, b^*) plots of 1269 Munsell color samples [25] under D65.

The 1964 $W^*U^*V^*$ object color space used in the CRI [6] is the same concept as CIELAB but is very outdated and nonuniform. Figure 13.10 shows the comparison of the two dimensional plots of 15 saturated color samples (from Munsell) under standard Illuminant A (2856 K blackbody) on $W^*U^*V^*$ and CIELAB. It shows that the red area on $W^*U^*V^*$ is extremely stretched compared to the yellow and blue region.

13.3.2 Color Rendering Index

Color rendering is defined by the CIE [5] as "effect of an illuminant on the color appearance of objects by conscious or subconscious comparison with their color appearance under a reference illuminant". The CIE color rendering index (CRI) [6] is widely used and the only internationally accepted metric for assessing the color rendering performance of light sources.

The details of procedure for the calculation are available in [6], but a brief introduction is provided in this section. The first step of the calculation is to calculate the color differences ΔE (in the 1964 $W^*U^*V^*$ object color space—now obsolete) of 14 selected Munsell samples when illuminated by a reference illuminant and when illuminated by a given light source (spectrum). The first eight samples are medium saturated colors, and the last six include four highly saturated colors (red, yellow, green, and blue), skin complexion, and leaf green. These samples are

Fig. 13.11 The 14 test color samples used in the calculation of CRI (color is not shown in printed version)

TCS01 TCS02 TCS03 TCS04 TCS05 TCS06 TCS07 TCS08

TCS09 TCS10 TCS11 TCS12 TCS13 TCS14
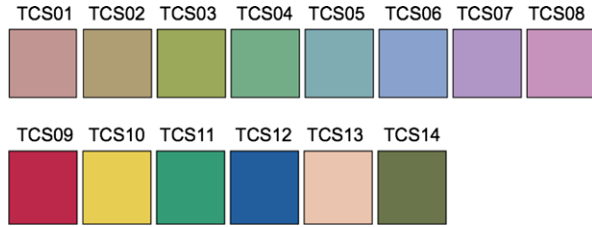
shown in Fig. 13.11. The reference illuminant is Planckian radiation for test light sources having a correlated color temperature (CCT) <5000 K, or a phase of daylight (one of daylight spectra at varied CCT; the formula is available in Ref. [1]) for test sources with CCT ≥ 5000 K. The chromaticities of the test samples are adjusted by the von Kries chromatic adaptation transformation. The *Special Color Rendering Indices* R_i for each color sample are obtained by

$$R_i = 100 - 4.6 \Delta E_i \quad (i = 1, \ldots, 14) \tag{13.11}$$

This gives the evaluation of color rendering for each particular object sample color. The *General Color Rendering Index* R_a is calculated as the average of R_i for the first eight color samples:

$$R_a = \sum_{i=1}^{8} R_i / 8 \tag{13.12}$$

The score for perfect color rendering (zero color differences) is 100, and the values of R_i or even R_a can be below zero (negative) if color differences are extremely large. Note that the term CRI is often used referring to R_a, but the CRI actually consists of 15 numbers; R_a and R_i ($i = 1$ to 14).

The values of $R_a > 80$ are generally considered acceptable for indoor lighting and $R_a > 90$ is generally considered excellent color rendering. The special color rendering indices R_i are not often used except R_9 (index for the saturated red sample), which is often examined as red is critical for skin tone and overall color rendering and also to supplement R_a to address some of its problems. Note that color rendering is often not critical for outdoor lighting and $R_a < 80$ can be well used for such applications.

13.3.3 Shortcomings of CRI

The CRI was first developed in the 1960s to evaluate the color-rendering performance of then-new fluorescent lamps. The formulae used in the current version of CRI date back to 1974 (CRI 2nd edition) using $W^*U^*V^*$ object color space and von Kries chromatic adaptation transform formula, which are now outdated and known to be very inaccurate. The number of test color samples (eight) used in the CRI R_a has been considered to be too few. In spite of these known shortcomings, the CRI served well for traditional light sources (discharge lamps) for over 35 years, and the

lighting industry did not find serious problems with the CRI, rejecting a recent revision attempt [26]. However, some serious problems of the CRI when applied to LED sources have been identified and needs for improvements are addressed [27–31] and also by CIE [7]. The CRI R_a score often does not correlate well with visual evaluation, particularly with RGB(A) LED sources, and such problems can occur for other types of narrowband source such as narrowband phosphor LEDs and quantum dots. Several aspects of problems of the CRI are explained below. These problems of the CRI should be noted when it is used.

13.3.3.1 CRI Penalizes Preferred Color-Enhancing Lights

The CRI R_a score does not correlate well with visual evaluation when the chroma (color saturation) of objects is increased by the light source. This is well demonstrated by the case of a neodymium lamp,[1] as shown in Fig. 13.12 (left). The lower chart shows CIELAB (a^*, b^*) plots of the 15 samples shown in Fig. 13.15 under the
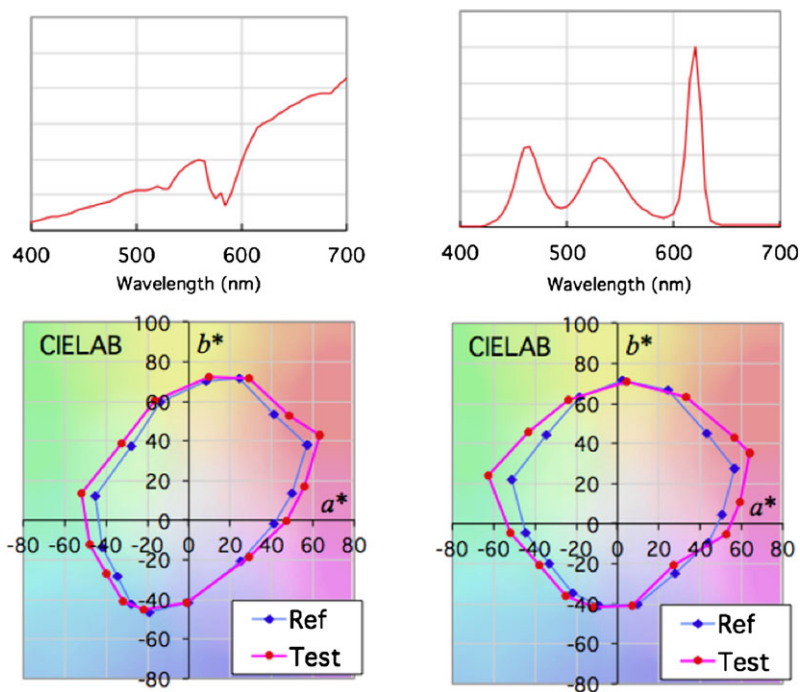
Fig. 13.12 Examples of color enhancement by a neodymium lamp and an RGB LED source

