Strained-Layer InGaAs(P)/InP Quantum Well Semiconductor Lasers Grown by Organometallic Vapour Phase Epitaxy
Cover figure: Schematic drawings of the shape of the unit cells in the active layer of an InGaAs(P)/InP quantum well semiconductor laser. In the middle figure, the InGaAs(P) quantum well (white) is grown lattice matched to its embedding layers, whereas in the outer figures, the InGaAs(P) quantum well compositions are varied to result in biaxial tensile (left) and biaxial compressive strain (right), respectively. The elastic lattice deformations result in modifications of the electronic band structure which facilitate the fabrication of improved performance devices. One of the parameters improved is e.g. the intensity of the optical powers emitted at a fixed drive current, as indicated schematically by the size of the output waves. The plane of the wave indicates that the polarization of the emitted light is affected by the strain as well.
Strained-Layer InGaAs(P)/InP Quantum Well Semiconductor Lasers Grown by Organometallic Vapour Phase Epitaxy

PROEFSCHRIFT

ter verkrijging van de graad van doctor aan de Technische Universiteit Delft, op gezag van de Rector Magnificus Prof. ir. K.F. Wakker, in het openbaar te verdedigen ten overstaan van een commissie, door het College van Dekanen aangewezen, op dinsdag 1 maart 1994 te 16.00 uur

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Petrus Johannes Adrianus THIJS

scheikundig ingenieur

geboren te Heeze.
Dit proefschrift is goedgekeurd door de promotoren:
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The work described in this thesis has been carried out in the Research Group of the Philips Optoelectronics Centre, Eindhoven, as part of the Philips Research programme.

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aan Leny, Hanneke en Maikel
aan mijn ouders
Preface

This thesis describes work carried out in the field of InP-based quantum well semiconductor lasers for optical fiber communication applications. Low-Pressure Organometallic Vapour Phase Epitaxy (LP-OMVPE), used as the crystal growth technique, runs through these studies like a continuous thread. The work has been carried out at Philips Research Laboratories since late 1986 in the research group "Technology of Optoelectronic Devices", headed by Ir. W. Nijman, and since 1991 in the research group of the Philips Optoelectronics Centre, headed by Prof. Dr. G.A. Acket.

Much work has been invested to gain insight in the effects of growth/switching in LP-OMVPE conditions on the materials/interfaces quality. A variety of electrical, optical, and structural characterization techniques have been used and combined with theoretical work. These studies have led to the first 1.5 μm wavelength buried heterostructure semiconductor lasers employing genuine quantum wells which were reported in 1987 (ECOC'87, held in Helsinki, Finland). However, further work has revealed that quantum well semiconductor lasers do not show as much improved performance as expected. This was not due to poor crystal quality, but it has been concluded that valence subband structure related intervalence band absorption and Auger recombination, which also limit the performance of bulk InGaAsP lasers, are not reduced significantly by applying quantum well active layers.

From theoretical work by Adams and Yablonovitch et al., it has become clear that a promising way for further reducing intrinsic loss mechanisms in long wavelength lasers is offered by the application of biaxially compressively strained quantum wells. Reductions of the in-plane hole effective mass, and of nonradiative recombination are predicted to boost the laser performance significantly. In 1989, we have reported a first, trendsetting, paper on the successful application of 1.8% compressively strained In$_{0.8}$Ga$_{0.2}$As/InGaAsP multiple quantum well (MQW) active layers in 1.5 μm wavelength lasers. Reliability studies have shown that these strained-layer MQW lasers are as reliable as lattice matched conventional bulk InGaAsP lasers. In 1990, we have fabricated and reported very high performance 1.5 μm wavelength tensile strained MQW lasers (SSDM’90, held in Sendai, Japan). The background of these improvements are reported in this thesis.

These papers acted as catalysts for work on strained-layer MQW lasers, not only for InGaAs(P) lasers emitting at 1.3 and 1.5 μm wavelength, but also in other group III-V materials systems. This is demonstrated by the number of reports on this topic at the 1992 IEEE Int. Semiconductor Laser Conference. About 40% of the papers dealt with strained-layer quantum well semiconductor
lasers, whereas at a similar conference two years earlier this was only about 5%. The success of strained-layer quantum well lasers including their excellent reliability is indicated by the fact that they are already commercially available, and are being applied in high end telecommunication systems.

Acknowledgement

The work described in this thesis would have been impossible without the contributions from many experts in various fields from the Philips Optoelectronics Centre, and from Philips Research Laboratories in Eindhoven and Paris. Without mentioning them all individually, I greatly acknowledge their experimental, theoretical, and technical contributions, and their mental support.

I would like to give my greatest thanks to Wim Nijman and Gerard Acket for the way they stimulated and supported me for many years. They urged me on writing this thesis.

I am indebted to Bart Verbeek and John Giling for putting confidence in me and their willingness to serve as my thesis advisors.

The board of directors of Philips Research Laboratories is acknowledged for giving me the opportunity to write this thesis.

The aid of the "Audio-Visuele Dienst" during the preparation of this thesis is greatly appreciated.

Prof. Dr. A.R. Adams, Dr. E.O. O’Reilly, and members of their research group at the Physics Department of the University of Surrey, England, are acknowledged for their (hydrostatic pressure) measurements leading to vivid discussions on the origins of the improved device performance. One of the many nice memories that remain is that we have been able to convince Prof. Adams, "the godfather of strain", of the benefits of tensile strain in semiconductor lasers.
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Chapter 1

Introduction

1.1. Lasers

A laser, an acronym for Light Amplification by Stimulated Emission of Radiation, is a device that can emit monochromatic and coherent electromagnetic radiation generated by stimulated emission. The operating principle is simple. Photons with frequency $\nu=(E_2-E_1)/h$ stimulate excited states $E_2$ to decay radiatively to the ground state $E_1$ such that the incident and emitted photons are of the same frequency and phase. The stimulated emission leading to amplification of the incident photons will only exceed the absorption associated with transitions from energy states $E_1$ to $E_2$, if the number of populated excited states $E_2$ is larger than that in the ground state $E_1$. This is called "population inversion" and must be accomplished by external pumping, either using light from another source or by an electrical current. In addition, to maintain a sufficient number of photons in the amplification medium required for the generation of the strong stimulated emission, an optical resonator is formed by using partially transparent mirrors perpendicular to the propagation direction of the electromagnetic radiation. All lasers, gas lasers, dye lasers, solid state lasers, and semiconductor lasers are based on the above principles. This work concentrates on electrically pumped semiconductor lasers. A brief historical review of the most important group III-V and II-VI compound semiconductor lasers is given in section 1.2. In section 1.3, we focus on the main topic of this thesis, i.e. semiconductor lasers fabricated using InGaAs(P) compounds grown on InP substrates for applications in optical fiber communication systems. Finally, in section 1.4 we outline the contents of this thesis.

1.2. Semiconductor Lasers

The radiative transitions in gas lasers, dye lasers, and solid state lasers occur between atomic or molecular energy levels, whereas in semiconductor lasers these transitions occur between electronic bands. The recognition that certain group III-V and II-VI compounds are direct semiconductors with efficient light emitting characteristics created a great technological challenge for the development of electrically pumped semiconductor lasers. The main features that
distinguish the semiconductor lasers from the other types of lasers are their:

1. small physical size (=0.3x0.3x0.1 mm³) which enables it to be incorporated easily into other instruments;
2. low driving voltage of typically several volts;
3. ability to modulate the optical output power by direct modulation of the electrical driving current;
4. possibility of integrating it monolithically with electronic and other optical components to form optoelectronic integrated circuits (OEIC’s);
5. semiconductor-based manufacturing technology, which lends itself to mass production, and
6. the possibility of tailoring the wavelength and output beam.

Various III-V and II-VI materials have been exploited for semiconductor lasers emitting in different wavelength ranges. The ability of III-V and II-VI compounds to form solid solutions opened the possibility to vary the composition in order to obtain the desired variations in bandgap and in refractive index, while maintaining lattice matching to the substrate. In figure 1.1, the solid dots represent the interrelationships between the bandgap $E_g$ and the lattice constant of binary III-V compounds containing Al, Ga, or In and P, As, or Sb, and II-VI materials containing Cd or Zn and S, Se or Te, respectively. Since the photon energy $E$ is approximately equal to the bandgap energy, the lasing wavelength $\lambda$ can be obtained using $E_g = hc/\lambda$, where $h$ is the Planck constant, and $c$ is the speed of light in vacuum. If $E_g$ is expressed in eV, the lasing wavelength in µm is given by $\lambda = 1.2389/E_g$.

Figure 1.1 shows for example that the bandgaps and lattice parameters of In$_x$Ga$_{1-x}$As$_y$P$_{1-y}$ (quaternary) compositions cover the area bound by connecting lines between the binaries InAs, InP, GaP and GaAs, and can be grown lattice matched to InP and GaAs substrates for certain combinations of $x$ and $y$. The interrelationships for the ternary Al$_x$Ga$_{1-x}$As follow the connecting line between the binary constituents GaAs and AlAs. Because of the fact that these binaries have almost identical lattice parameters, the ternary Al$_x$Ga$_{1-x}$As is lattice matched to GaAs for all $0 \leq x \leq 1$. There is a restriction on the bandgap range, however, because certain binaries are indirect semiconductors, where the minima in the conduction and the maxima in the valence bands are at different values of the wavevector $k$. The resulting indirect quaternary and ternary compositions from previous examples are bound (or shown) by the dashes lines.

III-V and II-VI compounds crystallize in the zinc blende structure which consists of two face cubic centered (fcc) crystal lattices, relatively shifted to each other over $\frac{1}{4}$ of the diagonal of the unit cell. For this crystal structure atomically
Figure 1.1. Bandgap of various III-V and II-VI semiconductors and mixed crystal systems versus the lattice parameter. The favourable wavelengths of 1.3 and 1.55 μm are indicated by arrows giving the compositions of the InGaAsP mixed crystals lattice matched (dotted line) to InP substrates.

flat crystal surfaces are obtained easily by cleaving the {110} planes which have the smallest cleavage energy. These cleavage planes are used as partially transparent (reflectivity about 32%) mirrors in the lasers. They are perpendicular to {001} planes which are by far the most frequently used substrate orientations for growth of the semiconductor laser structures.

The first semiconductor laser diodes studied were GaAs devices with diffused p-n junctions. They were reported by several groups in 1962 [1-4], only two years after lasing was achieved for the first time in a solid-state ruby laser [5]. The practical use of these semiconductor lasers was limited because the large heat dissipation due to the high threshold current density (~50-100 kA/cm²) prevented continuous wave (CW) operation at room-temperature. In these so-called homostructure semiconductor lasers (figure 1.2a), neither the carriers recombining at the p-n junction nor the generated photons are confined efficiently. After intensive research and using a conceptual breakthrough, Alferov et al. [6] and Hayashi et al. [7] achieved in 1970 the first CW operation at room temperature. This, and later breakthroughs, were mainly the result of
Figure 1.2. Schematic illustration of homostructure (a) and double heterostructure (b) semiconductor lasers with their typical dimensions. The dotted area represents the depletion region in the vicinity of the homojunction. The hatched area in (b) shows the thin (=0.1 to 0.2 μm) active layer of a semiconductor material whose bandgap is slightly lower, and its index of refraction is slightly higher than that of the surrounding cladding layers.

Improved crystal growth technologies for the fabrication of semiconductor laser structures (see chapter 3). Using liquid phase epitaxy (LPE) [8], low-defect-density structures consisting of stacked, thin layers (≥0.1 μm) of GaAs and Al_{x}Ga_{1-x}As compound semiconductors were fabricated. This permitted the utilization of the so-called double heterostructure (DH), (figure 1.2b)
consisting of a low bandgap and high refractive index active region, such as GaAs, sandwiched between n- and p-type doped materials with wider bandgap and lower refractive index, such as Al$_x$Ga$_{1-x}$As. The GaAs/Al$_x$Ga$_{1-x}$As material system provides these properties over the whole range of Al$_x$Ga$_{1-x}$As (0 ≤ x ≤ 1) compositions, while remaining lattice matched to GaAs. The simultaneous confinement of photons and carriers resulted in significantly improved device characteristics. By optimizing the material quality, device structure, and metallizations, the application of DH devices led to a reduction of the threshold current densities by two orders of magnitude to about 1 kA/cm$^2$ [9], as shown in figure 1.3.

![Graph showing the evolution of threshold current density of GaAs/AlGaAs lasers](image)

**Figure 1.3.** Evolution of the threshold current density of GaAs/AlGaAs lasers. The impacts of various improvements in the crystal growth technologies are clearly marked.

The next dramatic improvement is semiconductor technology came with the advent of the quantum well laser in 1975 [10, 11], made possible by
developments in growth technologies such as vapour phase epitaxy (VPE) and molecular beam epitaxy (MBE). These techniques, which will be discussed in chapter 3, facilitate reduction of the active layer thickness to values comparable to the De Broglie wavelength or the main free path of electrons in the layers (≈100 Å), while maintaining high material quality and low-defect density. As a result of the quantum confinement, these devices gave rise to a new field of semiconductor physics and resulted in improved performance with e.g. a significant reduction in the threshold current as shown in figure 1.3. Other effects such as the tailoring of the emission wavelength and higher levels for catastrophical optical degradation (COD) of the laser mirrors were achieved as well. It is found that the COD level is inversely proportional to the overlap of the optical mode with the active layer. This is called the optical confinement factor, Gamma. For quantum well lasers Gamma is small, in the order of one percent per quantum well, compared to 10-20% for conventional lasers.

In the mid 1980's, there was an interest in extending the emission wavelength beyond 880 nm, which corresponds to the bandgap of GaAs. This was achieved by adding Indium to GaAs quantum well active layers. These thin layers which despite their lattice mismatch to the substrate can be grown defect free, opened the new area of strained-layer quantum well devices. Initially there was some hesitation because the experience of the 1970's and 1980's taught that these devices were expected to suffer from a high degradation rate. Research results, however, indicated surprisingly improved reliability, and improved characteristics [12] compared to unstrained quantum well lasers as well. It was shown that the well known <100> dark line defects, which always occurred in degraded conventional (Al)GaAs/GaAs lasers were reduced in these strained-layer In_xGa_{1-x}As/GaAs lasers [13]. The most plausible explanation for this is the dislocation pinning as a result of the indium substitution [13].

The progress in GaAs semiconductor lasers, and challenging new applications, initiated the development of semiconductor lasers in various material systems emitting in different wavelength bands. Since the first demonstration of III-V semiconductor lasers in 1962, about every seven years a major breakthrough was achieved in this regard. In 1970 the first room temperature operation of a GaAs semiconductor laser was reported. The development of low loss optical fibers with dispersion and attenuation minima at 1.3 and 1.55 µm wavelength, respectively, triggered the research for semiconductor lasers emitting at these wavelengths. The first 1.3 µm wavelength laser, fabricated using the quaternary InGaAsP compound grown on InP substrate, was reported in 1977 [14]. The InGaAsP/InP materials and laser technology will be discussed in more detail in section 1.3.

The next breakthrough was achieved in 1985 with the first report of
room temperature CW operation of visible wavelength emitting semiconductor lasers fabricated using GaInP/AlGaInP structures grown on GaAs substrates [15, 16]. Figure 1.1 shows that this material system is bound by the binaries AIP, GaP, and InP. The bandgap energy can be tuned from 1.84 eV to about 1.96 eV, corresponding to emission wavelengths ranging from 675 to 630 nm. By using quantum well structures the emission wavelength could be reduced to about 600 nm. Organometallic vapour phase epitaxy (OMVPE) turned out to be a key technology in overcoming several significant difficulties associated with the growth of quaternary AlGaInP compound semiconductors.

The most recent breakthrough was achieved in 1991 with the demonstration of 77K pulsed operation of electrically injected II-VI compound blue light emitting semiconductor laser fabricated using CdZnSe quantum wells and ZnSSe cladding layers, both lattice mismatched to the GaAs substrate [17]. The clue for this breakthrough was the solution to problems associated with the p-type doping by using plasma source excited N-radicals in molecular beam epitaxy [18]. CW operation at 77K [19], and later pulsed operation to well above room temperature (394K) [20] were achieved by using lattice matched quaternary ZnMgSSe cladding layers which provide both enhanced electrical and optical confinement to the active layer.

In extrapolating this series of evolutions, around 1997-1998 a new class of materials may be due for applications in semiconductor lasers, possibly further extending the accessible wavelength range. These new materials might be III-V compound semiconductors, e.g. nitrides [21], or even plastics or polymers [22].

Over the past decade III-V semiconductor lasers opened up many significant opportunities such as optical fiber communications, compact disks and related optical data storage applications in particular, as shown in table 1.1. Initially considered as "unmarketable", in 1993 the worldwide semiconductor laser sales was about 40.10⁶ units, with revenues of about MS 250, representing about 25% of the total laser market [23]. A large majority of the semiconductor lasers fabricated (=80% in terms of number of units) is used in optical memories. Consequently, the unit price already dropped to around $ 1 to 2. A very important market for semiconductor lasers is the optical communication. Here, about 1% of the total number of units fabricated (that is about 400.000), represents about 50% (= M$ 120) of the total revenues in 1993 [23]. In next section the InGaAsP/InP materials and semiconductor lasers will be discussed in historical perspective.
Table 1.1. Review of the most important III-V semiconductor lasers.

<table>
<thead>
<tr>
<th>Substrate</th>
<th>Active/Cladding layer</th>
<th>Wavelength range (μm)</th>
<th>Applications</th>
</tr>
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<td>GaAs</td>
<td>(Al)GaAs/AlGaAs</td>
<td>0.70-0.88</td>
<td>Optical disk</td>
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<tr>
<td></td>
<td>(In)GaAs/AlGaAs</td>
<td>0.88-1.1</td>
<td>Free space communication</td>
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<td></td>
<td>(In)GaAs/InGaP</td>
<td>0.88-1.1</td>
<td>Pumping solid-state lasers</td>
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<td></td>
<td></td>
<td></td>
<td>Blue-light generation via frequency doubling</td>
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<tr>
<td>GaAs</td>
<td>(Al)GaInP/AlGaInP</td>
<td>0.60-0.67</td>
<td>Optical disk</td>
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<tr>
<td></td>
<td>InGaAsP/AlGaInP</td>
<td>0.65-0.88</td>
<td>Laser printer</td>
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<tr>
<td></td>
<td>InGaAsP/AlGaAs</td>
<td>0.65-0.88</td>
<td>Bar-code reader</td>
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<td>Displays</td>
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<td>Laser pointer</td>
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<td></td>
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<td></td>
<td>Substitution He-Ne laser (633 nm)</td>
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<td>InGaAsP/InP</td>
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<td>Optical fiber commun.</td>
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<tr>
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<td>(Al)GaInAs/AlGaInAs</td>
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<td>Photonic integr. circuit</td>
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<td>GaInAsSb/AlGaAsSb</td>
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</table>

1.3. InGaAsP/InP Semiconductor Laser Technology

Optical communication is the most important area responsible for the acceleration of a broad range of applications of optoelectronic devices. In the 1970’s, there was strong demand for long-distance, high capacity transmission systems. Optical fiber transmission systems were promising in order to satisfy the expansion of the channel capacity and for replacement of old equipment. Since widely available silica based optical fiber has an attenuation minimum at 1.55 μm wavelength (0.2 dB/km), and a zero-dispersion at 1.3 μm wavelength, the major development of semiconductor lasers for optical communication applications has been focused on the InGaAsP system grown on InP substrates. This material system has technological benefits over the alternative Al_xGa_yIn_{1-x-y}As /InP which contains aluminium. As a consequence, it is more difficult to handle
due to its higher reactivity towards oxygen, and the extremely large distribution coefficient of aluminium in LPE growth makes accurate composition control troublesome.

Many material parameters of In$_x$Ga$_{1-x}$As$_y$P$_{1-y}$ have been reported [24-27, and references therein]. The missing parameters for a specific quaternary composition Q(x,y) may be deduced from accurately known values of the binary compounds Q(AB) according

$$Q(x,y) = x y Q(\text{InAs}) + (1-x) y Q(\text{GaAs}) + x (1-y) Q(\text{InP}) + (1-x)(1-y) Q(\text{GaP})$$  \hspace{1cm} (1.1)

Using this relation, known as Vegard's law, it can be shown for example that for lattice matching to InP the As/Ga ratio in solid InGaAsP must be equal to 2.2. Under these conditions, the In$_x$Ga$_{1-x}$As$_y$P$_{1-y}$ room temperature bandgap $E_g$ can be continuously adjusted between 1.35 and 0.75 eV for y=0 and y=1, respectively [24]. Consequently, the entire wavelength region from 0.92 to 1.65 $\mu$m is covered with this material system. In addition, the ternary In$_{0.53}$Ga$_{0.47}$As ($E_g$=0.75 eV) can be used for the fabrication of photodetectors.

The work on InGaAsP/InP semiconductor lasers started in the early 1970's. Pulsed room-temperature operation of InGaAsP lasers was first reported in 1975 [28]. These lasers emitted at 1.1 $\mu$m wavelength and were grown by LPE. CW operation was achieved in 1976 by adopting a stripe geometry [29] with an active layer width of several micrometers reducing the threshold current and the heat dissipation accordingly. One year later the emission wavelength was extended to 1.3 $\mu$m [14]. It took until 1979 for the extension of the emission wavelength to the 1.55 $\mu$m wavelength band [30-32], the attenuation minimum of the optical fiber. Since then, much work has been performed on improving the characteristics of long wavelength semiconductor lasers. In the course of this work new advanced crystal growth technologies such as vapour phase epitaxy and molecular-beam-like techniques (chapter 3) were introduced for the fabrication of InGaAsP materials after solving the specific problems associated with the growth of indium and phosphorus containing materials.

However, despite all efforts, the common characteristic of the large variety of In$_x$Ga$_{1-x}$As$_y$P$_{1-y}$/InP semiconductor lasers fabricated using a range of different crystal growth technologies is that besides the change in emission wavelength, they all showed inferior characteristics compared to AlGaAs/GaAs lasers. In the so-called "long wavelength compounds", the nearly two times smaller bandgap compared to GaAs results in increased optical and carrier loss mechanisms. These loss mechanisms, specifically intervalence band absorption and Auger recombination (see chapter 2), combined with the smaller optical
confinement, due to the longer operating wavelength have stronger negative effects on the device characteristics than improvements expected over GaAs devices due to smaller effective masses in InGaAsP materials. Threshold current densities of good InGaAsP lasers are therefore generally higher by about a factor of two over those in GaAs. The best values reported are around 1 kA/cm$^2$ [33]. In addition, a stronger temperature dependence of the threshold current (by about a factor of 2), a smaller differential gain, and also a smaller and more temperature dependent external differential efficiency compared to the GaAs/AlGaAs lasers were observed. In one important point the InGaAsP/InP lasers excelled, however. This is their inherent slower degradation rate and the absence of facet degradation. This is of significant importance because for telecommunication lasers lifetimes in the order of 25 years are required.

Improvements in crystal growth technology facilitated the exploitation of quantum size effects also in the long wavelength materials system. The first genuine $\lambda=1.5$ µm buried heterostructure quantum well lasers were reported in 1987 by Thijs et al. [34]. With the application of quantum well active layers, the offset in performance between InP-based and GaAs-based semiconductor lasers even increased. This has its origin, as we will see, in the higher carrier concentrations at which quantum well lasers operate compared to the conventional bulk lasers. This significantly increases the intrinsic losses and gives an offset to some of the benefits of quantum wells. Materials studies have indicated that the quality of the quantum well structures were not the limiting factor.

Theoretical papers by Adams [35] and Yablonovitch et al. [36, 37] predicted that the performance of 1.5 µm quantum well lasers could be improved dramatically by growing the active material under compressive strain. As will be described amongst others in chapter 2, the strain-induced modifications of the valence band structure reduce the in-plane effective hole mass and accordingly the non-radiative loss mechanisms as well. The first successful demonstration of 1.5 µm wavelength compressively strained In$_x$Ga$_{1-x}$As/InGaAsP quantum well lasers was reported in 1989 by Thijs et al. [39]. Soon hereafter also improved performance 1.5 µm tensile strained quantum well lasers were reported. Initially these results were unexpected [35-37] but are now fully understood [40, 41], and are treated in this thesis.

Strained-layer lasers have attracted large interest in recent years. This topic was covered at the 1988 IEEE Semiconductor Laser Conference [42] only by strained In$_x$Ga$_{1-x}$As/GaAs devices emitting in the 1 µm wavelength region. At the 1990 semiconductor laser conference [43], the first papers reporting strained-layer quantum wells in the long wavelength InGaAsP material system appeared, and at the 1992 conference [44], a wide variety of strained-layer quantum well
lasers in all wavelength regions from blue to far infra-red covered 40% of the papers presented. The rapidly increased interest in strained-layer quantum well lasers in all materials systems is illustrated in figure 1.4. The impact of strained-layer quantum well active layers is so large that they are already used in the most advanced commercial telecommunication and visible lasers.

![Figure 1.4.](image)

**Figure 1.4.** Evolution of the number of publications on strained-layers quantum well lasers (literature scan Physical Abstracts, Nov. 1993). The papers are divided over: long wavelength, i.e. all lasers grown on InP substrates; 1 micron, i.e. InGaAs/GaAs; visible, i.e. InGaP/GaAs lasers; InAlGaAs; i.e. lasers in the 0.8 μm wavelength region, and blue, i.e. II-VI compound based lasers.
1.4. Outline of the Thesis

In this thesis, the low-pressure organometallic vapour phase epitaxial (LP-OMVPE) growth of InGaAsP/InP structures, the fabrication and the characterization of enhanced performance long wavelength ($\lambda = 1.5$ and $1.3$ $\mu$m) strained-layer InGaAs(P)/InGaAsP quantum well lasers are reported.

In chapter 2, basic parameters of semiconductor lasers are described, and the nonradiative loss mechanisms in conventional bulk InGaAsP lasers are reviewed. The main loss mechanisms, intervalence band absorption and Auger recombination, are introduced. Subsequently, the quantum well concept is introduced and the potentials for improvements of the semiconductor laser characteristics are treated. It is shown that a further improvement can be obtained by combining the quantum size effect with grown-in strain. The effects of this band structure engineering on the performance of long wavelength lasers is described.

The crystal growth technology has a large impact on the device performance. Therefore, in chapter 3, these techniques are reviewed with emphasis on phosphorus-containing materials. The LP-OMVPE is described in more detail as this technique is applied for the growth of the materials and devices studied in this work.

In chapter 4, the high quality of LP-OMVPE grown unstrained and strained-layer InGaAs(P)/InGaAsP quantum wells is demonstrated. The grown structures and interfaces are characterized by electrical transport measurements, by optical characterization using photoluminescence and photoluminescence excitation spectroscopy, and by structural analysis using transmission electron microscopy and high resolution X-ray diffractometry.

In chapter 5, the design, fabrication and characteristics of $1.5$ $\mu$m wavelength (strained-layer) In$_x$Ga$_{1-x}$As/InGaAsP quantum well lasers are reported. The effects of the magnitude and the sign of the strain on the device performance are studied, and the results are explained on the basis of both the reduced non-radiative recombinations and hole effective mass resulting from the strain-induced modifications of the band structure. Lifetests performed on $1480$ nm strained-layer In$_x$Ga$_{1-x}$As/InGaAsP multiple quantum well high power lasers confirm the high reliability of these lasers, even under severe operating conditions.

In chapter 6, the fabrication and characteristics of $1.3$ $\mu$m wavelength strained-layer InGaAsP/InGaAsP quantum lasers are reported. Using similar design rules as applied for the $1.5$ $\mu$m wavelength lasers, $1.3$ $\mu$m wavelength strained-layer quantum well lasers with improved performance are also demonstrated here.

The thesis is summarized in chapter 7.
1.5. References


Chapter 2

InGaAs(P)/InP Semiconductor Lasers: from Bulk to Strained-Layer Quantum Well

Abstract

Key-parameters for semiconductor lasers are presented. Analysis of the factors limiting the performance of long wavelength InGaAs(P) lasers employing bulk active layers shows that the most important loss-mechanisms are mainly associated with the structure of the valence subbands. These loss-mechanisms are intervalence band absorption and Auger recombination. Modifications to the valence subband structure by the application of strained-layer quantum wells, and the theoretically predicted improvements of the semiconductor laser parameters are reviewed.
2.1. Optical Gain and Static Device Parameters

A full theoretical description of the basic properties of semiconductor lasers can be found in text books [1-4]. Therefore, in this chapter we qualitatively relate the optical gain to the semiconductor band structure assuming bulk active layers with parabolic bands, and briefly summarize some important device parameters for semiconductor lasers.

Absorption and optical gain in semiconductors are caused by photon-induced transitions of electrons between the conduction band and the valence band. In quantifying the absorption and gain, we need to know the number of transitions that will occur per unit of time in the crystal in response to a given flux of photons in a given optical mode. The interactions of photons with electrons are described by the Einstein relations, while Fermi’s Golden Rule quantifies the transition probabilities [2]. Using these relations, the gain, \( G(E) \), or absorption, \( \alpha(E) \), where \( E \) is the photon energy, is expressed by [2, 4]

\[
G(E) = -\alpha(E) = \frac{e^2 \hbar}{2E c \varepsilon_0 m_0^2} \frac{n_g}{n^2} |M_T|^2 \rho_{\text{red}}(E)(f_c - f_v)
\]

where \( e \) is the electron charge, \( \hbar \) is Planck’s constant, \( n_g \) is the group index, \( n \) is the refractive index of the crystal, \( c \) is the speed of light in free space, \( \varepsilon_0 \) is the free space permittivity, \( m_0 \) is the electron mass, \( |M_T|^2 \) is the transition matrix element, \( \rho_{\text{red}}(E) \) is the reduced density of states, and \( f_c \) and \( f_v \) are the electron and hole Fermi functions. The assumption of conservation of momentum (\( \Delta k = k_{\text{initial}} - k_{\text{final}} - k_{\text{photon}} = 0 \)) in the radiative transitions between parabolic bands enables the identification of each conduction band state with a corresponding state in the valence band for a given photon energy. Consequently, the electron and hole density of states may be represented by a single reduced density of states function (per unit energy interval per unit volume) given by

\[
\rho_{\text{red}}(E) = 4\pi \left( \frac{2m_r}{\hbar^2} \right)^{3/2} \sqrt{E - E_g}
\]

where \( m_r = m_e m_h / (m_e + m_h) \) is the reduced effective mass with \( m_e \), \( m_h \) the electron, hole effective masses, respectively, and \( E_g \) is the bandgap energy. From eq.(2.1), we see that \( G(E) \) becomes positive for \( f_c > f_v \). Thus expressed in terms of energies, and using the concept of quasi energy levels \( E_{fc} \) and \( E_{fv} \), the requirement for optical gain at \( E = \hbar \nu \) is \( E_{fc} - E_{fv} > E > E_g \); in other words, the
Figure 2.1. Schematic illustration of the simultaneous confinement of charge carriers and the optical mode to the active layer of a double heterostructure semiconductor laser. The active layer (hatched), e.g., composed of In$_x$Ga$_{1-x}$As$_y$P$_{1-y}$, has a smaller bandgap and a higher refractive index than the InP cladding layers.

quasi-Fermi level separation in the conduction ($E_{\text{fc}}$) and valence ($E_{\text{fv}}$) bands must be wider than the bandgap energy ($E_g$) to achieve optical gain [5]. In order to obtain the above condition, we must create nonequilibrium conditions such that high electron and hole carrier densities in the conduction and valence bands can be simultaneously maintained. This can be achieved in the double heterostructure which is used in all today’s semiconductor lasers. This structure consists of a direct bandgap III-V material as active layer, embedded within materials with wider bandgaps and lower refractive indices, and n- and p-type doped on either side of the active layer, as shown in figure 2.1. For long wavelength semiconductor lasers, the active layer is composed of In$_x$Ga$_{1-x}$As$_y$P$_{1-y}$, a solid mixture of $x$yInAs+$x(1-x)$yGaAs$+(1-x)(1-y)$InP+$+(1-x)(1-y)$GaP binaries, and is lattice matched to the InP substrate and cladding layers. Under forward bias, or eventually by optical pumping, in the active layer a high density of electron-hole pairs can be achieved that, resulting from the index profile in the structure, efficiently overlaps with the optical field as shown in figure 2.1.

When due to carrier injection the quasi-Fermi level separation $E_{\text{fc}}-E_{\text{fv}}=E_g$, the active layer material becomes transparent for photon energies equal to the
Figure 2.2. Band edge transparency condition \((E_{fc} - E_{f'v} = E_g)\) illustrated for two idealized parabolic band structures of a bulk InGaAsP active layer. The carrier filling of each band is illustrated by the shaded overlap between the Fermi function and the density of states \((\rho_c(E) \text{ and } \rho_v(E) \text{ for electrons and holes, respectively})\). The "hole" Fermi function \(f'_v = 1 - f_v\) is used in this figure for clarity.

The carrier density required to provide this separation is known as the transparency carrier density, \(N_t\), and its magnitude is related to the electron and hole density of states \(\rho_c(E)\) and \(\rho_v(E)\) of the given material. To illustrate this, we can use the condition \(E_{fc} - E_{f'v} = E_g\) at transparency to draw a simple diagram shown in figure 2.2a. The shaded area, i.e. the overlap between the density of states- and the Fermi-functions, equals the total carrier density in the band. A bulk active layer with equal electron and hole masses is shown in figure 2.2a, whereas figure 2.2b shows a bulk active layer with heavier hole mass, respectively. Note that the asymmetry in the density of states (or effective masses) shifts both Fermi-functions towards the band with the lighter effective mass when maintaining equal numbers of carriers in both bands. This number is larger than for the symmetrical case, and consequently results in a larger \(N_t\). At injected carrier densities \(N\) exceeding \(N_t\) the gain becomes positive. For bulk lasers, the maximum optical gain given by eq. (2.1) may be simplified by the empirical relation.
\[ G = A \left( N - N_{D} \right) \]  

where \( A \) is the differential gain \( dG/dN \).

In the semiconductor laser, cleaved facets form a Fabry-Perot cavity and provide about 30% optical intensity reflection due to differences in the indices of refraction of the III-V semiconductor material (\( n = 3.2 \)) and air (\( n = 1 \)). The net round trip gain \( g_{R} \) of a Fabry-Perot resonator having a semiconductor active layer gain medium of length \( L \), and intensity reflection coefficients \( R_{1} \) and \( R_{2} \), is given by

\[ g_{R} = R_{1} R_{2} \exp \left[ 2L (g - \alpha_{t}) \right] \]  

where \( g \) is the modal gain and \( \alpha_{t} \) is the internal loss. The typical thickness of a semiconductor laser active layer is about 0.1-0.2 \( \mu \)m. Consequently, the optical field only partly overlaps the active layer as shown in figure 2.1d. The overlap-fraction is expressed by the optical confinement factor \( \Gamma \) and relates the mode gain \( g \) with the material gain \( G \) by

\[ g = \Gamma G \]  

Also, the internal loss, \( \alpha_{t} \), is distributed over the active layer and the surrounding cladding layers according to

\[ \alpha_{t} = \Gamma \alpha_{\text{act}} + (1 - \Gamma) \alpha_{\text{clad}} + \alpha_{s} \]  

where \( \alpha_{\text{act}} \) and \( \alpha_{\text{clad}} \) are the optical losses in the active and the cladding layers, respectively, and \( \alpha_{s} \) is the scattering loss at the interfaces between both layers. Optical loss in the active layer is caused by free carriers, and for long wavelength lasers, intervalence band absorption as we will show later on in this chapter. In the cladding layers there is mainly free carrier loss.

At the lasing threshold, the optical field is sustained after travelling one round-trip in the cavity, thus the net round-trip gain \( g_{R} \) is unity resulting in the threshold condition given by
\[ g_{th} = \alpha_i + \frac{1}{2L} \ln \frac{1}{R_1 R_2} = \alpha_i + \alpha_m \]  

(2.7)

where \( \alpha_m \) is called the mirror loss.

Combining eqs. (2.3), (2.5) and (2.7), gives for the carrier density at threshold

\[ N_{th} = N_{tr} + \frac{1}{\Gamma A} \left[ \alpha_i + \frac{1}{2L} \ln \frac{1}{R_1 R_2} \right] \]  

(2.8)

The current density is related to the carrier density by

\[ J = \frac{ed_{act} N}{\tau_s} \]  

(2.9)

where \( d_{act} \) is the thickness of the active layer, and \( \tau_s \) the carrier lifetime.

Combination of eqs. (2.8) and (2.9), and using \( J = \eta_i J_{th} \), where \( \eta_i \) is the internal efficiency with which electrons and holes recombine radiatively, yields for the threshold current density \( J_{th} \)

\[ J_{th} = \frac{J_{tr}}{\eta_i} + \frac{d_{act}}{\Gamma \beta \eta_i} \left( \alpha_i + \frac{1}{2L} \ln \frac{1}{R_1 R_2} \right) \]  

(2.10)

where \( J_{tr} \) is the current density to reach transparency, and \( \beta \) is the gain coefficient which is related to the differential gain \( \lambda \) by \( \beta = A \tau_s/e \). From eq. (2.10) we see that the threshold current consists of the transparency current density, \( J_{tr} \), and a contribution to overcome the internal and mirror losses. For minimum threshold current density it is therefore essential to minimize the internal and mirror losses, to achieve maximum gain at minimum carrier density, i.e. to obtain a large differential gain \( dG/dN \) for a structure with a given \( d_{act} \) and optical confinement factor \( \Gamma \), to minimize the transparency current density, and to maximize \( \eta_i \). In the active layer, the radiative recombinations are in competition with the non-radiative recombinations. Both processes can be characterized by their average lifetimes \( \tau_r \) and \( \tau_m \) which determine the internal efficiency according \( \eta_i = \tau_m / (\tau_r + \tau_m) \). The internal efficiency reaches 1 for \( \tau_m \gg \tau_r \), i.e. the radiative processes must be much more probable than the non-radiative. Looking at eq. (2.1), we find that, aside from some material constants, the rate at which we can change \([\rho_{red} f_e + \rho_{red} (1-f_e) - \rho_{red}]\) will determine how quickly the gain responds to changes in \( E_{ie} - E_{iv} \). From figure 2.2 we see that an increase in
$E_{F_{c}}-E_{F_{v}}$ has a much more pronounced effect on the band edge carrier density when $p(E)$ has a steep profile, i.e. when the effective mass is small. Additionally, as indicated in figure 2.2a, it would also be beneficial to have a symmetrical band structure, or $m_{c}=m_{v}$. Bulk group III-V semiconductors do not fulfill the requirements for minimum transparency current density and maximum differential gain. Later on in this chapter we will see that a best approach to the ideal band structure is obtained by using quantum well layers grown under biaxial strain.