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[1]Incandescent lamp with its glass envelope doped with neodymium, which absorbs much of the yellow portion of the lamp spectrum. This type of lamp is popularly sold in the USA and some other countries.

reference light and the test light. This lamp produces a slight increase of color gamut in the red and green region, and the color contrast of objects is slightly enhanced. Object colors under this light look more vivid, and this lamp is often preferred over normal incandescent lamps ($R_a = 100$). But this lamp is given a CRI score of only $R_a = 77$. Neodymium lamps are popularly sold in USA and this evidence suggests that this lamp deserves a much higher score. Figure 13.12 (right) shows an example made by a real RGB source, which is highly rated by visual evaluation but is given a CRI score of $R_a = 71$. The red-green contrast is said to be very important for overall color rendering of a light source [32], and enhancement of red and green is considered to be good trait of lighting source as long as it is not excessive. The R_a score tends to be much reduced for enhanced red due to very nonuniform $W^*U^*V^*$, which is extraordinarily stretched in the red region (see Fig. 13.10).

13.3.3.2 A High CRI Score Does Not Guarantee Good Color Quality

There are some cases where the CRI gives fairly high scores though some saturated object colors appear very poorly. An example in Fig. 13.13 shows a case of RGB white LED simulation optimized for highest luminous efficacy of radiation at CRI $R_a = 80$. This light would qualify for ENERGY STAR [33] and achieve high theoretical luminous efficacy, but the actual color rendering is very poor with $R_9 = -112$. Skin tones under this light look pale and very unhealthy. This problem

Fig. 13.13 An example of RGB spectrum (simulation) with very poor red rendering
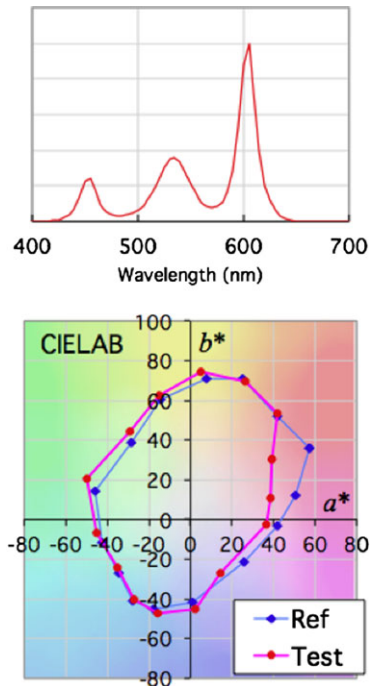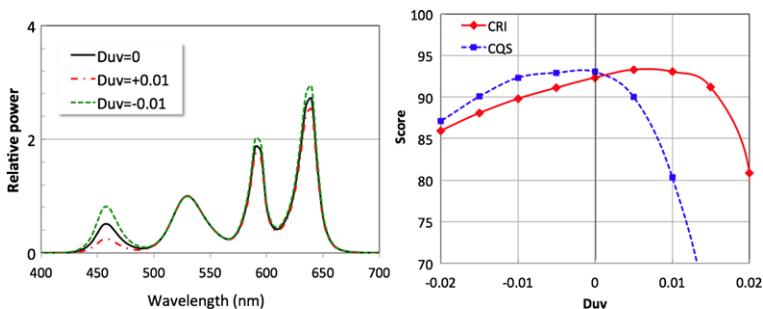
Fig. 13.14 Change of CRI R_a score with shifts of Duv on an RGBA spectrum

occurs due to the use of medium saturated test color samples in CRI, which fail to detect problems with saturated color samples under such narrow-band sources.

13.3.3.3 Problem with Duv Shifts

There are some cases where the CRI R_a score does not reflect the effects of Duv shifts correctly. The example in Fig. 13.14 (left) shows the case of an RGBA spectrum with a CCT of 2700 K with shifts of Duv from -0.02 to $+0.02$ (varied by only changing the ratios of four peaks). These spectra were created in the National Institute of Standards and Technology (NIST) Spectrally Tunable Lighting Facility [34]. NIST is conducting vision experiments on Duv shifts, and preliminary data indicates that yellowish shifts are disliked especially at low CCT ranges and slight shifts toward negative Duv are preferred. Figure 13.14 (right) shows that the CRI R_a score (solid curve) increases at positive Duv up to $+0.01$. This is the opposite of the visual observation. Color quality scale (CQS) scores (see next section), on the other hand, penalize the yellow shifts (dashed line). In this example, the CRI favors yellowish shifts, which does not reflect user perception and preference. This difference between the CRI and CQS comes from the chromatic adaptation transform. The von Kries formula used in the CRI is very outdated and considered not working accurately. The CQS uses one of the latest chromatic adaptation transforms, CMCCAT 2000 [35, 36].[2]

13.3.4 Color Quality Beyond CRI

The CRI assesses the fidelity of color rendition (color fidelity) with respect to a reference illuminant. It has long been known that color fidelity cannot serve as the

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[2]CMCCAT represents Color Measurement Committee (of the Society of Dyers and Colourists for predicting the degree of color inconsistency of surface colors) Chromatic Adaptation Transform.

sole indicator of color quality for lighting. There are other aspects of color quality, such as general color preference as well explained in [13, 15] and color discrimination [14], which influence the visual experience of color rendering. In particular, object color shifts that increase color saturation enhances visual clarity and color discrimination ability, and improves the subjective rating of color appearance of the illuminated scene. It is also known that perceived chroma (color saturation or colorfulness) of objects is reduced at lower light levels (Hunt Effect) [37]. Therefore, it seems that sources that enhance color slightly in indoor lighting conditions may achieve higher color fidelity to real daylight (which has much higher illuminance than indoor lighting). An attempt was made to measure this effect [38]. In the past, there have been attempts to account for these effects beyond color fidelity by various proposed metrics such as the Flattery Index [13], the Color Discrimination Index [14], the Color Preference Index [15], and more recently, the Feeling of Contrast Index [39], and the Gamut Area Index [12]. None of them, however, has been accepted as a standard, probably due to the fact that the CRI has been so widely used (in spite of its problems) and that the lighting industry does not want to use more than one number to rate color rendering of products.

Most recently, the Color Quality Scale [8] has been developed to address such needs providing a one-number output. The CQS was developed especially to solve the problems of the CRI for LED sources described in the previous section. The CQS is designed to provide a score that correlates well with visual evaluation of color rendering. For example, for the neodymium lamp shown in Fig. 13.12, CQS gives a score of 90 compared to the CRI score of 77. While this evaluation procedure is yet to be standardized, this metric and its associated analyses can be used as a design tool for color quality of SSL products. The CQS is basically a fidelity metric producing one number like R_a with a maximum score of 100, but the CQS has improvements including the ability to account for the direction of object color shifts, which deviate from a pure fidelity metric. The CQS score, therefore, agrees well with visual ratings for color-enhancing sources, while it works as an accurate fidelity metric for non-color-enhancing sources. The CQS is designed to evaluate the overall color quality of products for all types of light sources including narrowband LED sources and traditional lamps.

The detailed formulae of the CQS are available [8] but important features are explained below. The CQS has the same calculation structure as the CRI, with many improvements of components as below.

(1) 15 highly saturated test color samples selected from the 1200 Munsell color samples as shown in Fig. 13.15. They are: 5R4-14, 10R6-12, 7.5YR7-12, 5Y8-12, 2.5GY8-10, 7.5GY7-10, 2.5G6-12, 2.5BG6-10, 10BG6-8, 7.5B5-10,

Fig. 13.15 The 15 saturated test color samples used in the CQS (color is not shown in printed version)

2.5PB4-10, 7.5PB4-12, 5P5-10, 2.5RP6-12, 7.5RP5-12. This allows the metric to detect problems with rendering of saturated color objects as discussed in Sect. 13.3.3.2 (*A high CRI score does not guarantee good color quality.*) Doubling the number of samples also improves for the problem of the CRI, that allows gaming of the metric in spectral design.

(2) CIELAB object color space (see Sect. 13.3.1). This replaces the obsolete $W^*U^*V^*$ color space and mitigates the problem discussed in Sect. 13.3.3.1 (*CRI penalizes preferred color-enhancing light*). The plots of the 15 test color samples of the CQS on the CIELAB (a^*, b^*) graph provide intuitive information of color rendering. Examples have already been shown in Figs. 13.12 and 13.13.

(3) CMCCAT2000 chromatic adaptation transform [36]. This provides more accurate prediction of effects of Duv shifts on color rendering and ameliorates the problem discussed in the previous section (*Problem with Duv shifts*).

(4) The *saturation factor*, with which the increase of chroma is not penalized for a limited range of chroma increases ($\Delta C_{ab}^* < 10$). This overcomes the problem of the CRI unduly penalizing chroma-enhancing sources as discussed in Sect. 13.3.3. The CQS starts penalizing excessive chroma increases ($\Delta C_{ab}^* > 10$) when perceived color rendering starts appearing unnatural.

(5) Root-mean-square (RMS) averaging. This lowers the score if one or two of the 15 samples have very poor color rendering (example in Fig. 13.13), compared to simple averaging. This further addresses the issue discussed in Sect. 13.3.3 (*A high CRI score does not guarantee good color quality.*)

(6) The score is on a 0–100 scale, so that there can be no negative scores.

Note that the information above is on the current version of the CQS (version 9.0), which includes a few small improvements from the contents of publication [8]. Several of the 15 test samples in (1) were reselected for more uniform spacing in hue angle (listed above are the current sample sets). The ΔC_{ab}^* limit in (4) above was introduced based on the later experiments done at NIST. The CCT factor was dropped, as it has not been verified experimentally. The CQS calculation program is available from the author.

13.4 Luminous Efficacy of Radiation

The spectrum of light source affects not only color rendering but also energy efficiency (luminous efficacy) of light source. Thus color rendering is closely related to energy issues. A brief introduction on luminous efficacy is provided here.

The energy efficiency of a light source or a lighting product is evaluated by *luminous efficacy of a source* (often called simply *luminous efficacy*), which is the ratio of *luminous flux* (lumen) emitted by the source to the input electrical power (watt). The luminous efficacy of a source, η_v [lm/W], is determined by two factors:

$$\eta_v = \eta_e \cdot K \tag{13.13}$$

Fig. 13.16 Luminous
efficacy of monochromatic
radiation (*solid line*), plotted
with an example of LED
spectrum (*dashed line*)
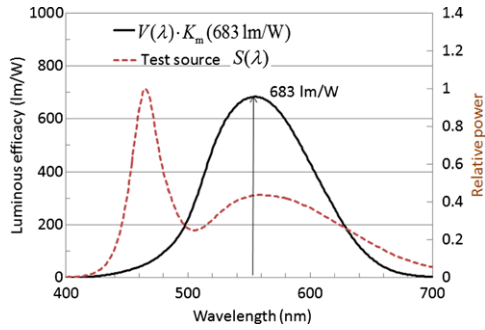
where η_e is the *radiant efficiency* of the source (ratio of radiant flux output to input electrical power; "external quantum efficiency" is often used with the same meaning). K is the *luminous efficacy of radiation* (ratio of luminous flux to radiant flux, denoted as LER in this chapter), and is determined by the spectral distribution $S(\lambda)$ of the source as given by,

$$K = \frac{K_m \int_\lambda V(\lambda)S(\lambda)\mathrm{d}\lambda}{\int_\lambda S(\lambda)\mathrm{d}\lambda}, \quad \text{where } K_m = 683 \; [\mathrm{lm/W}] \qquad (13.14)$$