With increasing injection current above threshold, the round-trip gain remains clamped to 1, and the excess carriers are consumed to build up the laser oscillation intensity. The optical power generated by stimulated emission is given by $P_{o}=[(I-I_{th})\eta_{o}h\nu]/e$, where $I$ and $I_{th}$ are the driving and the threshold currents, respectively, and $h\nu$ is the photon energy. Part of this power is dissipated inside the laser cavity, and the remaining is coupled out through the facets. These two powers are proportional to the effective internal and mirror loss. The output power $P$ is thus given by

$$P = \eta_{o}h\nu \frac{(I-I_{th})}{e} \frac{\alpha_{m}}{\alpha_{i}+\alpha_{m}}$$  \hspace{1cm} (2.11)

The external differential efficiency is the ratio of the variation in photon output rate that results from an increase in the injection current and is given by

$$\eta_{d} = \frac{d(P/h\nu)}{d[(I-I_{th})/e]} = \eta_{i} \left[ 1 + \frac{2\alpha_{i}L}{\ln \frac{1}{R_{1}R_{2}}} \right]^{-1}$$  \hspace{1cm} (2.12)

Figure 2.3 shows the typical optical output power per facet (mW) versus the drive current (mA) of a semiconductor laser. The output power increases sharply at the threshold current where eq. (2.12) corrected by $\frac{1}{2}h\nu/e$ gives the slope in units of mW/mA. For lasers with two cleaved facets the reflection coefficients $R_{1}$ and $R_{2}$ are identical. As a result, the semiconductor laser emits the same powers at the front and rear facets. This ratio can be changed by modification of the facet reflectivities by applying dielectric coatings. For asymmetric reflectivities the ratio of efficiencies is given by
Figure 2.3. Optical output power versus injection current of a narrow stripe InGaAsP/InP laser under continuous wave (CW) operation. The output power abruptly increases at the threshold current $I_{th}$; the slope of the curve gives the external differential efficiency $\eta_d$. The inset shows the multimode lasing wavelength spectrum (at 1.5 mW output power) typical for a Fabry-Perot laser.

\[
\frac{\eta_1}{\eta_2} = \frac{(1-R_1)}{(1-R_2)} \sqrt{\frac{R_2}{R_1}} \tag{2.13}
\]

The phase condition of a Fabry-Perot resonator is satisfied at an integer set of halve wavelengths (longitudinal modes) fitting into the cavity

\[
\frac{1}{2} m \lambda_c = L n_{\text{eff}} \tag{2.14}
\]

where $n_{\text{eff}}$ is the effective index of refraction and $m$ is the mode number (an integer). Assuming a negligible dispersion of $n_{\text{eff}}$, the wavelength difference between neighbouring Fabry-Perot modes, $\Delta \lambda_c$, is given by
\[ \Delta \lambda = \lambda_m - \lambda_{m-1} = \frac{\lambda_m^2}{2n_{\text{eff}}L} \]  

(2.15)

One mode will experience the largest gain and will therefore be the first to reach a round-trip gain of one at threshold. As the round-trip gain is clamped above threshold, other longitudinal modes are expected to remain below threshold. Because of the effects of the spontaneous emission and nonlinear gain mechanisms, the side modes will also reach threshold. This leads to multimode oscillation as depicted in the inset of figure 2.3, which is typical for Fabry-Perot lasers. For single frequency oscillation, a frequency selective grating element must be adapted inside the laser cavity. In a distributed feedback (DFB) laser, the grating region is built in a waveguide layer adjacent to the active layer, whereas in a distributed Bragg reflector (DBR) laser, the grating region is outside the active layer along the length of the cavity.

Variations in the active layer temperature modify the electron and hole Fermi-functions resulting in a decreased gain with increasing temperature. Consequently, the threshold current density increases. Apart from this effect, additional loss mechanisms may gather strength as will be shown later on in this chapter. The variation of the threshold current density with temperature is expressed by the empirical relation

\[ J_{\text{th},T_2} = J_{\text{th},T_1} \exp\left(\frac{T_2 - T_1}{T_0}\right) \]  

(2.16)

where \( J_{\text{th},T_1} \) and \( J_{\text{th},T_2} \) are the threshold currents at temperatures \( T_1 \) and \( T_2 \) respectively, and \( T_0 \) is the characteristic temperature expressed in K. A small value of \( T_0 \) represents a high temperature sensitivity of the threshold current density.

2.2. Dynamic Properties

2.2.1. Intensity Modulation

A unique feature of semiconductor lasers is that, unlike other lasers, the output power can be modulated directly by modulating the drive current. Rate equations describe the time dependence of the carrier (N) and photon densities (S). Assuming a laser emitting at only one wavelength, these can be written as
\[
\frac{dN(t)}{dt} = \frac{I}{eV_{\text{act}}} - \frac{N}{\tau_s} - v_g \Gamma G(N,S)S \\
\frac{dS(t)}{dt} = v_g \Gamma G(N,S)S - \alpha v_g S + \Gamma \beta BN^2
\]

(2.17)

where \( I \) is the drive current, \( e \) is the electron charge, \( V_{\text{act}} \) is the volume of the active layer, \( \tau_s \) is the carrier lifetime at threshold, \( G(N, S) \) is the active medium gain due to stimulated band-to-band transitions, \( \Gamma \) is the optical confinement factor, \( v_g = c/n_{\text{eff}} \) is the group velocity, \( \alpha = \alpha_c + \alpha_{\text{nl}} \) is the total cavity loss, \( \beta \) is the fraction of the spontaneous emission coupled into the optical mode, and \( B \) is the radiative recombination coefficient. Because the lasers are operated well above threshold, the gain \( G(N, S) \) contains the nonlinear gain: \( G(N, S) = G(N)/(1 + \epsilon_{\text{nl}} S) = G(N)(1 - \epsilon_{\text{nl}} S) \), where \( \epsilon_{\text{nl}} \) is the gain saturation factor describing the reduction of the gain with increasing photon density. The time dependence of the number of carriers is the accumulated effect of the injected carrier density minus the carriers lost in spontaneous and in stimulated recombination, respectively, as expressed by the upper equation of (2.17). The time dependence of the number of photons is the net result of the photons generated by the stimulated emission, the photons lost at the facets, and the contribution of the spontaneous emission to the laser mode, respectively.

When a small sinusoidal modulation current at an angular frequency \( \omega \) is superimposed to a dc current above the threshold, the frequency response \( R \) of the modulated light varies as \( [6, 7] \)

\[
R = \frac{dP(\omega)}{dI(\omega)} = R_i \times R_e
\]

\[
R_i \propto \frac{\omega^4}{(\omega^2 - \omega_r^2)^2 + \omega^2 \gamma^2}
\]

\[
R_e \propto \frac{1}{1 + (\omega/\omega_{\text{pe}})^2} \times \frac{1}{1 + (\omega/\omega_{\text{pc}})^2}
\]

(2.18)

where \( R_i \) is the intrinsic-, and \( R_e \) is the electrical frequency response due to RC rolloff and p-n junction capacitance. By squeezing the contact- and active layer dimensions, the device- and p-n junction capacitance can be reduced to flatten the electrical response contribution to large frequencies. \( R \) then equals \( R_i \), the intrinsic frequency response, determined by the rate equations. In the expression for \( R_e \), \( \omega/2\pi = f_r \) is the relaxation oscillation frequency, and \( \gamma \) is the damping rate obtained by fitting response data to the theoretical curve given by eq. (2.18).
This response curve is flat at small frequencies, shows a peak at the relaxation oscillation frequency $f_r$ due to a resonance between electron- and photon populations, and then steeply drops due to the damping. The relaxation oscillation frequency $f_r$ is given by [7]

$$f_r = \frac{1}{2\pi} \sqrt{\frac{v_g^2 \alpha \Gamma}{\frac{dG}{dN}} S}$$  \hspace{1cm} (2.19)

Using the relation between the internal photon density $S$ and the total output power $P$ emitted from both facets, and the relation between $P$ and the drive current above threshold, the resonance frequency can be rewritten as

$$f_r = \frac{1}{2\pi} \sqrt{\frac{v_g \alpha \Gamma}{h\nu V_{act} \alpha_m} \frac{dG}{dN} P} = \frac{1}{2\pi} \sqrt{\frac{v_g \Gamma \eta_i}{e V_{act}} \frac{dG}{dN} (1 - I_{th})}$$  \hspace{1cm} (2.20)

The relaxation oscillation frequency is proportional to the square root of the output power or the bias current above threshold, and furthermore, for maximum $f_r$, the confinement factor, and $dG/dN$ should be maximized, whereas $V_{act}$ should be minimized.

The damping factor $\gamma$ is given by [7]

$$\gamma = \frac{\Gamma B N^2}{S} + \frac{1}{\tau_s} + \frac{1}{v_g \Gamma} \frac{dG}{dN} S + \frac{\Gamma \alpha \varepsilon_{nl}}{S}$$  \hspace{1cm} (2.21)

At a few milliwatt output power, the spontaneous emission is negligible compared to the contributions of the other terms, resulting in the relation between the resonance frequency and the damping factor given by

$$\gamma = K f_r^2 + \frac{1}{\tau_s}$$  \hspace{1cm} (2.22)

with the $K$ factor expressed as

$$K = 4\pi^2 \frac{\varepsilon_{nl}}{v_g} \left( \frac{\varepsilon_{nl}}{dG/dN} + \frac{1}{\alpha} \right)$$  \hspace{1cm} (2.23)

Upon increasing the drive current, and thus the output power, the modulation
response finally becomes critically damped. The maximum -3dB modulation bandwidth is obtained by setting the intrinsic response function R_i equal to 0.5. This yields

\[ f_{-3dB} = \frac{2\pi \sqrt{2}}{K} \]  

(2.24)

### 2.2.2. Chirp and Linewidth

The current modulation of the laser not only produces an intensity modulation, as shown in previous section, but also modulation of the emission frequency, which is known as "chirp" of the laser. These are inherently combined and result from the fact that adding more carriers (current modulation) produces more gain (intensity modulation) and this produces a phase shift (frequency modulation). The relevant parameter here is Henry's alpha, \( \alpha_H \), or the linewidth enhancement factor, which is defined as the ratio of the change in effective refractive index (n_{eff}) with the carrier density (N) to the variation in optical gain (g) with the carrier density according to [8]

\[ \alpha_H = \frac{-4\pi}{\lambda} \frac{dn_{eff}/dN}{dg/dN} \]  

(2.25)

The linkage between an intensity modulation P(t) and the frequency modulation \( \Delta v_{ch} \) can be expressed mathematically by [9]

\[ \Delta v_{ch} = \frac{-\alpha_H}{4\pi} \left[ \frac{d}{dt} \ln P(t) + \kappa P(t) \right] \]  

(2.26)

with

\[ \kappa = \frac{2 \Gamma \varepsilon_{nl}}{V_{act} \eta_d h\nu} \]  

(2.27)

where \( \Gamma \) is the optical confinement factor, \( \varepsilon_{nl} \) is the nonlinear gain coefficient, \( V_{act} \) is the volume of the active layer, and \( \eta_d \) is the total differential efficiency. The chirp is a major concern in high bit rate intensity modulated transmission systems because it results in different arrival times at the detector for the different wavelength components composing the signal due to the chromatic
dispersion of the optical fiber. Similarly, the laser linewidth $\Delta v$ is an important parameter. Incoherent spontaneous emission coupled into the lasing mode affects the linewidth. The important parameter, also here, is the linewidth enhancement factor as shown in the expression of the laser linewidth given by [8]

$$
\Delta v = \frac{v_g^2 \hbar \nu g \alpha n_{sp} (1 + \alpha^2)}{8 \pi P}
$$

(2.28)

where $v_g$ is the group velocity of the light, $\hbar \nu$ is the photon energy, $g$ is the gain, $\alpha$ is the cavity loss, $n_{sp}$ is the spontaneous emission factor, and $P$ is the output power.

The expressions for the high speed and the spectral characteristics of semiconductor lasers clearly reveal the importance of the differential gain in these parameters. A reduction of the linewidth enhancement factor leads to an increase of the optical telecommunication capacity due to a reduction of the chirp and linewidth.

2.3. Loss Mechanisms in Bulk InGaAs(P) Semiconductor Lasers

AlGaAs/GaAs semiconductor lasers have been extensively studied and their characteristics are well documented [1-3]. The properties of InGaAsP/InP lasers may be predicted from the AlGaAs/GaAs results by making corrections for the differences in materials. However, the common characteristic of the large variety of bulk In$_x$Ga$_{1-x}$As$_y$P$_{1-y}$/InP double heterostructure semiconductor lasers emitting in the wavelength range from 1.1 to 1.65 $\mu$m, and fabricated by a whole range of growth techniques is, that despite the smaller carrier effective masses, they all suffer from a higher threshold current, a smaller differential gain, a stronger temperature dependence of the threshold current (i.e. smaller $T_0$-value), and also a smaller and more temperature dependent quantum efficiency compared to the AlGaAs/GaAs lasers [4]. Several mechanisms have been proposed to explain these effects which seem to be independent of the growth technique and device structure, suggesting the importance of intrinsic loss mechanisms within the laser which may be related to the InGaAsP band structure. As main loss mechanisms have been considered:

(i) carrier leakage across the heterobarrier into the confining layers due to the smaller carrier confinement;

(ii) non-radiative recombination at surface-states and at defects. The latter are more likely because lattice matching and composition control in InGaAsP/InP are much more difficult than in the AlGaAs/GaAs system;

(iii) intervalence band absorption causing optical losses, and
Auger recombination. These loss mechanisms will be discussed in more detail in next sections.

2.3.1. Carrier Leakage to the Confinement Layers

It was proposed that heterobarrier carrier leakage caused by diffusion and drift of thermally excited carriers into the cladding layers was significant in increasing the threshold current and in reducing the $T_0$-value of long wavelength lasers above room temperature [10]. For InGaAs(P)/InP heterojunctions, the discontinuity at the conduction bands, $\Delta E_c$, is smaller than at the valence bands ($\Delta E_c:\Delta E_v = 0.35:0.65$; chapter 4). Consequently, in various studies [11-13] it was shown that mainly electron leakage exists in InGaAs(P)/InP lasers. However, Ng et al. [14] reported that the electron leakage could be minimized by doping the p-type confinement layer higher than $3 \times 10^{17}$ cm$^{-3}$. This was confirmed in several later studies [15-17]. Furthermore, 1.5 $\mu$m wavelength InGaAsP/InP lasers generally show smaller $T_0$-values than lasers emitting at 1.3 $\mu$m wavelength despite the smaller electron confinement in the latter structures, which is inconsistent with the idea of heterobarrier carrier leakage. Additionally, Thijs et al. [18] and Zah et al. [19] recently demonstrated 1.5 $\mu$m wavelength tensile strained multiple quantum well lasers that are characterized by a very asymmetric band offset ($\Delta E_c:\Delta E_v = 0.05:0.95$; chapter 5) due to the strain-induced band structure modification as will be shown in section 2.4.2. The very small conduction band offset was expected to result in a very high temperature sensitivity of these lasers. Surprisingly, excellent high temperature characteristics, i.e. a record high operating temperature of 140$^\circ$C [18] with very regular $T_0$-values were measured [18, 19]. This was explained by an effect which has been considered only very recently; i.e. the injection-induced electrostatic confinement of carriers as proposed by Barrau et al. [20]. Due to the strong confinement of one type of carriers, the holes in this case, the Coulomb attraction localizes the other type of carriers (the electrons) in the active layer, and helps to reduce heterobarrier carrier leakage.

Leakage of carriers, mainly electrons, may also arise due to hot carriers in the active layer as a consequence of Auger recombination, in which the excess energy of the electron-hole recombination is used to excite carriers high into the bands, as will be shown in section 2.3.4. Monte Carlo simulations [21] indicate that leakage of these hot carriers is more important than of carriers excited thermally over the barrier. In order to artificially increase the effective barrier height for reduced carrier leakage, Iga et al. [22] have proposed the multiple quantum barrier, based on resonant reflection of electrons moving perpendicularly to a periodic potential energy structure. For long wavelength
lasers no experimental verification is reported yet. On the other hand, for visible wavelength GaInP/AlGaInP/GaAs lasers, improved $T_0$-values were claimed [23, 24] by the application of this multiple quantum barrier structure. The major difference with the long wavelength lasers is, however, that the AlGaInP material is known to be free of Auger recombination, so here the carrier leakage mainly results from thermal electrons due to the small difference in bandgaps between the cladding- and active layers, especially at short emission wavelengths below 630 nm.

2.3.2. Recombination at Defects and Surfaces

Defects at hetero-interfaces and in the active layer which may have been grown-in during epitaxy or occur during laser operation, cause non-radiative recombination via the continuum of states formed within the carrier diffusion length of the defects. The rate of non-radiative recombination is given by

$$R_d = A_{nr} N$$  \hspace{1cm} (2.29)

with $N$ the carrier density, and $A_{nr} = \sigma v N_i$ with $\sigma$ the capture cross section of the trap, $v$ the velocity of electrons or holes, and $N_i$ the trap density. The temperature dependence of this effect is too small to explain the observed variations in threshold current and differential efficiency with temperature [25].

States formed at the surface, e.g., at the laser facets, may also induce non-radiative recombination. The surface recombination rate is given by

$$R_s = S A N_{th}$$  \hspace{1cm} (2.30)

where $S$ is the surface recombination velocity, $A$ is the area, and $N_{th}$ is the threshold carrier density. Nijman et al. [26] observed no difference in the temperature dependence of the photoluminescence of InGaAsP single- and double heterostructures, and concluded that the $T_0$-problem was related to intrinsic InGaAsP properties rather than due to surface recombination.

2.3.3. Intervalance Band Absorption

Adams et al. [27, 28] proposed that intervalence band absorption could account for the increase in threshold current and decrease in differential quantum efficiency with increasing temperature. Intervalance band absorption involves the re-absorption of the laser radiation where the photon energy is used to excite electrons deep in the valence subbands away from $k=0$ into higher energy states.
in the valence band [27-31]. Assuming k-selection, three possible vertical transitions between different valence subbands are illustrated in figure 2.4. The lh-hh transition A, and also the so-lh transition B occur at relative large values of the wavevector \( \mathbf{k} \), and may be neglected on the basis of the small hole occupancies. The large density of states and the high hole occupancy in the heavy hole band near \( \mathbf{k}=0 \) produce large absorption at relevant wavelengths for the so-hh transition C in figure 2.4. For increasing emission wavelengths from 1 to 1.65 \( \mu \)m, the variations of the valence subband structure, e.g., the reduced bandgap, the increased spin split-off bandgap, \( \Delta_{so} \), and the valence subband dispersion determined by the effective masses, allow the intervalence band absorption to occur at smaller values of the wavevector \( \mathbf{k} \), where the hole occupancy is larger [32]. Consequently, the intervalence band absorption increases with increasing emission wavelength. Estimates for the intervalence band absorption coefficient, \( \alpha_{IVBA} \), differ considerably. Casey et al. [30] determined at 1.5 \( \mu \)m wavelength a linear dependence upon the hole concentration \( P \) in the active layer according

\[
\alpha_{IVBA} = 2 \times 10^{-17} P \quad (cm^{-1}) \tag{2.31}
\]

where \( P \) is expressed in \([cm^3]\). Recently, Hauser et al. [33] reported an intervalence band absorption of 120 \( cm^{-1} \) at a hole carrier density of 1.2x10^{18} cm^{-3} (i.e. \( \alpha_{IVBA}=1 \times 10^{-16} P \)) deduced from the temperature dependence of the threshold
carrier density and differential quantum efficiency of 1.65 μm lasers.

2.3.4. Auger Recombination

Since the pioneering work of Beattie and Landsberg [34], it is generally accepted that Auger recombination is an important non-radiative loss mechanism in direct, narrow bandgap semiconductors. The Auger recombination process involves three carriers on four states (mainly three electron and one hole, and one electron and three hole states). In this process, the excess energy of the electron-hole recombination is transferred to another carrier (electron or hole), which gets excited to a higher energy state in the band [34-39]. The excited (hot) carrier relaxes back to achieve thermal equilibrium by losing its energy to lattice vibrations, or is swept across the heterobarrier and contributes to the heterobarrier carrier leakage as discussed in section 2.3.1. Labelling the type of carriers in the respective bands C, H, L, and S, for conduction band, heavy hole, light hole and spin split-off valence bands, different band-to-band Auger processes can be distinguished which are represented in figure 2.5. In the CHCC mechanism, electron 1 recombines with hole 1', i.e. electron 1 makes

![Diagram](image)

**Figure 2.5.** Schematic illustration of three different band-to-band Auger processes. The excess energy of the electron-hole recombination, represented by the downward arrow, is transferred to another carrier (electron or hole), which gets excited to a higher energy state in the band (upward arrow).
a transition to the empty state $1'$, and the excess energy is transferred to electron 2, which is excited to state $2'$. The basis for the Auger effect is the Coulomb interaction between the electrons 1 and 2 in the conduction band. The CHSH process involves one electron and three hole states, two in the heavy hole and one in the spin split-off valence band. The CHLH process is similar to the CHSH process except that light holes instead of spin split-off holes are involved (figure 2.5).

The generally adopted formulation of the three-particle band-to-band Auger recombination rate is

$$R_{AR} = C_{AR} N^2 P \quad \text{or} \quad R_{AR} = C_{AR} N P^2 \quad (2.32)$$

depending whether the process is dominated by electrons (N) or holes (P), where $C_{AR}$ is the Auger coefficient. The band-to-band Auger processes are characterized by a strong temperature and bandgap dependence, i.e. the Auger rate decreases rapidly with decreasing temperature and with increasing bandgap. These dependencies arise from the laws of energy- and momentum conservation that the four particle states involved ($1$, $2$, $1'$, and $2'$ in figure 2.5) must satisfy. This may be seen in the following way for the CHCC process. The momentum and energy conservation laws give rise to a threshold energy $E_T$ for each process [35, 37]. If we assume $E_1 = E_2 = 0$, only holes with energies greater than

$$E_T - E_h = \Delta E_{CHCC} \quad (2.33)$$

can participate ($\Delta E_{CHCC}$ is a constant that depends on effective masses and bandgap energy). The number of such holes varies approximately as $\exp(-\Delta E_{CHCC}/kT)$ for nondegenerate semiconductors. Taking this exponential temperature dependence into the Auger coefficient this is expressed as

$$C_{AR} = C_{AR,0} \exp\left(\frac{-\Delta E_{CHCC}}{kT}\right) \quad (2.34)$$

where $C_{AR,0}$ is the temperature independent part of the Auger coefficient, and $\Delta E_{CHCC}$ is the activation energy. Due to the pinning of the carrier density at threshold, the Auger processes mainly affect the spontaneous emission and the threshold current but not the differential efficiency. Theoretical and experimental values at room temperature for the Auger coefficient $C_{AR,0}$ for 1.5 μm wavelength InGaAsP alloy cover a range from $10^{-29}$ to $10^{-28}$ cm$^6$s$^{-1}$ [28]. As a comparison, the Auger coefficient in GaAs is approximately $10^{-31}$ cm$^6$s$^{-1}$ [28].
For the CHCC process, the threshold energy $E_T$ is given by [35]

$$E_T = \frac{(2m_c + m_{hh})}{2m_c + m_{hh} - m_c(E_T)}E_g$$  \hspace{1cm} (2.35)$$

where $m_c$ and $m_{hh}$ are the conduction band and heavy hole band effective masses and $m_c(E_T)$ is the effective mass of the excited electron. (For true parabolic bands the carrier effective mass is independent of its energy). Combining eqs. (2.33) and (2.35) results in

$$\Delta E_{(CHCC)} = \frac{m_c(E_T)}{2m_c + m_{hh} - m_c(E_T)}E_g$$  \hspace{1cm} (2.36)$$

Substituting typical values for GaAs ($E_g = 1.42$ eV, $m_c = 0.067m_0$, $m_{hh} = 0.45m_0$) [4] yields $C_{AR} = 6.10^{-35}$ cm$^6$s$^{-1}$ at room temperature, whereas for In$_{0.53}$Ga$_{0.47}$As ($E_g = 0.75$ eV, $m_c = 0.041m_0$, $m_{hh} = 0.50m_0$) $C_{AR}$ becomes as large as $5.10^{-30}$ to $5.10^{-31}$ cm$^6$s$^{-1}$.

For the CHSH process $\Delta E_{(CHSH)} = E_T (E_g - \Delta_{so})$, with

$$E_T = \frac{2m_{hh} + m_c}{2m_{hh} + m_c - m_s(E_T)}(E_g - \Delta_{so})$$  \hspace{1cm} (2.37)$$

yields

$$\Delta E_{(CHSH)} = \frac{m_s(E_T)}{2m_{hh} + m_c - m_s(E_T)}(E_g - \Delta_{so})$$  \hspace{1cm} (2.38)$$

The activation energy for the CHLH process is similar to eq. (2.38), where $m_s(E_T)$ is substituted by the effective mass of the excited light hole $m_{gh}(E_T)$ [35]. The above equations show that the bandgap energy, $E_g$, is an explicit variable in the Auger recombination rate. Therefore, a meaningful comparison of the Auger recombination intensity can only be made for devices emitting at a same wavelength. Figure 2.5 clearly indicates that the band structure has a considerable influence on the probabilities of the Auger processes because the energy and the momentum must be simultaneously conserved. Based on the assumption of parabolic bands, in n-type bulk long wavelength InGaAsP materials the CHCC process dominates, whereas both the CHSH and CHLH processes are considered for p-type materials.

A second class of Auger processes where a deviation from $\Delta k = 0$ is compensated by the emission of absorption of a phonon are the phonon-assisted
Auger processes [37]. The phonon interactions P in the CHCCP, CHSHP and CHLHP processes take place in the heavy hole band, because heavy holes interact more strongly with phonons compared to electrons and light holes [4]. The lifting of the requirement of conservation of momentum for the four particle states involved, eliminates the threshold energy $E_T$, and also the strong bandgap dependence for the phonon-assisted Auger processes. However, the relative intensities of the various processes depend on the exact band structure. Haug [39] proposed that due to non-parabolic bands in 1.3 $\mu$m wavelength In$_{0.72}$Ga$_{0.28}$As$_{0.6}$P$_{0.4}$ the CHCC process should be negligible, whereas the CHSH band-to-band and phonon-assisted CHSHP processes should be more important and of comparable magnitude.

2.4. Improved Long Wavelength InGaAsP Lasers: from Unstrained to Strained-Layer Quantum Wells

2.4.1. Unstrained Quantum Well Lasers

Bulk III-V semiconductors are characterized by a conduction band with small effective mass, $m_e$, and at the $\Gamma$ point a degenerate valence band with large effective mass. This large asymmetry in the band structure is unfavourable for achieving optimum performance semiconductor lasers. In addition, due to this band structure, bulk InGaAsP suffers from loss mechanisms which limit the performance even more, especially at longer emission wavelengths as shown in previous section.

A logical continuation in the efforts striving for improved performance of long wavelength semiconductor lasers is the trend to shrink structures to smaller dimensions, similar to work performed in the AlGaAs/GaAs material system [40-42]. This has become possible also in the InGaAsP/InP material system by the application of advanced epitaxial growth techniques, as will be shown in chapter 4. For layer thicknesses in the order of the De Broglie wavelength (\(\lambda = h/p\approx 200-300\ \text{Å}, \ p=\text{momentum}\)) of confined carriers, their kinetic energy for motion normal to the direction of the interfaces (z-direction) becomes quantized into discrete energy levels. As depicted in figure 2.6, for every quantum number, one conduction band and two valence band levels, corresponding with the heavy holes (hh) and light holes (lh), exist. Considering these discrete states along the z direction and continuous states along the x and y directions, the confined energy levels assuming an infinitely deep quantum well are given by
$$E_{n,k_x,k_y} = E_n + \frac{\hbar^2}{8\pi^2 m_{i,\perp}} (k_x^2 + k_y^2)$$

(2.39)

where \(E_n=\hbar^2 n^2/(8m_{i,\perp} L_z^2)\) is the \(n\)-th confined carrier energy level, \(m_{i,\perp}\), and \(m_{i,\parallel}\) are the effective masses for this level in the direction perpendicular and parallel to the QW interfaces, respectively, and \(L_z\) is the quantum well width. Eq. (2.39) shows that the confinement energy \(E_n\) is determined by the effective mass of the carrier in the direction perpendicular to the quantum well. Due to the quantization of the kinetic energy of confined carriers, the density of states is modified from the well known three dimensional case

$$\rho_i^{3D}(E) = 4\pi \left(\frac{2m_i}{\hbar^2}\right)^{3/2} \sqrt{E}$$

(2.40)

to [4]

$$\rho_i^{2D} = \frac{4\pi m_{i,\perp}}{\hbar^2} \frac{1}{L_z}$$

(2.41)

A comparison of eqs. (2.40) and (2.41) shows that the density of states in a quantum well is independent of the carrier energy, and is determined by the effective mass in the quantum well plane (x-y plane). Thus, this effective mass is the relevant one in determining the characteristics of quantum well lasers. For unstrained quantum well lasers \(m_{i,\perp}\) and \(m_{i,\parallel}\) are identical. Figure 2.6 shows on the right-hand side the density of states for bulk, represented by the continuous dashed curve, and the density of states for quantum wells, represented by the solid step-like curves. The modification of the band structure by the quantum size effect has several attractive features for improving the performance of semiconductor lasers. In the now following derivation of the potentials of QW devices, the idealized band structure from figure 2.6 will be used ignoring broadening of the discrete energy levels in the \(z\)-direction due to carrier-carrier [43] and phonon scattering processes. Furthermore, Auger recombination and intervalence band absorption will be neglected as well.

As shown in figure 2.6, the effective bandgap of quantum wells \(E_g=E_g^{\text{bulk}}+E_n^{\text{c}}+E_n^{\text{a}}\) is a function of the well thickness \(L_z\), i.e. by using the same composition in the QW, the emission wavelength can be adjusted by changing the well width. Furthermore, due to the modified band structure the gain spectrum and the gain-current relations are modified as shown in figure 2.7.
Figure 2.6. Energy scheme (left) and density of states (right) for a quantum well of e.g., InGaAsP embedded within InP. Due to the confinement, for motion normal to the interfaces the kinetic energy of carriers is quantized into discrete energy levels. The density of states for unquantized carriers are shown by the dashed curves, the quantization changes them to the solid step-like curves.

For bulk active layers, the spectral gain bandwidth according eq. (2.1) increases with carrier density and above transparency the peak gain is linearly proportional to the injection current according eq. (2.3). Quantum well lasers show a steep onset of the gain at the bandgap energy, and plotted against the injection current, the gain becomes positive for smaller values of the injection current and shows a tendency to saturate at larger injection currents which is expressed by [44]

$$ G = J_n \beta \ln \frac{J}{J_n} \quad (2.42) $$

where $\beta$ is the differential gain, and $J_n$ and $J$ are the transparency and the injection current density. Quantum well active layers are attractive for semiconductor lasers because of the reduction of the transparency current density and the enhanced differential gain near transparency.

Assuming an idealized single quantum well with only one electron and hole carrier level, simple relations for the carrier density and the differential gain can be deduced as a result of the simple density of states function. Similar to shown in figure 2.2, the electron density in the quantum well is given by
$N = \rho_c kT \ln \left[ 1 + \exp \left( \frac{E_{fc} - E_c}{kT} \right) \right]$ \hfill (2.43)

where $\rho_c$ is the density of states given by eq. (2.41), $E_{fc}$ is the quasi Fermi-level, and $E_c$ is the lowest electron energy in the quantum well, respectively. For equal electron and hole effective masses $m_e = m_h = m$, at transparency the quasi-Fermi levels are located at the band edges. Thus, the transparency carrier area density per quantum well, the low bound of the threshold current density, is given by

$N^{3}_{tr} = \frac{4\pi m kT}{h^2} \ln \left( 2 \right)$ \hfill (2.44)

Assuming an effective mass of 0.041$m_0$, i.e. for a lattice matched In$_{0.53}$Ga$_{0.47}$As active layer [4], $N^{3}_{tr}$ becomes about $3.1 \times 10^{11}$ cm$^2$ at 300K. The transparency current density becomes as low as 50 A/cm$^2$ assuming a carrier lifetime $\tau_c$ of 1 ns. For asymmetric conduction and valence bands, the transparency carrier area density per quantum well can be approximately given by [45]
\[ N_{tr}^a = \left( \frac{m_e}{m_e + m_v} \sqrt{\frac{2m_v}{m_e + m_v}} \right) \frac{N_{tr}^s}{2} \]  \hspace{1cm} (2.45)

Assuming the same electron effective mass \( m_e = 0.041m_0 \), together with a hole effective mass of \( 0.70m_0 \) for an 80\AA wide \( \text{In}_{0.53}\text{Ga}_{0.47}\text{As} \) quantum well (section 2.4.2.2), the transparency current density increases by about a factor of three to about 150 A/cm².

Due to the step-like density of states, the gain per quantum well from the lowest energy subband transition is finite and given by [45]

\[ G_{\text{max}} = \frac{8\pi^3\hbar^3 \left| M_T \right|^2}{\varepsilon \nu h^3} m_e m_v \frac{1}{m_e + m_v L_z} \]  \hspace{1cm} (2.46)

For quantum wells embedded within a separate confinement heterostructure (section 5.4.2), the optical confinement factor \( \Gamma \) is proportional to the well thickness \( L_z \). Thus, the modal gain \( g = \Gamma G \) is independent of the well thickness, or expressed alternatively, a reduction in the optical confinement factor \( \Gamma \) is compensated for by an increase in the density of states (eq. 2.41). The maximum differential gain at transparency is given by

\[ \frac{dG}{dN} = \frac{G_{\text{max}}}{N_{tr}^s} = \frac{2\pi^2}{\ln (2)} \frac{\nu \left| M_T \right|^2 \sqrt{m_e m_v}}{\varepsilon \nu kT} \]  \hspace{1cm} (2.47)

The differential gain increases for reduced asymmetry between the electron and hole effective masses and becomes optimal for equal effective masses. The quantum confinement modifies the transition matrix element \( \left| M_T \right|^2 \) as well. Setting the reference value for the electron-hole recombinations in isotropic bulk layers to 1, the relative value of \( \left| M_T \right|^2 \) for e-hh recombinations in quantum wells becomes 1.5 [46]. Due to the increased differential gain, in quantum well lasers the relaxation oscillation frequency, and thus the modulation bandwidth will be enhanced, whereas the linewidth enhancement factor, and thus the linewidth and the chirp will be reduced. Furthermore, in bulk lasers the threshold current is proportional to \( T^{3/2} \), whereas in quantum well lasers \( J_{th} \propto T \) as indicated in eq. (2.44). This reduces the Auger current, and thus also the temperature sensitivity of the threshold current (or increases the \( T_0 \)-value) for quantum well
lasers. Finally, the overlap between the optical mode with the quantum wells is reduced which leads to larger output powers for the onset of catastrophic optical degradation (COD) of the facets.

For AlGaAs/GaAs quantum well lasers, large improvements over their bulk counterparts were obtained, as indicated by the realization of 1-mA-threshold current quantum well lasers [47]. However, for 1.5μm wavelength InGaAs(P)/InP quantum well lasers, the advantages obtained in comparison with conventional bulk InGaAsP semiconductor lasers were marginal, as will be reported in this thesis in chapter 5. This was attributed to optical and carrier loss mechanisms [48], which prevent laser operation at low carrier densities. As shown, significant improvements are expected by reduction of the valence band effective mass to values comparable to the electron effective mass, which is possible by growing the quantum wells under biaxial strain. In addition, for biaxial compressive strain, intervalence band absorption and Auger recombination were predicted to be reduced further enhancing the performance of the semiconductor lasers [49-51]. However, there was some reservedness in applying strained-layer quantum wells due to the feared negative effects on the reliability of these semiconductor lasers. In next sections, the strain-induced modifications of the band structure and the effects on the semiconductor laser parameters will be discussed.

2.4.2. Effects of Strain on Quantum Well Lasers

2.4.2.1. Critical Thickness Limitation

Defect-free heterostructures can be grown from lattice-mismatched materials up to thicknesses where the resulting strain can be accommodated by elastic tetragonal deformation of the unit cell, i.e. the critical thickness. Above this thickness, the elastic energy is released by the formation of misfit dislocations parallel to the layer interfaces acting as centers for non-radiative recombination and as sources for multiplication of defects. This will deteriorate the performance and the reliability of the devices. The critical thickness depends on the materials (elastic constants), on the degree of lattice mismatch, and on the substrate orientation. Studies of the critical thickness for the onset of misfit dislocations in a number of material systems such as InxGa1-xAs/GaAs [52, 53], InxGa1-xAs/InP [54], InxAl1-xAs/InP [55], GaAsxP1-x/GaAs [56], and Ge0Si1-x/Si [57] have turned out to be quite difficult and even controversial. Part of this controversy arises due to the different characterization techniques used to study the coherent growth in strained-layer structures. When growing strained-layer structures, the general trend is that the luminescence properties are more
Figure 2.8. Critical thickness according the Matthews and Blakeslee model of a single In$_{x}$Ga$_{1-x}$As epitaxial layer, buried within InP and grown on (001) oriented substrate, against the InAs mole fraction in the ternary layer.

sensitive than the electrical properties, and finally the morphology becomes affected. Therefore, although several theories have been proposed to predict the critical thickness, most designers of optoelectronic devices rely upon the model of Matthews and Blakeslee [56], which gives the most conservative estimate. Matthews and Blakeslee estimated the critical thickness, $d_c$, by equating the force exerted by misfit strain and the tension in a dislocation line. The result for a single buried layer was a transcendental equation

$$d_c = \frac{b}{4\pi \varepsilon_i} \frac{1-\nu \cos^2 \alpha}{(1+\nu) \cos \lambda} \frac{(\ln d - 1)}{b}$$  \hspace{1cm} (2.48)

where $b$ is the magnitude of the Burger’s vector of the dislocation, $\nu$ is the Poisson ratio, $\varepsilon_i$ is the in-plane strain given by eq. (2.49), $\alpha$ is the angle between the dislocation line and its Burger’s vector, and $\lambda$ is the angle between the slip direction and that direction in the plane of the epitaxial layer which is orthogonal to the line of the slip plane and the interface. All quantities except $b$ and $d_c$ in eq. (2.48) are dimensionless. For (001) oriented III-V compound semiconductors it has been shown that misfit dislocations are primarily of the $60^\circ$-type [58]. This implies $\alpha=\lambda=60^\circ$, $b=a_{\text{epilayer}}/\sqrt{2}$, and with $\nu=C_{12}/(C_{11}+C_{12})=0.33$ [59], where $C_{ij}$ are components of the elastic stiffness.
tensor, the critical thickness can be deduced. For In$_x$Ga$_{1-x}$As grown on InP, the materials of interest for 1.5 μm wavelength quantum well lasers the critical thickness of a single buried layer is shown in figure 2.8 against the InAs mole fraction. For a superlattice of strained-layer quantum wells embedded within barrier layers, the critical thickness is twice as large than in figure 2.8 since the strain is equally shared by adjacent layers. On the other hand, for a single strained-layer on a substrate the critical thickness is twice as small as shown in figure 2.8.

2.4.2.2. Strain Effects on the Band Structure

Lattice-mismatched epitaxial layers grown coherently on a thick substrate are subjected to a biaxial in-plane strain. For (001) substrate orientations, the epitaxial lattice experiences a tetragonal distortion, resulting in a simple form of the strain tensor $\varepsilon_{ij}$. Its only non-vanishing components are

$$\varepsilon_{1} = \varepsilon_{xx} = \varepsilon_{yy} = \frac{a_x - a_o}{a_o}$$  \hspace{1cm} (2.49)

and

$$\varepsilon_{\perp} = \varepsilon_{zz} = -2 \frac{C_{12}}{C_{11}} \varepsilon_{1}$$  \hspace{1cm} (2.50)

where $z$ is chosen along the growth direction and $x$, $y$ are in the growth plane, $a_o$ and $a_x$ are the lattice parameters of the substrate and of the relaxed epitaxial layer, respectively, and $C_{ij}$ are the components of the elastic stiffness tensor. For $a_x > a_o$, i.e. $\varepsilon_1 < 0$, the epitaxial layer is under biaxial compressive strain, which will be simply referred to as compressive strain in this thesis, whereas for $a_x < a_o$ we have biaxial tensile strain, further on referred to as tensile strain. The biaxial strain breaks the cubic symmetry of the lattice as shown in figure 2.9 for In$_x$Ga$_{1-x}$As embedded within InP, and has consequences mainly for the valence subband structure. For compressive strain, i.e. $x > 0.53$, the lattice is compressed in the x-y plane, and consequently the lattice parameter is elongated in the z-direction. For tensile strain, $x < 0.53$, the reverse is occurring, i.e. the lattice is stretched in the x-y plane, and consequently the lattice parameter is reduced in the z-direction. According Vegard’s law, for In$_x$Ga$_{1-x}$As grown on InP, the strain in the direction parallel to the interface is given by
\[ \varepsilon_1 = 6.9(0.53-x) \% \] (2.51)

In In$_x$Ga$_{1-x}$As/InP the in-plane strain can be systematically varied between 3.2% compression for \( x=1 \), i.e. InAs/InP, and 3.7% tension for \( x=0 \), i.e. GaAs/InP. The total strain can be resolved into a purely hydrostatic component

\[ \varepsilon_{hy} = \frac{\Delta V}{V} = \varepsilon_{xx} + \varepsilon_{yy} + \varepsilon_{zz} = 2\varepsilon_1 + \varepsilon_{\perp} \] (2.52)

and a purely axial component

\[ \varepsilon_{ax} = \varepsilon_{zz} - \varepsilon_{xx} = \varepsilon_{\perp} - \varepsilon_1 \] (2.53)

The hydrostatic component affects the bandgap, \( E_g \), by

\[ \delta E_g = a \left( 2\varepsilon_1 + \varepsilon_{\perp} \right) \] (2.54)

where \( a \) is the hydrostatic deformation potential. The shifts of the conduction- and valence band edges with hydrostatic pressure are expressed similarly to eq. (2.54), where the prefactor becomes the conduction band or valence band hydrostatic deformation potential, respectively. The difference in hydrostatic deformation potentials for the conduction- and valence bands directly affects the band offsets by the application of strain. As a result, the conduction band discontinuity is increased for compressive and decreased for tensile strain as was verified experimentally Cavicchi et al. [60]. At large tensile strains this leads to a negative conduction band offsets \( \Delta E_c \). In these so-called type II quantum wells the electrons are confined to the barrier/separate confinement heterostructure layers, whereas the hole carriers remain confined to the wells.