K_m is *the maximum spectral luminous efficacy*, and its value, 683 lm/W (for monochromatic radiation at 555 nm) comes from the international definition of the candela. While various other terms are used in the LED industry, the terms introduced above are the ones officially defined internationally [5]. Figure 13.16 illustrates the calculation of the LER. This LED spectra has an LER value of $K = 294$ (lm/W). The LER of common discharge lamps range from ~250 to ~380 lm/W. The maximum LER for white light with acceptable color rendering is considered to be around 400 lm/W. The LER is entirely dependent on the SPD of the light source and represents the luminous efficacy that would be possible if the conversion of electricity to light could happen without any losses.

13.5 Color Characteristics for Single Color LEDs

Chromaticity, as described in Sect. 13.2.1, is used for all LED devices including white and single color LEDs. In addition to chromaticity, the characteristics below are used for single-color LEDs.

13.5.1 Dominant Wavelength λ_d

The dominant wavelength is commonly used to specify the color of single-color LEDs. It is defined in [5] as follows:

The wavelength of a monochromatic stimulus that, when additively mixed in suitable proportions with the specified achromatic stimulus, matches the color stimulus considered in the 1931 (x, y) chromaticity diagram.

Fig. 13.17 Illustration of dominant wavelength
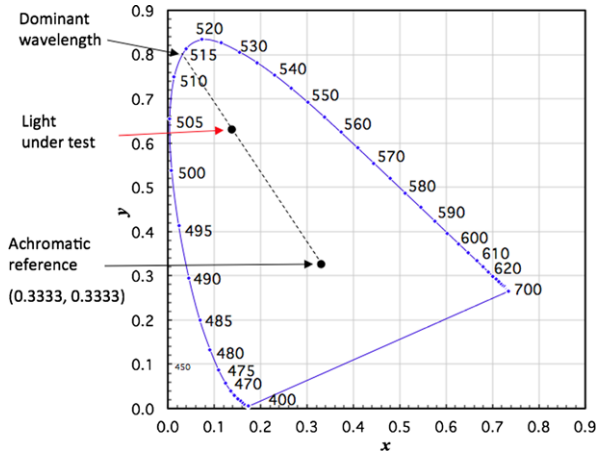
Fig. 13.18 An example of dominant wavelength and peak wavelength
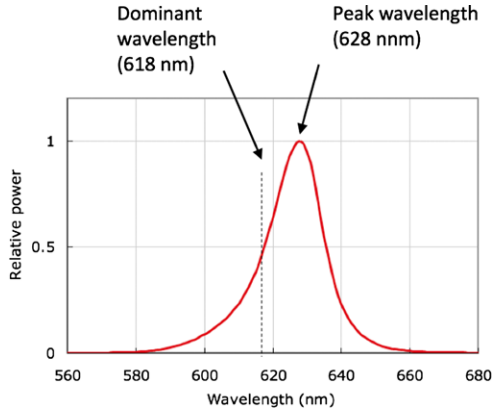
This is illustrated in Fig. 13.17. For characterizing LEDs, the reference achromatic stimulus should be an equal-energy spectrum (sometimes called Illuminant E), which has the chromaticity coordinates $x_E = 0.3333$, $y_E = 0.3333$ [40]. It should be noted that dominant wavelengths of red LEDs are shorter than their peak wavelengths as shown in Fig. 13.18, and longer for blue LEDs. This is due to the slope of the color matching functions in the red and blue region, where deeper red and blue contribute less to color perception. The difference can be greater than 10 nm for deep red and deep blue LEDs.

13.5.2 Centroid Wavelength λ_c

The centroid wavelength λ_c is defined in [40] and is calculated as the "center of gravity wavelength" by

$$\lambda_c = \int_{\lambda_1}^{\lambda_2} \lambda \cdot S(\lambda)d\lambda \Big/ \int_{\lambda_1}^{\lambda_2} S(\lambda)d\lambda \qquad (13.15)$$

The diminishing tails of LED spectrum (including noise and stray light) may affect the value of centroid wavelength significantly, which should be considered as an uncertainty in a given measurement.

13.5.3 Peak Wavelength λ_p

As defined in Ref. [40], the wavelength at the maximum of the spectral power distribution is known as the peak wavelength λ_p. The (absolute) spectral distribution is usually normalized at this wavelength rather than at an arbitrary wavelength, to give a relative spectral distribution with a maximum value of unity. See an example in Fig. 13.18.

13.6 Future Considerations on Color Quality for White LED Developments

By studying the CRI, people may be led to believe that white LED spectra should mimic the spectrum of the sun or a blackbody for best color rendering. While such spectra would certainly give high CRI values and provide excellent color rendering, they would not be very energy efficient. The main driving force for solid-state lighting is for huge energy savings on a national or global scale. The challenge in developing LEDs for use as illumination sources is to provide the highest possible energy efficiency while achieving the best color rendering possible for intended applications. These two factors tend to be in a trade-off relationship, and the balance between the two is often a compromise. However, there are some ways to design light sources for which both merits are high.

Phosphor white LEDs are currently dominant in the market and play important roles in introducing solid state lighting with competitive energy efficiency and affordable cost that meet current needs. However, there is the potential of achieving much higher energy efficiency and excellent color rendering using other technologies such as RGB, RGBA, and narrowband phosphors, quantum dots, etc., even though these technologies still have technical challenges and cost issues. Narrowband combinations, having deep valleys in the spectrum, may not be considered as good color rendering sources, but if the peaks are selected properly, they can provide excellent color rendering, which has been well demonstrated at the NIST facility [34].

Figure 13.19 shows the spectra and color analyses of examples of various types of white LEDs currently available. From the left, (1) is an example of a typical warm white phosphor LED lamp found in the market. This example has an LER of 305 lm/W, and the LER of other similar products range from 300 to 320 lm/W. While these products are meeting current needs, higher color rendering is desired. The main issue is that the red object rendering of these products is not sufficient, as
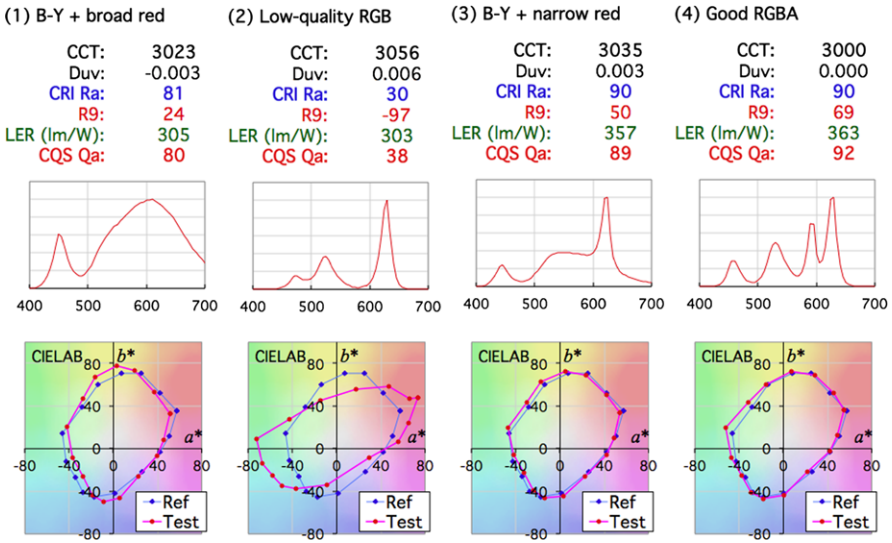
Fig. 13.19 Color rendering analyses of examples of real white LED products

can be seen on the CIELAB chart. To improve this, a deeper red phosphor would be needed, but the LER will then drop further. (2) is an example of a typical low-quality RGB product found in the market, exhibiting very poor color rendering and low LER of ∼300 lm/W. (3) is a hybrid product using phosphor and a red LED peak, having much increased LER of 357 lm/W and high CRI of $R_a = 90$, and is considered one of solutions that overcome the limitations of broadband phosphor technology. (4) is an RGBA spectrum with selected single-color LED packages commercially available, showing LER of 363 lm/W and $R_a = 90$ and $Q_a = 92$. Examples (3) and (4) show ∼20 % higher LER (lm/W) than the typical phosphor warm white LED, while achieving much higher scores of the CRI and CQS.

Further analyses not limited to currently available LEDs show results of some RGB model (3000 K) with LER of 378 lm/W with $R_a = 86$, $R_9 = 90$, and an RGBA model (3000 K) with LER 382 lm/W and $R_a = 91$, $Q_a = 90$. Such LED sources can realize very high luminous efficacy as well as excellent color rendering. These models use LED peaks such as 547 nm or 576 nm. High power LEDs with such peak wavelengths are currently not available. Development of efficient high power LEDs in such deep green and yellow regions will make such high performance RGB or RGBA systems possible.

Fluorescent lamps evolved from broadband to narrowband throughout their long history to better achieve both high efficacy and good color rending. As discussed above, RGB or RGBA technologies, though they still have technical challenges, have theoretically higher luminous efficacies. As well as having no Stokes loss, they have a potential of reaching 20 to 30 % higher luminous efficacy than the limit of the broadband phosphor LED technology. Narrowband phosphors replacing broadband phosphors would also enable significant improvements in this respect. Narrowband

components may play important roles in the future developments of SSL, especially when broadband phosphor technology matures.

LED sources with slightly increased color gamut as discussed in Sect. 13.3.3 are known to have a merit of general preference. Such color-enhancement effects can be created with narrowband sources and may facilitate important new applications of SSL. The effects of negative Duv as discussed in Sect. 13.2.2 need to be studied further, and such light sources may also find popularity in the future.

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Chapter 14
Emerging System Level Applications for LED Technology

Robert F. Karlicek Jr.

Abstract This chapter describes possible future system level applications of LED technology from the viewpoint of emerging trends in lighting related human health, communications, and display technologies. Currently, almost every aspect of LED technology from substrates to systems applications is undergoing rapid technical evolution, so projections about future system level applications are highly speculative, and this assessment presumes that solutions to long standing LED issues like droop and poor green/yellow performance will ultimately be found. Future system level applications will also be tightly convolved with the system level incorporation of new types of light sensors and embedded processing capabilities so that feedback loops between the light source, the environment and the control system can be closed. Future systems level applications will also be tightly convolved the development of new features and services needed to extend business revenue models of lighting companies as the progress in LED system reliability drives future applications in lighting to the point that bulb/socket commodity business models begin to fail, and business models based on the offering of new lighting features and services are developed.

14.1 Introduction

The development of highly efficient LED technology, particularly for the generation of visible light has progressed rapidly, especially since the development of high brightness GaN based LEDs in the early 1990's [1]. Even as LED based lighting systems are now beginning to penetrate conventional lighting markets at commercially significant levels [2], there is still considerable, basic and developmental research focused on almost every fundamental aspect of the LED technology supply chain, from substrate type and device design to fixture level packaging and control systems architectures. While the main driver for this research is the substantial global lighting market and the realization of improved energy efficiency possible with LED

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based solid state lighting (SSL) relative to the widely used incandescent bulb, the
high efficiency and lower cost of some conventional lighting systems (i.e. fluores-
cent based technologies) will continue to limit adoption of SSL technologies for
illumination until the cost of SSL technology decreases significantly. Beyond cost,
the current widespread cultural acceptance of existing illumination system designs
and technology specific capabilities (e.g. CCT decrease with dimming for incandes-
cent) will continue to represent a challenge for SSL penetration in some applica-
tions. Finally, the very high reliability of LEDs (in properly designed systems) is
approaching the point where the long life of LED based illumination systems may
ultimately lead to the demise of well established bulb/socket based lighting systems.