The axial strain affects the valence subband positions according to

\[ \delta E_v = 2b \left( \varepsilon_{\perp} - \varepsilon_1 \right) \] (2.55)

where \( b \) is the tetragonal shear deformation potential. Analogous to the work of Pollak and Cardona [61] for uniaxial stress, the Hamiltonian matrix incorporating the biaxial strain can be expressed as
\[
\begin{pmatrix}
0 & 0 & 0 & 0 \\
0 & -E_g - \delta E_g - \frac{1}{2} \delta E_v & 0 & 0 \\
0 & 0 & -E_g - \delta E_g + \frac{1}{2} \delta E_v & \frac{\delta E_v}{\sqrt{2}} \\
0 & 0 & \frac{\delta E_v}{\sqrt{2}} & -E_g - \delta E_g - \Delta_{so}
\end{pmatrix}
\]

(2.56)

The Hamiltonian operates on the basis states

\[
u_c = |s \uparrow\rangle
\]
\[
u_{hh} = \frac{1}{\sqrt{2}} |(x + iy) \uparrow\rangle
\]
\[
u_{lh} = \frac{1}{\sqrt{6}} |(x + iy) \downarrow\rangle - \frac{2}{3} |z \uparrow\rangle
\]
\[
u_{so} = \frac{1}{\sqrt{3}} |(x + iy) \downarrow\rangle + \frac{1}{\sqrt{3}} |z \uparrow\rangle
\]

(2.57)

where \(u_c\) and \(u_{ij}\) (ij=hh, lh, and so) are the conduction and valence band Bloch functions, respectively, the designations s and x, y, and z refer to the conduction- and valence band wavefunctions, transforming like atomic s- and p-functions under the operations of the tetrahedral group [62], and \(\uparrow\) and \(\downarrow\) correspond to spin up and spin down. Choosing E=0 at the bottom of the conduction band, from the eigenvalues of the strain Hamiltonian, the energies of the conduction and valence bands at \(k=0\) are given by [63, 64]
$E_c = 0$

$E_{hh} = -E_g - \delta E_g - \frac{1}{2} \delta E_v$

$E_{jh} = -E_g - \delta E_g + \frac{1}{4} \delta E_v - \frac{1}{2} \Delta_{so} + \frac{1}{2} \sqrt{\Delta_{so}^2 + \delta E_v \Delta_{so} + \frac{9}{4} \delta E_v^2}$

$E_{so} = -E_g - \delta E_g + \frac{1}{4} \delta E_v + \frac{1}{2} \Delta_{so} - \frac{1}{2} \sqrt{\Delta_{so}^2 + \delta E_v \Delta_{so} + \frac{9}{4} \delta E_v^2}$

**Figure 2.9.** Modification of the shape of the unit cell for In$_x$Ga$_{1-x}$As with $x > 0.53$ (left), and $x < 0.53$ (right) grown on InP. The bottom part shows the valence subband structure for the unstrained In$_x$Ga$_{1-x}$As by the dashed curves, and for the strained ternary by the curves with the symbols representing the respective subbands.
It is seen that the off-diagonal elements of eq. (2.56) produce the mixing between the lh and so valence subbands. For zero strain, the above equations describe the well known bulk band structure at the Γ point. For unstrained \( \text{In}_x\text{Ga}_{1-x}\text{As} /\text{InP} \) the dashed lines in the bottom part of figure 2.9 show the valence subband structure as a function of \( x \). The effects of the biaxial strain on the band structure at the Γ point of bulk \( \text{In}_x\text{Ga}_{1-x}\text{As} \) grown epitaxially on InP, calculated using eqs. (2.58), are shown in figure 2.9 as well. The material parameters required can be estimated by linear interpolation between the values for GaAs and InAs. For compressive strain \( (x>0.53) \), the hh levels form the top of the valence band, whereas for tensile strain the lh levels are on top. Furthermore, for compressive strain the bandgap is increased compared to unstrained material of the same composition, whereas it is decreased for tensile strain, and for both signs of the strain the energy splitting between the spin split-off band and the valence band maximum is increased.

Pikus and Bir [66] have shown for the first time for Si and Ge, which have a crystal lattice similar to the III-V compounds, that due to the tetragonal lattice distortion the lh-hh valence subband mixing is modified yielding strongly anisotropic valence band dispersions for the two signs of the strain. Figure 2.10 shows the resulting dispersions of the band structure for bulk layers where bandgap variations due to composition variations are ignored. In a strain-free structure, the band structure is isotropic in the three directions. For compressive strain, the in-plane hh effective mass is significantly reduced, whereas a heavy mass is maintained in the z-direction. In the case of tensile strain, the in-plane lh effective mass remains small, although it is increased from its zero-strain value. This has important implications for the performance of semiconductor lasers as will be shown.

From section 2.4.2.1, it is clear that the thickness for defect-free growth of strained-layer structures is limited to dimensions for which the quantum size effect has to be considered. Figure 2.11 shows the emission wavelength versus the thickness of the \( \text{In}_x\text{Ga}_{1-x}\text{As} \) quantum wells embedded within \( \text{InGaAsP} \) \( (\lambda_g=1.25 \ \mu\text{m} \text{ for } x>0.53, \text{ and } \lambda_g=1.15 \ \mu\text{m} \text{ for } x<0.53) \) using a four-band Kane-type \( \text{k.p} \) Hamiltonian [67]. By using the compressively strained quantum wells, the emission wavelength can be tuned to over 2 \( \mu\text{m} \). The increase in emission wavelength is due to the strong decrease of \( E_g \) with increasing InAs mole fraction (figure 2.9). This enables the application of strained-layer \( \text{In}_x\text{Ga}_{1-x}\text{As/InP} \) for emission wavelengths which were only accessible by the less mature antimony-system of interest for applications in chemical sensing, light detection and ranging (LIDAR), and medical treatment. For telecommunication lasers emitting at 1.5 \( \mu\text{m} \), we consider \( \text{In}_x\text{Ga}_{1-x}\text{As/InGaAsP} \) quantum wells with
Figure 2.10. Schematic representation of the band dispersions of In$_x$Ga$_{1-x}$As under compression, no strain, and tension on InP substrate, (ignoring composition effects). In the unstrained situation (middle figure), the heavy hole (hh) and light hole (lh) bands are degenerate at the zone centre and spin split-off (so) band lies $\Lambda_{so}$ lower in energy. The bandgap is indicated by $E_g$. Under compressive strain (left figure), the hydrostatic component of the strain increases the bandgap, while the axial component splits the degeneracy of the valence band maximum, where the hh levels become on top. The hh effective mass is large in the growth direction (z), and is reduced in the in-plane direction (x, y). Under biaxial tensile strain (right), the bandgap is reduced by the hydrostatic pressure component, and the axial component brings the lh levels on top of the valence band. The in-plane lh effective mass remains small, although it is increased from its zero-strain value.

Simultaneous variations in the InAs mole fraction $x$ and the well thickness as shown in figure 2.11. The quantum well states were evaluated using an eight-band Kane-type $\mathbf{k.p}$ Hamiltonian [68]. It takes full account of the coupling between the lowest electron, heavy hole, light hole, and spin split-off bands. The coupling in all other bands is included perturbatively. Figure 2.12 shows three examples of the in-plane energy dispersions for the valence subbands in the quantum wells grown under 1.6% tensile strain (left), lattice matched (middle),
Figure 2.11. Emission wavelength of In$_x$Ga$_{1-x}$As/InGaAsP QWs calculated as a function of QW thickness. For $x \geq 0.53$, the InGaAsP cladding layers were assumed to have $\lambda_s = 1.25 \, \mu m$ composition, whereas for $x \leq 0.53$ $\lambda_s = 1.15 \, \mu m$ InGaAsP was assumed. The solid lines represent c-hh transitions, dashed lines c-lh transitions.

and 1.2% compressive strain (right). InAs mole fractions and wells widths of $x = 0.32$ and $L_z = 160$ Å, $x = 0.53$ and $L_z = 80$ Å, and $x = 0.70$ and $L_z = 25$ Å, respectively, were chosen to maintain the emission wavelength fixed at 1.5 μm. For the unstrained In$_{0.53}$Ga$_{0.47}$As quantum well the heavy hole band forms the top of the valence band, and has a large effective mass of 0.7$m_0$ due to the mixing with the light hole states away from k=0. For compressive strain the application of thin wells enhances the strain-induced valence subband splitting. As depicted in figure 2.12, the hh$_1$-lh$_1$ subband splitting is increased largely. This eliminates the bandmixing, resulting in a small effective hole mass of 0.15$m_0$. For tensile strain, the quantum confinement counteracts the strain-induced modification of the valence band. As a result, a (near-degenerate) valence band maximum with large effective hole mass is obtained for quantum well widths used typically (approximately 80 Å or less). The remedy for obtaining also here a light hole effective mass is to reduce the quantum confinement effect largely by increasing the well width. Figure 2.12 (left) shows the valence subband structure of 1.6% tensile strained In$_{0.32}$Ga$_{0.68}$As with a thickness of 160 Å. The result is an uppermost valence subband with light hole character and a small effective mass of 0.20$m_0$. The band structures deduced here are used as input for the prediction of the performance of $\lambda = 1.5 \, \mu m$ wavelength strained-layer quantum well lasers which will be reported in next section.
Figure 2.12. Calculated band structure of In$_x$Ga$_{1-x}$As/InGaAsP for 1.6% tensile strained (x=0.32), lattice matched (LM, x=0.53), and 1.2% compressive strain (x=0.70) in left, middle and right figure, respectively. The well width $L_z$ is chosen for emission at 1.5 μm wavelength. The solid and dashed lines denote the hh and lh subbands, respectively. The wavevector is taken along [210] which is intermediate between [100] and [110]; the [001] is assumed to be perpendicular to the plane of the well.

2.4.2.3. Characteristics of Strained-Layer In$_x$Ga$_{1-x}$As/InGaAsP Quantum Well Lasers; Theoretical Predictions

The consequences of the strain-induced band structure modifications shown in figure 2.12 were for compressive strain in theory first recognized as being advantageous for semiconductor lasers by Adams [49] and Yablonovitch et al. [50, 51]. Experimental results demonstrating beneficial effects as well for relatively wide tensile strained quantum wells were reported later for the first time by Thijs et al. [69, 70], and were readily understood [20, 68, 71].

From the Bloch functions (eq. (2.57)), the enhancement of the transition matrix element and the polarization of the laser emission in (strained-layer) quantum wells can be explained. In bulk lasers with a degenerate valence band maximum, equal probabilities for recombinations of s-like conduction band
states with p-like states in the valence band for x, y and z directions exist. Assuming a coordinate geometry within the laser with the z-direction perpendicular to the layer-interfaces, and the x direction perpendicular to the facets, the s-p \_i (i=x, y, z) overlaps produce electric fields in the respective directions. The electric field in the x-direction has no reflection and escapes from the laser as spontaneous emission. In the y-directions the optical confinement factor is larger than in the z-direction. Consequently, the s-p \_y recombinations will first reach threshold and the laser emission is polarized in the y-direction, which is called transverse electric (TE). In unstrained quantum well lasers, due the differences in effective mass, the hh-lh degeneracy is lifted (figure 2.12, middle figure), with the hh levels becoming on top. In the limit situation there is no overlap between s-p \_i; this implies that every one out of two carriers can contribute to the TE gain, compared to one out of three in bulk. This corresponds to an enhancement of the transition matrix element to 1.5, setting the reference for isotropic bulk layers at 1. In practice, however, in unstrained quantum wells the hh-lh splitting is relative small, leading to hh-lh bandmixing. This introduces s-p \_z recombinations reducing the transition matrix element. The limit situation is approximated better in compressively strained quantum wells where the hh-lh splitting is much larger (figure 2.12, right figure). In tensile strained quantum wells, the lh levels form the valence band maximum. In this case, the s-p \_z contribution of electron-light hole recombinations with the electric field vector of the optical mode in the z-direction is such dominant that despite the smaller optical confinement factor it reaches threshold first. Now the magnetic field is parallel to the y-direction and this polarization is called transverse magnetic (TM). The s-p \_y recombinations produce TE gain. The transition matrix element for electron-light hole recombinations is 2 for TM gain and 0.5 for TE gain, respectively.

The gain and spontaneous emission spectra of 1.5 μm wavelength strained-layer In\_xGa\_1-xAs/InGaAsP SQW lasers were calculated by Krijn et al. [68] as a function as a function of carrier density and strain. The temperature was assumed 77K which also justifies the ignorance of Auger recombinations and interband absorption. Intraband relaxation processes were accounted for by a lifetime broadening of the spectra with a Lorentzian corresponding with a relaxation time of 10\(^{-13}\) s. At threshold, \(\Gamma G_{th} = \alpha_{loss}\) gives the relation between the material gain, \(G_{th}\), and the sum of the internal and the mirror losses, \(\alpha_{loss}\). Results for the threshold carrier density as a function of strain are depicted in figure 2.13a. Both compressive strain (x>0.53) and tensile strain (x<0.53) result in considerable reduction of \(N_{th}\), especially at larger \(\alpha_{loss}\). The larger well widths in the case of tensile strain ensure a low \(N_{th}\) even for large \(\alpha_{loss}\). Considering a gradual increase in tensile strain, initially the threshold carrier density increases,
Figure 2.13. Calculated threshold carrier density (a), and threshold current density (b) at 77K as a function of the strain in In₅ₓGa₁₋₅ₓAs quantum wells. Results are shown for three values of $\alpha_{\text{gas}}$ (given in units of cm⁻¹). The solid and dashed curves refer to the TE and TM gain, respectively.
while the transverse electric (TE) gain still exceeds the transverse magnetic (TM) peak gain, until about 0.3% tensile strain where the TM gain becomes larger. At the point where the TE and TM branches cross, the uppermost valence subband changes from hh to lh character: its in-plane effective mass increases dramatically and even changes sign, i.e. the bandgap becomes indirect. With further increase in tensile strain $N_{th}$ decreases. Results for $J_{th}$ at 77K as a function of strain are depicted in figure 2.13b. Compared to the conventional unstrained quantum well lasers, $J_{th}$ is reduced considerably for both compressive and tensile strain, especially for large $\alpha_{loss}$. For compressive strain there seems to be an optimum around 0.7%. The optimum is the result of a trade off between the benefits from a small effective hole mass and the disadvantages of both the reduced optical confinement, and a saturation of the gain that takes place at smaller carrier densities. For $\alpha_{loss}$ exceeding 20 cm$^{-1}$, $J_{th}$ is significantly lower for tensile strained wells compared to the unstrained wells. Here the larger well widths giving a larger optical confinement are beneficial.

Since, as a result of the reduced hole effective mass, the carriers in strained-layer quantum wells are confined to smaller values of the $k$-vector, a large reduction in the number of transitions involved in Auger recombination and intervalence band absorption is expected [49]. This, in combination with smaller threshold current densities are expected to reduce the temperature sensitivity of the threshold current, i.e. to an increased $T_0$-value. Consequently, Heasman et al. [72] predicted $T_0$-values of about 140K for 1.5 μm wavelength compressively strained In$_x$Ga$_{1-x}$As quantum well lasers. The reduced intervalence band absorption is expected to improve the external differential efficiency of the strained-layer quantum well lasers.

The smaller threshold current densities for the strained-layer quantum well lasers are a result of the larger gain produced in these devices. The more symmetrical band structure is also beneficial for the differential gain. As a result, theoretical calculations indicate a significant reduction of the linewidth enhancement factor [73], and consequently the laser linewidth and chirp. Also, a significant enhancement of the modulation bandwidth up to 90 GHz was predicted [74].

Furthermore, the valence band maximum can be tuned to be heavy or light hole, resulting in TE or TM polarized emission. For compressive strain the conduction band offset is increased [60]. This reduces the electron leakage to the barrier layers and thus contributes to improvements of the $T_0$-values as well. With increased InAs mole fractions the InGaAs bandgap decreases. In figure 2.11 is shown that the emission wavelength can be tuned to about 2 μm which is of interest for several applications.
Table 2.2. Summary of effects of biaxial strain on the performance of long wavelength quantum well lasers.

<table>
<thead>
<tr>
<th>Effect of biaxial strain:</th>
<th>Effect on laser parameter:</th>
</tr>
</thead>
<tbody>
<tr>
<td>Reduction effective hole mass</td>
<td>$J_{th}$, $dG/dN$, $f_r$, $\alpha_h$</td>
</tr>
<tr>
<td>Reduction Auger recombination</td>
<td>$T_0$</td>
</tr>
<tr>
<td>Reduction Intervalance band absorption</td>
<td>$\alpha_i$, $\eta_d$</td>
</tr>
<tr>
<td>Valence band maximum: hh or lh</td>
<td>TE or TM polarization control</td>
</tr>
<tr>
<td>Compression: increased $\Delta E_c$</td>
<td>$T_0$, $J_{th}$</td>
</tr>
<tr>
<td>Accessible bandgap window extended</td>
<td>$\lambda$ up to 2.0 $\mu$m</td>
</tr>
</tbody>
</table>

Figure 2.14. 1.5 $\mu$m wavelength strained-layer In$_x$Ga$_{1-x}$As/InGaAsP quantum well laser map. The shaded areas are unfavourable due to the critical thickness limitation and the type II quantum wells. The remaining area is divided by another unfavourable area due to the hh-lh subband crossing. The InAs mole fractions and the quantum well widths required for 1.5 $\mu$m quantum well lasers are indicated by the dashed curve, fitted through circles (TE polarization) and crosses (TM polarization).

The predicted improvements of $\lambda=1.5$ $\mu$m semiconductor lasers resulting from the reshaping of the valence band structure by the application of strained-
layer quantum wells is summarized in table 2.2.

The combinations of strains and well thicknesses that are accessible for the fabrication of strained-layer \( \text{In}_x\text{Ga}_{1-x}\text{As/InGaAsP} \) QW lasers are summarized in a map shown in figure 2.14. The dashed top corners are not accessible due to the critical thickness limitation, and for small InAs mole fractions the quantum wells become type II. The remaining area is by the hh-lh crossing divided into two parts which represent the combinations of InAs mole fractions and well widths resulting in TE or TM polarized devices, respectively. At the boundary, the hh-lh valence subband levels crossing results in a degradation of the laser performance due to the formation of an unfavourable band structure. The dashed curves fitted through the circle and cross symbols represent the required well widths and InAs mole fractions in the \( \text{In}_x\text{Ga}_{1-x}\text{As} \) quantum wells for TE and TM polarized 1.5 \( \mu \text{m} \) wavelength lasers.

2.5. Conclusions

It is indicated that conventional bulk InGaAsP long wavelength semiconductor lasers are burdened with loss mechanisms related to the band structure which limit their performance. The effects of quantization of the band structure by applying quantum wells, including the effects of strain have been discussed. This analysis shows the high potential of strained-layer \( \text{In}_x\text{Ga}_{1-x}\text{As/InGaAsP} \) quantum wells for applications in long wavelength lasers with predicted dramatically improved performance.

2.6. References

[67] G.W. 't Hooft, Program "Quantumwell", version 92/01/03.


Chapter 3

Epitaxial Growth Techniques

Abstract

The epitaxial techniques available for the growth of group III-V compound semiconductors are briefly discussed. Their strengths and weaknesses with respect to the growth of InGaAs(P)/InP structures are emphasised. Organometallic Vapour Phase Epitaxy (OMVPE) which is used for the growth of materials and devices reported in this thesis is discussed in greater detail.
3.1. Introduction

Group III-V compound (opto)-electronic devices are composed of layered structures having various compositions, doping levels ranging from $10^{14}$ to $10^{20}$ cm$^{-3}$, with different doping elements for n-type, p-type, and semi-insulating behaviour. Optimum performance devices require layers controlled in thickness from the micrometer to the (sub)-nanometer range, of high crystalline quality, free of defects (native, dislocations, interfacial), with abrupt and flat interfaces. This can be achieved by epitaxy which comes from the Greek "epi" + "taxis", the former meaning outer, upon, or attachment to, and the latter meaning arrangement or order. Thus, epitaxy is the growth of crystalline layers on a single crystalline substrate where the atoms in the grown layers are arranged in the same order as in the crystal lattice of the substrate. Epitaxy can be carried out at temperatures significantly lower than the melting point, which is very important for achieving high interfacial and crystalline perfectness. However, a limit is set to the minimum growth temperature by the chemistry and physics of the growth process.

Several epitaxial techniques are currently available for the growth of semiconductor materials, including the eldest techniques, Liquid Phase Epitaxy (LPE), and Chloride Vapour Phase Epitaxy (CVPE), as well as Hydride Vapour Phase Epitaxy (HVPE), Organometallic Vapour Phase Epitaxy (OMVPE), Molecular Beam Epitaxy (MBE), and related techniques like Gas Source Molecular Beam Epitaxy (GSMBE), Organometallic Molecular Beam Epitaxy (OMMBE), and Chemical Beam Epitaxy (CBE). Each technique has specific strengths and weaknesses (summarized in table 3.1) which are related either to its principle or to the materials to be grown.

In this chapter, the epitaxial techniques are briefly reviewed with emphasis on InGaAs(P)/InP growth, the material system used in this work for the fabrication of long wavelength semiconductor lasers. The OMVPE technique, which is used in this study, is discussed in greater detail.

3.2. Liquid Phase Epitaxy (LPE)

In LPE, epitaxial layers are grown on a single crystalline substrate by direct precipitation from the liquid phase. The single-batch-process apparatus uses a graphite boat (figure 3.1) with a recess to hold the substrate, and wells that hold the separate melts for each layer. For InGaAsP/InP growth the melts are indium-rich solutions containing the dissolved layer constituents derived from solid InAs, GaAs, and InP. High purity materials for the element and compound chemicals, the quartz reactor, and the graphite boat are used, while
Table 3.1. Overview of sources, typical working pressures, mass-transport, strengths and weaknesses of epitaxial techniques

<table>
<thead>
<tr>
<th>Technique</th>
<th>Sources</th>
<th>Pressure (mbar)</th>
<th>Gasflow (mass-transport)</th>
<th>Strengths</th>
<th>Weaknesses</th>
</tr>
</thead>
<tbody>
<tr>
<td>LPE</td>
<td>elemental III, compound III-V</td>
<td>--</td>
<td>--</td>
<td>Simple High purity Planarization</td>
<td>Scale economics Inflexible Nonuniformity (composition, thickness)</td>
</tr>
<tr>
<td>CIVPE</td>
<td>elemental III, group V chlorides</td>
<td>1000-10⁻²</td>
<td>viscous flow (boundary layer)</td>
<td>Simple High purity</td>
<td>No Aluminium alloys Non-abrupt interfaces</td>
</tr>
<tr>
<td>HVPE</td>
<td>elemental III, vapour V</td>
<td>1000-10⁻²</td>
<td>viscous flow (boundary layer)</td>
<td>Simple</td>
<td>No Aluminium alloys Hazardous precursors</td>
</tr>
<tr>
<td>OMVPE</td>
<td>vapour III, V</td>
<td>1000-10⁻²</td>
<td>viscous flow (boundary layer)</td>
<td>Most flexible Abrupt interfaces Simple reactor High purity</td>
<td>Expensive reactants Hazardous precursors</td>
</tr>
<tr>
<td>MBE</td>
<td>solid III, V</td>
<td>≤10⁻⁹</td>
<td>molecular flow</td>
<td>Simple process Good uniformity Abrupt interfaces In-situ monitoring</td>
<td>Phosphorus alloys difficult &quot;Oval&quot; defects Capital intensive Low throughput</td>
</tr>
<tr>
<td>GSMBE</td>
<td>solid III, vapour V</td>
<td>≤10⁻⁵</td>
<td>molecular flow</td>
<td>See MBE</td>
<td>Capital intensive Cracker cells</td>
</tr>
<tr>
<td>OMMBE</td>
<td>vapour III, elemental V</td>
<td>≤10⁻⁵</td>
<td>molecular flow</td>
<td>See MBE</td>
<td>See GSMBE</td>
</tr>
<tr>
<td>CBE</td>
<td>vapour III, V</td>
<td>≤10⁻⁴</td>
<td>molecular flow</td>
<td>See MBE</td>
<td>See GSMBE/OMVPE</td>
</tr>
</tbody>
</table>
The system cleanliness is maintained by flushing the reactor with purified carrier gas, generally hydrogen or nitrogen. A resistance-heated furnace provides the high temperature required for the growth. After thermal equilibrium (at about 650°C), the growth is initiated when the substrate is brought into contact with the subsequent melts carefully controlled for the proper concentration to achieve growth. The supersaturation is controlled by lowering the temperature from the initial thermal equilibrium temperature. The desired layer thickness is achieved by programming the temperature decrease and growth time. In LPE the growth rate is demonstrated [1] to be limited by diffusion of the constituents of the solution to the substrate surface, provided the free convection in the solution is suppressed. Crystal growth by LPE using a graphite boat as shown in figure 3.1 was introduced by Nelson [2]. This enabled the fabrication of the first room temperature operated double heterostructure AlGaAs/GaAs laser [3, 4]. Since then, LPE has been successfully applied for the fabrication of various types of semiconductor devices including injection lasers, light emitting diodes, solar cells, photodetectors, and transistors in various materials systems containing mixtures of Al, Ga, or In, and As, P, or Sb. Until the mid 1980's, InP based optoelectronics was dominated by LPE. Extensive work was performed on the InGaAsP liquid-solid phase diagram, both theoretically [5, 6] and experimentally [1, 7-9]. The first room temperature CW operated InGaAsP lasers were grown by LPE in 1976 [10].

The advantages of the LPE technique lie in the relative simplicity of the equipment, the low operating cost, the safe operation, and the excellent optical material quality obtained routinely. In this sense, LPE certainly benefits from the
gettering effect of the liquid phase. The major drawbacks of the LPE technique are the limited homogeneities in both composition and thickness, which maximize the useful wafer size to several square centimeters. Furthermore, extremely thin layers with thicknesses in the order of 100 Å or less cannot be grown reproducibly. This is due to the rather high growth rates and the limited degree of abruptness in compositional changes between different layers caused by solution drag-over and/or by melt-back. The latter is the (partial) dissolution of epitaxial layers, which occurs especially for larger wavelength (≥ 1.55 μm bandgap wavelength) InGaAsP grown on InP. This implies that the growth of quantum wells by LPE is practically impossible. LPE is a (near)-equilibrium technique in which effects of the crystal surface orientation become evident. The tendency of crystal surfaces to minimize the surface energy gives LPE the unique feature of perfect flat groove filling which still is applied to date in several index guided semiconductor lasers (see section 5.2).

3.3. Vapour Phase Epitaxy (VPE)

Epitaxial layers of III-V compounds can also readily be grown by transporting gaseous components to the substrate by means of a carrier gas. The most commonly used molecules are chlorides, hydrides and organometallic compounds together with generally hydrogen as the carrier gas. VPE using organometallic precursors (OMVPE) was used for the growth of materials and devices in this study and will be discussed more extensively in section 3.5. VPE techniques using chlorides and hydrides [11-13] are based on the temperature dependence of the thermodynamic equilibrium between chlorides and the elements. In chloride-VPE (ClVPE), trichlorides of group V elements (PCl₃ and AsCl₃) are passed over elemental group III sources in the source zone of the reactor cell, which is set at high temperature (750-850°C) to form metal chlorides and elemental group V vapours. Downstream in the reactor cell, due to the lower temperature (650-750°C) set there, these components react near the substrate surface such that decomposition and epitaxial growth occur simultaneously. With ClVPE, layers with impurity levels <10¹⁵ cm⁻³ can be grown. This technique is therefore very well suited for the growth of electronic devices.

In the hydride-VPE (HVPE) technique, the group III metal chlorides are formed by passing HCl gas over elemental Gallium or Indium (figure 3.2). Together with group V elements supplied as hydrides, the deposition of the molecules and epitaxial growth on the substrate take place at a lower temperature than in the source zone. Note that HCl gas can also be used to remove material from the substrate by selecting a suitable temperature in the
Figure 3.2. Schematic illustration of a vapour phase epitaxy (VPE) dual-chamber growth reactor (ref. 14). In the upper part of the reactor cell InGaAsP and in the lower part sulphur-doped InP layers are grown.

deposition zone. This enables an in-situ etching procedure preparing an extremely clean substrate surface prior to the growth. The majority of optoelectronic devices which have been synthesized by VPE were grown by the hydride method.

The VPE technique offers a larger flexibility in the introduction of doping (elements and concentrations) and in the control of the chemical composition (e.g. graded layers) than LPE. In addition, the substrate area is more easily scaled up. However, the chloride- and hydride VPE techniques are not universal tools for growing all III-V semiconductor materials. They are well suited for the growth of Ga, In, As, and P-containing materials but not for the growth of Al-containing materials due to the extreme corrosivity of the Al-chlorides. Furthermore, due to the "hot-wall" nature of the process, the gas phase composition cannot be changed abruptly because no valves can be used for switching the gas flows. Alternating layers can thus only be grown by physically moving the substrate back and forth between different chambers in the reactor cell where the appropriate gas phase compositions are flowing as indicated in figure 3.2.
3.4. Molecular Beam Epitaxy (MBE) and Related Techniques

MBE is essentially a refined form of ultra-high vacuum (P=10^{-8} to 10^{-10} mbar) evaporation in which directed neutral molecular beams impinge on a heated (T=500-600°C) substrate to form an epitaxial film [16]. The apparatus required for achieving this is rather complex; the essential parts of the system for the growth of III-V components are illustrated schematically in figure 3.3.

![Diagram of MBE system](image)

**Figure 3.3.** Schematic illustration of a molecular beam epitaxy (MBE) system for the epitaxial growth of AlGaAs (ref. 15). As sources, effusion cells containing Ga, As, Al, Mn and Sn are used. The high vacuum conditions enable in-situ analysis techniques like electron diffraction, mass spectrometry, Auger analysis, etc.

The MBE growth chamber is equipped with a number of effusion cells, each filled with one pure element, e.g. Ga(1), As(2), and Al(3) for the growth of Al_{x}Ga_{1-x}As, and Mn(4) and Sn(5) for p-type and n-type doping, respectively. The composition of the epitaxial layer is controlled by the temperature of the effusion cells and by opening and closing of shutters in front of the effusion cells which permit a rapid exchange of the beam species in order to abruptly alter the film composition and/or doping. Inside the MBE growth chamber, cryopanels, i.e. large cold surfaces cooled at liquid nitrogen temperature for gettering the re-evaporated gaseous compounds from the substrate, maintain a sufficiently low background pressure. The high vacuum conditions enable in-situ observations during the growth for surface analytical techniques, including e.g.
reflection high-energy electron diffraction (RHEED) to monitor the surface structure, residual gas analyzer, and other characterization techniques. While MBE is ideally suited for fundamental investigations of the surface processes prior and during the growth, it has limitations as a fabrication technique. The need for UHV equipment makes it expensive and sensitive to failures. Due to the limited amounts of source materials that can be loaded into an effusion cell, the UHV system has to be opened regularly which requires subsequent time consuming bake-out procedures to re-establish optimum growth conditions. Furthermore, MBE gives rise to typical morphological defects, so-called "oval defects", due to the spitting of small gallium droplets from the effusion cell. The equilibrium vapour pressures of the group III elements are relatively low compared to the beam equivalent pressures arriving at the substrate. As a result, the sticking probability is essentially unity for Al, Ga, and In at typical growth temperatures. The vapour pressures of the group V elements, however, are relatively high and as a result the sticking probability is essentially zero. For As-containing materials, it follows that stoichiometric materials are always obtained if it is provided in excess to the growth interface, the excess As being lost by desorption. The extremely large vapour pressure of P poses serious problems, however. Despite using cryopanels, the P degrades the vacuum in the MBE chamber, and ultimately collects in the vacuum pumps. In addition, the growth of alloys containing both As and P is particularly difficult. Therefore, for III-V compounds MBE is almost uniquely used for the growth of As containing materials in which group III elements have been substituted. High quality (Al)GaAs [17-19] and AlGaInAs, the alternative material system for long wavelength lasers [20], were demonstrated. In the production of AlGaAs/GaAs semiconductor lasers, ROHM, a Japanese components manufacturer, is presently the only one to use MBE, whereas it is used in several other companies for the fabrication of electronic components.

The substitution of the solid group V sources by gaseous sources (group V hydrides) in combination with high temperature crackers, referred to as gas source MBE (GSMBE) was used first by M.B. Panish [21]. With this technique high quality InGaAsP/InP [22], and recently also excellent results on InGaAs(P)/InP MQW lasers were obtained [23]. Similarly, elemental group III sources were replaced by group III organometallic precursors in the organometallic (OM)MBE. Veuhoff [24] pioneered with the substitution of both the solid group III and group V elements by group III organometallics and group V hydrides. This technique, referred to as Chemical Beam Epitaxy (CBE), is nowadays the most widely used MBE-related technique for the growth of InP-based optoelectronic devices [25], and is a challenge for organometallic vapour phase epitaxy (OMVPE) and GSMBE.
3.5. Organometallic Vapour Phase Epitaxy (OMVPE)

OMVPE is a cold-wall process based on the chemical reaction of organometallic group III precursors together with group V precursors, which may be hydrides or organometallic compounds, on a heated substrate with subsequent epitaxial growth. This technique solves the main practical problems encountered in the chloride- and hydride-VPE, i.e. the precursors are thermodynamically stable (or show very large activation energy for decomposition) at room temperature, and are therefore not corrosive, including the aluminium precursors. As a consequence, pneumatic valves can be applied for rapid switching of the vapour flows, enabling rapid compositional changes in the reactor cell required for growing abrupt interfaces. Furthermore, OMVPE is well suited to deal with phosphorus containing materials. Compared to LPE, OMVPE provides larger size growth, with excellent morphology, control of thickness, composition, and doping, and is upscalable to multi-wafer reactors [26, 27].

OMVPE was introduced by H.M. Manasevit [28] in 1968 with the growth of GaAs films on glass and Al₂O₃ substrates in an open tube reactor (sic) from trimethylgallium and arsine. The overall reaction equation is \( \text{Ga(CH}_3)_3 + \text{AsH}_3 \rightarrow \text{GaAs} + 3 \text{CH}_4 \). In a relative short period of time, Manasevit and co-workers [29] explored the growth of virtually all III-V compounds containing combinations of the group III elements Al, Ga, or In, and the group V elements N, P, As, or Sb, demonstrating the versatility of the OMVPE technique. This initiated a large world-wide research effort on OMVPE which was summarized in several excellent reviews [30-33]. Despite the fact that the full details of the reaction steps involved are not yet known, today OMVPE is well established and has become the major production technique for III-V compound semiconductors. However, it took some years before the control of the growth of InGaAsP/InP was obtained. Difficulties arise especially from the marked difference in the ease with which arsine (\( \text{AsH}_3 \)) and phosphine (\( \text{PH}_3 \)) are decomposed and incorporated into the epilayer and from premature side-reactions between indium organometallic precursors with \( \text{PH}_3 \). The latter problem was solved by reducing the reactivity by using adduct precursors [32], or by using normal precursors and reducing the contact time between these precursors by injecting them separately into the reactor and by increasing the flow of the gas phase either by increasing the flow of by reducing the reactor pressure. Semiconductor lasers in the InGaAsP system grown by OMVPE were first reported in 1980 [34], and in the same year lower threshold current densities than for LPE grown lasers were reported [35]. Since then, improvements have been obtained both in low-pressure and atmospheric-pressure OMVPE. The OMVPE technique offers a high degree of uniformity and allows the growth of complex layer sequences with very
abrupt compositional interfaces (chapter 4) for high performance devices as will
be demonstrated in this work for (strained-layer) InGaAs(P)/InP quantum well
(chapter 4) semiconductor laser diodes (chapters 5 and 6).

3.5.1. OMVPE Growth System

A simplified sketch of the OMVPE growth system used in this study is
shown in figure 3.4. It can be grouped into five major parts:

a) gas manifold, including the welded stainless steel tubing, and
   instruments for the control of concentration and composition (mass flow
   controllers, pressure controller and pneumatic valves);

b) quartz reactor cell in which the epitaxial growth occurs;

c) reactor outlet with the low-pressure pumping system and exhaust with
   the scrubber;

d) electronics and computer, and

e) safety system.

In the gas manifold, the gas phase concentrations and the compositions
are controlled. Group III organometallic precursors are held in stainless steel
bubblers within thermostatic baths to control their vapour pressures. A precise
amount of palladium-diffused hydrogen carrier gas, metered by a mass flow
controller, is saturated with organometallic vapour by passing it through the
bubbler. This flow is diluted with hydrogen and sent to a pneumatic four-way
valve over a pressure controller, which stabilizes the pressure in the bubbler. In
the four-way valve the hydrogen flow containing the organometallic precursor
can either via the run-line be sent into the mixing chamber prior to the reactor
cell or via the vent-line be bypassed to the reactor cell into the exhaust. Each
channel has its own four-way valve from which all outlets to the reactor cell are
combined into one common run-line. The system used in this study is equipped
with organometallic channels for trimethyl precursors of aluminium, gallium, and
indium. Several channels are present in duplicate such that one can be used for
the growth while the other, which is stabilized at the appropriate flow required
for the next epitaxial layer, is directed towards the vent-line. Simply inverting
the gasflows in vent and run-lines by switching the four-way valves enables
extreme abrupt changes of the concentrations in the reactor cell with consequent
abrupt compositional changes in the grown layers as will be shown in chapter
4. The manifold is designed for minimal dead volume by permanently purging
all lines. Diethylzinc is used as the p-type dopant source and has a separate line
towards the mixing chamber.

The hydrides (pure AsH₃ and PH₃, and H₂S diluted to 1% in hydrogen
as n-type dopant source) are introduced in metered amounts by a mass flow controller without carrier gas. After introduction into the manifold the hydrides are diluted with carrier gas. Each group V channel has a separate line with vent-run option; these are combined into one hydride run-line towards the reactor cell. Also here, twin-sources are present to enable abrupt changes of the group V gas phase compositions. The n-type dopant source has a separate line towards the mixing chamber.

The reactor cell is made of quartz and has a horizontal design. It contains a graphite susceptor which can hold two 2-inch wafers. The susceptor is radio frequency inductively heated to the growth temperature, which ranges from 600 to 650°C. Inside the reactor cell dead volumes are eliminated, variations in the cross sectional area are smooth, and the gas flow is controlled to be laminar. This is essential for the elimination of memory effects and vortices to assure abrupt compositional changes in the gas phase upon switching.
The reactor is operated at reduced pressure (50 to 100 mbar) using a rotary pump with a throttle valve for pressure control. Both, the reactor cell as well as the gas manifold are held at reduced pressure.