These challenges in the lighting space can be contrasted with the incredibly rapid
penetration of LED technology in mobile video displays where the basic LED capa-
bilities of efficiency, size, and robustness are essential to the performance of these
sophisticated devices. Coupled with what seems to be an insatiable desire for people
everywhere to network, remain connected, access video information and function in
an increasingly untethered world, LEDs have become a basic part of a complex
engineered system that provides a valuable service that was never before possible.
In spite of their critical role in mobile communications in this market, LEDs sim-
ply provide white light, and the wider fundamental capabilities (i.e. spectral control
properties and digital signaling capabilities) of LED technology are largely unused.
As in the case of mobile video, the future value of new markets enabled by LEDs
will largely be driven by system level capabilities that provide unique, value added
solutions for overcoming challenges in energy efficiency, human health and well-
being, and productivity. This chapter examines some of the system considerations
and applications, some now emerging, where broader advantages of LED system
level technology can play a unique, enabling role in developing new illumination
applications.

14.2 Advanced Lighting Systems

Solid state lighting is at its core, a transition from electric to electronic lighting,
much as the transistor was a transition from electric to electronic switching. At
the semiconductor technology level, comparisons are even being made between the
well known Moore's Law [3] for transistor technology and Haitz's Law[1] for LED/IC
roadmapping efforts [4]. While visible LEDs were first demonstrated in 1962 [5],
the real breakthrough in LED technology required the demonstration of high bright-
ness GaN based LEDs about 30 years later. Chronologically, one might say that
the {transistor–systems–market} ecosystem is running about four to five decades
ahead of its {semiconductor diode SSL} sibling, so one can consider the evolution
of advanced lighting systems to still be in a rather early developmental phase.

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[1]Haitz's Law was first mentioned in at LED conferences 2000. A more detailed description can be
viewed at Nature Photonics 1:23, 2007.

While terms like "Smart" or "Intelligent" are increasingly being applied to SSL technologies, most of these technologies focus on a merger of basic communications technology (wireless or wired) to effect rather rudimentary control of lighting functions like dimming and remote controlled on/off switching. While elaborate lighting control systems exist today, they are rarely selected by lighting designers because they cannot easily adapt to post installation interior design changes without recalibration or programming changes. The US Department of Energy's (DOE) recent data on lighting installations suggests that fewer most US lighting installations lack even simple dimming controls in spite of the potential energy savings that automatic controls could bring to lighting systems [6].

Perhaps one of the most important drivers for advanced LED based lighting systems development stems from the long lifetime of properly designed LED systems. In spite of current limitations of some LED driver lifetimes and photothermal degradation of certain types of encapsulation and secondary optics in LED fixtures (both are addressable issues with continued research), future LED system lifetimes will be measured in decades and will outlive the design life of both commercial and residential interior spaces.[2] In such a scenario, LED systems may be sold more as durable goods or as an integral part of construction systems, with revenue models for LED system manufacturers depending increasingly upon providing new features and services derived from perceived or real market needs and consumer benefits. This vision is already emerging as companies seek to gain a competitive advantage in the nascent SSL market space by offering lighting with embedded operating systems [7] and simple video functions [8, 9] mirroring the development of the evolving smart TV market space.

Presuming that continued research on the materials and device designs used to manufacture LEDs will ultimately address common LED maladies like droop (nonthermal) and green (and yellow) gap issues, then efficient lighting systems with embedded intelligence that can control color dynamically and be modulated at high speed at any wavelength will open a wide range of new system level capabilities that offer new functions and services that go well beyond the simple mapping of legacy communications and control technologies to LED replacement bulb and fixture technology. Identification of possible future lighting system applications outlined here involve the integration of LED lighting systems with impact on human health and well being, improved and expanded access to wireless data, and the integration with video systems. These services will be constrained, to some degree, by the ability to provide these services at a reasonable energy cost relative to the benefit provided. This view is captured schematically in Fig. 14.1, and three of the non-energy related future system level applications are briefly described here.

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[2]Based on numerous conversations with lighting designers, the estimated average interior lighting design lifetime for retail and commercial spaces is between 7 and 10 years, approximately equal to or less than the lifetime of well designed SSL systems.

Fig. 14.1 Future illumination systems will balance energy efficiency with new features and services in areas of human health and well being, communications and integrated video, illumination capabilities
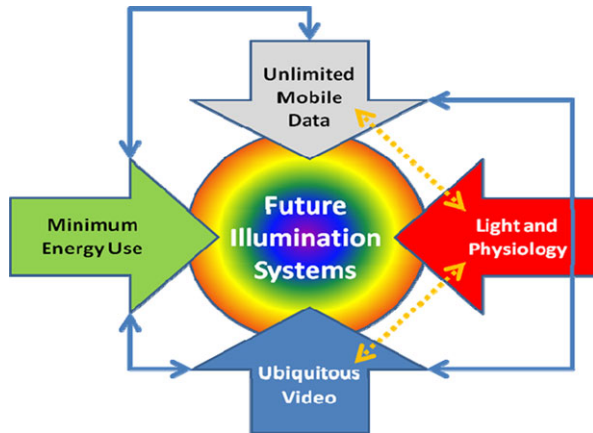
14.2.1 New Applications in the Field of Human Health and Wellbeing

Throughout the history of lighting, illumination has been inextricably involved with human well being and productivity [10], with extensive research on electric lighting design's impact on human performance starting shortly after the introduction of the incandescent light bulb [11]. The Lighting Handbook published by Illumination Engineering Society of North America (IESNA) [12], provides basic guidelines on lighting design and human factors illumination considerations, including some discussion of lighting applications using LED technology. Most of this guidance pertains to how humans respond to nominally conventional white lighting sources in terms of safety, comfort and satisfaction with the quality of light. The first wave of SSL products now available seek to replicate or marginally improve on the performance of existing white light incandescent or fluorescent sources, with the control of spectral output limited to what can be achieved at a reasonably high efficacy with phosphor converted blue or near UV LED sources. SSL spectral output is typically static (except during the initial warm-up period), though lighting systems with dynamic tuning between cool white and warm white LEDs have been available for several years, and color tunable bulbs that are wirelessly programmable for a wide variety of color functions have become available recently.[3] Ideally, LED systems should offer better spectral control of illumination and can enable a much wider range of color and gamut tuning capabilities, but it is not clear that these features will lead to broader or more rapid adoption of solid state lighting. Studies of artificial lighting and human satisfaction can be difficult to perform and outcomes can be impacted by notions that artificial light (perhaps regardless of the source) is of lower quality or less desirable than natural daylight [13].

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[3]New color tunable bulb from Philips permits the interface of smart devices and the bulb through a wireless connection, see http://www.meethue.com/en-US. Accessed 29 October, 2012.

In spite of issues with quantifying customer acceptance of different types of lighting sources, the spectral flexibility available with more advanced SSL systems will offer new capabilities with regard to improving human health and well being, with the benefits being in the area of circadian regulation, education and improved worker productivity. Research has shown that exposure to sunlight can have an impact on hospitalized patient health and wellbeing [14]. Tailoring the circadian color temperature of solid state illumination systems is now being explored in a hospital setting to study the impact on patient sleep.[4] It will be interesting to see just how much impact spectral control of lighting in healthcare environments will impact patient recovery, and how these findings would migrate to improved worker health and on-the job performance, where lighting quality is already a major factor in worker productivity and job satisfaction [15].

Research on light and sleep has a long history, but the discovery of non-visual light receptors in the human retina [16] and their role in setting and altering human circadian rhythm and sleeping patterns, coupled with the relationship with sleep problems and disease [17], suggest that future lighting systems could play an important role in human sleep management and health. Correlations of lighting and shift work has been linked to serious health issues like cancer [18] and heart disease ([19], and references therein), the possible impact lighting on shift worker health (and by extension to the general population) represents an important area of research. Several lighting products that use blue LEDs for personalized circadian rhythm management are already available on the market and undergoing serious evaluation [20]. These systems run "open loop", since circadian function is part of a closed loop control system where one of the elements of the control loop involves internal circadian regulation mechanisms. Ideally, the use of light to manage circadian entrainment would use some form of sensors to provide information on light dose and the human circadian status, and these sensors are under development.[5] It is probably unlikely to expect that most people would try to actively manage their circadian entrainment, but applications in relatively controlled healthcare environments (hospitals and eldercare facilities) are quite feasible.

Beyond impacting circadian rhythm and sleep, the timing, wavelength and magnitude of the lighting dose can impact the alertness and performance of humans for some time after the light exposure has occurred [21]. Also, new studies are emerging that suggest new approaches for control of light spectral content may be helpful for circadian management in shift workers. Recent studies on subtracting a narrow band of blue wavelengths from conventional lighting systems suggest that subtractive spectral methods may be useful for circadian rhythm management may have an impact on the health and alertness of shift workers [22]. Lighting spectral content is also being studied with regard

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[4]Philips Study at http://www.newscenter.philips.com/main/standard/news/press/2011/20111122-healwell.wpd. Accessed 3 April 2012.

[5]A research tool developed at Rensselaer's Lighting Research Center, http://www.lrc.rpi.edu/programs/lightHealth/projects/Dimesimeter.asp. Accessed 17 November 2012.

to impact on education and learning [23, 24]. Beyond simple illumination, studies on health related impacts of near UV radiation (common in sunlight) suggest that, while there are health concerns at much shorter UV wavelengths, certain health benefits may result from exposure to portions of the longer wavelength UVA region ([25], and references therein) and these considerations could be important in the design of solid state lighting systems with optimized UV output.

Human factors studies based on the new spectral tuning capabilities of solid state lighting systems are in their very early stages and much more work remains to be done. Initial studies, however, already indicate that systems capable of dynamically adjusting spectral content will be an important part of advanced lighting systems for optimizing human health and wellbeing. Given the globally soaring costs of health care, well designed lighting systems with dynamic spectral tuning capabilities may open lucrative new markets for solid state lighting systems.

14.2.2 Illumination with Communication

Communicating with light is commonplace (i.e. fiber optic communications), but lighting that communicates with visible light represents a new application uniquely enabled by solid state lighting, as LEDs can be modulated at relatively high speed. Referred to as Visible Light Communication (VLC) or more recently LiFi (like the RF based WiFi), this sort of functionality has been under development for some time [26]. Most VLC efforts focus on the wireless transmission of data at high bit rates as a means to augment or even replace radio frequency (RF) based wireless communications due to rapidly increasing demand for bandwidth and the limited regulated RF spectrum available to meet demand [27]. It should be possible to significantly expand RF based wireless capacity by moving to RF systems operating at 60 GHz, but communications at this radio frequency are line of sight (just as is VLC) and 60 GHz signals are more strongly absorbed by a wider range of common materials (similar to light). These limitations require special considerations for the design and implementation of 60 GHz wireless technology [28], many of which are similar to challenges faced by proponents of VLC systems. Lighting systems designed to supply high speed wireless data and illumination simultaneously could relax demand for more sophisticated high frequency wireless systems and provide wireless connectivity in RF restricted locations and provide a somewhat higher level of security than is possible with RF wireless services. The concept is shown in Fig. 14.2.