As the extremely toxic hydrides, arsine and phosphine, do not completely decompose during their residence time in the reactor cell, the waste gases have to be detoxified prior to releasing the effluents into the environment. A wet chemical scrubber reducing the arsine and phosphine content below 20 ppb is used.

To allow reproducible growth of complicated layer sequences, all functions, including flow-, pressure-, and temperature control and the gas phase switching are controlled by a dedicated computer; manual control is in principle possible also. In OMVPE extremely toxic and pyrophoric precursors are used together with hydrogen. Therefore, a hardware controlled safety system is included comprising an eight channel hydride detector and fire detection with automated extinguisher. In case of an emergency, e.g. a hydride leak, the input gases are stopped, the complete reactor is shut down and purged with nitrogen.

3.5.2. Pyrolysis of Group III Organometallic Precursors

The term organometallic precursor includes a broad class of compounds that contain metal-carbon bonds or coordination compounds of metals and organic molecules. The organometallic compounds of interest for the growth of III-V compound semiconductor layers are usually simple metal alkyls containing methyl, ethyl or iso-butyl ligands, because these compounds have a sufficient vapour pressure near room temperature. These compounds are generally liquids at room temperature, though trimethylindium is a solid up to 88°C.

The organometallic precursors are not stable at high temperatures, i.e., they pyrolyse when heated above a certain temperature. This effect is used in OMVPE to release the ligands from the metal atom. In OMVPE, the decomposition reactions will be in part homogeneous reactions, i.e., they occur entirely in the gas phase, and in part heterogeneous, i.e., they occur at a solid surface. The homogeneous pyrolysis of some organometallic precursors in various carrier gases has been intensively studied [32]. As an example, the temperature ranges for the pyrolysis of trimethyl precursors of gallium and indium and of triethylgallium for typical residence times in an OMVPE reactor cell in H₂ carrier gas are given in table 3.2.
Table 3.2. Pyrolysis of metal alkyls in H₂ carrier gas (after [36, 37])

<table>
<thead>
<tr>
<th>Reactant</th>
<th>Pyrolysis starting</th>
<th>Pyrolysis completed</th>
</tr>
</thead>
<tbody>
<tr>
<td>TMGa (=Ga(CH₃)₃)</td>
<td>370°C</td>
<td>460°C</td>
</tr>
<tr>
<td>TEGa (=Ga(C₂H₅)₃)</td>
<td>220°C</td>
<td>330°C</td>
</tr>
<tr>
<td>TMIn (=In(CH₃)₃)</td>
<td>250°C</td>
<td>340°C</td>
</tr>
</tbody>
</table>

As shown in table 3.2, these alkyls are completely decomposed above 460°C. Due to the strongest metal-carbon bonds present in TMGa, it shows the highest pyrolysis temperature. The weaker indium-carbon bonds in TMIn, due to the larger indium atom, are broken at lower temperatures. The homogeneous part of the decomposition process is elucidated by using different carrier gases. Figure 3.5 shows the temperature dependence of the pyrolysis of TMIn in various carrier gases. The decomposition proceeds at lower temperature in H₂ and D₂ because these carrier gases can take part in the reaction. In H₂, methane, and in D₂, D-labelled methane are formed. On the other hand, He cannot take part in the TMIn decomposition reaction which proceeds therefore more difficult in this ambient. Similarly, the presence of group V hydrides are shown to reduce the pyrolysis temperature [32].

![Graph showing percent decomposition vs temperature for different carrier gases](image)

**Figure 3.5.** In(CH₃)₃ decomposition versus temperature showing the influence of various carrier gases. The decomposition is faster in H₂ and D₂ because here the formation of the stable methane or D-labelled methane are possible. The decomposition is slowest in a He ambient because it does not take part into the reaction.
A complete pyrolysis does not imply that we have free metal atoms but rather that the original organometallic precursor concentration is zero. They have reacted to intermediate components which are adsorbed at the crystal surface where final heterogeneous reaction(s) take place with as a final result that the metallic atom becomes incorporated into the crystal. Although in an OMVPE process no equilibrium will exist at the crystal surface, modelling using the thermodynamical equilibrium state can be helpful in revealing the most important species. For TMGa and TMAI the most important species participating in the growth process at the substrate surface are estimated to be monomethylgallium (MMGa) and monomethylaluminium (MMAI) with activation energies of 28.61 kcal/mole and 37.91 kcal/mole [38] for their formation from the trimethyl precursors, respectively. The importance of heterogeneous reactions is demonstrated by the fact that the pyrolysis temperature is reduced by the presence of solid III-V compound materials [32].

Figure 3.6. Decomposition of PH₃ versus temperature. Curve (a) 60 cm² silica tube with N₂ (□), H₂ (●), and D₂ (■) carrier gas; curve (b) 300 cm² silica packing with D₂ carrier gas; curve (c) 60 cm² InP coating with D₂ carrier gas (ref. 39).

3.5.3. Pyrolysis of Group V Hydrides

The sources for arsenic and phosphorus used from the earliest days of OMVPE, arsine (AsH₃) and phosphine (PH₃), are also used in this work. Pyrolysis studies have shown that the hydrides are more stable than the organometallic precursors. Similar to the group III organometallic precursors, the pyrolysis of AsH₃ is accelerated by the presence of GaAs material; the pyrolysis was found to start at 400°C and to be completed at 550°C.

PH₃ is more stable than AsH₃. Recent studies have shown that PH₃
pyrolysis starts at 650°C and is completed above 920°C, irrespective whether the carrier gas is N₂, H₂ or D₂, as shown by curve a in figure 3.6. The pyrolysis temperature of PH₃ is significantly reduced by increasing e.g. the silica packing in the reactor, and by adding solid InP, as shown by curves b and c in figure 3.6, respectively. Moreover, the addition of TMIn to the reaction mixture was also found to lower significantly the decomposition temperature of PH₃ as shown in figure 3.7.

![Graph showing PH₃ pyrolysis versus temperature](image)

**Figure 3.7.** PH₃ pyrolysis versus temperature showing the effect of TMIn addition to the vapour phase. Experiments were performed in InP-coated SiO₂ tube with surface area of 50 cm². Curve a (▲): PH₃ alone at a concentration of 15% in D₂; curve a (□): 15% PH₃ in D₂ with TMIn added to PH₃/TMIn ratio of 47. Curve b: 15% PH₃ in D₂ with TMIn added to PH₃/TMIn ratio of 4.2. Curve d: similar to (b) with PH₃/TMIn ratio of 2.1. The data labelled (c) are measured under similar conditions to those labelled (b) using a reactor tube with 1200 cm² inner-tube area. (ref. 40).

### 3.5.4. OMVPE Growth Mechanism and Growth Parameters

The principle of OMVPE is rather simple. The vapour phase in the reactor cell has a higher chemical potential than the solid phase at the growth temperature. This yields the supersaturation needed for the crystal growth. However, a complete understanding of the process requires knowledge about the hydrodynamics and every individual step with the chemical species involved. The chemistry of the OMVPE process is very complicated and is not yet known in full detail, especially when using several different precursors such as e.g. TMGa+TMIn+AsH₃+PH₃ for the growth of InGaAsP. The kinetics of the growth process can be divided into the following sequential steps as illustrated schematically in figure 3.8:
1. mixing and transport of the reactants from the inlet to the heated reaction zone,
2. gas phase reactions leading to reactive radicals and byproducts,
3. transport of precursors and radicals towards the substrate by diffusion,
4. adsorption of these components on the substrate surface,
5. surface diffusion towards reactive surface sites,
6. incorporation of film constituents into the crystal lattice via chemical reactions,
7. desorption of reaction (by)-products off the crystal surface, and
8. mass transport of products away from the deposition zone towards the exhaust.

The step that proceeds most difficult determines the overall growth rate. The hydrodynamic part of the process is described by the three-dimensional Navier-Stokes partial differential equations. Analytical solutions for the geometry of the reactor cell are next to impossible, but fortunately the experimental conditions may be chosen such that physically justifiable simplifications are allowed [41]. By using a properly designed reactor cell, proper gas flow conditions, and by placing the substrates at a certain distance downstream the leading edge of the susceptor, it can be shown that above the substrates the velocity and temperature profiles have completely developed [41, 42], that is they have become

![Diagram](image)

**Figure 3.8.** Overview of the sequential steps in the growth process: (1) mixing and transport of the reactants from the inlet to the heated reaction zone, (2) gas phase reactions leading to reactive radicals and byproducts, (3) transport of precursors and radicals towards the substrate by diffusion, (4) adsorption of these components on the substrate surface, (5) surface diffusion towards reactive surface sites, (6) incorporation of film constituents into the crystal lattice via chemical reactions, (7) desorption of reaction (by)-products off the crystal surface, and (8) mass transport of products away from the deposition zone towards the exhaust.
independent of the position along the flow direction. Important figures here are the Reynolds number (Re), which should be smaller than 2300 for laminar flows, and the Grashof number (Gr), which is the Reynolds number-analogue for free convection, which should be smaller than 1700 for gasflows free of convection. Convective instabilities may be present at atmospheric pressure in large height reactor cells and will disappear at reduced (<0.1 bar) pressures or at higher operating pressures by using small height (~1 cm) reactor cells. With the transport problem solved, it was shown that the growth of elemental and binary materials can be fully described in different regimes by introducing the chemical boundary layer concept [43, 44]. The chemical boundary layer thickness is defined as the region in the reactor cell in the hot zone just above the susceptor in which the chemistry takes place. The thickness seems to be independent of the concentration of the reactants, and also of the order of the reaction. The temperature in this layer may be assumed to be constant and equal to the temperature of the susceptor, while because of the laminarity of the flow, the flow velocity can be approximately set to zero. The thickness of the chemical boundary layer is related to the thickness of the developed temperature profile $\delta_r$ but reduced by a factor $E_a/2kT$ where $E_a$ is the activation energy of the reaction involved [43].

For the growth of GaAs from trimethylgallium and arsine, monomethylgallium (MMGa) [38] and arsenicmonohydride (AsH) are assumed to be the important species. Using the rate constants for the reaction of MMGa with AsH at a step site at the substrate surface, the GaAs growth rate has been deduced as a function of the growth temperature [43, 44]. Figure 3.9 shows the excellent agreement with the experimentally measured growth rate of GaAs [45]. In this curve three regions can be distinguished on the basis of the growth temperature, indicating that there are three different limiting steps governing the growth:

a) at low temperatures ($T \leq 550^\circ C$) the growth exhibits a temperature-activated growth rate dependence with an activation energy of about 20 kcal/mole. The growth-rate limiting step is attributed to the slow decomposition of TMGa in the gas phase. This is indicated by the high concentration of impurities, mainly carbon, found in the materials grown in this temperature range.

b) at high temperatures ($T \geq 850^\circ C$) the growth rate is temperature-deactivated. This cannot result from a chemical reaction limitation but must rather originate from desorption of adsorbed gallium and/or arsenic species.

c) the mid-temperature range ($850^\circ C \leq T \leq 550^\circ C$) is typically utilized in the OMVPE growth of III-V semiconductors, and is characterized by a
nearly temperature independent growth rate. In this regime, the growth is controlled by the diffusion of growth species through the gas phase towards the substrate surface. Material with the best morphological, optical, and electrical characteristics are grown in this temperature region. For the growth of InP at atmospheric pressure using TMIn and PH₃, this diffusion limited regime was found to range from 550-700°C [46].

In the diffusion limited regime, the highest growth rate is obtained which is a linear function of the group III input pressure, and is independent of the AsH₃ partial pressure under conditions of excess AsH₃. The group V/III ratio is generally chosen \( \gg 1 \) (typically 50-250) to prevent evaporation of the highly volatile group V elements at the growth temperature. Moreover, the V/III ratio

![Figure 3.9. GaAs growth rate from TMGa and AsH₃ as a function of temperature. Experimental points (■) are from Reep et al. [45]. Theoretical curves are calculated by Croon et al. [43] (dashed) and van Sark [44] (solid) using a chemical boundary layer concept. At low temperature the growth rate is kinetically limited by the TMGa decomposition in the gas phase, a maximum growth rate is obtained in the middle temperature range, and is diffusion limited, whereas at high temperatures the growth rate is limited by desorption of adsorbed species from the substrate surface.](image)
controls the group V vacancy concentration, and plays an important role in the adsorption/desorption of impurities at/from the substrate surface. Furthermore, in the diffusion limited regime the growth rate is a function of the lateral position of the substrate, that is towards the end of the susceptor the gas phase is depleted resulting in a reduced growth rate. Ways to reduce these effects are the use of a tilted susceptor or more ideally the use of substrate rotation. The OMVPE growth at low-pressure gives the ability to enlarge the gas velocity which results in improved homogeneity of thickness and composition, and additionally the reduced pressure is also effective to minimize premature side-reactions which are known to be a problem for the combination of TMIn or TEIn (triethylindium) with PH$_3$.

The above modelling using the chemical boundary layer concept has been shown so far for the growth of Si, GaAs [43, 44] and for Al$_x$Ga$_{1-x}$As with x=0.14 [44] but not yet for other combinations of Al, Ga, In, As or P.

3.5.5. OMVPE Growth of InGaAs(P)/InP

The overall reaction equation for the growth of In$_x$Ga$_{1-x}$As is written as

\[ x \ln(CH_3)_3 + (1-x) \text{Ga(CH}_3)_2 + \text{AsH}_3 \xrightarrow{\text{temperature}} \text{In}_x\text{Ga}_{1-x}\text{As} + 3 \text{CH}_4 \]

For the growth of InGaAsP also PH$_3$ is added to the reactant flow. Table 3.3 summarizes the gas phase concentrations, reactor pressure, growth temperature, growth rate and V/III ratio for the growth of 1.55 µm bandgap wavelength In$_{0.58}$Ga$_{0.42}$As$_{0.9}$P$_{0.1}$ and InP typically used in this study.

The overall reaction equation for In$_x$Ga$_{1-x}$As growth suggests a linear relation between the group III vapour and solid concentrations. As shown, in figure 3.10, the Ga fraction in solid In$_x$Ga$_{1-x}$As grown at 625°C, 50 mbar, is found to depend linearly on the TMGa/(TMGa+TMIn) ratio in the vapour phase. The solid In$_x$Ga$_{1-x}$As ternary composition was determined from the lattice parameter on thick relaxed and thin (strained or pseudomorphic) layers. This shows that the elastic energy accumulated in the mismatched layers is no obstacle in the OMVPE growth as a result of the large supersaturation. In the temperature range 600-625°C a constant group III vapour-solid distribution was observed. Addition of PH$_3$ to the reactant flow required for the growth of InGaAsP does not influence the In/Ga gas phase to solid distributions.

On the other hand, due to the large difference in the pyrolysis of AsH$_3$
Table 3.3. LP-OMVPE growth conditions for $\text{In}_{0.58}\text{Ga}_{0.42}\text{As}_{0.9}\text{P}_{0.1}$ (bandgap wavelength=1.55 μm) and InP used in this work.

<table>
<thead>
<tr>
<th>Parameter</th>
<th>$\text{In}<em>{0.58}\text{Ga}</em>{0.42}\text{As}<em>{0.9}\text{P}</em>{0.1}$</th>
<th>InP</th>
</tr>
</thead>
<tbody>
<tr>
<td>$\text{P(IndCH}_3)_3$</td>
<td>$4.8 \times 10^{-3}$ mbar</td>
<td>$4.8 \times 10^{-3}$ mbar</td>
</tr>
<tr>
<td>$\text{P(Ga(IndCH}_3)_3$</td>
<td>$3.6 \times 10^{-3}$ mbar</td>
<td>--</td>
</tr>
<tr>
<td>$\text{P(AsH}_3$</td>
<td>$3.3 \times 10^{-1}$ mbar</td>
<td>--</td>
</tr>
<tr>
<td>$\text{P(PH}_3$</td>
<td>1.11 mbar</td>
<td>1.20 mbar</td>
</tr>
<tr>
<td>$\text{V/III ratio}$</td>
<td>171</td>
<td>250</td>
</tr>
<tr>
<td>Reactor pressure</td>
<td>50 mbar</td>
<td>50 mbar</td>
</tr>
<tr>
<td>Substrate temperature</td>
<td>625°C</td>
<td>625°C</td>
</tr>
<tr>
<td>Growth rate</td>
<td>250 Å/min</td>
<td>150 Å/min</td>
</tr>
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</table>

Figure 3.10. Ga fraction in solid $\text{In}_{1-x}\text{Ga}_x\text{As}$ versus the $\text{TMGa/(TMGa+TMIn)}$ fraction in the vapour phase. The Ga fractions were deduced from the lattice parameters measured by X-ray on (relaxed) bulk and (strained-layer) QW structures grown at 625 °C on InP substrates.
Figure 3.11. As/(As+P) fraction in lattice matched InGaAsP grown at 625°C on InP versus the AsH3/(AsH3+PH3) in vapour phase. The InGaAsP composition was deduced from the room temperature bandgap, determined by PL, and the lattice parameter deduced from X-ray measurements.

and PH3, the AsH3/(AsH3+PH3) ratios in the gas phase and the As/(As+P) ratios in the solid phase show a strong non-linear relation as shown in figure 3.11. In addition, this relationship is a strong function of the temperature. The quaternary InGa1-xAsxP1-y solid compositions were deduced from its room temperature bandgap measured by photoluminescence and the lattice parameter using the relations $E_g = 1.35 - 0.72y + 0.12y^2$ (eV) [47] and $(\Delta a)/a = 0.07115(x-1) + 0.03443y - 0.00212xy$. The latter equation can be deduced using Vegard’s law.

Using LP-OMVPE growth, excellent homogeneity in terms of thickness and composition were obtained across a 2-inch wafer. Excluding an outer edge of 5 mm on the 2-inch wafer, the homogeneity in thickness is typically better than ±3% without substrate rotation. Photoluminescence emission wavelengths vary typically ±3 nm across a 2-inch wafer for 1.3 and 1.55 μm InGaAsP. Figure 3.12 shows the photoluminescence emission wavelengths of an In0.55Ga0.45As/InGaAsP multiple quantum well structure across a 2-inch wafer. The small variation in emission wavelength confirms the excellent homogeneity in composition and thickness across the wafer. In next chapter, the growth and characterization of InGaAs(P)/In(GaAs)P quantum structures will be discussed in more detail.
Figure 3.12. Photoluminescence wavelengths across a 2-inch wafer of In$_{0.53}$Ga$_{0.47}$As/InGaAsP multiple quantum well structure. No substrate rotation was used; gasflow was in y-direction.

3.6. References


Chapter 4

Characterization of LP-OMVPE Grown InGaAs(P)/In(GaAs)P Interfaces

Abstract

Unstrained and up to 1.8% biaxially compressed InGaAs(P)/In(GaAs)P heterointerfaces and quantum well structures grown by low-pressure organometallic vapour phase epitaxy (LP-OMVPE) have been studied using transport measurements, photoluminescence (PL), photoluminescence excitation (PLE) spectroscopy, transmission electron microscopy (TEM), and high resolution X-ray diffractometry (HR-XRD). Shubnikov-de Haas measurements reveal the existence of a two-dimensional electron gas at InP-to-InGaAs(P) ($\lambda_g=1.65, 1.55, and 1.3 \mu m$) heterojunctions, with mobilities among the highest ever reported. This indicates the high quality of the materials and the abruptness of, and the crystalline quality at the interfaces. The abrupt and flat interfaces at both sides of the unstrained and strained-layer quantum wells were confirmed by direct comparison of low-temperature PL spectra of structures grown simultaneously on InP substrates with small ($\leq 0.2^\circ$) and larger ($\geq 2^\circ$) misorientations from the (001) plane. Multiple-line PL spectra from thin quantum wells grown on substrates with small misorientations could be assigned to inter-well-width variations of integer number of atomic layers by the k.p calculation using a consistent set of band structure parameters, including the band offsets deduced by PLE. Quantum wells with thicknesses down to several monolayers are shown to have very narrow PL linewidths (e.g. 8.8 meV for a 6 Å thick well of InGaAsP), and to show quantum shifts corresponding with 90% of the difference in bandgaps between well and barrier layers. For strained-layer quantum wells the quality of the interfaces was found to depend on the substrate misorientation, i.e. for strains of 1.8% and misorientations larger than 0.2°, the PL spectra are broadened resulting from 3D-like growth as revealed by TEM, whereas the corresponding QWs grown on (001) InP substrates with smaller misorientations show narrow PL linewidths.

For the formation of high quality abrupt interfaces, the gas phase switching is extremely important, especially in structures where the highly volatile group V element has to be replaced in the subsequent layers. PL and TEM measurements indicate and HR-XRD measurements on InGaAs/InP superlattice structures confirm the exchange of group V elements during growth pauses at the interfaces, introduced to purge the reactor cell to prevent group V element carry-over to the subsequent layer.
4.1. Introduction

For optimum performance high speed electronic devices [1] (e.g. high electron mobility transistor (HEMT), modulation doped field effect transistor (MODFET), and heterobipolar transistor (HBT)), and optoelectronic devices (e.g. semiconductor lasers, photodetectors, modulators), the abruptness and flatness of heterointerfaces are of ultimate importance, especially when a large number of interfaces is involved, like in multiple quantum well (MQW) structures. High quality AlGaAs/GaAs heterojunctions and quantum wells were grown by atmospheric-pressure organometallic vapour phase epitaxy (AP-OMVPE [2], low-pressure (LP)-OMVPE [3], molecular beam epitaxy (MBE), and related techniques [4]. This confirms the feasibility of the abrupt group III element substitution in the subsequent layers which are lattice matched for all aluminium fractions. In InGaAs(P)/InP heterojunctions, however, the concentrations and substitution on both the group III and the group V element sublattices have to be controlled. Much work on InP-based materials grown by a variety of epitaxial techniques like AP-OMVPE [5, 6] or LP-OMVPE [7], hydride-VPE [8], chloride-VPE [9], MBE [10], gas source MBE (GSMBE) [11], and chemical beam epitaxy (CBE) [12] has been performed in order to optimize the growth conditions with emphasis on the relation of interfacial structures with gas switching conditions. Here, we report on the assessment of LP-OMVPE grown InGaAs(P)/In(GaAs)P interfaces using transport measurements, optical characterization using (low-temperature) photoluminescence (PL) and photoluminescence excitation (PLE) spectroscopy, transmission electron microscopy (TEM), and high resolution X-ray diffraction (HR-XRD) measurements. Results from this study will be compared with data reported in the literature.

4.2. Transport Measurements

4.2.1. Introduction

At an abrupt heterojunction the charge transfer from the wide bandgap semiconductor into the narrow bandgap semiconductor, e.g. from n'-doped layers of InP into n-InGaAs(P), results in bending of the conduction- and valence bands as shown in figure 4.1a. Electrons are confined in the potential well at the heterointerface in the InGaAs(P) leading to the formation of a two-dimensional electron gas (2-DEG) [13]. Here, the physical separation of the electrons and the ionized donors results in the most desirable property of the 2-DEG, i.e. an electron mobility significantly larger than in bulk materials. The carrier densities
Characterization of InGaAs(P)/In(GaAs)P interfaces

Figure 4.1. a) Schematic band structure of an In$_{0.53}$Ga$_{0.47}$As/InP heterojunction. b) Sample geometry used in magnetotransport measurements.

and mobilities can be deduced from magnetotransport measurements using a sample geometry shown in figure 4.1b. In the conventional Van der Pauw measurement, a small constant magnetic field $B$ is applied perpendicularly to the 2-DEG, and from the voltage drop $V_{xx}$ the carrier mobility may be deduced [14]. This measurement does not give proof of the 2-DEG-nature of the conduction, however. At low temperatures ($kT<\hbar\omega_c/2\pi$), the kinetic energy of the electrons becomes quantized at $E_n=(n-\frac{1}{2})\hbar\omega_c/2\pi$, where $\omega_c=eB/m_e$ is the cyclotron frequency, $h$ is Planck's constant, $e$ is the electron charge, and $m_e$ is the electron effective mass. The quantum numbers $n=1, 2, ..., k$ label the Landau levels [15] which correspond classically to the motion of electrons in cyclotron orbits. Only Landau levels up to the Fermi energy are occupied, and the number of occupied Landau levels $N_c-2\pi E_F/\hbar\omega_c$ decreases with increasing magnetic field. Every time when a Landau level crosses the Fermi energy, the carriers can contribute to the conduction and then the system will act as a metallic conductor. Subsequently, the Landau level is depopulated when the magnetic field is increased, resulting in a sharp drop of the conductivity to behave like an isolator when the level becomes completely emptied. Therefore, upon sweeping the magnetic field, the magnetoresistance $R_{xx}=V_{xx}/I$ shows strong oscillations: the Shubnikov-de Haas (SdH) oscillations [16]. For magnetic fields with $R_{xx}=0$, that is when a Landau level coincides with the Fermi energy, the quantized Hall resistance $R_{xy}=V_{xy}/I$ shows plateaus, and jumps to a higher resistivity when the
Fermi energy is in between two Landau levels. At the plateaus the quantized Hall resistance is independent of the carrier mobility, sample geometry, and impurity concentration and is given with an accuracy better than $1:10^8$ by: $R_{xy} = h/ie^2$ where $i$ is an integer [17]. The SdH effect appears regularly periodic when $R_{xx}$ is plotted against the inverse magnetic field $B$. This is used to determine the areal carrier density which together with the resistivity is used to deduce the carrier mobility.

The carrier mobility and the sharpness of the SdH oscillations are mainly influenced by the purity of the epitaxial layers (ionized impurity scattering), the alloy composition (alloy scattering), the optical phonon scattering, especially at higher temperatures, and the quality of the heterointerface (abruptness and flatness), as this determines the shape of the energy well. However, a quantitative relation between the scatter potentials and the interface irregularities is unknown, and therefore, the transport measurements only give a qualitative indication of the interface quality. In the InGaAs(P)/InP material system magnetotransport studies were performed using structures grown by various epitaxial techniques [18-38].

4.2.2 Results and Discussion

The samples for this study were grown on (001) semi-insulating InP substrates by LP-OMVPE using conditions reported in section 3.5.5, and consisted of a 0.2 µm InP buffer layer followed by a 0.5-1.0 µm thick layer of lattice matched InGaAs(P). Both layers were not intentionally doped, which typically resulted in n-type background doping levels of low-$10^{14}$ and $1-2\times10^{15}$ cm$^{-3}$ for thick layers of InP and InGaAs(P), respectively. For $\text{In}_{0.53}\text{Ga}_{0.47}\text{As}$, excellent electron mobilities were routinely obtained with typically $\mu_{300K} > 10,000$ and $\mu_{77K} > 80,000$ cm$^2$/Vs. In figure 4.2, the temperature dependence of the Hall mobility and the sheet electron concentration is shown for two $\text{In}_{0.53}\text{Ga}_{0.47}\text{As/InP}$ samples grown in different epitaxial runs using an optimized gas switching procedure with a growth interruption at the interface. With decreasing temperature, the Hall mobility monotonically increases down to around 40K. Below this temperature, the mobility is fairly constant. This indicates the formation of a 2-DEG in which the electrons are separated from the ionized impurities. For conduction in a bulk layer, the electron mobility would pass through a maximum around 60-70K, where the impurity scattering becomes dominant causing the mobility to decrease according to a $T^{3/2}$ law [40]. Hall mobilities in the 2-DEG as high as 164,000, 103,000 and 11,600 cm$^2$/Vs were measured at 4, 80, and 300K, respectively. A maximum mobility as high as
172,000 cm$^2$/Vs was measured at 20K. Below 80K the sheet electron concentration was nearly constant at 3.7x10$^{11}$ cm$^{-2}$ and increased to 4.4x10$^{11}$ cm$^{-2}$ at 300K. In another sample not shown here, a maximum 4K mobility of 171,000 cm$^2$/Vs was measured. The latter Hall mobilities at low temperatures (T<80K) are amongst the highest reported up to now for InP-to-In$_{0.53}$Ga$_{0.47}$As heterojunctions grown by any epitaxial technique (see figure 4.6). The Van der Pauw measurements in the 4-300K temperature range shown in figure 4.2 were analyzed to estimate the major mobility limiting contributions $\mu_i$ using Matthiessen’s rule

$$\frac{1}{\mu_{\text{measured}}} = \sum_i \frac{1}{\mu_i}$$

(4.1)

The temperature dependencies of the three major contributions were taken as follows: $1/\mu_{ii}$, the ionized impurity scattering contribution proportional to $T^{16}$ [39], $1/\mu_{po}$, the optical phonon scattering contribution proportional to $T^2$ [40], and $1/\mu_{al}$, the alloy scattering is assumed to be temperature independent for a 2-DEG [6]. Fitting of the experimental data to eq. (4.1) yields $\mu_{ii}=230,000$ and $\mu_{io}=470,000$ cm$^2$/Vs at 4K. The value of the $\mu_{al}$ is in excellent agreement with theoretical values [40, 41] calculated assuming a defect-free In$_{0.53}$Ga$_{0.47}$As random alloy. This confirms
the negligible contribution of the interface-roughness-scattering-mobility limitation in eq. (4.1) which would imply an interface roughness less than 2 atomic layers (=1 monolayer) according theoretical work reported by Takeda et al. [39].

In order to confirm the 2-DEG, quantum Hall measurements at high magnetic field were performed. The experimental set-up consisted of a dilution refrigerator mounted in a Bitter magnet. At 0.2K the magnetoresistance and the quantized Hall resistance were measured in magnetic fields up to 20 T perpendicular to the InP-to-InGaAs interface. The minimum magnetic field could be controlled to 0.5 T. Hall bridges were prepared using standard lithography, wet-chemical etching and metallization using AuGeNi. For ohmic contacts at low temperatures the samples required illumination. The measurements, however, were performed in the dark.

The magnetoresistance and the quantized Hall measured at 0.2K up to 20 T are shown in figure 4.3. Figure 4.4 shows a magnification at smaller magnetic fields. From the oscillations in the magnetoresistance a carrier concentration of 4.4x10^{11} cm^{-2} is deduced. The latter value is slightly higher than the one obtained from the Van der Pauw measurements because of the sample illumination. The electron mobility extracted from the SdH measurements is 161,000 cm^2/Vs which is in good agreement with the low field mobilities. Sometimes large differences between the Hall- and SdH mobilities were reported which could be due to parallel conduction in the In_{0.53}Ga_{0.47}As top layer [24] or to population of a second subband. In our sample, the magnetoresistance becomes zero for the first time at a magnetic field as low as 1 T, and the high electron mobility causes very sharp oscillations with large amplitude at larger magnetic fields. Also, the magnetoresistance is zero for integer filling factors at stronger magnetic fields. This is evidence for the absence of parallel conduction in the structures grown for the present study. In figure 4.4 a second, slow oscillation is visible at low fields (indicated by the arrows on the B-axis), indicating that the electron concentration is high enough to populate two subbands. With this in mind, and the fact that no back-gating was used, the absence of parallel conduction becomes even more remarkable. This confirms the low impurity levels of the epitaxial layers. The slow oscillation in the magnetoresistance is not the only feature indicating the population of a second subband. Also, the Hall resistance exhibits some steps which are due to the fact that there are two series of Landau levels instead of one. Due to the difference in effective mass, these Landau levels depend on the magnetic field in different ways, which makes the Fermi level jump between the two series of Landau levels. The presence of two series of Landau levels explains the additional peaks in the magnetoresistance at 1.2 and 2.9 T, which do not fit in the 1/B periodicity of the oscillations.
Figure 4.3. Magnetoresistance $R_{xx}$ and quantized Hall resistance $R_{xy}$ of a 2-DEG formed at an InP-to-In$_{0.55}$Ga$_{0.47}$As heterojunction versus the magnetic field up to 20 T perpendicular to the heterointerface.

Figure 4.4. Magnification of magnetoresistance $R_{xx}$ and quantized Hall resistance $R_{xy}$ of figure 4.3. at low magnetic fields.

InP/InGaAsP ($\lambda_g=1.55$ and 1.3 $\mu$m) heterojunctions grown with a similar gas switching sequence as the one used in the growth of InP/In$_{0.53}$Ga$_{0.47}$As heterojunctions were also studied. The measurements were performed using different equipment with a maximum magnetic field up to 8 T. Due to the larger bandgap of the InGaAsP and consequently decreased well depth, lower carrier concentrations in the 2-DEG may be expected. Furthermore, for the InP-to-InGaAsP
heterojunctions smaller electron mobilities may be expected due to the higher effective mass and the increased alloy scattering in the quaternary material. Figure 4.5 shows the nicely developed oscillations in the magnetoresistance $R_{xx}$ and the stepped quantized Hall resistance $R_{xy}$ measured at 30 mK as a function of the magnetic field up to 8 T perpendicular to the InP/In$_{0.58}$Ga$_{0.42}$As$_{0.9}$P$_{0.1}$ ($\lambda_e$=1.55 µm) heterointerface. The observation of a decreasing magnetoresistance with increasing magnetic field from 0 to 0.25 T when the oscillations start, is an indication of the absence of parallel conduction [42] also in this structure. From a plot of the indices of minima in the magnetoresistance versus the inverse magnetic field, two sets of points on straight lines through the origin were obtained. This clearly indicates that two subbands are populated. For the case the sample was not illuminated, sheet electron concentrations of 2.18$x10^{11}$ and 3.9$x10^{10}$ cm$^{-2}$ for the first and the second subband were deduced, respectively. The total carrier sheet concentration is, within the experimental error, the same as the concentration deduced from the Van der Pauw measurements at low fields, being 2.61$x10^{11}$ cm$^{-2}$. Electron mobilities as high as 86,000 cm$^2$/Vs [31] were deduced at low temperature. From the temperature dependent Van der Pauw measurements, $\mu_{ei}$ = 1.1$x10^2$ cm$^2$/Vs was obtained for the 1.55 µm InGaAsP alloy using the same fitting procedure from eq. (4.1).

Shubnikov-de Haas oscillations were observed also from the 2-DEG formed at the InP-to-In$_{0.72}$Ga$_{0.28}$As$_{0.6}$P$_{0.4}$ ($\lambda_e$=1.3 µm) heterointerface. Hall mobilities of 3300, 32,700 and 39,500 cm$^2$/Vs were measured at 300, 75 and 49K, respectively,
with a sheet carrier concentration of $2.5 \times 10^{11}$ cm$^2$.

Figure 4.6 summarizes reported 4K electron mobilities in a 2-DEG formed at InP/InGaAs(P) heterojunctions grown by various epitaxial techniques. An electron mobility as high as 700,000 cm$^2$/Vs at InP/In$_{0.53}$Ga$_{0.47}$As heterojunctions was claimed by Razeghi et al. [19]. However, this value seems unlikely high as it is over a factor of two larger than the theoretical predictions [39, 41]. Moreover, it has never been reproduced, not even in strained-layer InP-to-In$_x$Ga$_{1-x}$As (x>0.53) heterojunctions where due to the increased InAs mole fraction in the alloy, the electron effective mass is reduced significantly [38].

In conclusion, the observation of two-dimensional electron gases with high electron mobilities at InP-to-InGaAs(P) ($\lambda_e$ =1.65, 1.55, and 1.3 μm) heterointerfaces is indicative for abrupt and flat (within several atomic layers) heterointerfaces, and for the high crystalline quality of the epitaxial layers grown in this study. As shown in figure 4.6, the electron mobilities reported in this work using LP-OMVPE are among the highest ever observed for InP/InGaAs(P) grown by any epitaxial technique.
4.3. Optical Characterization

4.3.1. Introduction

With photoluminescence (PL) the radiative recombination of optically excited carriers in a semiconductor in thermal equilibrium is studied. In a heterostructure, e.g. a quantum well (QW) embedded within barrier or cladding layers, carriers may be generated in the different layers of the structure depending upon the excitation photon energy. Electrons and holes generated in or collected to the quantum well as hot particles, rapidly thermalise down to their lowest confined states, $c_i$ for electrons in the conduction band and $hh_i$ or $lh_i$ for heavy or light holes in the valence band, respectively, depending upon the width of the quantum well and the sign and magnitude of the strain. At low temperatures, the electrons and holes typically form excitons with a further lowering of the energy by typically less than 10 meV. The excitons decay to detectable photons with energies corresponding to the $n=1$ electron and hole energy differences minus the exciton binding energies. The relation between the photon energy and the quantum well width for a perfect square QW was discussed in section 2.4. However, in a real structure this idealized picture of the square quantum well generally does not hold. Mainly due to problems associated with the abrupt substitution of group V elements owing to their high vapour pressures, the interfaces may deviate from the idealized picture, i.e. they may be not flat or compositionally abrupt. Due to the excitonic nature of the recombination process, especially at low temperatures, the recombination energy is an integration of the transition energies occurring within the dimensions of the excitonic diameter. The excitonic diameter is the Bohr radius analogue in solids corrected for the effective masses and dielectric constant; in InGaAs(P) materials the excitonic diameter is about 200 Å. Figure 4.7 schematically shows different compositional and topographical transitions between a quantum well and the embedding barrier- or cladding layers together with their expected lineshapes of low temperature PL spectra. In figure 4.7A, atomically abrupt and flat QW interfaces are shown as assumed in the idealized case. These quantum wells are expected to emit narrow linewidth PL-lines, whose peak-wavelengths accurately fit with the theoretically calculated values assuming square wells. In figure 4.7B, at the bottom and the top of the QW, compositionally graded but flat interfaces are assumed. They may arise due to non-abrupt changes in the gas phase composition, due to modification of the surface composition or due to incorporation of materials evaporated from the susceptor or reactor wall deposits during growth pauses at the interfaces, or due to intermixing in the solid state. Similarly to the situation depicted in figure 4.7A, the excitons in the QW of figure 4.7B experience constant potential energies
Figure 4.7. Effects of interface structure between QW and barrier layers on the lineshape of low temperature PL spectra for abrupt and flat (A), abrupt, flat and graded (B), abrupt and stepped with lateral dimensions smaller (C) and larger (D) than the excitonic diameter.

resulting in narrow linewidth PL emission lines. The exact transition energy in a graded QW depends on the bandgaps of the graded materials, however. When the bandgap of the graded material is in between the quantum well and the barrier bandgaps, the effective thickness of a graded QW is larger than of the square QW, leading to a smaller transition energy, i.e. a larger emission wavelength compared to a square well. A QW having atomically abrupt interfaces which are not flat but with thickness variations over lateral steps smaller than the excitonic diameter is shown in figure 4.7C. Such fluctuations are known to act as local trapping centers for the excitons. Hence, once localized, the excitons decay with peak PL energies representing an average of the width variations sampled over the excitonic diameter which results for the integrated PL signal in a single line emission with increased linewidth. In figure 4.7D, the sizes of the lateral thickness variations between the atomically abrupt and flat interfaces are larger than the excitonic diameter. In this case, the PL spectrum is expected to exhibit a multiple peak emission spectrum with energy separations corresponding to the energy differences between the lowest confined states in the distinct parts of the QW with different
thickness. The relative intensities in the doublets depend on the areal distribution of the different QW widths within the area of sampling. The linewidth of each individual peak is expected to be similar to the one observed from the quantum well of figure 4.7A. Mixed forms of the situations presented in figure 4.7 are also possible. Finally, lateral alloy fluctuations can also trap excitons which would add to the width fluctuation phenomena. The alloy broadening is minimal for binary compounds and increases with increasing number of chemical elements in the alloy. Furthermore, with decreasing well width, the exciton wave function becomes increasingly confined to the quantum well. Therefore, apart from the effects discussed above, the PL linewidth is also a function of the well width and the materials employed in the well and barrier layers.

With photoluminescence excitation (PLE) spectroscopy optical absorptions are examined by detecting the resultant photoluminescence from energy (or wavelength) adjustable excitation. The reason to apply PLE besides PL is that also information about the higher quantum states is obtained, whereas in PL usually only the lowest state is involved. The optical absorption coefficient reflects the step-like density of states in the QW. In addition, at the threshold of each transition, excitonic effects give rise to a resonance peak in the absorption coefficient. Therefore, the PLE spectrum is a step-like function with the onset of the step at the bandgap energy, and at larger energies peaks due to subband transitions are superimposed. For square and infinitely deep quantum wells only radiative recombinations between conduction band and valence band (hh, lh) states with the same quantum number are allowed. The PLE spectrum gives the opportunity to fit all the observed levels where the well width and the band offsets are the only free fitting parameters.

4.3.2. Experimental

The structures grown for the PLE measurements consisted of a relatively thick In$_{0.58}$Ga$_{0.42}$As$_{0.9}$P$_{0.1}$ ($\lambda_p=1.55$ µm at 300K) quantum well of about 70 Å to ensure the presence of several confined levels, and a 2000 Å thick reference layer. This provides an internal standard for confirmation of composition and lattice matching, and serves as a reference for the precise determination of spectral shift due to the quantum confinement. The quaternary layers were embedded within InP. All epitaxial layers were nominally undoped and had background doping levels of low $10^{14}$ cm$^{-3}$ and low $10^{15}$ cm$^{-3}$ for InP and InGaAsP, respectively.

For the PL measurements, multiple single quantum well (MSQW) structures, i.e. samples with a stack of single quantum wells with decreasing well thickness in the growth direction, composed of lattice matched In$_{0.55}$Ga$_{0.47}$As,
In\textsubscript{0.58}Ga\textsubscript{0.42}As\textsubscript{0.9}P\textsubscript{0.1}, or compressively strained In\textsubscript{x}Ga\textsubscript{1-x}As were grown on (001) InP substrates. Generally, several InP substrates with various misorientations towards either nearest \{111\} or \{110\}, respectively, were placed side by side on the susceptor. The quantum wells were embedded within InP or smaller bandgap barrier layers of InGaAsP, with thicknesses (=250 Å) sufficient to ensure isolation of the confinement levels in the individual quantum wells. In lattice matched MSQW structures also a thick reference layer was grown with the same composition as employed in the QWs. As shown in chapter 3, figure 3.11, the solid-vapour distribution of arsenic is much larger than for phosphorus. Consequently, abrupt transitions from phosphorus-containing materials to arsenic-containing materials are more easily achieved than vice versa [43, 44]. The high mobility 2-DEG with two populated subbands indicates the high quality of the lower InP-to-InGaAs(P) interface. Therefore, throughout the growth experiments of the lattice matched \textsubscript{0.58}Ga\textsubscript{0.42}As\textsubscript{0.9}P\textsubscript{0.1}/InP quantum well structures, the gas phase switching procedure at the InP-to-InGaAsP interface (=lower interface) was kept the same to the one used in the samples for the transport measurements (section 4.2.2). The switching at the InGaAsP-to-InP (=upper interface) was varied. To prevent arsenic-drag-over from the quaternary to the InP layer, generally the growth is interrupted to exchange the group V ambient in the reactor cell. Figure 4.8 schematically shows the gas switching procedure used at the InGaAsP-to-InP interfaces. The growth of the InGaAsP is stopped by switching the group III precursors to the vent-line. Arsine is switched to the vent-line one second later, and the InGaAsP surface becomes exposed to phosphine only. The growth of the InP layer is started by switching the TMIn towards the reactor cell after 1, 5, or 15 seconds. For comparison also structures were grown without growth interrupt, by directly switching the gas phase composition from InGaAsP-to-InP.

The photoluminescence measurements were carried out at low temperature (=2-4K) by excitation with 514.5 nm light from an Ar-ion laser. An excitation density of about 5 W/cm\textsuperscript{2} was used. The emission was dispersed with a 0.75 m monochromator with 5 Å resolution and detected by a liquid nitrogen cooled Ge detector. The spectra were recorded using phase-sensitive detection.

In the PLE measurements, the energy of the exiting light from a 250 Watt halogen lamp was scanned using a 0.25 m grating monochromator and a suitable set of filters. The detection was set at the low energy side of the QW luminescence and PLE spectra were corrected for the wavelength dependence of the excitation power.

The PL and PLE spectra were analyzed with the \textbf{k.p} theory which describes the electron and hole confinement states in a unified manner [45].
Figure 4.8. Schematic of the gas phase switching scheme used at the InGaAsP-to-InP interfaces. (Filled bar represents switched towards the reactor cell).

Table 4.1. Parameters used in the k.p theory. Data were taken from [47]; for the quaternary well material the values were interpolated between the two nearest compositions listed. An estimate of the k.p parameter \( P^2 \) for the In\(_{x}\)Ga\(_{1-x}\)As\(_y\)P\(_{1-y}\) alloy has been given by Pearsall [47], based on a linear interpolation between the measured values for InP, InAs and GaAs, and a correction for the alloy disorder effects.

<table>
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<th>Parameter</th>
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<th>In(<em>{0.55})Ga(</em>{0.42})As(<em>{0.9})P(</em>{0.1})</th>
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<td>( \Delta_{\omega} )</td>
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<tr>
<td>( P^2/m_0 )</td>
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<td>22.9 eV</td>
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</table>
Material parameters used for In$_{0.58}$Ga$_{0.42}$As$_{0.9}$P$_{0.1}$/InP quantum wells are listed in table 4.1, and a similar variation of the exciton binding energy with the quantum well thickness as observed in the GaAs/AlGaAs system was assumed [46]. The spin split-off effective mass was calculated from

\[
\frac{1}{m_{so}} = \frac{1}{2} \left( \frac{1}{m_{th}} + \frac{1}{m_{th}'} - \frac{2\Delta_{so} P^2}{3E_g (E_g + \Delta_{so})} \right)
\]  

(4.2)

4.4. Results and Discussion

4.4.1. Photoluminescence Excitation Spectroscopy

In figure 4.9A, the 4K photoluminescence spectra of an In$_{0.58}$Ga$_{0.42}$As$_{0.9}$P$_{0.1}$/InP quantum well (L$_z$=72 Å) and its corresponding bulk reference layer show intense emissions at 0.897 eV and 0.844 eV, respectively, with linewidths of 7.5 and 3.5 meV. The narrow linewidths indicate the high quality of the epitaxial layers. At low excitation densities, the luminescence of the thick reference layer exhibits a low-energy tail due to residual impurities originating from p-type doped materials deposited on the susceptor in previous growth runs. The energy gap $E_g$ of the In$_x$Ga$_{1-x}$As$_y$P$_{1-y}$ was determined from the luminescence peak position of the reference layer. A 4-meV correction for the exciton binding energy was applied, estimated from the effective masses. In the corresponding PLE spectrum of the quantum well, shown in figure 4.9B, peaks due to excitonic transitions associated with the confined electrons, c$_n$, heavy- and light holes, hh$_n$, and lh$_n$, and free holes, fh, are indicated. A Stokes shift of 18 meV is observed between the QW luminescence and the first PLE peak. This indicates an extra binding energy of the excitons probably due to the residual impurities. The arrows in the PLE spectrum mark the calculated energies for the observed transitions using the parameters listed in table 4.1. A best fit was obtained for $\Delta E_g/\Delta E_g=0.35$, and for a well width L$_z$=72±2 Å, in excellent agreement with the value of 71 Å estimated from the growth rate. The accuracy of the value determined for the discontinuity is estimated at 10%. All transitions could be fitted to the observed peaks within 3 meV.

The conduction band offset could be deduced even more directly from the transition observed between the free holes and the confined electrons (fh-c$_1$) in figure 4.9B. The effective masses affect this value only via the electron confinement energy, which is small compared to the energy gap. If the low-energy side
Figure 4.9. A) 4K photoluminescence spectra of 72 Å In_{0.59}Ga_{0.42}As_{0.9}P_{0.1}/InP single quantum well and its 2000 Å thick In_{0.59}Ga_{0.42}As_{0.9}P_{0.1} reference layer; excitation power density about 1W/cm². B) 4K photoluminescence excitation spectrum of a 72 Å In_{0.59}Ga_{0.42}As_{0.9}P_{0.1}/InP single quantum well. The detection was set at 0.89 eV, the excitation power density was about 0.01 W/cm², and the resolution is better than 1 meV. The arrows indicate the calculated transition energies using ΔE_{c}/ΔE_{g}=0.35, the bandgap of the bulk quaternary layer deduced from part A of this figure, and a quantum well thickness of 72 Å.
onset of this transition is used, $\Delta E_r/\Delta E_g=0.37\pm0.01$ is obtained. This is in good agreement with the value deduced in the fitting procedure of the confinement levels. These results were reproduced accurately from a second, separately grown, similar sample.

The conduction band discontinuities deduced in this work are in good agreement with most data reported in the literature as shown in table 4.2. The only exception is the 70:30 band offset ratio reported by Ogura et al. [49].

Westland et al. [52] reported a conduction band discontinuity of $0.4\Delta E_g$ obtained from absorption data of 150 Å $\text{In}_{0.53}\text{Ga}_{0.47}\text{As/InP}$ multiple quantum wells. Their analysis of the spectra, however, was based on a simpler theory. Using our model, we have been able to fit their data with the same accuracy as our own. The peak assignment of Westland and co-workers is in agreement with our calculations, except for the peak assigned as $\text{hh}_2-c_4$, which is within our model assigned to be a $\text{lh}_1-c_3$ transition. In fact our calculations show that the fourth electron level is not confined for a well width less than $\approx200$ Å. The fitted discontinuity is $\Delta E_r/\Delta E_g=0.34\pm0.03$, and the fitted well width of 150±4 Å agrees well with their transmission electron microscopy result of 154±10 Å. Similar absorption data were reported by Skolnick et al. [53]. Their data can only partly be fitted using our model and set of material parameters. This may be due to the fact that their material was not lattice matched, so that strain plays a role, and the material parameters are different.

<table>
<thead>
<tr>
<th>Material</th>
<th>$\Delta E_r/\Delta E_g$</th>
<th>Technique</th>
<th>Reference</th>
</tr>
</thead>
<tbody>
<tr>
<td>$\text{In}<em>{0.53}\text{Ga}</em>{0.47}\text{As/InP}$</td>
<td>0.39</td>
<td>Capacitance-Voltage</td>
<td>Forrest [48]</td>
</tr>
<tr>
<td>$\text{In}<em>{0.53}\text{Ga}</em>{0.47}\text{As/InP}$</td>
<td>0.70</td>
<td>Capacitance-Voltage</td>
<td>Ogura [49]</td>
</tr>
<tr>
<td>$\text{In}<em>{0.53}\text{Ga}</em>{0.47}\text{As/InP}$</td>
<td>0.42</td>
<td>Admittance Spectr.</td>
<td>Lang [50]</td>
</tr>
<tr>
<td>$\text{In}<em>{0.53}\text{Ga}</em>{0.47}\text{As/InP}$</td>
<td>0.38</td>
<td>Photoconductivity</td>
<td>Skolnick [51]</td>
</tr>
<tr>
<td>$\text{In}<em>{0.53}\text{Ga}</em>{0.47}\text{As/InP}$</td>
<td>0.40</td>
<td>Absorption</td>
<td>Westland [52]</td>
</tr>
<tr>
<td>$\text{In}<em>{0.53}\text{Ga}</em>{0.47}\text{As/InP}$</td>
<td>0.40</td>
<td>Absorption</td>
<td>Skolnick [53]</td>
</tr>
<tr>
<td>$\text{In}<em>{0.58}\text{Ga}</em>{0.42}\text{As}<em>{0.9}\text{P}</em>{0.1}/\text{InP}$</td>
<td>0.35-0.37</td>
<td>PLE (This work)</td>
<td>Montie [54]</td>
</tr>
</tbody>
</table>
4.4.2. PL of Lattice Matched InGaAsP/InP Quantum Wells; Effect of Gas Phase Switching

In this section the effect of the growth interruption time (or better the phosphine purge time as indicated in figure 4.8) at the InGaAsP-to-InP interface on the PL spectra is reported. Figure 4.10 shows the 1.5 K PL spectra of the InGaAsP/InP MSQW structures, consisting of six quantum wells and a reference layer, as shown in the inset, grown simultaneously on 0.2° misoriented (spectrum a) and 2° misoriented (spectrum b) (001) InP substrates by directly switching the gas phase composition from the InGaAsP to the InP composition at the upper interfaces (i.e. t=0 in figure 4.8). The reference layer shows the longest wavelength PL emission and the QWs show emission lines at decreasing wavelength with decreasing well width due to the increasing quantization effect. A remarkable difference between the spectra (a) and (b) becomes evident for the thinner wells. Spectrum (b) (structure grown on 2° misoriented substrate) shows single line PL emissions for each well, whereas spectrum (a) (structure grown on 0.2° misoriented substrate) shows single line emissions for the thicker wells and doublet emissions from the narrower (L≤38 Å) wells. The relative intensities of the doublet-peaks were found to depend on the position of measurement on the wafer. The PL-peak assignment was checked by measuring PL spectra from structures in which one of the QWs was left out. When compared to the spectra of figure 4.10, the (multiple-line) PL emission of the QW left out in the growth was indeed absent. From figure 4.10a, it is clear that the splitting of the doublets increases with decreasing well width. Calculations of the energy levels show that the doublet PL-peak emission from one well can neither originate from higher order (n=2, 3, ....) electron to heavy hole transitions because higher order states are cut-off for thin (≤50 Å) wells nor from the split electron-heavy hole and electron-light hole transitions because the observed splittings are too small. On the other hand, the striking difference observed between PL spectra from MSQW structures grown in a same run on 0.2° and 2° misoriented (001) InP substrates could be caused by thickness variations within one QW as shown schematically in figures 4.7C and 4.7D. This can be made plausible as follows: a 0.2° misorientation from the (001) plane corresponds to atomically flat (001) terraces of ≈800 Å in length for monolayer steps (≈2.9 Å in height) or terraces of ≈400 Å in length for half-monolayer steps. Lateral displacements of the steps at the upper and lower interfaces of the QW over distances larger than the excitonic diameter (≈200 Å) give the possibility to distinguish between different thicknesses within one QW. The narrow linewidths of the peaks in the doublets of figure 4.10a indicate the flat interfaces. On the contrary, the terraces on 2° misoriented substrates are ≈80 Å for monolayer steps or ≈40 Å for half-monolayer steps.
Characterization of InGaAs(P)/In(GaAs)P interfaces

Figure 4.10. 1.5K photoluminescence spectra of In\textsubscript{0.38}Ga\textsubscript{0.42}As\textsubscript{0.5}P\textsubscript{0.5}/InP quantum well structures as shown in the inset which were grown simultaneously on 0.2° (a) and 2° (b) misoriented (001) InP substrates. The InGaAsP-to-InP interfaces were continuously grown. The PL spectrum of the structure grown on 0.2° misoriented substrate shows doublet emissions corresponding to monolayer (=3Å) well width variations over areas larger than the excitonic diameter. Each quantum well grown 2° misoriented substrates exhibits a single line emission because the lateral dimensions of the thickness variations within one well are smaller than the excitonic diameter.

This is clearly smaller than the excitonic diameter and will result in single peak PL emission with larger linewidths for thin (≤50 Å) QWs. The observed linewidths of the PL emission support this hypothesis. As shown in figure 4.10, the linewidths of the PL emissions from QWs grown on the 2° misoriented substrates are larger
compared to the linewidths of the PL emission from the corresponding QWs grown on 0.2° misoriented substrates, except for the 58Å well where the PL linewidth on the 0.2° misoriented substrate may be broadened due to non resolved splitted peaks. The QW-widths calculated from the energy shifts in figure 4.10a, using the k.p theory and the band offset deduced in section 4.4.1, are shown in the inset. For the doublet emissions the given thicknesses correspond to the short emission wavelength. All energy splittings correspond indeed with 3 Å, i.e. one monolayer, well width variations.

PL spectra of similar MSQW structures grown with a 1 second phosphine purge, i.e. t=2 seconds in figure 4.8, at the upper InGaAsP-to-InP interface essentially showed similar behaviour as described above.

Figures 4.11a and b show PL spectra of InGaAsP/InP MSQW structures grown on 0.2° and 2° misoriented (001) InP substrates, respectively, with a phosphine purge time of 5 seconds (t=6 seconds in figure 4.8) at the InGaAsP-to-InP interfaces. The main difference between both PL spectra is again the multiple-line emission from the narrower QWs grown on 0.2° misoriented (001) InP substrate and single line emissions from the QWs grown on 2° misoriented substrates. The QW widths estimated from the emission wavelength shifts are indicated in the inset; for the multiple-line emissions the given QW widths correspond to the shortest wavelength PL-peak. Figure 4.11a shows an intense single peak at 1447 nm due to excitonic transitions with a linewidth of 3.8 meV from the reference layer. For the 6 Å well the emission peak shifted from 1447 nm to 942 nm, corresponding to an energy upshift of 459 meV which represents 81% of the difference in bandgaps between the quantum well and InP cladding layer. The 6 Å well of InGaAsP shows a linewidth of only 8.8 meV which is among the smallest values ever reported for InGaAsP of this thickness grown by any epitaxial technique (see figure 4.13). In this study a maximum wavelength shift from 1449 nm for the bulk InGaAsP layer down to 905 nm (ΔE=514 meV=90% of the bandgap difference) was observed from an InGaAsP well of about 4 Å in thickness grown on a 0.2° misoriented (001) InP substrate. The PL emissions of the 9, 14 and 32 Å QWs show resolved double line spectra, whereas the emission of the 22 Å QW shows a triple line spectrum. In this case, according the k.p calculation, the double and triple emission lines correspond to a difference in QW thickness of 1.2-1.6 Å. This equals to about a half-monolayer distance, i.e. the neighbouring group III to group V atom planar distance in the [001] direction. These observations are consistent with data reported by Wang et al. [5] for single QWs of InGaAs/InP grown by AP-OMVPE with a similar gas phase switching procedure but a phosphine purge time of 2 min. As shown in figure 2 of ref. 5, the calculated splitting for one monolayer thickness fluctuation is twice the
Figure 4.11. 2K photoluminescence spectra of In$_{0.58}$Ga$_{0.42}$As$_{0.9}$P$_{0.1}$/InP quantum well structures as shown in the inset grown on 0.2° (a) and 2° (b) misoriented (001) InP substrates, respectively, using a 5-second phosphine purge at the InGaAsP-to-InP interfaces. The multiplet emissions from the 9, 14, 22, and 32 Å wells of spectrum (a) correspond to half monolayer (=1.5 Å) well width variations over areas larger than the excitonic
diameter. Each well in spectrum (b) again exhibits a single line photoluminescence emission because the lateral sizes of the thickness variations within one narrow well are smaller than the excitonic diameter.

observed splitting. In InGaAsP/InP heterostructures four different atomic planes in the [001] direction can be distinguished. These are the planes with In and P atoms for InP and the planes with mixed In+Ga and As+P atoms for the InGaAsP, respectively. If the InGaAsP-InP interfaces are abrupt it would be possible to obtain QWs differing by half-monolayers. This could be achieved if atomic planes are not completely filled when the gas phase is switched. However, due to the high group V element over-pressure, a partial exchange of the group V element of the surface layer during the growth pause is more likely which would result in our case in the substitution of arsenic by phosphorus. It will be unlikely that half-monolayer well width variations will be the cause of the peak splitting as will be shown further on in this chapter.

The PL spectrum of the corresponding MSQW structure grown on the $2^\circ$ misoriented (001) InP substrate (figure 4.11b), using the 5-second phosphine purge at the InGaAsP-to-InP interface, showed sharp single peak emissions for each QW. In this spectrum, a wavelength shift from 1463 nm to 922 nm corresponding to an energy shift of 497 meV ($=0.87\Delta E_g$) can be observed. The linewidth increases from 5.2 meV for the reference layer to about 16.7 meV for the 12, 7, and 5 Å wide QWs which is larger than observed in spectrum of the structure grown on $0.2^\circ$ misoriented substrate.

By increasing the phosphine purge time at the InGaAsP-to-InP interface to 15 seconds, the PL spectra of MSQW structures grown on $2^\circ$ off (001) InP substrates again show only single line emissions for each quantum well. The linewidths of the PL-peaks are significantly broader than of the corresponding MSQW structure grown with the 5-second purge time. MSQW structures grown on $0.2^\circ$ misoriented substrates all showed multiple-line emissions from the narrow ($\leq 40$Å) wells as shown in figure 4.12. This spectrum shows up to four resolved emissions in the multiplets from a single narrow QW. The energy splittings in the multiplets were deduced to result from 1.5 Å, i.e. half monolayer or single atomic layer thickness fluctuations in the well width. The well widths shown in the inset of figure 4.12 correspond with the shortest wavelength emission of the multiplets. The linewidths observed in this study for the phosphine purge times $\leq 5$ seconds at the InGaAsP-to-InP interfaces compare very favourably with the best data reported in the literature as shown in figure 4.13, and demonstrate the high quality interfaces achievable by LP-OMVPE. For narrow wells the smallest PL linewidths are obtained on $0.2^\circ$ misoriented (001) InP substrates.
Figure 4.12. 1.5K photoluminescence spectrum of an In$_{0.58}$Ga$_{0.42}$As$_{0.9}$P$_{0.1}$/InP quantum well structure as shown in the inset grown on a 0.2° misoriented (001) InP substrate using a 15-second phosphine purge at the InGaAsP-to-InP interfaces. Up to four resolved emission lines are observed near 1150 nm wavelength. All splittings in this spectrum were deduced to correspond to half-monolayers (=1.5 Å) well width variations.

The PL spectra of thin In$_{0.58}$Ga$_{0.42}$As$_{0.9}$P$_{0.1}$/InP QWs consistently show single line emissions for QWs grown on 2° misoriented substrates, whereas multiple-line PL emissions were observed from the QWs grown on 0.2° misoriented substrates for all growth interruption times used. Similar observations were reported by other researchers [5, 55-58].

The inter-well-width variations deduced from the energy splittings decrease stepwise for increasing phosphorus purge time at the InGaAsP-to-InP interfaces. Splittings were observed which correspond to monolayer (=3 Å) thickness variations for interfaces grown with ≤1 second phosphine purge time, whereas for phosphine purge times larger than 5 seconds the energy splittings correspond to half-monolayer (=1.5 Å) well width variations. Furthermore, for increasing phosphine purge times the number of emissions in the multiplets from one well increases, and also the linewidths of the multiplet peaks and of the single peak emissions from QWs grown on 2° misoriented substrates clearly increase. The increased number of peaks indicates additional trapping of the excitons due to either thickness or lateral compositional variations. This may result from the modification of the InGaAsP surface under the phosphine ambient during the growth pause where
Figure 4.13. Photoluminescence FWHM linewidths at 2-10K versus InGaAs(P)/In(GaAs)P quantum well widths grown by the techniques given within brackets on (001) ± 0.2° InP substrates unless indicated otherwise. The broadening due to monolayer well width variations and the alloy ($\text{In}_{0.53}\text{Ga}_{0.47}\text{As}$) broadening [66] are indicated by the dashed and the solid lines, respectively. InGaAsP ($\lambda = 1.55$ and $1.3$ μm)/InP SQW: (●) this work [62], (LP-OMVPE); (■) this work, 2° misorientation [62], (LP-OMVPE); (▲) Panish et al. [11], (GSMBE); (∗) Razeghi et al. [63], (LP-OMVPE); (⊗) Tsang et al. [63], (CBE); InGaAsP/InP MQW: (●) this work [65], (LP-OMVPE); (■) Panish et al. [11], (GSMBE); InGaAs/InP SQW: (○) Tsang et al. [12], (CBE); (∨) Wang et al. [5], (AP-OMVPE); (∆) Kamei et al., [55], (LP-OMVPE); InGaAs/InGaAsP SQW: (♦) this work [33], (LP-OMVPE); (◇) this work, 2° misoriented substrate [33], (LP-OMVPE).

the substitution of arsenic by phosphorus atoms, ultimately resulting in the formation of lattice mismatched In$_{0.58}$Ga$_{0.48}$P, is most likely. Similarly, at the lower InP-to-InGaAsP interface the InP surface may be converted to InAs during an arsine purge. However, the formation of flat layers of InAs at the lower InP-to-InGaAsP interfaces and flat layers of InGaP at the upper InGaAsP-to-InP interfaces do not explain the differences in the PL spectra observed here for the increased phosphine purge times. The transition layers may effectively modify the well width (InAs increases, and InGaP decreases the well width) resulting in a shift of the emission wavelength. The increased number of emissions in the multiplets and the broadening of the PL lines could only be explained by island-wise exchange of the group V element which becomes unlikely, however, to explain the quartet
emissions as observed in figure 4.12. Inferred from PL data, the phenomenon of (partial) group V exchange at more than one atomic plane has been previously proposed for atmospheric pressure [59] and low pressure [55, 56] OMVPE growth. Similar substitutions were suggested by Meijer et al. [60] based on X-ray measurements. Using the same technique, significant substitution at the interfaces was also detected in GSMBE grown InGaAs/InP quantum wells [61].

Another observation generally made for InGaAs(P)/InP interfaces, namely that in cross-sectional transmission electron micrographs the InGaAs(P)-to-InP interfaces show undulations, whereas the InP-to-InGaAs(P) interfaces are flat [43, 44] does not fit into the above picture. These effects are clearly shown in figure 4.14. TEM micrograph (a) shows abrupt and flat interfaces at both sides of the InGaAsP quantum well where the lower interface was grown using a 5-second arsine purge time and at the upper interface the gas phase composition was directly

![Image of TEM micrograph](image)

**Figure 4.14.** Cross-sectional transmission electron micrographs of In$_{0.55}$Ga$_{0.42}$As$_{0.9}$P$_{0.1}$/InP multiple single quantum well structures with the InGaAsP-to-InP interfaces grown continuously (a), and grown using a 5-second phosphine purge (b). In (a) the upper interfaces are clearly flatter than in (b). (For both structures the lower InP-to-InGaAsP interfaces were grown using a 5-second arsine purge).
switched from InGaAsP to InP. In TEM micrograph (b), the lower interface was
grown again with 5 seconds arsine purge, but at the upper interface a 5-second
phosphine purge was used resulting in more undulated interfaces most obvious
for the narrower wells. Even larger undulations at the upper interfaces were
observed for the 15-second phosphine purge time. This indicates that during the
phosphine purge the surface becomes undulated. This purge should thus be
minimized to obtain the highest quality interfaces as shown in figure 4.14, and
can be inferred from the PL linewidths shown in this section. The swelling
undulation with increasing phosphine purge time at the upper InGaAsP-to-InP
interface is consistent with the growing number of emission lines in the multiplets.

4.4.3. PL of Lattice Matched In$_{0.53}$Ga$_{0.47}$As/InGaAsP Quantum Wells

For applications in semiconductor lasers, active layers consisting of
InGaAs(P) quantum wells embedded within quaternary InGaAsP barrier- and
separate confinement layers are desired for enhancement of the optical confinement
factor. Lattice matched In$_{0.53}$Ga$_{0.47}$As MSQW structures embedded within 300 Å
thick In$_{0.77}$Ga$_{0.23}$As$_{0.5}$P$_{0.5}$ (λ$_e$=1.2 μm) barrier layers were grown on 0.2° and 2°
misoriented (001) InP substrates in the same growth run. At the InGaAs-to-
InGaAsP interfaces a 5-second purge with mixed arsine+phosphine at a
concentration appropriate for the growth of the lattice matched In$_{0.77}$Ga$_{0.23}$As$_{0.5}$P$_{0.5}$
barrier layers was used. The grown structure is shown schematically in the inset
of figure 4.15. An InP capping layer of about 1000 Å thickness was grown on
top of the structure to reduce the surface recombination. In the 2K PL spectra
similar characteristics are observed as in the PL spectra shown for InGaAsP/InP
MSQW structures, i.e. thin QWs of In$_{0.53}$Ga$_{0.47}$As grown on 2° misoriented
substrates show single line PL emissions for each well, whereas for the
corresponding structures grown on 0.2° misoriented substrates doublet emissions
for thin QWs are observed as shown in figure 4.15. The In$_{0.53}$Ga$_{0.47}$As QW-widths
were deduced from the quantum shifts using the same band offsets as obtained
for In$_{0.58}$Ga$_{0.42}$As$_{0.9}$P$_{0.1}$/InP heterojunctions. In the inset, for the multiple-line
emissions again the widths corresponding with the shortest emission wavelengths
are given. The exact lineshape of the 35 Å QW is obscured due to the water
absorption lines in the quartz window of the Ge detector. However, the emissions
of the 19 and 9 Å QWs are clearly splitted in narrow double lines, whereas the
thinnest QW of 6 Å shows a single line emission. The peak splittings correspond
to 3.7 Å (i.e. about one monolayer) well width variations for a 5-second mixed
phosphine and arsine purge at appropriate concentrations required for the growth
of 1.2 μm bandgap wavelength InGaAsP. The high quality of the InGaAs/InGaAsP
Figure 4.15 2K photoluminescence spectrum of In$_{0.55}$Ga$_{0.47}$As/InGaAsP quantum well structure as shown in the inset grown at 625°C on 0.2° misoriented (001) InP substrate using a phosphine purge time of 5 seconds at the ternary-to-quaternary interfaces. The 9 and 19 Å quantum wells show doublet emissions corresponding to monolayer well width variations. The exact shape of the emission from the 35 Å wide well is obscured by absorption due to water in the quartz window of the detector.

QWs is indicated by the narrow emission linewidths of about 12.5 meV for the thinnest wells. Moreover, the recombination of carriers in the In$_{0.55}$Ga$_{0.47}$As QWs is very effective also in this structure, as indicated by the absence of emission from the barrier layers.

During the 5-second growth pause at the In$_{0.55}$Ga$_{0.47}$As-to-In$_{0.77}$Ga$_{0.23}$As$_{0.5}$P$_{0.5}$ interfaces, the ternary surface is exposed to an arsine+phosphine mixture appropriate for the growth of 1.2 μm bandgap wavelength InGaAsP. This variation in the gas phase composition is less abrupt than during the growth of InGaAsP/InP structures. Consequently, less exchange at the surfaces is expected, resulting in less undulation for the ternary-to-quaternary interfaces. This is indeed confirmed by the fact that for the same duration of the growth pause, monolayer well width variations are observed for ternary-to-quaternary interfaces, whereas quaternary-to-binary interfaces show half-monolayer well width variations. Our nonintegral value of about 0.5 monolayers in InGaAsP/InP is therefore most likely due to the continuous distribution function of thicknesses which arises from the sampling
of a large number of small flat islands.

For good control of the emission wavelength, the thickness of the quantum wells has to be controlled accurately. Figure 4.16 shows the well thickness deduced from the PL-spectra versus their growth times for In$_{0.58}$Ga$_{0.42}$As$_{0.9}$P$_{0.1}$/InP and In$_{0.55}$Ga$_{0.47}$As/In$_{0.77}$Ga$_{0.23}$As$_{0.5}$P$_{0.5}$ ($\lambda=1.2~\mu$m) quantum wells. For different growth times and different purge times a good linear relation between the well thickness and the growth time were obtained, indicating a constant growth rate. The observation that these lines do not extrapolate to zero thickness for zero growth time might be explained by the group V exchange at the interfaces. Due to the non lattice matched layers at the interface, strain-effects on the bandgap and on the band offset should be taken into account in the k.p calculation (section 4.5). The fact that this thickness offset also depends on the growth rate indicates that it may be partially related to the growth apparatus as well.

![Graph showing well widths versus growth time](image)

**Figure 4.16.** Well widths versus growth time. (○) InGaAs/InGaAsP QWs; (●, ∇, and ▲) InGaAsP/InP QWs with phosphine purge times of 0, 5 and 15 seconds, respectively.

### 4.4.4. PL of Strained-Layer In$_x$Ga$_{1-x}$As/In(GaAs)P Quantum Wells

In this section, 1.5K and 300K photoluminescence and TEM studies on strained-layer In$_x$Ga$_{1-x}$As ($x=0.7$ and 0.8, resulting in 1.2 and 1.8% compressive
strain, respectively)/In(GaAs)P QW structures are reported. The effects of the strain and the substrate misorientation from the (001) plane on the interface quality are studied.

In the 1.5K PL spectra of multiple single quantum well structures of 1.8% biaxially compressed InGaP grown on 0.2° misoriented (001) InP substrates typically the number of narrow PL emission lines is larger than the number of quantum wells grown. In this case, the thick reference layer cannot be used because the estimated critical thickness is around 200 and 100 Å for InAs mole fractions of 0.7 and 0.8, respectively [67]. Figure 4.17 shows a PL spectrum with eleven emissions from six 1.8% biaxially compressed InGaP quantum wells with nominal well widths of 6, 9, 12, 18, 24, and 33 Å, as measured by cross-sectional TEM, embedded within InP. The accuracy of TEM is 3 Å corresponding to one monolayer. The well widths deduced from the PL emissions in the spectrum using the k·p theory, and taking the strain effects into account [68] by adding the appropriate terms to the Kane Hamiltonian were: 6.5, 9, 11.5, 14, 17, 20, 23, 26, 29, 30.5 and 32 Å, respectively. These data indicated in figure 4.17 on top of

![Figure 4.17](image-url)

Figure 4.17. 1.5 K photoluminescence spectrum of six 1.8% biaxially compressed InGaP quantum wells grown on 0.2° misoriented (001) InP substrate showing eleven emissions. The numbers on top of the peaks indicate the well widths.
the PL-peaks agree excellently with the thicknesses obtained from TEM. For the narrower wells the PL-peak splittings correspond to monolayer well width variations, whereas for the thickest well half-monolayer well width variations were obtained from the peak splitting. In figure 4.18 the interface abruptness and the integrity of the narrowest quantum wells in this structure, 6 and 12 Å in thickness, are shown in a high resolution cross-sectional TEM micrograph. The FWHM of low temperature PL emissions of In$_{0.8}$Ga$_{0.2}$As/InP quantum wells of this work are plotted together with literature data [70, 71] versus the quantum well thickness in figure 4.19. These data clearly indicate the high quality 1.8% compressively strained QWs grown in this work. According LeCorre et al. [70], the increased linewidths for well thicknesses larger than 95 Å can be attributed to the relaxation of the strained QW lattice. This is in excellent agreement with the critical thickness deduced according Matthews and Blakeslee [67] shown in figure 2.8.
Figure 4.19. 2K PL linewidths of 1.8% compressively strained In$_{0.8}$Ga$_{0.2}$As/InP quantum wells versus the quantum well thickness. (●) this work [69]; (○) LeCorre et al. [70]; and (▼) Uchida et al. [71].

The 1.8% compressively strained In$_{0.8}$Ga$_{0.2}$As/InP MSQW structures were also grown on (001) InP substrates off-oriented by 0°, 0.5°, 1.0°, 1.5°, 2.0°, 5.0°, and 7.5°, respectively. The MSQW structures on the exact (001), 0.5° and 2.0° and on the 0.2°, 5.0°, and 7.0° misoriented substrates were grown in two different growth runs; the low temperature PL spectra are shown in figure 4.20. In the latter run, only the thickest quantum well was grown with slightly reduced TMGa fraction in the vapour phase as reflected by the longer wavelength PL emission. The strained-layer MSQW structures grown on the exact and 0.2° misoriented (001) InP substrates show PL spectra with many well resolved narrow linewidth PL emissions similar to the PL spectrum shown in figure 4.17. The PL spectrum of the strained MSQW structure grown on the 0.5° misoriented substrate shows some broadening of the emission lines, which is more pronounced for longer wavelength emissions from the thicker wells. This effect propagates to the shorter wavelength emissions, i.e. to thinner wells, as the misorientation increases to 1.0°, 1.5° (not shown) to 2°, where the 1250-1500 nm wavelength window peaks cannot be discriminated anymore. The slight undulation in the PL spectra near 1400 nm wavelength is due to the absorptions from water in the quartz ware in the optical path of the PL-equipment. The PL spectra of the strained-layer MSQW structures grown on 5 and 7.5° misoriented substrates show a long wavelength peak and
Figure 4.20. 1.5K photoluminescence spectra of 1.8% biaxially compressed In$_{0.8}$Ga$_{0.2}$As/InP QW structures, as shown in the inset, grown on InP substrates with misorientations from the (001) plane as indicated. Defects-free QW structures grown on nominally (001) and 0.2$^\circ$ misoriented substrate showed narrow, resolved PL emissions corresponding to inter-well-width variations of one monolayer. For larger substrate misorientations the PL emissions are broadened due to the formation of defects and undulated interfaces.

a continuous emission band at shorter wavelength. Similar effects were observed for 3-well In$_{0.8}$Ga$_{0.2}$As/InP structures, where the thickest well was only 24 Å and the total thickness of the QWs was reduced by a factor of two compared to the MSQW structure shown in figure 4.20. The differences in the PL spectra have to be ascribed to the different substrate orientations, apparently resulting in different structural perfection of the quantum wells. This behaviour is in sharp contrast with lattice matched InGaAsP/InP QWs which showed narrow line PL spectra
for misorientations up to 7.5° from the (001) plane; the maximum misorientation used.

The In₀.₈Ga₀.₂As/InP MSQW structures grown on 2° misoriented substrates were analyzed in cross section by TEM. These structures were found to contain a high density of dislocation clusters and showed wavy interfaces in sharp contrast to the TEM picture shown in figure 4.18. The thicker wells varied more than a factor of two in thickness. This explains the broad PL emissions as shown in figure 4.20. These observations suggest, in this case of the growth of 1.8% biaxially compressed In₀.₈Ga₀.₂As/InP QWs, 2D growth on less than 0.7° misoriented substrates and 3D-like growth on substrates with larger misorientations. The deterioration of the PL spectra starting from the long wavelength side indicates that the onset of 2D-like growth shifts to narrower well widths with increasing substrate misorientations from the (001) plane. Similar observations were reported by Grodzinsky et al. [72] for the growth of strained In₀.₁₈Ga₀.₈₂As on non-planar GaAs substrates. No attempts were undertaken in our study by optimizing the growth conditions, e.g. growth temperature or V/III ratio, to suppress the 3-D growth.

For the fabrication of long wavelength strained-layer multiple quantum well laser diodes, InGaAsP rather than InP barrier layers are preferred, as the former maximize the optical confinement factor. The 1.5K PL spectra from a single quantum well and a four quantum well structure of 1.8% biaxially compressed In₀.₈Ga₀.₂As, of about 30 Å in thickness, embedded within lattice matched InGaAsP (λₙ=1.3 μm), are shown in figure 4.21. The short wavelength emissions result from the InGaAsP barrier layers, and the much larger intensity multiple PL emissions in curves A and B result from the SQW and 4 QW, respectively. Both structures emit at the same wavelength, demonstrating the excellent control of the well width and composition. The multiple-lines observed in these strained ternary-quaternary heterostructures indicate that also for these structures flat interfaces over areas larger than the excitonic diameter are feasible.

The effect of the strain on the quality of InₓGa₁₋ₓAs/InGaAsP interfaces for growth on misoriented substrates was studied in greater detail [73]. InₓGa₁₋ₓAs (x=0.7, corresponding to 1.2% compressive strain, and x=0.53 corresponding to zero strain, as a reference) QWs were grown on (001), and the A and B faces of (111), (311), and (511) oriented InP substrates. These substrate orientations correspond to 54°, 25°, and 15° misorientation from the (001) plane, respectively. The substrate surface of (N11) oriented substrates with N>1 can be described as a stepped surface comprising (001) treads and (111) risers [68]. The surface fraction of (111) risers increases with decreasing N, and consist of (111)A planes, i.e. planes with the group III atoms at the surface, for (N11)A oriented substrates, whereas for the (N11)B substrates the risers are (111)B planes with the group V atoms
Figure 4.21. 1.5K photoluminescence spectra of a single quantum well (A) and a multiple quantum well (B) of 30 Å wide wells of In$_{0.8}$Ga$_{0.2}$As separated by InGaAsP, grown on 0.2° off (001) InP substrates. The multiple-line emissions from the single quantum well structure indicate monolayer well width variations with flat interfaces over areas larger than the excitonic diameter.

at the surface. MQW structures shown schematically in figure 4.22 were grown in this study. For strained as well as unstrained MQW structures grown on (111)A and B, and (311)A substrates no good morphology nor good PL data were obtained. Similar results were reported for the growth of In$_{0.55}$Ga$_{0.47}$As/InP heterostructures on (111) faces on profiled substrates [75]. The In$_{0.55}$Ga$_{0.47}$As MQW structures grown on (001) and (511)B substrates emit at 1590 nm and 1596 nm, respectively, whereas the structures grown on (311)B and (511)A emit at 1585 nm. A similar but more pronounced wavelength variation was observed for the strained In$_{0.7}$Ga$_{0.3}$As MQW structures, where the (001) and (511)B emit at 1580 nm and the (311)B and (511)A emit at 1567 nm. At the moment it is not clear whether these variations in emission wavelengths reflect modifications in hole effective mass, resulting in different confinement energies, or whether they have to be attributed to variations in the growth rate, composition or both. Figure 4.23 shows the PL linewidth (full width at half maximum) and the PL intensity versus the misorientation from the (001) plane for the strained and the unstrained MQW
Figure 4.22. Schematic of In$_{x}$Ga$_{1-x}$As/InGaAsP MQW structure.