Dual use (illumination with communication) lighting systems being explored with current phosphor converted white LED systems typically use filtered detections systems to receive only the modulated blue LED emission because the radiative lifetimes of efficient phosphors are too long to permit high speed data transmission. Recently, relatively high bit rate data transmission with white LEDs and without the use of a blue filter using special modulation and digital filtering has been demonstrated [29]. When efficient color tunable lighting systems using RGBY LEDs become available, VLC applications can be partitioned by color (coarse wavelength

Fig. 14.2 Dual use
(illumination and
communications) will enable
new lighting services as well
as sensing technologies
enabling autonomous,
dynamic optimization
illumination patterns

division multiplexing). Using these sorts of multi-wavelength experimental systems, high speed data links using LED systems have been demonstrated [30]. Advantages of VLC over RF based wireless systems include security and unrestricted allocation of bandwidth, though many challenges need to be worked out on topics of mobility, integration with fixture designs (combinations of spot and diffuse lighting), and integration with RF uplink channels. Preliminary standards for VLC applications are under development.[6] Ultimately, integrated combinations of VLC and RF wireless communications will mostly likely be needed to meet the seemingly endless growth in demand for wireless data and mobile connectivity [31].

Solid state lighting systems that can illuminate an area and simultaneously transmit and receive data will enable other new and interesting functionality. For example, the use of modulated lighting for indoor localization (indoor GPS) is now being commercialized.[7] Drawing on concepts of time-of-flight sensing and the use of structured light for coarse 3D mapping of surfaces [32], it will be possible to design illumination systems equipped with embedded light detector networks to perform light-only based occupancy sensing and localization of occupants. When sources in an illumination system simply transmit identification codes and setting information (dimming, color point, etc.), networked receivers in the illumination system could monitor scattered light to characterize the rough geometry of the illuminated space and determine the when, where and type of illumination required based on the environment.[8] The ultimate goal of such systems is the development of autonomous, adaptive lighting control capability that will enable much more intuitive, self calibrating and self commissioning illumination systems. In some sense these types of

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[6]See the IEEE802.15.7 draft standard for visible light communications.

[7]A new company has developed indoor light-based communications services for a variety of uses, see www.bytelight.com.

[8]An IR camera technology highlights the concept, see [33].

Fig. 14.3 Moving Sky Panel using dynamically controlled RGB LEDs to simulate a skylight (courtesy of Shenzhen Neon my Technologies Co., Ltd.). Continued improvements in LED efficiency and packaging technology will enable the creation of energy efficient, high resolution video—Illumination dual function systems

applications are similar to earlier work where simple camera networks have been applied to issues of lighting control [34], but have seen very limited application, possibly for privacy concerns.

More broadly, equipping illumination systems with specialized sensors to enable low resolution light field mapping will create lighting systems which, for the first time, will "see" where the emitted light is going, coarsely image the surfaces that scatter or reflect it, and localize the illuminated objects in its field of view. Reflected light without a digital signature would suggest daylight and color tunable systems could adjust the spectral output for circadian matching (or not, depending on the illumination requirements of the occupants detected by the lighting system), assuming the digital light detection system can adequately resolve color and directionality. Lighting systems could also autonomously interact with other light emitting systems with their own digital light signature (e.g. projection or video systems), adapting the illumination as required. As lighting continues its migration to the electronic age, even more novel applications of communicative illumination systems will emerge.

14.2.3 Illumination with Display Capability

As mentioned earlier, LED technology has had a significant impact on display technology, enabling technologies like tablet computers and thin form factor LCD displays. Evolving development of gaming, immersive digital environments and new lighting technologies is already leading to the fusion of illumination and display technology. Like the integration of illumination and communications, this trend has been under development for some time, and includes the use of LCD display technology for virtual windows and skylights [35], and the use of color tunable LED lighting panel technology for illumination from an "open sky" source [36, 37]. An early design of this concept is shown in Fig. 14.3. The integration of display and illumination technology, if accomplished with inorganic LEDs in an emissive display format, will require significant improvements in the operating efficiency of green and yellow LED devices, and this has proven difficult for the LED die developers, though these wavelengths might start to receive more research attention as blue

LED performance gains saturate. Significant technical development of both display and LED technology will be needed to fully develop this fusion, but it is likely that future lighting systems will more closely approximate natural lighting (and can display the scenery that accompanies it). Display manufacturer Sony has recently described an experimental inorganic LED display format (no OLED, no LCD) [38], and as both lighting and display technologies are under pressure to be more energy efficient, it is expected that more illumination capable video systems will be developed. This trend will be analogous to trends observed in telephony, where what was once known as "plain old telephone service" (or POTS) transitioned to smart mobile communications systems with a fusion of communications, computation and display technology.

14.3 Summary

The rapid development of LED technology, largely driven by the need for efficient illumination, is a truly disruptive technology that will have wide technological, commercial and social applications beyond simple illumination. Increasingly, lighting suppliers will need to explore systems that deliver new services from lighting if only to maintain a viable business strategy as simple replacement bulb/socket technologies gradually become increasingly anachronistic (if not extinct) and the long service life of properly designed SSL systems drive new models for revenue generation. This is already occurring in the marketplace as lighting suppliers seek to leverage the capabilities of solid state lighting, offering dynamic lighting concepts tied to warm/cool color tuning capabilities simply to differentiate SSL offerings relative to highly efficient but spectrally static fluorescent lighting systems where the differential energy efficiency savings are small.

    As developers explore the new spectral, spatial and temporal properties afforded by advanced LED based illumination systems it is likely that a very wide range of advanced system features and services will be developed and some of those possibilities are described in this chapter. Industry is already testing some of these concepts by developing dynamic lighting systems that offer applications based illumination recipes, tunable lighting on or off the black body curve, indoor localization services with digital lighting, and even fixtures that can dynamically alter the fixture beam pattern electronically. Increasingly, the effective use of advanced solid state lighting systems will require advanced sensor capabilities and standardized control and communications protocols which will take some time to develop. The evolving Internet of Things (IoT) concepts already connecting consumer products to a growing network will enable lighting systems to seamlessly interface with building and grid systems while providing illumination of vastly improved quality relative to that available with incumbent incandescent and tube technologies. Ultimately, SSL will enable unforeseen systems capabilities and services much as the introduction of the transistor did over half a century ago.

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