Figure 4.23. 300 K photoluminescence linewidths and intensities of strained-layer (SL) In$_{0.7}$Ga$_{0.3}$As/InGaAsP and lattice matched (LM) In$_{0.53}$Ga$_{0.47}$As/InGaAsP MQW structures versus the misorientations of the substrates from the (001) plane. Misorientations towards (111)B are assigned positive and misorientations towards (111)A negative ((311) = 25° misorientation, (511) = 15° misorientation from (001)).

structures. Except for the strained MQW structure grown (511)A, the 300K PL linewidths and intensities are not significantly affected by the substrate orientation. The linewidths are about 50 meV for the unstrained and 36 meV for the strained MQW structures. Reducing the optical excitation density (by a factor of 5) reduced the PL linewidth to a value as low as 29.5 meV for the strained and 45 meV for the unstrained MQW structures. The PL emission of the strained MQW structures is more intense than of the unstrained MQW structure. The PL measurements
clearly indicate that high luminescence, device quality strained and unstrained In\textsubscript{x}Ga\textsubscript{1-x}As (x≤0.7)/InGaAsP MQW structures can be grown on (511)A and B, and (311)B InP substrates which have misorientations of 15° and 25° towards the (001) plane.

4.5. Layers at the Interfaces Studied by High Resolution X-ray Diffraction on Superlattices

In section 4.4.2., low-temperature PL spectra and high resolution transmission electron microscopy indicated an exchange of group V atoms at the In\textsubscript{x}Ga\textsubscript{1-x}As\textsubscript{y}P\textsubscript{1-y} surface during growth pauses, introduced to purge the reactor-cell with the group V hydride(s) before starting the growth of the subsequent layer. The exchange of the group V atoms induces a variation of the lattice parameter. The accumulated thickness of many identical periods in a superlattice rather than just a single quantum well provides the opportunity for characterization techniques such as high resolution X-ray diffraction (HR-XRD) to elucidate compositional variations of these "transition layers" at the interfaces.

In order to test the above hypothesis, 12-period InGaAs/InP superlattice structures, grown using different switching procedures of the gas phase, were measured by HR-XRD. For a fixed switching sequence at the lower InP-to-InGaAs interfaces using a one second arsine purge, the second step in the gas phase switching at the upper InGaAs-to-InP interfaces, i.e. the phosphine purge time effectively, was increased from 1 to 5, and finally to 15 seconds. Equivalently, for a fixed phosphine purge time of 1 second at the InGaAs-to-InP interfaces, superlattice structures were grown using arsine purge times at the InP-to-InGaAs interfaces of 1, 5, and 15 seconds.

The repetitive period, P, consisting of one well and one barrier layer including the interface layers, and the averaged lattice parameter, a\textsubscript{0}, in the period for a 004 reflection are given by

\[
\sin \theta_j = \frac{\lambda}{2} \left( \frac{4}{a_0} + \frac{j}{P} \right)
\]  

(4.3)

where \( \theta_j \) and \( j \) are the diffraction angle and the order of the satellite peak, respectively, and \( \lambda \) is the X-ray wavelength. The averaged lattice parameter \( a_0 \) determines the position of the zeroth order (\( j=0 \)) peak. The number of monolayers in one period is given by \( N=2P/a_0 \) with \( a_0 \) given by
Figure 4.24. 004 HR-XRD spectrum of 12-period InGaAs/InP superlattice grown with arsine and phosphine purge times of 1 second at the InP-to-InGaAs and InGaAs-to-InP interfaces, respectively. The presence of the high number of satellite peaks up to over the 30th order, together with the additional periodicity of the envelope, indicate the excellent periodicity and the abrupt interfaces.

\[ a_0 = \frac{n_w a_w + n_b a_b}{n_w + n_b} \]  \hspace{1cm} (4.4)

where \( n_w \) and \( n_b \) are the numbers of monolayers, and \( a_w \) and \( a_b \) are the lattice constants in the well- and barrier layers, respectively. Figure 4.24 shows as an example a (004) HR-XRD satellite spectrum of a 12-period In\(_{x}\)Ga\(_{1-x}\)As/InP superlattice grown with an optimized gas switching scheme at the interfaces. Satellite peaks up to over the 30th order confirm the excellent periodicity in this structure. The additional periodicity of the envelope, induced by the difference
in the InP and InGaAs structure factors, indicates a well to barrier thickness ratio of about 1:4. This spectrum, as all others in this work, was recorded using a computer controller HR-XRD diffractometer (Philips HR-1) using CuKα1 radiation monochromated by a Bartels monochromator [76]. The presence of interface layers with different compositions, one at the InP-to-InGaAs(P) interface of \( n_{b\rightarrow w} \) monolayers with a lattice spacing \( a_{b\rightarrow w} \), and one at the InGaAs-to-InP interface of \( n_{w\rightarrow b} \) with \( a_{w\rightarrow b} \), can be implemented in eq. (4.4) resulting in

![Graph showing diffraction spectra](image)

**Figure 4.25.** 004 HR-XRD diffraction spectra of 12-period InGaAs/InP superlattices grown with an arsine purge time of 1, 5 and 15 seconds at the InP-to-InGaAs interfaces, spectra (a), (b), and (c), respectively. The phosphine purge time at the InGaAs-to-InP interfaces was fixed at 1 second. With increasing arsine purge time the diffraction peaks shift to negative Θ-values indicating the formation of a compressively strained interface layer with increasing thickness, due to increased arsenic substitution.
\[ a_0 = \frac{n_{b \rightarrow w} a_{b \rightarrow w} + n_w a_w + n_{w \rightarrow b} a_{w \rightarrow b} + n_b a_b}{n_{b \rightarrow w} + n_w + n_{w \rightarrow b} + n_b} \]  

(4.5)

The formation of interface layers may shift the position of the satellite peaks, whereas provided that during the switching no material evaporates nor is deposited, the formation of interface layers is not expected to change the period of the

Figure 4.26. 004 HR-XRD diffraction spectra of 12-period InGaAs/InP superlattices grown with phosphine purge times of 1, 5 and 15 seconds at the InGaAs-to-InP interfaces, spectra (a), (b), and (c), respectively. The arsine purge time at the InP-to-InGaAs interfaces was fixed at 1 second. With increasing phosphine purge time the diffraction peaks shift to positive Θ-values indicating the formation of a tensile strained interface layer with increasing thickness, due to increased phosphorus substitution.
superlattice.

The HR-XRD spectra depicted in figures 4.25 and 4.26 show the influence of the increased arsine purge time at the lower interfaces, and the increased phosphine purge time at the upper interfaces, respectively. In the spectra, the substrate diffraction peaks are aligned at $\Delta \Theta = 0$, and are labelled with "substrate." The order of the satellite peaks is indicated by the numbers on top. The zeroth order peak was determined without ambiguity by measuring spectra using various reflections [77]. Figure 4.25 clearly shows a shift of the zeroth order satellite peak to negative $\Delta \Theta$-values (that is to smaller diffraction angles) which according eq. (4.3) implies an increase of the averaged lattice spacing. On the other hand, the spectra in figure 4.26 where the phosphine purge time was varied, show a shift of the zeroth order satellite peak to larger diffraction angle $\Theta$ which implies a reduction of the averaged lattice parameter. For all spectra in figures 4.25 and 4.26, the spacings between the subsequent orders of the satellite peaks are the same within the error of the determination of the peak position ($\pm 5$ arc sec, corresponding to $\pm 2$ Å). A constant averaged satellite period confirms the independency of the superlattice period on the switching procedure. This demonstrates the excellent control of the layer thickness, down to monolayer level, in the LP-OMVPE growth. The spacing of the satellite peaks corresponds to a 271 Å period, which consists of 213 Å InP and 58 Å InGaAs.

The shifts of the zeroth order satellite peaks are attributed to the formation of surface layers during the purge time with the other group V hydride during the gas switching. When the InP surface is exposed to arsine, the top-phosphorus-atoms are exchanged by arsenic resulting in the formation of InAs. The shift of the zeroth order superlattice peak for $t_{As} = 5$ seconds in respect to the $t_{As} = 1$ second corresponds with the formation of about 1.5 monolayer of InAs on InP. For the longer purge time $t_{As} = 15$ seconds, the shift corresponds to an additional 1.2 monolayer InAs. The shifts of the zeroth order satellite peaks with increasing phosphine purge time from 1 to 5, and from 5 to 15 seconds correspond to the complete exchange of As to P in the first, and in an additional 1.5 monolayer of the InGaAs-layers, respectively. We have assumed a complete group V substitution in the surface layers; for partial substitutions, the thicknesses over which substitutions take place increase correspondingly.

The HR-XRD measurements demonstrate the formation of highly strained interfacial layers during the growth pauses, which should be considered in $k.p$ calculations for an accurate estimate of the QW thickness. In addition, the material formed at the interface may affect the band offsets.
4.6. Conclusions

In summary, LP-OMVPE grown unstrained and strained-layer InGaAs(P)/In(GaAs)P interfaces were characterized by transport measurements, by photoluminescence excitation spectroscopy (PLE) and photoluminescence (PL) measurements, by transmission electron microscopy (TEM), and by high resolution X-ray diffraction (HR-XRD) measurements. During growth pauses at the interfaces, introduced to purge the reactor-cell with the new group V hydride required for the growth of the subsequent layer, exchange of the group V element occurs in the top layer(s) of the crystal. In case of the presence of gallium in the crystal, some slight roughening of the surface is indicated by the PL spectra and by TEM measurements. Therefore, InP-to-InGaAs(P) interfaces are less affected by the growth pauses than InGaAs(P)-to-InP interfaces.

Using optimized gas switching schemes, abrupt and flat unstrained and strained-layer InGaAs(P)/In(GaAs)P interfaces were demonstrated in our zero-dead volume, very abruptly switching manifold LP-OMVPE reactor. We recall the high 4K mobility two dimensional electron gas with two populated subbands at InP-to-InGaAs(P) heterojunctions showing $\mu=171,000$ cm$^2$/Vs at InP-to-In$_{0.55}$Ga$_{0.45}$As, $\mu=86,000$ and $\mu=39,500$ cm$^2$/Vs at InP-to-InGaAsP ($\lambda_{s}=1.55$ and 1.30 $\mu$m), respectively. Furthermore, the excellent interface quality, also at the InGaAs(P)-to-In(GaAs)P interfaces, is demonstrated by low-temperature PL-spectra measured on structures grown on substrates with various misorientations from the (001) plane in the same growth run. Among the narrowest linewidths ever reported (8.8 meV for a 6 Å well of InGaAsP) with multiple line spectra resulting from integer monolayer inter-well-width variations over lateral sizes larger than the excitonic diameter. For substrates with $\geq 2^\circ$ misorientations, the lateral sizes of the flat terraces are smaller than the excitonic diameter, and consequently single line PL emissions were observed consistently. For strained-layer InGaAs/In(GaAs)P quantum wells similar results were demonstrated for $\leq 0.2^\circ$ misoriented (001) InP substrates. In contrast to the lattice matched quantum wells, for quantum wells grown under strain, the interface quality depends critically on both the amount of strain applied and the substrate misorientation from the (001) plane.

4.7. References


Characterization and Performance of $\lambda=1.5$ μm (Strained-Layer) InGaAs(P) Quantum Well Lasers

Abstract

The fabrication, characterization, and performance of semiconductor lasers emitting in the 1.48-1.55 μm wavelength band employing In$_x$Ga$_{1-x}$As$_y$P$_{1-y}$/In(GaAs)P quantum wells grown in the range from 2.1% biaxial tensile to 1.8% biaxial compressive strain are reported. For the compressively strained, as predicted theoretically, as well as for tensile strained quantum well lasers, improved performance compared to unstrained quantum well and bulk active layer devices is observed. The improvements in the laser parameters: threshold current (density), external differential efficiency, output power, characteristic temperature, and linewidth enhancement factor can be interpreted on the basis of reduced intrinsic loss mechanisms (Auger recombination and intervalence band absorption), and reduced effective hole mass in long wavelength strained-layer quantum well lasers brought about by the modified band structure.

The results obtained from this work are compared with those reported in the literature on strained-layer long wavelength InGaAs(P) quantum well lasers. Finally, long-term reliable operation under severe conditions of strained-layer In$_x$Ga$_{1-x}$As ($x=0.8$ and 0.7)/InGaAsP multiple quantum well lasers is demonstrated.
5.1. Introduction

High quality (strained-layer) InGaAs(P)/In(GaAs)P quantum well (QW) structures grown by low-pressure organometallic vapour phase epitaxy (LP-OMVPE) were reported in chapter 4. Similar structures are now employed as active layers in (multiple) quantum well semiconductor lasers emitting at 1.5 μm wavelength. In section 5.2, the device structures used in this work and their fabrication are reported. The characteristics of unstrained InGaAsP MQW lasers are summarized in section 5.3. The marginal improvements of these lasers over conventional bulk active layer lasers, together with the conviction that this was not caused by the limited crystalline quality but were rather due to intrinsic effects, triggered work on strained-layer InGaAs(P) quantum well lasers. The design of the active layers and their characteristics are reported in section 5.4. This chapter is concluded in section 5.5.

5.2. Device Structures and Fabrication

In this study three device structures were fabricated. Mainly for materials evaluation purposes 50 μm oxide stripe lasers were used. For practically applicable devices index guided lasers with either the hybridly LP-OMVPE/liquid phase epitaxial (LPE) grown Double Channel Planar Buried Heterostructure (DCPBH) or the completely LP-OMVPE grown Semi-Insulating Planar Buried Heterostructure (SIPBH) were used.

The wafers for the 50 μm oxide stripe lasers are fabricated in a single LP-OMVPE growth step. A double heterostructure (DH) consisting of an active layer embedded within n-type and p-type InP confining layers, and capped with a p'-InGaAs(P) contact layer is grown on a n-type InP substrate. The active layer structures used in this work will be reported in sections 5.3. and 5.4.2. For the device fabrication, on the p-side of the wafer a SiO₂ film is deposited by chemical vapour deposition. Using conventional lithography and wet chemical etching the SiO₂ is opened over 50 μm wide stripes with a period of 300 μm. After thinning the wafer to about 100 μm, metallizations are sputtered on both sides of the wafer (Au/Ge/Ni for the n-electrode, and Pt, Ta, Pt, Au for the p-electrode, respectively). Subsequently, the metallizations are alloyed, laser chips are cleaved and mounted p-side down on copper heatsinks on TO-5 headers. In the oxide stripe lasers the lateral optical confinement is provided by the current flow, i.e. by the lateral gain distribution. This leads to an unstable optical mode which is often reflected by non-linearities in the light-current characteristics, and therefore these lasers are mainly used for material evaluation (e.g. threshold current density) purposes. In addition, the threshold currents are generally too large for continuous wave (CW) operation.
Figure 5.1. Sketch of double channel planar buried heterostructure (DCPBH) laser chip.

For practically applicable lasers, low lasing threshold, and stable fundamental lateral mode operation with an optical beam that efficiently can be coupled into the optical fiber are required. This is achieved using buried heterostructure (BH) lasers, fabricated in multiple stages of epitaxy and etching to laterally confine the optical mode and the injection current.

The fabrication of the DCPBH device structure (figure 5.1) involves two epitaxial growth steps. In the first step, three layers - n-InP, active layer and p-InP - are grown by LP-OMVPE on a two-inch n-type InP substrate. Subsequently, two channels, about 3 μm deep and 10 μm wide, leaving a mesa of about 1.5 μm in width, are wet chemically etched at a pitch of 300 μm along [110] into the wafer using a photoresist mask. After removal of the mask and cleaning of the wafer, the device structure is finished in a second growth by LPE which allows maskless selective growth of epitaxial layers, embedding the mesa with p-type and n-type InP current blocking layers, respectively. The third and fourth layer, the p-InP confinement and p'-InGaAs(P) contact layers, respectively, are grown on the entire wafer for planarization. After the growth, the wafer is thinned to about 100 μm and both sides are metallized and alloyed. Laser chips are cleaved and mounted p-side down on copper heatsinks on TO-5 headers.

The third device structure is the SIPBH (figure 5.2) which is fabricated entirely by LP-OMVPE in three growth steps. In the first step, three layers - n-InP, active layer, and p-InP - are grown successively on a two-inch n-type InP substrate. Then, using a SiO₂ mask, all grown layers are wet chemically etched away except mesas with 300 μm period along [110], about 1-2 μm wide and 2-3
μm high, which become the active stripes. With the SiO₂ mask still on top of the mesas, a semi-insulating Fe-doped InP layer and a n-InP layer are grown selectively. After removing the mask, p-InP and p⁺-InGaAs layers are grown on the entire wafer in the third LP-OMVPE growth. The n-type InP layer is introduced in the second epitaxial growth to block Fe-Zn interdiffusion [1, 2] in order to preserve the high resistivity \(10^8 \ \Omega \cdot \text{cm}\) of the Fe-doped InP layer [3, 4]. The wafers are processed using metallization and mounting procedures similar to the DCPBH laser. The SIPBH laser structure is of interest firstly because it is fabricated completely by LP-OMVPE which is a 2-inch wafer technology allowing cost-efficient laser fabrication. Secondly, the parasitic capacitance of the SIPBH laser is significantly smaller than for BH structures employing reversely biased p-n junctions for current blocking. The latter is of special interest to eliminate electrical frequency response limitations for high speed applications as expressed by eq. (2.18).

Both BH lasers can be designed to show output beams of 25±5° in the directions parallel and perpendicular to the plane of the active layer, facilitating fiber coupling efficiencies of 60-70%.

5.3. Characterization of Unstrained InGaAsP MQW Lasers

The first 1.5 μm wavelength InGaAs(P)/InP optoelectronic devices exploiting the quantum size effects were reported in the mid-1980’s [5-8]. Typically, the threshold currents were higher than 1 ampere because no appropriate current confinement structure was applied. Moreover, several of
these devices were not real quantum well lasers; e.g. Dutta et al. [7] used 300 Å thick "quantum wells" without supplying evidence for quantization effects. The devices reported by Miyamoto et al. [8] apparently showed a blue shift in the spontaneous emission spectra due to quantization effects, but lasing was only obtained at a wavelength corresponding to the bandgap of the thick confinement layers. The first genuine low threshold current 1.5 μm wavelength buried heterostructure MQW lasers were reported by Thijs et al. [9] in 1987. MQW active layers consisting of six 90-Å-thick wells of InGaAsP (λg=1.55 μm) and five 200 Å InP barrier layers embedded within InGaAsP (λg=1.28 μm) separate confinement heterostructure (SCH) layers were employed, in devices with the DCPBH lateral current confinement. The genuine quantum well nature of the active region is experimentally verified by room temperature PL on wafers and electroluminescence from lasers. The n=1 electron-hole transitions show a shift from the bulk band edge at λ=1.55 μm to λ_{self}=1.47 μm, and under intense PL excitation a second PL emission at 1.35 μm due to higher energy quantum well levels becomes visible (figure 5.3a). A similar wavelength spectrum is observed in the electroluminescence of short cavity length (L=200 μm) DCPBH lasers as shown in figure 5.3b. The light-current characteristics at various temperatures of 650 μm cavity length SCH-MQW-DCPBH lasers are shown in figure 5.4. The

**Figure 5.3.** a) Room temperature photoluminescence spectra, and b) electroluminescence spectra of InGaAsP/InP MQW structure at various photo excitation densities and injection currents, respectively. The undulations in the photoluminescence spectra around 1380 nm are due to absorptions of water in the quartz-ware lenses.
threshold current is about 35 mA and the lasing wavelength is 1.48 μm. Hence, the devices lase from the c_1-hh_1 recombinations. This also demonstrates the thermal stability of the InGaAsP/InP quantum wells against interdiffusion during the heating in the LPE regrowth process. At room temperature the T_g-value is around 120K but decreases to 45K between 10 and 70°C heatsink temperature. At high output powers and elevated temperatures (e.g. around 20 mW/60°C) the output power is no longer linear with the current. Apparently, gain saturation from the ground level recombination becomes important. Upon further increasing the drive current, the lasing wavelength switches from 1.5 to 1.4 μm. Short cavity length (L=200μm) SCH-MQW-DCPBH lasers start to lase at 1.4 μm. This implies that the gain from the ground level recombination of the six InGaAsP QW-structure is smaller than 800 cm\(^{-1}\) (assuming an internal loss of 15 cm\(^{-1}\), a facet reflectivity of 32%, and an optical confinement of 1.5% per QW).

Subsequently, lasers employing twelve 90 Å InGaAsP (λ_g=1.55 μm) quantum wells and eleven 200 Å InP barrier layers were fabricated. Figure 5.5 shows a scanning electron microscope (SEM) picture of the stained InGaAsP/InP MQW active layer in cross section. The devices emit at 20°C from the n=1 quantized state with a CW threshold current as low as 20 mA, show an external
1.5 μm (Strained-Layer) InGaAs(P) QW Lasers

Figure 5.5. Scanning electron microscope (SEM) picture of stained facet of InGaAsP/InP MQW-DCPBH laser.

Efficiency of 0.2 mW/mA per facet and maximum output powers of 25 mW and 15 mW per facet for pulsed and CW operation, respectively. The lower saturation value of the optical output, compared to laser diodes with bulk active layers, is most likely caused by an increase in voltage drop across the MQW resulting in increased leakage current. The $T_0$-value was 60K at room temperature. In figure 5.6a a set of light-current characteristics show that CW operation remains possible up to 75°C. As a comparison, figure 5.6b shows the CW light-current characteristics between 10 and 60°C of a 1.5 μm conventional bulk InGaAsP active layer laser. The $T_0$-value of this device is 59K between 10 and 60°C. An improvement of the QW device characteristics was obtained by applying high reflection coatings on the laser facets. This yields threshold currents of 12 mA, CW operation to over 100°C and a maximum front facet CW
output of 40 mW at room temperature. Devices with single and triple InGaAsP/InP quantum well active layers, sandwiched between separate confinement layers of InGaAsP ($\lambda=1.3 \mu m$) were found to lase from the separate confinement layers. This shows that the gain of these QW active layers is too small to overcome the internal and mirror losses.

From this paragraph and data reported in the literature [10], it may be concluded that InGaAsP/InP MQW laser diodes have not shown similar improvements in performance as reported for GaAs/AlGaAs QW lasers (e.g. sub-mA threshold current devices [11], improved high temperature operation and external differential efficiency). This is attributed to the increased Auger recombination, the strong intervalence band absorption, the smaller optical confinement, and the smaller conduction band discontinuity compared to the GaAs/AlGaAs system. The performance of the 1.5 $\mu$m InGaAsP MQW devices is not expected to be limited by the materials- and interface quality as was also indicated in chapter 4 and will be confirmed in next section. This initiated work on strained-layer InGaAs(P) quantum well lasers which are reported in next sections.
5.4. Strained-Layer In$_x$Ga$_{1-x}$As Quantum Well Lasers

5.4.1. Introduction

In chapter 2, enhanced performance was predicted for QW lasers with the active material deliberately grown in a state of strain resulting in removal of the cubic symmetry of the lattice. In combination with the quantum size effect, a modified valence band structure is obtained resulting both in reduced hole effective mass and non-radiative loss mechanisms. Additionally, by using mismatched compositions, the band discontinuities are affected and the accessible wavelength window is extended beyond 1.65 μm wavelength. The effects of the biaxial strain and the predicted consequences for the laser parameters are summarized in table 5.1.

<table>
<thead>
<tr>
<th>Effect of biaxial strain:</th>
<th>Effect on laser parameter:</th>
</tr>
</thead>
<tbody>
<tr>
<td>Reduction hole effective mass</td>
<td>$J_{th} \gamma$, dG/dN$\gamma$, $f_r \gamma$, $\alpha_r \gamma$</td>
</tr>
<tr>
<td>Reduction Auger recombination</td>
<td>$T_\eta \gamma$</td>
</tr>
<tr>
<td>Reduction Intervalance band absorption</td>
<td>$\alpha_i \gamma$, $\eta_i \gamma$</td>
</tr>
<tr>
<td>Valence band maximum: hh or lh</td>
<td>TE or TM polarization</td>
</tr>
<tr>
<td>Compression: increased Δ$E_c$</td>
<td>$T_\eta \gamma$, $J_{th} \gamma$</td>
</tr>
<tr>
<td>Accessible bandgap extended</td>
<td>$\lambda$ up to 2.0 μm</td>
</tr>
</tbody>
</table>

It is the main purpose of this section to assess the characteristics of 1.5 μm wavelength strained-layer InGaAs(P) quantum well lasers and to compare the results with the initial expectations. In addition, the reliability issue of strained-layer quantum well lasers is addressed since this has been a major point of concern.

There are several difficulties to single out the pure effects of strain on the laser performance. For example, as indicated in chapter 2, the strain affects both the bandgap and the band offsets [12-14]. So, in keeping the quantum well
Figure 5.7. Calculated emission wavelengths of In$_x$Ga$_{1-x}$As/InGaAsP QWs as a function of QW thickness. For $x \geq 0.53$ the InGaAsP cladding layer was assumed to have $\lambda_e = 1.25$ μm composition, whereas $\lambda_e = 1.15$ μm InGaAsP was assumed for $x < 0.53$. Solid curves denote c-hh transitions, and dashed curves c-lh transitions.

width fixed, both the lasing wavelength (figure 5.7) and the carrier confinement will vary upon varying the strain. Since the loss mechanisms (Auger recombination and intervalence band absorption) both depend on the emission wavelength, the pure effects of strain will be obscured. On the other hand, by adjusting the quantum well width for different strains, the emission wavelength may be fixed but the optical confinement factor, being approximately proportional to the quantum well width in a separate confinement heterostructure, strongly varies. In addition, the current injection efficiency may vary [15] for different well widths, as well as the number of confined quantum well levels and their dispersions [16]. Nevertheless, we have chosen for the latter fixed-wavelength-approach since for applications in optical fiber communication systems the main interest is in optimum performance lasers emitting in a specific narrow wavelength band, namely around 1.55 and 1.3 μm because of the minima in attenuation and dispersion of the optical fiber, respectively, and at 1.48 μm for pumping Er$^{3+}$-doped fiber amplifiers (EDFAs).
5.4.2. Design of $\lambda=1.5$ $\mu$m Strained-Layer In$_x$Ga$_{1-x}$As Quantum Well Lasers

In this work In$_x$Ga$_{1-x}$As/InGaAsP single and multiple quantum well (SQW and MQW) lasers were studied with InAs mole fractions in the ternary QW varying from $x=0.22$ to $x=0.80$, corresponding with 2.1% tensile to 1.8% compressive strain, respectively. To enhance the optical confinement factor of the ternary wells, Separate Confinement Heterostructures (SCH) or step-GRaded INdex-SCHs (GRINSCH) of InGaAsP lattice matched to InP were applied. In compressively strained In$_x$Ga$_{1-x}$As ($x>0.53$) SQW lasers a symmetrical GRINSCH composed of lattice matched 1.0, 1.1, and 1.2 $\mu$m bandgap wavelength InGaAsP layers was used, as shown schematically in figure 5.8a. For structures with the In$_x$Ga$_{1-x}$As wells under tensile strain ($x<0.53$), the smallest bandgap quaternary InGaAsP had 1.15 $\mu$m emission wavelength to compensate for the reduced conduction band offset [12]. For MQW lasers various (GRIN)SCH structures were applied; figure 5.8b shows an example of a two-step GRINSCH composed of $\lambda_g=1.1$ and 1.25 $\mu$m InGaAsP with a thickness of

![Diagram](image)

**Figure 5.8.** Schematic conduction band diagrams of graded index separate confinement heterostructure (GRINSCH) compressively strained single quantum well (a) and four quantum well (b) active layer.
typically 750-1500 Å. To decouple the quantum wells while maintaining efficient carrier injection, the barrier thickness was typically chosen 125 Å [15]. In order to keep the emission wavelength fixed near 1.5 μm, the quantum well thickness must be reduced for increasing InAs mole fraction (=increasing compressive strain) whereas the well thickness must be increased for decreasing InAs mole fraction (=increasing tensile strain) as indicated in figure 5.7. The InAs mole fractions x, the resulting strains, the well widths L_z, the optical confinement factors Γ for the TE and TM modes as calculated using an effective index program, and the room temperature PL emission wavelengths of a set of In_xGa_1-xAs SQW laser structures grown for this study are summarized in table 5.2.

A remark must be made regarding the room temperature PL spectra. In_xGa_1-xAs/InGaAsP QW structures with the wells grown either lattice matched or under compressive strain, always show single line room temperature PL emissions with increasing intensity and decreasing FWHM linewidth for

<table>
<thead>
<tr>
<th>x</th>
<th>strain</th>
<th>L_z (Å)</th>
<th>Γ_TE (%)</th>
<th>Γ_TM (%)</th>
<th>λ_{PL} (nm)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.22</td>
<td>2.1%T</td>
<td>160</td>
<td>2.35</td>
<td>1.89</td>
<td>1504</td>
</tr>
<tr>
<td>0.32</td>
<td>1.6%T</td>
<td>120</td>
<td>1.71</td>
<td>1.38</td>
<td>1523</td>
</tr>
<tr>
<td>0.37</td>
<td>1.2%T</td>
<td>100</td>
<td>1.41</td>
<td>1.13</td>
<td>1518</td>
</tr>
<tr>
<td>0.40</td>
<td>0.9%T</td>
<td>75</td>
<td>1.04</td>
<td>0.84</td>
<td>1487</td>
</tr>
<tr>
<td>0.44</td>
<td>0.6%T</td>
<td>90</td>
<td>1.26</td>
<td>1.01</td>
<td>1524</td>
</tr>
<tr>
<td>0.49</td>
<td>0.3%T</td>
<td>95</td>
<td>1.33</td>
<td>1.07</td>
<td>1535</td>
</tr>
<tr>
<td>0.53</td>
<td>0</td>
<td>80</td>
<td>1.11</td>
<td>0.90</td>
<td>1565</td>
</tr>
<tr>
<td>0.57</td>
<td>0.3%C</td>
<td>80</td>
<td>1.11</td>
<td>0.90</td>
<td>1576</td>
</tr>
<tr>
<td>0.65</td>
<td>0.8%C</td>
<td>45</td>
<td>0.61</td>
<td>0.49</td>
<td>1556</td>
</tr>
<tr>
<td>0.70</td>
<td>1.2%C</td>
<td>25</td>
<td>0.33</td>
<td>0.27</td>
<td>1515</td>
</tr>
</tbody>
</table>
Figure 5.9. Room temperature photoluminescence spectra of tensile strained In$_x$Ga$_{1-x}$As/InGaAsP QW structures with increasing strain and well width from top to bottom. The spectra can be deconvoluted (solid curves) in emissions resulting from e-lh, and e-hh transitions, respectively. Zero strain, $L_z = 80$ Å; 1.2% tensile strained In$_{0.37}$As$_{0.63}$As, $L_z = 100$ Å, $\Delta E_{\text{hh-lh}} = 27$ meV; 1.6% tensile strained In$_{0.32}$Ga$_{0.68}$As, $L_z = 120$ Å, $\Delta E_{\text{hh-lh}} = 107$ meV; 2.1% tensile strained In$_{0.22}$Ga$_{0.78}$As, $L_z = 160$ Å, $\Delta E_{\text{hh-lh}} = 123$ meV.

Increasing strain up to 1.8%, the maximum strain used in this study. Generally, the PL intensity is used as a qualitative measure for the materials quality. For tensile strained QWs, however, the modification of the band structure
substantially affects the PL lineshape and intensity. The latter could be mis-interpreted to result from inferior materials quality. For structures with the In$_x$Ga$_{1-x}$As well grown under tensile strain, the PL spectra generally show a doublet due to electron-light hole (c-lh$_1$), and most probably due to electron-first heavy hole (c-hh$_1$) recombinations because of the large density of the heavy hole states. With both increasing strain and quantum well width, the PL-peak separation increases due to the increased valence subband splitting resulting from the reduced quantum size effect, as shown in figure 5.9a to d. However, in this sequence also the PL intensity decreases due to the reduced conduction band offset resulting in reduced electron confinement to the wells, and therefore also in reduced overlap with the holes states. Figure 5.10 shows, as an example, the calculated band structure [18] of a 110 Å thick 1.6% tensile strained In$_{0.32}$As$_{0.68}$As QW embedded within $\lambda_g$=1.15 μm InGaAsP. The electron confinement amounts only 13 meV, whereas the lh$_1$ levels are strongly confined (225 meV). In the photoluminescence measurements the carrier density in the wells is much smaller than under electrical injection required to start the lasing action. Consequently, the large number of holes strongly confined to the valence band induces additional electron confinement in the conduction band [18] and enhances the radiative electron-light hole recombination in the tensile strained quantum wells as will be indicated later on in this chapter.

![Figure 5.10](image)

**Figure 5.10.** Calculated band structure of 110 Å 1.6% In$_{0.32}$Ga$_{0.68}$As/InGaAsP ($\lambda_g$=1.15 μm) quantum well.
5.4.3. Threshold Current Density versus Strain

The strain dependence of the threshold current density \( (J_{th}) \) has been studied using 1.5 \( \mu m \) wavelength \( In_{0.32}Ga_{0.68}As \) SQW and MQW 50 \( \mu m \) oxide stripe lasers. For SQW lasers particularly, the threshold gain given by \( G_{th} = (\alpha_i + \alpha_m)/\Gamma \), with \( \alpha_i \) the internal loss, \( a_m \) the mirror loss and \( \Gamma \) the optical confinement factor, easily exceeds several hundreds cm\(^{-1}\), because of the small optical confinement factor. Due to the gain saturation related to the constant density of states of the lowest subband of the quantum well, the variation of the gain \( G \) with the injected current density \( J \) is best approximated by [19, 20]

\[
G = J_{th} \beta \ln \frac{J}{J_{th}}
\]  

(5.1)

where \( J_{th} \) is the transparency current density and \( \beta \) is the slope of the gain-current characteristic at the origin, i.e. \( \beta \) is proportional to the differential gain \( (\mathrm{d}G/\mathrm{d}N) \) in the low gain region of the characteristic. From eq. (5.1), together with \( g_{th} = \alpha_i + 1/L \ln(1/R) \), where \( L \) is the cavity length and \( R \) is the mirror reflectivity, and with \( J = J_{th} \eta_i \), where \( \eta_i \) is the internal efficiency, the threshold current density is expressed as

\[
\ln J_{th} = \ln \frac{J_{th}}{\eta_i} + \frac{\alpha_i}{\beta J_{th}} + \frac{1}{\Gamma \beta J_{th} L} \ln \frac{1}{R}
\]  

(5.2)

When only the first subband is populated, it has been shown that \( J_{th} \) is independent of the quantum well width \( L \), and \( \beta \leq L_e^{-1} \) [21]. Furthermore, since \( \Gamma \leq L_e \), eq. (5.2) shows that \( J_{th} \) should be independent of \( L_e \).

Recently, Tiemeijer et al. [22] demonstrated for a wide range of the gain in quantum well semiconductor laser amplifiers, the validity of the net gain \( (=g-\alpha_i)-current \) relation: \( g - \alpha_i = g_0(J_{th})/(J + J_{th}) \), where \( g_0 \) is the gain at complete inversion. This relation solves the physically unrealistic limits of the logarithmic gain-current relation (eq. 5.1) for \( J = 0 \) and \( J = \infty \). However, in the regime of interest for quantum well lasers, that is near transparency, both expressions give similar results [22].

In figure 5.1, threshold current densities at 20°C plotted on a logarithmic scale against the inverse cavity length of \( In_{0.6}Ga_{0.4}As \) SQW oxide stripe lasers show the linear behaviour as deduced in eq. (5.2). Minimum threshold current densities as low as 92 A/cm\(^2\) for 1.1 cm cavity length \( In_{0.32}Ga_{0.68}As \) (1.6% tension) SQW lasers [23, 24], and 147 and 160 A/cm\(^2\) for 4.5 mm cavity length \( In_{0.65}Ga_{0.35}As \) (0.8% compression) and \( In_{0.7}Ga_{0.3}As \) (1.2% compression) SQW lasers [25, 26], respectively, were measured. These values
are among the lowest reported so far, and make sub-mA threshold current lasers in the 1.5 \( \text{\textmu m} \) wavelength range feasible. This is of interest for bias-free modulation and has important implications for the integration of lasers with electronic drive circuitry. The monotonically decreasing threshold current densities up to large cavity lengths, as observed in figure 5.11, demonstrates that the threshold current is loss limited rather than transparency limited [22].

The strain dependence of the threshold current density is depicted more clearly in figure 5.12. Here, the threshold current densities per well for infinite cavity length lasers extrapolated using eq. (5.2) have been plotted versus the magnitude of the strain in the well [23, 26]. In this figure also data from multiple quantum well lasers are included. For certain compressive as well as tensile strain values, significantly lower threshold current densities compared to the conventional unstrained QW devices are observed. The origins of these reductions in threshold current will be discussed next.
**Figure 5.12.** Strain dependence of threshold current density per quantum well at infinite cavity length of 1.5 μm wavelength In$_x$Ga$_{1-x}$As lasers.

### a) Compressive Strain Branch

With increasing compressive strain, the in-plane hole effective mass monotonically decreases and the hh-lh valence subband energy splitting increases. This is expected to reduce the loss-mechanisms (Auger processes and intervalence band absorption). All effects will contribute to the observed reduction in the threshold current with increasing compressive strain. However, for large compressive strains the In$_x$Ga$_{1-x}$As QW-widths become very narrow as indicated in table 5.2. Despite that the modal gain was theoretically predicted to be independent of the quantum well width, in practice for narrow wells, interwell-width variations and decreased current injection efficiencies [15] broaden and reduce the gain. Consequently, a larger carrier density in the well is required, which additionally enhances the Auger recombination both leading to the slight increase in threshold current density as observed in figure 5.12 for 1.8% compressively strained In$_{0.8}$Ga$_{0.2}$As wells. An increase in threshold current density due to crystalline defects can be excluded since the well thicknesses are
smaller than the critical values as indicated in figure 2.14. For 1.5 μm wavelength InₓGa₁₋ₓAs quantum well lasers, the optimum compressive strain is around 1%. By applying larger bandgap strained-layer InGaAsP quaternary material in the well, wider quantum wells must be applied for 1.55 μm wavelength emission. It is expected and indeed demonstrated [27, 28] that the increase in threshold current observed in figure 5.12 for narrow, highly strained InₓGa₁₋ₓAs is absent.

b) Tensile Strain Branch

Figure 5.12 shows the second branch for tensile strain with also a significantly reduced threshold current density. This was originally not predicted theoretically, but could be readily explained [16, 29] after the experimental demonstration. In the strain-region of the lh-hh valence subband crossing, i.e. for tensile strains up to about 0.9%, bandmixing effects increase the in-plane hole effective mass. For certain combinations of tensile strain and QW width, the band structure even becomes indirect [17, 30-32]. The resulting heavy hole mass increases both the threshold current and the non-radiative losses such that it prevents the 0.3, 0.6 and 0.9% tensile strained SQW lasers from lasing at room temperature in our case. With both increasing tensile strain and quantum well width, the lh-hh valence subband separation increases. This reduces the lh-hh bandmixing and consequently results in reduction of the hole effective mass and of the intensity of non-radiative loss mechanisms leading to a lower threshold current density. The increasing lh-hh valence subband splitting with increasing tensile strain and quantum well width can be monitored by room temperature PL measurements as shown in figure 5.9. In addition, the TM polarized emission of the tensile strained QW lasers confirms that lasing occurs from electron-light hole recombinations. A minimum threshold current density as low as 88 A/cm² is extrapolated for 1.6% tensile strained In₀.₃₂Ga₀.₆₈As SQW lasers. The increased threshold current for the 2.1% tensile strained SQW laser is ascribed to two effects. First, this structure has a large well width grown under large strain (160 Å x 2.1% strain) which is above the critical thickness. Indications for the presence of misfit dislocations were obtained from Nomarski interference contrast microscopy. In addition, high resolution X-ray diffraction measurements showed an in-plane lattice parameter differing from the substrate lattice parameter demonstrating the relaxation of the QW lattice. Secondly, taking into account the band structure shown in figure 5.10, the conduction band offset in this 2.1% tensile strained SQW structure will be smaller or may even be negative resulting in a type II band structure where the electrons are confined in the barrier layers and the holes in the well.
Figure 5.13. Summary of reported threshold current densities [27, 28, 34-44] versus the strain in 1.5 μm wavelength InGaAs(P) QW lasers.

Figure 5.13, shows a compilation of threshold current densities per quantum well of infinite cavity length 1.5 μm strained-layer InGaAs(P)/InP single and multiple quantum well lasers versus the strain in the well, as extrapolated according eq. (5.2) from data reported in this work and in the literature [27, 28, 34-44]. A W-shaped curve similar to the one shown in figure 5.12 [23], with significantly lower threshold current densities compared to unstrained QW lasers is observed for both signs of the strain. In agreement with the arguments given above concerning the formation of an unfavourable band structure for certain combinations of the tensile strain and of the quantum well width, there has indeed been no report on low threshold current lasers with strains in the quantum wells ranging from zero to about 1% tension.

However, it has to be noted that threshold current densities per quantum well deduced from multiple quantum well data should be interpreted with some caution. If the well thicknesses in a multiple quantum well structure differ, unequal pumping of the wells is achieved, i.e. the well with the smallest (effective) bandgap provides most of the gain to overcome the losses. If the gain in this well is saturating, the next well is "switched on" to contribute to the gain, etc. [33]. This leads to an underestimation of the threshold current density per quantum well. We have checked this by deliberately growing two strained In$_{0.65}$Ga$_{0.35}$As/InGaAsP MQW structures, one with four wells each differing one monolayer in thickness, and one with four identical wells. The FWHM PL
linewidth of the structure with different QWs was about 30% broader and showed due to sequential bandfilling effects a much stronger wavelength shift of 21 nm upon increasing the PL excitation intensity compared to 4 nm shift for the structure with the identical quantum wells under the same excitation conditions. As expected, lasers (1 mm cavity length) employing QWs of different thickness in the active layer showed about 30% smaller threshold current density than the lasers with identical wells.

5.4.4. Strain-Effect on the Threshold Current Density: Hole Effective Mass or Non-Radiative Recombination?

The significant reductions in the threshold current densities observed in figure 5.12 may be explained by the reduced hole effective mass and/or non-radiative losses such as Auger recombination and intervalence band absorption. In order to gain more insight in the origin of these improvements, threshold current densities were measured as a function of temperature. Reduction of the temperature significantly reduces the effect of the Auger processes and intervalence band absorption, and at low temperature (e.g. at 77K), except for almost perfect degenerate valence band structures, the threshold current will be mainly determined by the effective masses. At the valence subband crossing, the band structure has become unfavourable for lasing action, i.e. the band structure may be indirect or direct with a large hole effective mass. As a consequence, a large carrier population at high k-values leads to enhanced non-radiative recombination, even at 77K [16]. The threshold current densities at various temperatures of as-cleaved 1 mm cavity length 50 μm oxide stripe lasers with strains in the InGaAs single quantum wells in the range of 1.6% tension to 1.2% compression are plotted in figure 5.14a. As expected, the largest temperature effect is observed for strains ranging from 0.9% tension to 0.3% compression. These lasers fail to lase at room temperature, whereas the 77K threshold current densities of the -0.3, 0, and 0.3% strained lasers have become very similar to the devices with larger strains which show significantly lower thresholds at room temperature. This demonstrates the freezing out of the non-radiative recombination and loss mechanisms. Even at 77K, a maximum in the threshold current density is still observed around 0.8% tensile strain, in the transition region of the two branches of the W-shaped curve with different polarizations of the emitted light demonstrating the crossing of the lh-hh states. Except for the threshold current densities in this transition region, the remaining 77K values fit well with values calculated [16], assuming the absence of Auger recombination and intervalence band absorption, and a total cavity loss of about 15-20 cm⁻¹, including 11.5 cm⁻¹ mirror loss. This indicates that the effective
1.5 μm Strained-Layer InGaAs QW Lasers

![Graphs showing threshold current densities for InGaAs/InP and InGaP/GaAs lasers.](image)

**Figure 5.14.** Strain dependence of threshold current densities at various temperatures for (A) 1 mm cavity length 1500 nm wavelength GRINSCH In$_x$Ga$_{1-x}$As/InGaAsP SQW 50 μm oxide stripe lasers, and (B) 633 nm wavelength In$_x$Ga$_{1-x}$P/AlGaInP/GaAs DQW lasers.

masses determine the threshold current density.

As a comparison, in figure 5.14b the threshold current densities at various temperatures have been plotted versus the grown-in biaxial strain in 633 nm wavelength In$_x$Ga$_{1-x}$P/AlGaInP double quantum well lasers [45, 46]. Also for these lasers a maximum in the threshold current density is observed near the lh-hh valence band crossing at small tensile strains, although less pronounced than for the long wavelength lasers. In contrast to the 1500 nm wavelength lasers, the In$_x$Ga$_{1-x}$P lasers are known to be free of Auger recombination and intervalence band absorption. This is confirmed by an almost constant temperature dependence of threshold current density independent of the strain, and indicates that in these lasers the strain-induced modification of the effective mass is the dominant effect. The ratios of the 293K and 77K threshold current densities are about ten. Similar ratios for 1500 nm SQW lasers are observed only for the largest strains applied (>1% compression and 1.6% tension) indicating that for these ranges of biaxial strain the non-radiative losses and intervalence band absorption have been reduced to a large extent.
5.4.5. Evidence for Reduced Intervalance Band Absorption

Compressively strained 1.5 μm wavelength In$_{x}$Ga$_{1-x}$As/InGaAsP (x>0.53) quantum well lasers exhibit very large differential efficiencies [47] which indicates a reduced internal loss $\alpha_i$. Moreover, the temperature dependence of the differential efficiency is reduced. As the carrier density becomes pinned above threshold, the most likely candidate responsible for this improvement is the reduction of the intervalence band absorption as discussed in section 2.3.3.

The internal loss $\alpha_i$, and the internal efficiency $\eta_i$ are correlated with the inverse differential efficiency $\eta_d$ according to

$$\eta_d^{-1} = \eta_i^{-1} \left(1 + \frac{\alpha_i}{\alpha_m}\right)$$

(5.3)

![Graph](image)

**Figure 5.15.** Inverse external differential efficiency of 50 μm oxide stripe 1.5 μm wavelength GRINSCH In$_{0.7}$Ga$_{0.3}$As/InGaAsP SQW lasers as a function of cavity length. An internal efficiency $\eta_i$ and an internal loss $\alpha_i$ of 0.54 and 3.8 cm$^{-1}$ were deduced from the intercept and slope, respectively.
where \( a_n = \frac{\ln(1/R_1 R_2)}{2L} \), with \( L \) the cavity length, and \( R_1 \) and \( R_2 \) the facet reflectivities. Figure 5.15 shows the inverse external differential efficiency of 1.2% compressively strained In\(_{0.5}\)Ga\(_{0.3}\)As SQW 50 \( \mu \)m oxide stripe lasers versus the cavity length. From the intercept and slope, an internal efficiency of 54% and an internal loss as low as 3.8 \( \text{cm}^{-1} \) were deduced [24]. Generally, the internal efficiencies of SQW lasers are smaller than for MQW lasers because of the relative large probability for the carriers to escape from the SQW which cannot be recaptured, and therefore are lost for the stimulated recombination process [15]. The internal efficiency can be improved by optimizing the GRINSCH as demonstrated by Tsang et al. [38] who reported an internal efficiency of 78% for 1% compressively strained In\(_{0.65}\)Ga\(_{0.35}\)As SQW broad area lasers with an internal loss of also 3.8 \( \text{cm}^{-1} \). By increasing the number of quantum wells the internal efficiency indeed increases. For an In\(_{0.5}\)Ga\(_{0.3}\)As/InGaAsP four quantum well SIPC BH laser an internal efficiency of over 85% and an internal loss of 8-9 \( \text{cm}^{-1} \) were observed [24]. In this case, part of the increased internal loss compared to the 3.8 \( \text{cm}^{-1} \) observed on the 50 \( \mu \)m oxide stripe SQW lasers may be introduced due to the formation of the narrow stripe in the SIPC BH lasers giving rise to some scattering at the sidewalls of the chemically etched active layer, while part of the increase may also be due to the larger number of quantum wells. DCPBH lasers employing four 1.8% compressively strained In\(_{0.8}\)Ga\(_{0.2}\)As quantum well lasers, with 1.3 \( \mu \)m bandgap InGaAsP barrier- and separate confinement layers, show internal losses of 13 \( \text{cm}^{-1} \) and internal efficiencies close to 100% as shown in figure 5.16. For six QW structures these figures are 14 \( \text{cm}^{-1} \) and 90% (not shown), respectively. A maximum differential efficiency as high as 80% was observed from 265 \( \mu \)m cavity length lasers employing four 1.8% compressively strained quantum wells [47]. This is an improvement of over a factor of two compared to conventional bulk InGaAsP lasers for which an internal efficiency and the internal loss of 62% and 67 \( \text{cm}^{-1} \) were deduced (figure 5.16). Generally, DCPBH lasers exhibit somewhat larger internal loss than SIPC BH lasers (e.g. 13 \( \text{cm}^{-1} \) versus 8-9 \( \text{cm}^{-1} \) for the four quantum well lasers) because in DCPBH lasers a larger fraction of the optical mode is propagating in p-type InP which has a larger free carrier absorption than the iron-doped InP employed in the SIPC BH. These simple measurements clearly indicate the reduced internal loss in 1.5 \( \mu \)m compressively strained In\(_x\)Ga\(_{1-x}\)As/InGaAsP most probably resulting from significantly reduced intervalence band absorption. This is indicated by the fact that the internal loss is not linearly proportional to the number of quantum wells employed in the active layer. The large values of \( \eta_i \) indicate that the intensity of the radiative recombinations is much larger than of the non-radiative recombinations, or \( \tau_i > \tau_n \) (section 2.1).

Relative small internal losses of about 9 \( \text{cm}^{-1} \) and 11 \( \text{cm}^{-1} \) were observed
Figure 5.16. Inverse differential external efficiency versus the cavity length of 1.5 μm wavelength DCPBH lasers employing a bulk InGaAsP and a 1.8% compressively strained In$_{0.8}$Ga$_{0.2}$As/InGaAsP MQW active layer. The strained-layer MQW laser showed an $\eta_d$ as high as 80% at 265 μm cavity length. The $\eta_i$ and $\alpha_i$ for the bulk and MQW lasers were deduced to be 0.62 and 67 cm$^{-1}$, and close to 100% and 13 cm$^{-1}$, respectively.

for 1.6% tensile strained In$_{0.32}$Ga$_{0.68}$As SQW and MQW lasers, respectively, which also indicates a small intervalence band absorption. Tensile strained QW lasers show more constant internal efficiencies of about 60-70%, independently of the quantum well number which reflects the larger heterobarrier carrier leakage due to the small electron confinement as shown in figure 5.10.

Hydrostatic pressure measurements which have proven to be a useful tool for investigating loss mechanisms in semiconductor lasers [48] were performed [49] to confirm the above hypothesis. Hydrostatic pressure only increases the direct bandgap at about 10 meV/kbar without affecting the subband dispersion. Therefore, the point in k-space at which intervalence band absorption processes can be active moves to larger k-values where the hole carrier density and hence the absorption are less probable. Assuming that the carrier
concentration in the active layer becomes pinned at threshold, Auger recombination has no effect on the external differential efficiency \( \eta_d \). So, changes in \( \eta_d \) are primarily due to variations in intervalence band absorption. These changes can be carefully controlled by hydrostatic pressure and their magnitude provides a measure of the strength of intervalence band absorption.

Unmounted chips with cavity lengths ranging from 500 to 1000 \( \mu m \) were immersed in an iso-pentane:n-pentane=1:1 mixture in a high pressure Cu-Be cell [50] with a sapphire window, and were measured using short current pulses at a low repetition rate to avoid heating of the devices. The temperature was kept at 295±1K and the light output was detected by a calibrated Ge photodiode outside the cell. Observation of the light-current characteristics up to 1.5 times the threshold current yielded the variation the differential efficiency with pressure. By comparison, measurements were performed under the same conditions on bulk InGaAsP DH and on MQW lasers employing unstrained In\(_{0.51}\)Ga\(_{0.47}\)As, 1.8% compressively strained In\(_{0.46}\)Ga\(_{0.54}\)As, and 1.6% tensile strained In\(_{0.32}\)Ga\(_{0.68}\)As quantum wells, respectively. All lasers emitted at 1.5 \( \mu m \) wavelength. Figure 5.17 shows the normalised \( \eta_d \) against the hydrostatic pressure. As can be seen there is a dramatic increase in \( \eta_d \) for the bulk InGaAsP and the unstrained MQW devices. Similar changes have already been discussed.

**Figure 5.17.** Normalised efficiency against hydrostatic pressure for bulk InGaAsP (+), unstrained In\(_{0.51}\)Ga\(_{0.47}\)As/InGaAsP (\( \Delta \)), 1.8% compressively strained In\(_{0.46}\)Ga\(_{0.54}\)As/InGaAsP (\( \circ \)), and 1.6% tensile strained In\(_{0.32}\)Ga\(_{0.68}\)As/InGaAsP (\( \bullet \)) MQW lasers operating at 1.5 \( \mu m \) wavelength.
elsewhere [51], and indicate the reduction of the intervalence band absorption by the hydrostatic pressure. By contrast, for the first time in any laser investigated, almost no change in \( \eta_d \) with pressure in both the 1.6% tensile and 1.8% compressively strained quantum well lasers were observed. This may be interpreted by assuming that modification of the valence band structure by the grown-in strain has already removed intervalence band absorption and hence the application of hydrostatic pressure is ineffective.

Additionally, studies performed independently by Fuchs et al. [52] demonstrate the insensitivity of the absorption on the hole density in compressively strained 1.5 \( \mu \)m wavelength InxGa\(_{1-x}\)As/InGaAsP QW lasers. This also leads to the conclusion that intervalence band absorption has become negligible.

Measurements of the internal loss and the hydrostatic pressure measurements confirm that the high differential efficiencies observed for strained-layer MQW lasers result from a significant reduction of the intervalence band absorption.

### 5.4.6. Temperature Dependence of the Threshold Current and High Temperature Characteristics

In previous sections, the band structure modifications resulting from the deliberately grown-in strain and the quantum size effect were shown to reduce the intervalence band absorption. Auger recombination is generally believed the dominant cause of the poor temperature characteristics of long wavelength lasers. Reduction of Auger recombination should therefore increase the characteristic temperature, \( T_0 \), given by the empirical relation

\[
T_0 = (T_2 - T_1) \left\{ \ln \frac{J_{th}(T_2)}{J_{th}(T_1)} \right\}^{-1}
\]

(5.4)

where \( J_{th}(T_1) \) and \( J_{th}(T_2) \) are the threshold current densities at \( T_1 \) and \( T_2 \). The initial expectations for the strain-induced improvement of the \( T_0 \)-value ran high with values as high as 140K were predicted for 1.5 \( \mu \)m wavelength compressively strained MQW lasers [119]. However, a simple analysis [53] shows that incomplete removal of the Auger recombination still limits largely the maximum achievable \( T_0 \). The threshold current density can be expressed as
\[ J_{th} \propto (A_{nr} + SA)N_{th} + BN_{th}^2 + C_{AR}N_{th}^3 \]  \hspace{1cm} (5.5)

The first term \((A_{nr} + SA)N_{th}\) expresses the trap- and surface mediated non-radiative recombination current density \((A_{nr} = \sigma v N_t\), \(\sigma\) is the capture cross section of the trap, \(v\) is the velocity of electron or holes, \(N_t\) is the trap density, \(S\) is the surface recombination coefficient, \(A\) the area at which surface recombination takes place, and \(N_{th}\) the threshold carrier density). The second term \(BN_{th}^2\) is the bi-molecular radiative recombination current density \(J_R\), with \(B\) the radiative recombination coefficient, and \(C_{AR}N_{th}^3\) represents the three carrier non-radiative Auger current density \(J_{AR}\) with \(C_{AR}\) the Auger coefficient. The mono-molecular non-radiative processes depend weakly on the temperature (section 2.3.2), and are therefore ignored in this analysis. For an ideal QW laser, the radiative recombination coefficient \(B\) is proportional to the inverse temperature, whereas \(N_{th} \propto T\). For this analysis, the threshold carrier density is taken proportional to \(N_{th} \propto T^{1+x}\) \((x \geq 0)\), where \(x\) accounts for any non-ideality factors such as carriers occupying higher subbands, carrier spillover into the barrier material, or intervalence band absorption, which all may increase the temperature sensitivity of the threshold carrier density beyond linearity. The temperature dependence of the Auger coefficient is expressed by

\[ C_{AR} = C_{AR,0} \exp\left(\frac{-\Delta E}{kT}\right) \]  \hspace{1cm} (5.6)

with \(C_{AR,0}\) the Auger coefficient, and \(\Delta E\) the activation energy of the Auger process determined by the band structure. Combining eqs. (5.4), (5.5), and (5.6), together with \(B \propto T^x\), and \(N_{th} \propto T^{1+x}\), the \(T_0\)-value can be deduced as

\[ T_0 = \frac{T \left(1 + \frac{J_{AR}}{J_R}\right)}{1 + 2x + (3 + 3x \frac{\Delta E}{kT}) \frac{J_{AR}}{J_R}} \]  \hspace{1cm} (5.7)

In the ideal case, i.e. \(J_{AR}/J_R = 0\), \(\Delta E = 0\), and \(x = 0\), a \(T_0\)-value of 300K is expected around room-temperature. For GaAs/AlGaAs and InGaAs/GaAs MQW lasers \(T_0\)-values close to this prediction have been reported. In the case of incomplete removal of the Auger recombination, e.g. for \(J_{AR}/J_R = 3\), and assuming \(x = 0\) and \(\Delta E = 0\), a maximum \(T_0\) of 120K around room temperature may be expected. If \(J_{AR} > J_R\), a maximum \(T_0\)-value of 100K around room temperature may be
expected. From the above considerations it is also clear that the $T_0$-value is not a fixed figure as expressed in eq. (5.4) but depends on the temperature interval considered, and on the threshold gain or threshold carrier density because $J_{AR}/J_{K} \propto N_{th}$. An additional complication for buried heterostructure lasers arises because at threshold the devices may suffer from a leakage current which is less temperature dependent than the threshold current, and therefore artificially enhance the $T_0$-value.

![Graph showing characteristic temperature $T_0$ versus cavity length for various InGaAs quantum well lasers.](image)

**Figure 5.18.** Characteristic temperature $T_0$ (deduced between 20 and 50°C) versus the cavity length of 1.5 μm wavelength compressively strained In$_{0.8}$Ga$_{0.2}$As MQW lasers. As a reference, the $T_0$-values of lattice matched In$_{0.53}$Ga$_{0.47}$As MQW and a bulk InGaAsP laser are also shown. All lasers have the DCPBH device structure.

Figure 5.18 shows the $T_0$-values deduced from the threshold currents at 20°C and 50°C versus the cavity length of 1.8% compressively strained In$_{0.8}$Ga$_{0.2}$As quantum well lasers with four, six and eight wells in the active layer all with the same DCPBH device structure. The $T_0$-value increases for smaller threshold gain, or equivalently for a smaller carrier density per quantum well, i.e. for a fixed number of quantum wells the $T_0$-value increases with increasing...
cavity length and for a fixed cavity length, the $T_0$ increases with increasing QW number. Remarkable is the saturation of the $T_0$-values around 100K. For comparison, in figure 5.18 also $T_0$-values are shown of long cavity unstrained In$_{0.55}$Ga$_{0.45}$As MQW and bulk InGaAsP DCPBH lasers. Their $T_0$-values are lower by about 30-35K than those of the strained MQW devices. In figure 5.19, $T_0$-values of 1.5 μm wavelength lasers are plotted versus the strain. Data represented by the hourglass symbols are literature reports and show a wide scatter because the $T_0$-values were deduced over various temperature ranges, from devices with different lateral current confinement structures, quantum well numbers, and cavity lengths. Our data deduced from SQW and MQW lasers with identical device structure and cavity length are accentuated by the solid and dashed line, respectively, as a guide to the eye. We observe minimum $T_0$-values for unstrained QW lasers and improved values for both signs of the strain. This indicates a reduction of the Auger recombination for both signs of the strain. Figure 5.19 shows that no $T_0$-values over 100K have been reported supporting

![Figure 5.19. $T_0$-values of 1.5 μm wavelength lasers versus the strain in the In$_x$Ga$_{1-x}$As quantum wells; (⊕) In$_x$Ga$_{1-x}$As/InGaAsP SQW lasers, 77≤T≤300K, (+) In$_x$Ga$_{1-x}$As /InGaAsP 4QW lasers, 293≤T≤323K, (X) literature reports for devices employing SQW and MQW active layers with various cavity lengths and device structures by Alcatel, AT&T, Bellcore, BNR, BTRL, NEC, NTT, Philips and Univ. of Southern California.](image-url)
the arguments given above.

Reduction of Auger recombination in 1.5 μm wavelength strained-layer MQW lasers was also indicated by studies of the hydrostatic pressure dependence of the threshold current. In a loss-free laser, one would expect the threshold current to increase with pressure as the bandgap increases with pressure [48]. This is indeed observed for GaAs lasers [48]. By contrast, all the 1.5 μm wavelength devices show decreased threshold currents with increasing pressure. Figure 5.20 shows that the decrease in threshold current was largest for the bulk InGaAsP lasers by about 25% in 6 kbar [49]. Smaller reductions, by about 15% in 6 kbar for the 1.8% compressively strained In$_{0.8}$Ga$_{0.2}$As MQW devices [49], and about 9% in 4.5 kbar for the 1.6% tensile strained In$_{0.32}$Ga$_{0.68}$As MQW devices [54] were observed. This indicates that the non-radiative current was already reduced in the strained MQW devices and hence less reduction is observed with hydrostatic pressure. The experimentally observed pressure dependencies of the threshold current could be fitted (solid curves) by assuming coefficients for intervalence band absorption and Auger

![Figure 5.20](image)

**Figure 5.20.** Normalised threshold currents (T=22°C) of 1.5 μm wavelength lasers employing bulk InGaAsP (●), 1.8% compressively strained In$_{0.8}$Ga$_{0.2}$As 4QW (●), and 1.6% tensile strained In$_{0.32}$Ga$_{0.68}$As 4QW (●) active layers against hydrostatic pressure. The curves through data points of bulk and compressively strained MQW lasers are fits, the curve for tensile strain is a guide to the eye.
recombination of 50 cm\(^{-1}\) (per 10\(^{18}\) holes) and 1x10\(^{28}\) cm\(^6\)/s for bulk lasers, and zero cm\(^{-1}\) and 3.5x10\(^{29}\) cm\(^6\)/s for the compressively strained devices. The Auger coefficients deduced are in excellent agreement with data reported by Zou et al. [55] for 1.5 µm wavelength 1.8% compressively strained InGaAsP MQW lasers. According to estimates, the Auger coefficients for compressively strained QW lasers should be ten orders of magnitude less for the CHHS process, and three orders of magnitude for the CHCC process compared to bulk devices [56, 57]. Instead, we observed a reduction of the Auger coefficient by about a factor of three.

The \(T_0\)-value is not a fixed figure as indicated above. From the relative threshold currents at various temperatures, the temperature dependence of the \(T_0\)-value can be deduced. For the high temperature characteristics of semiconductor lasers not only the magnitude of the \(T_0\)-value but also its temperature dependence is important. This is different for compressively strained MQW, tensile strained MQW and bulk InGaAsP lasers. Figure 5.21 shows normalised threshold currents plotted on a logarithmic scale versus the temperature of various cavity length 1.8\% compressively strained In\(_{0.4}\)Ga\(_{0.6}\)As 4QW lasers, of

\[\frac{I_{th, T}}{I_{th, 20°C}}\]

Temperature (°C)

**Figure 5.21.** Normalised threshold currents (\(I_{th}\))/\(I_{th}(20°)\)) against temperature of 250, 500, and 1000 µm cavity length 1.8\% compressively strained In\(_{0.4}\)Ga\(_{0.6}\)As 4QW lasers (curves a, b, and c, respectively), of a 1200 µm cavity length bulk InGaAsP laser (curve d), and of a 500 µm cavity length 1.6\% tensile strained In\(_{0.32}\)Ga\(_{0.68}\)As 4QW laser (curve e).
a 500 μm cavity length 1.6% tensile strained In$_{0.32}$Ga$_{0.68}$As 4QW lasers, and of a 1200 μm long bulk InGaAsP active layer laser, respectively. The bulk active layer device can be characterised by a fairly constant $T_0$-value between 20-90°C. Compared to the bulk device, the compressively strained MQW lasers show a much stronger temperature dependence, i.e. the $T_0$ degrades with increasing temperature. For example, the 500 μm cavity length compressively strained MQW device shows smaller normalised threshold currents, i.e., higher $T_0$-values than the bulk laser up to 90°C, whereas the 1 mm cavity length device shows lower normalised threshold currents than the bulk device up to over 100°C. If the effect of the compressive strain is to reduce the in-plane hole effective mass, it will increase the activation energy for the band-to-band Auger processes which will therefore become more temperature sensitive. Assuming the CHSH the dominant process, the activation energy for the Auger recombination is given by (eq. 2.38)

$$\Delta E_{CHSH} = \frac{m_{so}}{2m_{hh} + m_c - m_{so}} (E_g - \Delta_{so})$$  \hspace{1cm} (5.8)

The heavy hole, $m_{hh}$, spin split-off, $m_{so}$, and conduction band effective masses, $m_c$, were taken to be 0.15$m_0$, 0.12$m_0$, and 0.037$m_0$, respectively, and the direct bandgap $E_g$ and the spin split-off bandgap $\Delta_{so}$ of the well material were taken to be 0.79 and 0.39 eV, respectively. With the Auger coefficient of $3.5\times10^{-29}$ cm$^6$/s this leads to the observed temperature dependence of the threshold current with temperature for the compressively strained MQW devices. Similar consideration may be given to phonon-assisted Auger processes [120], which become more temperature dependent as well with decreasing hole effective mass.

Tensile strain was expected to significantly reduce the $T_0$-values because of the reduced conduction band discontinuity leading to excessive heterobarrier carrier leakage. For 1.6% tensile strained In$_{0.32}$Ga$_{0.68}$As/InGaAsP ($\lambda_s=1.15$ μm) quantum wells, the conduction band discontinuity nearly vanishes as shown in figure 5.10. Opposed to this prediction, the $T_0$-value was found to increase with increasing tensile strain as shown in figure 5.19. This indicates that carrier spillover into the barrier/SCH layers must be of minor importance. In these tensile strained MQW lasers the bandgap difference is nearly completely present in the valence band giving a strong confinement to the holes (figure 5.10) which causes an electrostatically induced confinement of the electrons [18]. As shown in figure 5.21, the relative threshold currents of the 500 μm cavity length tensile strained MQW laser is significantly smaller, or the $T_0$-value is larger, than for the compressively strained MQW laser with the same cavity length. For 1.6% tensile strain the hole effective mass is calculated to be reduced to 0.2 $m_0$ [16].
Therefore, similar to compressive strain, the activation energy for Auger recombination is expected to be increased. However, a large advantage for the tensile strained QW devices arises from their larger well thickness, by about a factor of 5 to 7, compared to the compressively strained quantum wells. This reduces the threshold carrier density in the tensile strained devices by the same amount and the non-radiative Auger current by the third power of this thickness ratio. This reduction of the Auger recombination in tensile strained devices is also indicated by the temperature dependence of the threshold current at different hydrostatic pressures [58] as shown in figure 5.22. The bulk InGaAsP device clearly shows a decreased temperature sensitivity at a higher pressure. This indicates a reduction of the Auger recombination by the application of hydrostatic pressure which in turn has reduced the temperature sensitivity of the threshold current resulting in an increased $T_0$ from 68 to 95K in the temperature range 150-300K. On the other hand, for the tensile strained MQW laser, the

![Threshold Current Density vs Temperature](image)

Figure 5.22. Temperature sensitivity of the threshold current density of 1.6% tensile strained In$_{0.32}$Ga$_{0.68}$As MQW and bulk InGaAsP lasers at different hydrostatic pressures. The $T_0$-value of the bulk InGaAsP laser improves from 68 to 95K by applying 4.6 kbar hydrostatic pressure, whereas the tensile strained MQW device is virtually independent on hydrostatic pressure; $T_0$-values are 95 and 104K at zero and 4.9 kbar, respectively.
Figure 5.23. CW light-current characteristics up to 140°C heatsink temperature of a 500 μm cavity length 1.6% tensile strained In_{0.33}Ga_{0.68}As MQW SIBPH laser with as-cleaved facets. The laser shows lowest order TM polarized emission at 1.5 μm wavelength.

temperature dependence of the threshold current remains approximately constant at the two pressures yielding T_0-values of 95 and 104K at zero and 4.9 kbar hydrostatic pressure, respectively. Both observations, the smaller threshold current and its smaller temperature dependence support the conclusion that the introduction of tensile strained quantum wells has reduced the Auger recombination.

The stronger temperature dependence of the characteristic temperature for compressively strained lasers limits their maximum CW operating temperature to about 120 °C, whereas for tensile strained lasers a record maximum CW operating temperature for 1.5 μm wavelength lasers up to 140 °C was observed as shown in figure 5.23 [59].

5.4.7. Realization of Sub-mA Threshold Current λ=1.5 μm Lasers

The low threshold current density strained-layer quantum well material facilitates the fabrication of low threshold current lasers which are of interest for pre-bias-free modulation and for applications in low-thermal crosstalk interconnects. In order to achieve low threshold currents, narrow stripe, short cavity length, low leakage current devices are required. Additionally, in order to obtain the low threshold current density operation similar to that of long
cavity length lasers shown in section 5.4.3, the laser facets should be high reflectivity coated to minimize the cavity loss.

SIPBH lasers employing 1.2% compressively strained In$_{0.7}$Ga$_{0.3}$As double quantum well and 1.6% tensile strained In$_{0.32}$Ga$_{0.68}$As SQW active layers were fabricated. These structures showed minimum threshold current densities for 50 μm oxide stripe lasers as shown in figure 5.12. CW light-current characteristics up to 100°C of 150 μm cavity length high-reflectivity (R$_f$=92%, R$_c$=98%) coated compressively and tensile strained In$_{x}$Ga$_{1-x}$As quantum well lasers are shown in figures 5.24 and 5.25, respectively. At 10°C, both lasers show threshold currents as low as 0.8 mA, and at 10 mA drive current the front facet output powers are about 1 mW. At 60 °C, the threshold currents are still the same at 2.5 mA for both lasers, whereas at 100 °C the tensile strained SQW laser shows a lower threshold current than the compressively strained laser: 7.5 mA versus 10 mA. This again confirms the finding that the temperature dependence of tensile strained QW lasers is smaller than of compressively strained QW lasers. By further increasing the front-facet reflectivity to 98%, the threshold current of the tensile strained SQW laser still further decreased to as low as 0.62 and 0.72 mA at 0°C and 10°C [23], respectively, as shown in the inset of figure 5.25. For

![Diagram](image)

**Figure 5.24.** CW light-current characteristics up to 100°C heatsink temperature of 150 μm cavity length HR/HR' (R$_f$=92%, R$_c$=98%) coated 1.2% compressively strained In$_{0.7}$Ga$_{0.3}$As 2QW-SIPBH laser. At 10°C the device shows a threshold current as low as 0.8 mA.
both signs of the strain in the quantum wells the threshold currents reduced to as low as 115 μA at 93K as shown in figure 5.26. At 70K, the compressively strained device showed a minimum threshold current as low as 96 μA. These data also demonstrate the small leakage current of the SIPBH device structure. The threshold currents reported in this section are the lowest for 1.5 μm
1.5 μm Strained-Layer InGaAs QW Lasers

wavelength lasers as shown in the summary given in table 5.3.

![Threshold Current vs Temperature Graph](image)

**Figure 5.26.** Temperature dependence of threshold current of 1.2% compressively strained In$_{0.7}$Ga$_{0.3}$As 2QW (C) and 1.6% tensile strained In$_{0.32}$Ga$_{0.68}$As SQW (T) SIPBH lasers. At 93K both devices show threshold currents as low as 115 micro-amperes; at 70K, the compressively strained device shows 96 μA threshold current. As a reference, threshold currents of AlGaAs (λ=0.85 μm) and strained InGaAs/GaAs (λ=0.96 μm) lasers are given [60].

**Table 5.3.** Summary of sub-mA CW threshold current 1.5 μm wavelength lasers around room temperature. All devices were fabricated on (001) InP substrates, except in ref. [63] (311)B InP substrates were used.

<table>
<thead>
<tr>
<th>$I_{th}$ (mA)</th>
<th>Strain</th>
<th>Structure</th>
<th>Reference</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.62</td>
<td>tension</td>
<td>BH, InP-Fe block</td>
<td>P.J.A. Thijs et al. [22]</td>
</tr>
<tr>
<td>0.80</td>
<td>compression</td>
<td>BH, InP-Fe block</td>
<td>P.J.A. Thijs et al. [22]</td>
</tr>
<tr>
<td>0.88</td>
<td>compression</td>
<td>BH, reversed pn</td>
<td>J.S. Osinski et al. [61]</td>
</tr>
<tr>
<td>0.90</td>
<td>compression</td>
<td>BH, InP-Fe block</td>
<td>H. Temkin et al. [62]</td>
</tr>
<tr>
<td>0.90</td>
<td>compression</td>
<td>BH, InP-Fe block</td>
<td>P.J.A. Thijs et al. [63]</td>
</tr>
<tr>
<td>0.98</td>
<td>compression</td>
<td>BH, reversed pn</td>
<td>C.E. Zah et al. [64]</td>
</tr>
</tbody>
</table>
5.4.8. Linewidth Enhancement Factor and Linewidth

The linewidth enhancement factor, $\alpha_{\text{lt}}$, is a key parameter that determines the spectral properties of semiconductor lasers and semiconductor laser amplifiers. It characterizes the spectral linewidth, $\Delta v = (1 + \alpha_{\text{lt}}^2)$ [65], and the frequency chirp under modulation, $\Delta v_{\text{chirp}} = (1 + \alpha_{\text{lt}}^2)^{1/3}$ [66], due to fluctuations in the carrier density altering the refractive index. A reduction of the linewidth enhancement factor therefore leads to an enhancement of the optical telecommunication capacity due to the reduction of the linewidth and frequency chirping. The linewidth enhancement factor is defined as the ratio of the variation in effective refractive index, $n_{\text{eff}}$, with the carrier density, $N$, to the variation in optical gain, $g$, with the carrier density by

$$\alpha_{\text{lt}} = \frac{-4\pi}{\lambda} \frac{dn_{\text{eff}}}{dN} \frac{dg}{dN}$$  \hspace{1cm} (5.9)

where $\lambda$ is the wavelength of the light. The variations of the refractive index and of the modal gain with the carrier density may be deduced from the spontaneous emission spectrum below threshold. The mode gain is determined according the Hakki and Paoli [67] method from

$$g = \frac{1}{L} \ln \left( \frac{\sqrt{r} - 1}{\sqrt{r} + 1} \right) + \frac{1}{L} \ln \frac{1}{R} + \alpha_i$$  \hspace{1cm} (5.10)

where $r$ is the modulation depth, i.e. the ratio of an intensity maximum and the adjacent minimum in the spontaneous emission spectrum, and $L$, $R$ and $\alpha_i$ are the cavity length, mirror reflectivity, and internal loss, respectively. The variation in the index of refraction is determined from

$$dn_{\text{eff}} = \frac{\lambda}{2L\Delta\lambda} d\lambda$$  \hspace{1cm} (5.11)

where $\Delta\lambda$ is the Fabry-Perot mode spacing and $d\lambda$ is the wavelength shift with dN variation in injected carrier density. The variation in the index of refraction consists of the anomalous dispersion due to band-to-band electronic transitions, $n_{\text{ad}}$, the plasma term due to the free carrier plasma effect, $n_{\text{pl}}$, and the thermal contribution, $n_\text{t}$. In our measurements, the latter is eliminated by using small injection currents preventing heating of the epi-side-down mounted devices on large Cu heatsinks. In TE polarized lasers (bulk, unstrained MQW, and compressively strained MQW active layer devices) $n_{\text{pl}}$ arises from carriers in the
quantum wells, which can move in the plane of the active layer or of the quantum wells, parallel to the E-vector of the optical field, and due to carriers present in the barrier/SCH layers. These carriers produce according Newton’s second law a polarization proportional to the electrical field and reduce the refractive index by [68]

\[ n_{pl} \approx \frac{-e^2 \lambda^2}{8 \pi \varepsilon_0 c^2 n_e m_r} \]  

(5.12)

where \( e \) is the electron charge, \( \varepsilon_0 \) is the permittivity in vacuum, \( c \) is the speed of light, \( n_e \) is the refractive index of the quantum well or the barrier/SCH layer, and \( m_r \) is the reduced effective mass given by

\[ m_r = \frac{m_e m_i}{m_e + m_i} \]  

(5.13)

Tensile strained quantum well lasers show TM polarized emission where the E-vector of the optical field is perpendicular to the plane of the active layer. The quantum confinement eliminates the carrier movement in the wells parallel to the E-vector of the optical field, and thus also the contribution to the plasma effects is eliminated in these lasers [69].

Based on calculations, it was predicted that the linewidth enhancement factor would be reduced compared to bulk active layer devices by using unstrained MQW active layers. A further decrease by using active layers employing quantum wells grown under compressive strain was predicted [Table 5.1, 70, 71]. Linewidth enhancement factors deduced according to the procedure given above are plotted in figure 5.27 versus the detuning, i.e. the transition photon energy \( E_T \) minus the bandgap photon energy \( E_L \), for bulk InGaAsP, unstrained \( \text{In}_{0.53}\text{Ga}_{0.47}\text{As} \), 1.2% compressively strained \( \text{In}_{0.7}\text{Ga}_{0.3}\text{As} \), and 1.6% tensile \( \text{In}_{0.32}\text{Ga}_{0.68}\text{As} \) MQW lasers, all emitting at 1.5 \( \mu \)m wavelength. At the gain peak wavelength, i.e. \( E_T = E_L \), linewidth enhancement factors as low as 1.7 for compressively strained MQW, and even as low as 1.5 for tensile strained MQW lasers were observed, respectively. This is a significant improvement over unstrained \( \text{In}_{0.53}\text{Ga}_{0.47}\text{As} \) MQW and bulk InGaAsP lasers showing \( \alpha_{th} \)-parameters of 3 and 5, respectively. As indicated in figure 5.27, a further reduction of the linewidth enhancement factor can be obtained by increasing the wavelength detuning to positive \( E_T - E_L \) values as is possible in distributed feedback lasers. A smallest linewidth enhancement factor as low as 0.8 is deduced for compressively strained MQW lasers at 20 meV detuning. Linewidth enhancement factors around 2 at the lasing wavelength of 1.5 \( \mu \)m compressively
Figure 5.27. Linewidth enhancement factors versus detuning $E_T - E_L$ of bulk InGaAsP (■), unstrained In$_{0.53}$Ga$_{0.47}$As/InGaAsP($\lambda_g = 1.15$ μm) 4QW (○), 1.2% compressively strained In$_{0.7}$Ga$_{0.3}$As/InGaAsP($\lambda_g = 1.15$ μm) 4QW (●), and 1.6% tensile strained In$_{0.32}$Ga$_{0.68}$As/InGaAsP($\lambda_g = 1.15$ μm) 4QW (○) lasers.

strained MQW lasers were also reported by Dutta et al. [72] and Kikuchi et al. [73].

At the gain peak wavelength, the linewidth enhancement factor of the tensile strained MQW laser is smaller than for the compressively strained device (1.5 versus 1.7 as shown in figure 5.27). Two effects can be put forward as possible explanations: first, in tensile strained MQW lasers there is no contribution of the plasma effect arising from the quantum wells, and the number of carriers spilled over into the barrier/SCH layers is sufficiently small, as indicated in section 5.4.6, that this does not deteriorate the linewidth enhancement factor. Secondly, the differential gain of 1.6% tensile strained MQW devices is larger than for 1.8% compressively strained MQW devices as indicated in figure 5.28 comparing the gain of semiconductor laser amplifiers versus the injection current. However, the plasma contribution to the refractive index arising from carriers in the barrier/SCH layers may be responsible for the floor in the linewidth enhancement factor of tensile strained MQW lasers upon detuning. This effect is illustrated in figure 5.29 where the linewidth enhancement factors are plotted versus the detuning of two compressively strained MQW lasers having different barrier/SCH compositions. One, represented by the open circles, employs 1.3 μm wavelength InGaAsP
1.5 μm Strained-Layer InGaAs QW Lasers

![Graph showing net gain and single pass gain versus current for different layer structures.]

**Figure 5.28.** Gain at the peak wavelength versus the drive current for laser amplifiers employing bulk InGaAsP and In$_x$Ga$_{1-x}$As/InGaAsP MQW active layers with the wells grown under 1.8% compressive and 1.6% tensile strain. The net gain includes 2 x 3 dB coupling loss.

barrier/SCH layers, and the other, represented by the filled circles, employs 1.15 μm wavelength InGaAsP barrier/SCH layers. The heterobarrier carrier leakage gives a saturation at a linewidth enhancement factor at about 2.5 for the device with the 1.3 μm barrier layers, whereas the device employing the 1.15 μm InGaAsP barriers, as a result of the negligible heterobarrier carrier leakage, shows a monotonically decreasing linewidth enhancement factor with the detuning [69].

The reduction of the linewidth enhancement factor is reflected in narrow linewidth single mode lasers as shown in figure 5.30. A minimum linewidth of 300 kHz was observed from the compressively strained MQW lasers employing the 1.15 μm wavelength InGaAsP barrier layers, whereas the device with the 1.3 μm wavelength barrier layers showed a minimum linewidth of 2.5 MHz [74]. For comparison, also a bulk InGaAsP laser is shown with a minimum linewidth
Figure 5.29. Linewidth enhancement factor versus detuning of 1.2% compressively strained In$_{0.7}$Ga$_{0.3}$As 4QW lasers employing $\lambda_g=1.3$ μm InGaAsP barrier/SCH layers (○), and $\lambda_g=1.15$ μm InGaAsP barrier/SCH layers (•).

Figure 5.30. Linewidth against output power of 1 mm cavity length 1.5 μm wavelength DFB lasers employing bulk InGaAsP (■), 1.2% compressively strained In$_{0.7}$Ga$_{0.3}$As 4QW lasers employing $\lambda_g=1.3$ μm InGaAsP barrier/SCH layers (○), and $\lambda_g=1.15$ μm InGaAsP barrier/SCH layers (•).

of 13.2 MHz demonstrating the clear advantages of the compressively strained MQW lasers.

Figure 5.31 gives a summary of the best linewidths reported so far. A minimum linewidth as low as 170 kHz was observed by Okai et al. [75] in unstrained MQW distributed feedback (DFB) lasers employing a so-called corrugation-pitch-modulated grating to facilitate a homogeneous internal optical
field distribution by reducing the spatial hole burning effect. By reducing the parasitic noise from the laser current source and by using special measuring conditions in vacuum, the 170 kHz linewidth reduced to 56 kHz [76]. Without these precautions, Bissessur et al. [77] demonstrated 70 kHz linewidth for compressively strained MQW-DFB lasers. The present linewidths are sufficiently small for demanding coherent telecommunication systems such as 2.5 Gbit/s wide frequency shift keying (FSK) which requires a linewidth per laser (identical transmitter and local oscillator linewidth) of about 500 kHz.

In addition, reduced chirp widths have been reported. In a comparative study Hirayama et al. [86] reported a reduction of the FWHM of the chirp width of 16.1 Å, to 8.0 Å and 6.1 Å for 1.5 μm wavelength bulk, unstrained MQW and compressively strained MQW lasers under 10 Gbit/s modulation, respectively. A minimum chirp width as low as 2.0 Å under 10 Gbit/s modulation was reported recently for compressively strained MQW-DFB lasers by Matsui et al. [87].

![Graph](image)

**Figure 5.31.** Summary of reported linewidths versus the output power of unstrained In$_{0.33}$Ga$_{0.47}$As MQW (●), and compressively strained In$_{x}$Ga$_{1-x}$As MQW (○) lasers (A: Hitachi’92 [76], B: Alcatel’92 [77], C: Hitachi’90 [75], D: OKI’91 [78], E: OKI’91 [79], F: NEC’90 [80], G: Philips’92 [81], H: Fujitsu’90 [82], I: Bellcore’91 [83], J: CNET’91 [84], K: Philips’91 [85]).

### 5.4.9. High Power Characteristics

The output power $P$ of a semiconductor laser can be written as
\[ P = \eta_i(T_j) \frac{h \nu [I-I_{th}(T_j)]}{e} \frac{\alpha_m}{\alpha_m + \alpha_i} = \eta_d(T_j) [I-I_{th}(T_j)] V_j \]  \hspace{1cm} (5.14)

where \( \eta_i(T_j) \) is the internal efficiency, \( h \nu \) is the photon energy, \( I \) and \( I_{th}(T_j) \) are the drive and threshold current, \( e \) is the electron charge, \( \alpha_i \) and \( \alpha_m \) are the internal and mirror loss, \( \eta_d(T_j) \) is the external differential efficiency, and \( V_j \) is the junction voltage. The parameters with index \( (T_j) \) depend on the junction temperature \( T_j \) which can be inferred from

\[ T_j = T_{amb} + R_{th} \left\{ V_j [I_{th}(T_j) + (I-I_{th}(T_j))] [1 - \eta_d(T_j)] + R_s I^2 \right\} \]  \hspace{1cm} (5.15)

where \( T_{amb} \) is the ambient temperature, \( R_{th} \) is the thermal resistance, and \( R_s \) is the series resistance. For maximum front facet output power, the front facet external differential efficiency should be maximized by applying a low reflectivity coating, whereas the rear facet should be high reflectivity coated. The maximum output power of a semiconductor laser is either limited by catastrophic optical degradation (COD) of the front facet, by roll-over of the light-current characteristic due to the device heating, due to gain-saturation (section 5.4.10), or simply due to breaking down of the current blocking structure of the device. Fortunately, for InP-based lasers the COD level is estimated to be as large as 100 MW/cm² [88]. This corresponds to output powers in the order of 1 W for narrow stripe devices. Following eqs. (5.14) and (5.15), and using up to date parameters, maximum CW output powers around 0.5 W at room temperature are predicted [121]. With increasing cavity length, both the thermal- and series resistance decrease but also the external differential efficiency does. So, for maximum output power, the cavity length must be optimized. Moreover, a device structure which confines the current to the active layer up to very high driving conditions is required. From the preceding sections it is clear that strained-layer quantum well lasers are excellent candidates for high power applications. By the incorporation of strain, the threshold current, the temperature sensitivity of both the threshold current and the external differential efficiency are reduced. Also the internal loss is significantly reduced. This enables the fabrication of long cavity length lasers without serious penalty on the differential efficiency. Figure 5.32 shows the CW output powers of coated \((R_s=4\%, R_a=98\%)\) compressively strained In_{0.8}Ga_{0.2}As/InGaAsP 4QW DCPBH lasers with cavity lengths of 500, 750 and 1500 \( \mu \)m, mounted p-side down on Cu-heatsinks. With increasing cavity length, the external differential efficiency decreases, but due to the fact that the L-I characteristic rolls-over at larger drive currents, the maximum output power increases to 200 mW for 1500 \( \mu \)m cavity
Figure 5.32. CW output power versus the drive current of AR-HR coated strained In\textsubscript{0.8}Ga\textsubscript{0.2}As/InGaAsP 4QW-DCPBH lasers. An output power as high as 200 mW was obtained from 1500 μm cavity length devices.

length lasers [89]. A record CW output power as high as 325 mW [24] was obtained from 1000 μm cavity length AR-HR coated compressively strained In\textsubscript{0.7}Ga\textsubscript{0.3}As/InGaAsP MQW-SIPBH lasers as shown in figure 5.33. This significantly larger output power compared to the DCPBH devices reflects the larger external differential efficiency and the excellent current blocking of the SIPBH lasers. Assuming a spotsize of 0.5x1.5 μm\textsuperscript{2} (=near field dimensions), the optical power density at the front facet is about 43 MW/cm\textsuperscript{2}. Despite this high power density never sudden failures due to COD were observed in agreement with ref. 88.

Tensile strained QW lasers also show high power operation although somewhat lower output powers than for compressively strained devices were measured. From 1.6% tensile strained In\textsubscript{0.32}Ga\textsubscript{0.68}As/InGaAsP 4QW SIPBH lasers a maximum CW output power of 150 mW [56] was observed, whereas SQW lasers showed CW output powers up to 220 mW [23] due to the lower internal loss. These data represent the highest output powers from TM polarized lasers.
Figure 5.33. CW light-current characteristic of a 1000 μm cavity length AR-HR coated GRINSCH-In\textsubscript{0.7}Ga\textsubscript{0.3}As MQW-SIPBH laser showing an output power as high as 325 mW.

reported to date. All data reported were measured on devices mounted on Cu-heatsinks. A reduction of the thermal resistance with a consequent improvement of the maximum output power may be expected by using diamond heatsinks instead of copper.

The high power lasers emitting at 1480 nm wavelength find their application as pump laser in Er\textsuperscript{3+}-doped fibers for all optical amplifiers. Pump lasers with fiber coupled output powers in excess of 100 mW at 500 mA drive current, and operating up to 70°C case temperature were realized.

A summary of CW output powers reported for narrow stripe BH semiconductor lasers emitting around 1.5 μm wavelength is given in figure 5.34. These data clearly demonstrate the potential of the recently developed strained-layer QW lasers, especially those with the quantum wells grown under compressive strain, for high power applications.
Figure 5.34. Summary of CW output powers from narrow stripe 1.5 \textmu m wavelength BH semiconductor lasers. (OKI’87 [90], NEC’89 [91], Philips’89 [89], Furukawa’90 [93], OKI’90 (140 mW) [92], AT&T’90 (140mW) [94], OKI’90 (180 mW) [95], BTRL’90 [96], AT&T’90 (206 mW) [97], NEC’90 [98], OKI’91 [99], SEI’91 [100], Philips’91 [24], Philips’92 (TM polarization) [23], SEI’92 [101]).

5.4.10. High Speed Characteristics

By a small signal analysis using two coupled rate equations (one for the carrier density, and the other for the photon density in the cavity), the high speed dynamics of a semiconductor laser can be modelled, and the modulation bandwidth is determined by its relaxation oscillation frequency $f_r$ and the damping rate $\gamma$. The relaxation oscillation frequency can be written as

$$f_r = \frac{1}{2\pi} \sqrt{\frac{v_g}{h \nu} \frac{\Gamma \alpha}{V_{act} \alpha_m} \frac{dG}{dN} P} = \frac{1}{2\pi} \sqrt{\frac{v_g}{e V_{act}} \frac{\Gamma \eta_i}{dN} \frac{dG}{dN} (I-I_{th})} \quad (5.16)$$

where $v_g$ is the group velocity in the cavity, $\alpha$ is the total cavity loss, $\Gamma$ is the optical confinement factor, $dG/dN$ is the differential gain, $h\nu$ is the photon energy, $V_{act}$ is the active layer volume, $\alpha_m$ is the mirror loss, $P$ is the emitted optical output power, $\eta_i$ is the internal efficiency, and $I$ and $I_{th}$ are the drive current and threshold current, respectively. Above threshold, when the photon density in the cavity far exceeds the spontaneous emission coupled into the optical mode, the damping rate $\gamma$ due to the nonlinear gain varies linearly with $f_r^2$ according to [102].
\[ \gamma = K f_r^2 + \frac{1}{\tau_s} \]  
(5.17)

where the proportionality constant is called the K-factor, and \( \tau_s \) is the carrier lifetime. The maximum intrinsic -3dB modulation bandwidth is determined solely by the K-factor according

\[ f_{-3db} = \frac{2\pi \sqrt{2}}{K} \]  
(5.18)

In the series of devices employing bulk, unstrained MQW, and strained-layer MQW active layers, the differential gain and the optical output power are both increased, and the threshold current and the optical output power are both increased, and the threshold current is reduced as shown in the preceding sections, while the reduced \( \Gamma \) for QWs can be compensated for by using a larger number of quantum wells. For maximum \( f_r \), also the active layer volume should be minimized which implies in practice that the cavity length is minimized. A large number of quantum wells in the active layer is then beneficial to overcome the large mirror losses. For strained-layer quantum wells, strain-compensation, i.e. the wells and barrier/SCH layers grown under opposite strain as will be shown in chapter 6 for \( \lambda=1.3 \text{ \mu m} \) wavelength strained-layer MQW lasers, offers the opportunity to increase the number of quantum wells without penalty on the reliability. Theoretically -3dB bandwidths as large as 90 GHz for compressively strained In\(_{0.4}\)Ga\(_{0.6}\)As/InP MQW laser were predicted [71, 103], provided the RC parasitics are small enough to eliminate this limitation.

We estimated the maximum intrinsic modulation bandwidth from relative intensity noise (RIN) spectra of our 1.8% biaxially compressed In\(_{0.4}\)Ga\(_{0.2}\)As 4QW DCPBH and 1.6% tensile strained In\(_{0.3}\)Ga\(_{0.7}\)As 4QW SIPBH lasers. The RIN measurements are a parasitic free means of determining the potential modulation performance of a laser and this method was applied since our devices were not designed for minimized RC parasitics. For these measurements 250 \( \mu \text{m} \) cavity length devices were used. Figure 5.35 shows the damping factor \( \gamma \) versus \( f_r^2 \) obtained by fitting RIN spectra to the standard rate equation solution [104]. K-factors of 0.22 ns and 0.58 ns, resulting in maximum modulation bandwidths of 40 GHz and 15 GHz, significantly smaller than the values predicted theoretically, were obtained for the tensile and compressively strained MQW lasers, respectively. Small K-factors were obtained, despite the small number of quantum wells, the rather thick barriers (\( \approx 200 \text{ \AA} \)), and the low doping levels used (nominally undoped barrier layers in the tensile strained laser and only \( 5 \times 10^{17} \text{ cm}^{-3} \) p-type modulation doped barrier layers in the compressively strained
Figure 5.35. Plot of the damping constant \( \gamma \) versus the square of the relaxation oscillation frequency for strained-layer In\(_{x}\)Ga\(_{1-x}\)As MQW lasers. Mainly due to the larger differential gain, the tensile strained (TS) MQW laser exhibits a smaller K factor and, therefore, a larger maximum intrinsic modulation bandwidth than the compressively strained (CS) MQW device.

MQW lasers). The difference in K-factor for the two types of lasers is mainly due to a difference in differential gain of over a factor of two (\( f_0^2 = 3.6 \text{ GHz}^2/\text{mW} \) and 7.7 \( \text{GHz}^2/\text{mW} \) for the compressively and tensile strained MQW lasers, respectively). For both types of strain, the carrier lifetime was about 0.2 ns, as can be deduced from figure 5.35. (This is an additional indication that the reduced threshold current for tensile strain, reported in section 5.4.3, is not due to larger carrier lifetimes.) Typical K-factors for conventional bulk InGaAsP lasers are estimated to be about 0.35 ns [102], corresponding to a maximum modulation bandwidth of about 25 GHz.

Direct measurement of the modulation response even yielded smaller -3dB bandwidths than extrapolated from the K-factors using eq. (5.18). The K-factor used to determine the maximum bandwidth in bulk laser diodes [102] was recently shown not to yield a linear relationship of damping factor/(resonance frequency)\(^2\) for strained-layer MQW lasers operated at high output powers. In addition, at high output powers, required to achieve large resonance frequencies (eq. 5.16), the \( f_0/\sqrt{P} \) relation is sublinear. This is introduced into eq. (5.16) by writing the gain as
\[ g = \frac{g_0}{(1 + \varepsilon_{nl}S)} \equiv g_0(1 - \varepsilon_{nl}S) \]  

(5.19)

where \( g_0 \) is the gain solely determined by the material parameters, \( \varepsilon_{nl} \) is the gain compression factor, and \( S \) is the photon density in the cavity. The physical origins of the gain compression factor have been attributed to various phenomena like cavity standing wave dielectric gratings [105], spectral hole burning [106], and carrier heating [107]. Recently, it has been shown that, particularly in quantum well lasers, the carrier transport in the SCH-layers is also an important, and often the dominant effect in the high speed performance [108].

Table 5.4. Listing of structure of active layer and device, observed -3dB bandwidth, \( f_{3dB} \) (GHz), at the drive current given within brackets, and estimated intrinsic bandwidth, \( f_{3dB}^{int} \) (GHz).

<table>
<thead>
<tr>
<th>Structure of active layer and device</th>
<th>( f_{3dB} ) (GHz)</th>
<th>( f_{3dB}^{int} ) (GHz)</th>
<th>Ref.</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.98 ( \mu \text{m} ) compressively strained InGaAs/GaAs 4QW, modulation doped, FP</td>
<td>30 (114 mA)</td>
<td>63</td>
<td>[109]</td>
</tr>
<tr>
<td>1.55 ( \mu \text{m} ) compressively strained InGaAs/InGaAsP 7QW, modulation doped, FP</td>
<td>25 (180 mA)</td>
<td>40</td>
<td>[110]</td>
</tr>
<tr>
<td>1.3 ( \mu \text{m} ) bulk InGaAsP, FP</td>
<td>24</td>
<td>44</td>
<td>[111]</td>
</tr>
<tr>
<td>1.5 ( \mu \text{m} ) tensile strained InGaAs/InGaAsP 4QW, FP</td>
<td>--</td>
<td>40</td>
<td>this work [24]</td>
</tr>
<tr>
<td>1.3 ( \mu \text{m} ) compressively strained InGaAsP/InGaAsP 10QW, FP</td>
<td>20 (90 mA)</td>
<td>32</td>
<td>[112]</td>
</tr>
<tr>
<td>1.55 ( \mu \text{m} ) compressively strained InGaAs/InGaAsP 8QW, DFB</td>
<td>19 (71 mA)</td>
<td>--</td>
<td>this work</td>
</tr>
<tr>
<td>1.55 ( \mu \text{m} ) unstrained InGaAs/InGaAsP 16QW, modulation doped, FP</td>
<td>17 (190 mA)</td>
<td>--</td>
<td>[113]</td>
</tr>
</tbody>
</table>
Table 5.4 shows the record -3dB bandwidths at the operating currents within brackets, and the estimated intrinsic -3dB bandwidths. This table shows that the bandwidths, despite improvements compared to bulk lasers, are still behind the theoretical predictions. In addition, most bandwidths were realized at impractical large drive currents using Fabry-Perot lasers which cannot be used in telecommunication systems because of their multimode spectrum. By increasing the number of compressively strained InGaAs QWs to eight, we were able to increase the -3dB bandwidth in DFB lasers to 19 GHz at 71 mA drive current.

5.4.11. Reliability of λ=1480 nm Strained-Layer InGaAs MQW High Power Lasers

High power pump lasers are key-components in Erbium\(^3\)-doped fiber amplifiers (EDFAs) used as booster-, regenerator-, and pre-amplifiers in optical communication systems. Optical pumping at 980 nm using a strained-layer InGaAs/GaAs laser is more efficient in terms of dB gain per milliwatt pump power, and shows lower noise figures, by about 1 dB in practice [114]. On the other hand, a pump source at 1480 nm wavelength has the advantages of a significantly wider pumping bandwidth (15 nm versus 3 nm at 980 nm for 3 dB gain compression [114]), an inherent slower degradation rate, and immunity to so-called "sudden-failures" due to the formation of dislocation networks and/or catastrophic optical degradation (COD) of the facets because this laser consists of InP-based materials [88]. Recently, reliable high power operation of 1480 nm unstrained InGaAs MQW pump lasers for 2000 hours at 70°C [115] and of strained-layer InGaAs(P) MQW pump lasers for 10,000 hours at 40°C [25] and 600 hours at 35°C [116] were reported.

In this section, the performance and reliability at high operating temperatures and high output powers of 1480 nm strained-layer InGaAs/InP MQW pump lasers is reported. In total 233,000 hours of testing at 70°C heatsink temperature (100°C junction temperature) and 80-90 mW output power without any device failure were recorded. From these measurements a projected median lifetime as large as 3.25x10^5 hours at 70°C with a small standard deviation of 0.43x10^5 hours is deduced.

The base wafers for the 1480 nm pump lasers were grown by low-pressure organometallic vapour phase epitaxy (LP-OMVPE) on 2-inch n-type InP substrates using precursors and conditions given in section 3.5.5. The active layers consist of four 0.8% compressively strained InGaAs quantum wells of about 3 nm in thickness, embedded within a two-step graded index separate confinement heterostructure (GRINSCH) of lattice matched InGaAsP. The
excellent quality and homogeneity of the 1480 nm strained InGaAs MQW wafers is indicated by the intense room temperature photoluminescence (PL) emission with full width at half maximum linewidths as low as 40 meV. The standard deviation of the PL wavelength across the 2-inch wafer (excluding an outer edge of 5 mm) is as low as 2 nm, which facilitates good control of the emission wavelength of the lasers at 1480 nm. Subsequently, using procedures described in section 5.1, DCPBH lasers were fabricated. Devices of 1 mm in
length with the facets coated to $R_{\text{front}}=5\%$ and $R_{\text{rear}}=95\%$, to enhance the front facet output power, were mounted junction down on silicon heatsinks on carriers.

The 1480 nm strained-layer MQW pump lasers show threshold currents of typically 30 mA at room temperature. At 325 mA drive current typically 100 mW output power is obtained, which increases to 160 mW at 500 mA drive current. This implies a differential efficiency of 0.34 mW/mA with good linearity of the light-current characteristics as a result of the long laser cavity. The 1480 nm pumping bandwidth (3dB gain compression) is about 15 nm [114], so the emission wavelength is an important parameter. Figure 5.36 shows the room temperature wavelength distributions measured at 80 mW CW output power for lasers mounted from three different wafers. Averaged emission wavelengths of 1482.2±7.9 nm (231 lasers), 1482.3±5.4 (111 lasers), and 1482.0±4.5 nm (75 lasers), were measured for the number of lasers given within brackets. Twelve pump lasers from each wafer were subjected to a burn-in at 100°C heatsink temperature, 500 mA drive current. For the 1 mm cavity length pump lasers mounted on silicon heatsinks a thermal resistance of about 35-40°C/W was deduced, implying about 130°C junction temperatures during these tests. All lasers passed the stress-test with an averaged increase in threshold current of 2%. For further evaluation of the reliability, 24 lasers (12 samples from two different wafers), which did undergo neither a burn-in nor any other screening, were lifetested for 9500 hours at 70°C, 500 mA constant drive current fixing any possible degradation due to carrier injection. In this lifetest the junction temperature was approximately 100°C, and under these conditions the lasers emit CW output powers ranging from minimum 60 up to 80-90 mW for the majority of the devices. Figure 5.37 shows the normalised threshold currents versus testtime; an averaged increase of 2% was observed during 9500 hours. In the analysis, one laser was excluded because after 4500 hours its monitored output power was significantly higher than its initial value, whereas the threshold current remained the same. Its was verified that this was caused by a defective detector. Figure 5.38 shows the cumulative lognormal distribution of lifetimes, defined in this case as the time required for a 50% reduction in output power from the initial value. For the extrapolation, the output decreasing rate was assumed to be constant at the value observed in the 9500 hours test time. A projected median lifetime at 70°C is estimated to be as large as 3.25x10^5 hours (37 years) with a small standard deviation of only 0.43x10^5 hours. In the lifetests, however, the reduction in output power is observed to saturate with increasing test time resulting in longer lifetimes in practice. Another definition of lifetime often used is concerned with the increase in threshold current. In the case of the end-of-life definition as the time required for 50% increase in the
Figure 5.37. Normalised threshold currents at 70°C heatsink temperature of strained-layer InGaAs MQW lasers versus testtime.

Figure 5.38. Cumulative lognormal distribution of lifetimes of 1480 nm strained-layer InGaAs MQW lasers operated at 70°C heatsink temperature, 500 mA drive current. Using a reduction of the output power by 50% as end of life definition, a median lifetime of $3.25 \times 10^5$ hours (over 37 years) is obtained.

initial threshold current, we obtained at 70°C a median lifetime of $2.54 \times 10^5$ hours with a dispersion of 0.31. Long wavelength InGaAs(P)/InP lasers are insensitive to mirror degradation which is in sharp contrast to the GaAs-based
lasers, including 980 nm strained InGaAs/GaAs pump lasers [117, 118]. Figure 5.39 shows the decreasing rate of the output power versus the output power of the individual pump lasers at 70°C. Within the range of testing from 60-90 mW, no correlation is found between these figures. In addition, the independence of the degradation rate on the output power was confirmed up to 150 mW CW output power (20°C, 500 mA drive current, 500 hours testtime). None of the 20 lasers tested failed, and the variations in threshold current, differential efficiency, and drive current for 100 mW output power were less than 1%.

![Graph](image)

**Figure 5.39.** Decreasing rate of the output power versus the output power of AR-HR coated 1480 nm strained-layer InGaAs MQW lasers. The absence of a correlation confirms the insensitivity of long wavelength lasers to facet degradation.

In this section, the fabrication of 1480 nm lasers employing strained-layer InGaAs/InGaAsP MQW active layers with excellent performance and wavelength control was demonstrated. In total 233,000 hours of testing at 70°C heatsink temperature and under high power operation without any device failure were recorded. A projected median lifetime as large as 3.25×10^5 hours at 70°C with a small standard deviation of 0.43×10^5 hours is deduced. These data clearly demonstrate the applicability of 1480 nm strained-layer InGaAs/InP MQW lasers as reliable pump sources in EDFAs.
5.5. Conclusions

We have fabricated 1.5 μm wavelength tensile (x<0.53) and compressively (x>0.53) strained In$_x$Ga$_{1-x}$As/InGaAsP quantum well lasers showing significantly improved performance compared to unstrained (x=0.53) quantum well and conventional bulk InGaAsP lasers. For compressive strain, the intervalence band absorption is eliminated and the Auger recombination is reduced as a result of the strain-induced modification of the band structure. Consequently, the enhanced TE polarised gain results in reduced threshold current density, enabling the fabrication of buried heterostructure (BH) lasers with sub-milliamperes (0.8 mA) threshold currents, with improved differential efficiencies up to 80%, showing extremely high CW output powers (325 mW), and reduced linewidth enhancement factors ($\alpha_{nl}$=0.8) resulting in narrow linewidth lasers.

For lasers with the quantum wells in the active layer grown under both tensile strain and with a relative large thickness (>100 Å), studies of the effect of externally applied hydrostatic pressure on the laser characteristics indicate that both the Auger recombination and intervalence band absorption are reduced significantly, and that the carrier spillover into the barrier/SCH layers is the main remaining loss mechanism. For tensile strained quantum well lasers also extremely low threshold current densities (92 A/cm$^2$), low threshold currents (0.62 mA), high power operation up to 220 mW for narrow stripe BH lasers, and a small linewidth enhancement factor ($\alpha_{nl}$=1.5) were demonstrated.

For both signs of the strain, due to the incomplete removal of the Auger recombination, the $T_0$-values are not increased as much as predicted. Between 20-60°C, typical $T_0$-values for unstrained, tensile strained, and compressively strained multiple quantum well lasers range from 40-60K, 70-80K, and 90-100K, respectively. For larger temperature intervals, the increase in threshold current is smallest for tensile strained quantum well devices because due to their large well thickness, these devices operate with the smallest carrier concentrations in the well. This is reflected by the record CW operating temperature of 140°C compared to about 120°C maximum operating temperature for compressively strained QW devices both with as-cleaved facets.

The improvements achieved by the application of strain suggest enhancement of the differential gain, and together with the large output powers achieved, this was expected to result in enhanced modulation bandwidths. However, the experimental data are still far behind the theoretical predictions due to different factors causing gain saturation such as carrier transport times that were not taken into account in the theory up to now.

Reliability studies demonstrate that up to 1.8% compressively strained MQW lasers remain stable under severe operating conditions. At 70°C heatsink
temperature 0.8% compressively strained MQW devices deliver up to 90 mW output power and show a median lifetime as large as 37 years, as deduced by extrapolation.

5.6. References

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Chapter 6

High Performance λ=1.3 μm Strained-Layer InGaAsP/InP Quantum Well Lasers

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Abstract

Compressively and tensile strained InGaAsP/InP multiple quantum well (MQW) Fabry-Perot (FP) and Distributed Feedback (DFB) lasers emitting at 1.3 μm wavelength are reported. For both signs of the strain, improved device performance over bulk InGaAsP and lattice matched InGaAsP/InP MQW lasers was observed. Tensile strained MQW lasers show TM polarised emission, and with one facet high reflectivity (HR) coated the threshold currents are 6.4 and 12 mA at 20 and 60°C, respectively. At 100°C, over 20 mW output power is obtained from 250 μm cavity length lasers, and HR coated lasers show minimum thresholds as low as 6.8 mA. Compressively strained InGaAsP/InP MQW lasers show improved differential efficiencies, CW threshold currents as low as 1.3 mA and 2.5 mA for high reflectivity-coated single- and multiple quantum well active layers, respectively, and record CW output powers as high as 380 mW for HR-AR coated devices. For both types of strained quantum wells, compensation of the total strain applied by oppositely strained barrier- and separate confinement layers, results in higher intensity, narrower linewidth photoluminescence emissions, and reduced threshold currents. Furthermore, the strain-compensation is shown to be effective in improving the reliability of strained MQW structures with the quantum wells grown near the critical thickness. Linewidth enhancement factors as low as 2 at the gain peak wavelength were measured for both types of strain. Distributed feedback lasers employing either compressively or tensile strained InGaAsP/InP MQW active layers both emit single mode output powers of over 80 mW and show narrow linewidths of 500 kHz.

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6.1. Introduction

Strained-layer InGaAs(P)/InP single- and multiple quantum well (SQW- and MQW) semiconductor lasers emitting in the 1.48-1.55 μm wavelength window have been intensively studied recently. It has been demonstrated that devices employing InGaAs(P) quantum wells grown under compressive or tensile strain show superior performance over both lattice matched InGaAs(P)/InP quantum well and bulk InGaAsP active layer double heterostructure (DH) lasers. Strained-layer quantum well lasers feature significant reduction in threshold current density [1-6], enabling the realization of sub-milliampere threshold current buried heterostructure (BH) lasers [2, 7-9], improved differential efficiencies [10], extremely high CW output powers [1, 9-12], reduced linewidth enhancement factors [13, 14], and improved high temperature operation [15, 16]. For compressively strained MQW lasers, these improvements were predicted on the basis of strain-induced valence band modifications resulting in a reduced in-plane hole effective mass as well as reduced intervalence band absorption and Auger Recombination [17-19]. Measurements of the threshold current and the differential efficiency as a function of externally applied hydrostatic pressure [20], performed on 1.8% compressively strained In$_{0.8}$Ga$_{0.2}$As/InP MQW lasers [21], and gain and absorption measurements performed independently on compressively strained MQW lasers [22] indeed give strong indications that the intervalence band absorption is switched off. The temperature sensitivity of the threshold current turned out to be reduced only slightly by the application of 30 Å wide 1.8% compressively strained In$_{0.8}$Ga$_{0.2}$As QWs, however [21]. For 1.5 μm wavelength tensile strained InGaAs MQW lasers theoretically no improved performance was predicted. Only recently, after its experimental demonstration [2], the reduced threshold densities for tensile strained InGaAs MQW lasers were confirmed by calculations using an eight-band Kane-type k.p Hamiltonian without taking Auger Recombination and intervalence band absorption into account [23].

Measurements of the light-current characteristics as a function of temperature and externally applied hydrostatic pressure [24] indicate that for 120 Å wide 1.6% tensile strained In$_{0.3}$Ga$_{0.7}$As/InP MQW lasers both the Auger Recombination and the intervalence band absorption are reduced, and that heterobARRIER carrier leakage is the main loss mechanism in these lasers [25].

Semiconductor lasers emitting at 1.3 μm wavelength with similarly improved performance as observed for the 1.5 μm strained-layer InGaAs(P)/InP MQW lasers are of interest for several applications in optical fiber communication systems. Low threshold lasers are important for bias-free modulation, and low power consumption modules in local area networks. Low power consumption of lasers also leads to low thermal crosstalk integrated
devices for use in parallel optical interconnects for switching and computer networks. High output power, single mode lasers with extremely linear light-current characteristics are key components in analog cable television (CATV) systems. Considering the significant improvements obtained recently at 1.5 μm wavelength, strained-layer MQW lasers at 1.3 μm, which were reported for the first time only very recently [26-28], are promising candidates for these applications.

In this chapter, some design considerations for enhanced performance λ=1.3 μm strained-layer InGaAsP/InP quantum well lasers are given in section 6.2. In section 6.3, the fabrication and device structures of compressively as well as tensile strained InGaAsP/InP quantum well Fabry-Perot (FP) and distributed feedback (DFB) lasers are described. Details on the characterization of the InGaAsP/InP strained-layer MQW structures grown in this study are reported in section 6.4, and subsequently, the device characteristics and their reliability are reported in section 6.5. Finally, conclusions are given in section 6.6.

6.2. Design Considerations of Enhanced-Performance λ=1.3 μm Strained-Layer Quantum Well Lasers

Strained-layer ternary In_xGa_{1-x}As quantum wells will be unlikely to yield high performance 1.3 μm wavelength lasers. For example, curve (a) in figure 6.1 shows the calculated bandgap wavelength versus the well thickness of 1% compressively strained In_{0.68}Ga_{0.32}As quantum wells embedded within lattice-matched InGaAsP (λ_0=1.25 μm) barrier- and separate confinement layers. In order to obtain a lasing wavelength near 1.3 μm, the quantum well thickness has to be adjusted to 6 Å, i.e. only 2 monolayers. By reducing the bandgap wavelength of the InGaAsP barrier- and separate confinement layers to 1.05 μm, the In_{0.68}Ga_{0.32}As quantum well width has to be adjusted to 15 Å, as shown by curve (b) in figure 6.1. Due to the small optical confinement, together with the strong broadening of the gain spectrum in case of monolayer inter-well-width variations in these narrow quantum wells, these lasers are expected to show limited performance. For tensile strain, the In_xGa_{1-x}As quantum wells will also have to be grown relatively narrow, so that the upper valence subbands are brought together again by the quantum size effect, eliminating the potential advantages of the tensile strain. Therefore, for lasers at 1.3 μm wavelength, strained-layer quaternary InGaAsP quantum wells will be more suitable. The dashed curve (c) in figure 6.1 shows the well width dependence of the bandgap wavelength of 1% compressively strained InGaAsP (λ_0=1.35 μm for the unstrained bulk) quantum wells embedded within InGaAsP (λ_0=1.05 μm) barrier- and separate confinement layers. For this combination of materials, the well
Figure 6.1. Bandgap wavelength as a function of well thickness for 1% compressively strained $\text{In}_{0.68}\text{Ga}_{0.32}\text{As}$ quantum wells embedded within $\lambda_z = 1.25\ \mu\text{m}$ (curve a) or $\lambda_z = 1.05\ \mu\text{m}$ InGaAsP (curve b), and 1% compressively strained InGaAsP ($\lambda_z = 1.35\ \mu\text{m}$ for unstrained bulk) embedded within $\lambda_z = 1.05\ \mu\text{m}$ InGaAsP (curve c).

thickness has to be adjusted to about 50-55 Å in order to obtain lasing near 1.3 \mu m wavelength. Figure 6.1 suggests that for the 1.3 \mu m wavelength many other combinations of compositions (strains) and well widths of the InGaAsP quantum wells may be chosen. However, from the results obtained on 1.5 \mu m wavelength strained-layer In$_{x}$Ga$_{1-x}$As/InGaAsP/InP MQW lasers, some general indications for the most optimal combinations can be deduced.

The compressive strain lifts the heavy hole (hh)-light hole (lh) degeneracy at the valence band maximum (as shown in section 2.4.2.2.). The in-plane heavy hole effective mass is reduced resulting in reduced threshold current density and increased differential gain because of the reduced density of states at the band edge. This, combined with a large quantum size effect, which further enhances the valence subband splitting and consequently reduces the in-plane hh effective mass even more, is demonstrated to lead to further reductions of the threshold current and in the amount of non-radiative recombinations resulting in enhanced performance lasers. For 1.5 \mu m wavelength, compressive strains in the InGaAs(P) quantum wells larger than 1% and about 40-90 Å well widths were demonstrated to yield low threshold [1-8], highly efficient [1, 10] and reliable lasers [1].

On the other hand, an even more favourable valence subband structure for improved performance lasers arises for bulk tensile strained layers. The light hole levels form the top of the valence band, which makes the conduction and
valence bands real mirror reflections of one another about the center of the bandgap. This, together with the increased transition dipole moment between the conduction band and light hole valence band states [29], ensures a low threshold current density, high differential gain, and high total gain for the TM mode. However, for tensile strain, contrary to the situation for compressive strain, the quantum size effect deteriorates the performance of lasers [1, 2, 8, 16]. With decreasing well width, the lh-hh splitting decreases and the in-plane light hole effective mass increases. Therefore, for high performance tensile strained MQW lasers relatively thick (=120-160 Å), highly strained (≥1%) quantum wells should be used [1, 2, 9, 15, 16].

6.3. Fabrication and Device Structure

The wafers for this study were grown by low-pressure organometallic vapour phase epitaxy (LP-OMVPE) at 625°C using trimethyl alkyls of gallium (TMGa) and indium (TMIN), together with pure arsine (AsH₃) and pure phosphine (PH₃) as precursors. Hydrogen sulphide (H₂S), diluted to 1% in hydrogen, and diethyl zinc (DEZn) were employed as n-type and p-type dopants, respectively. Palladium-diffused hydrogen was used as the carrier gas. On 2-inch (001)±0.2° oriented n-type InP substrates double heterostructures were grown comprising 1% compressively strained InGaAsP (λₕ=1.35 μm for the unstrained bulk) or 1% tensile strained InGaAsP (λₕ=1.44 μm for the unstrained bulk) quantum wells in the active layer. The strain was applied by changing the Ga/In ratio and its magnitude was calculated from a calibration curve of the Ga/In ratio in the solid versus the TMGa/TMIN ratio in the vapour. The calibration curve was deduced from room temperature photoluminescence and X-Ray measurements performed on lattice-matched InGaAs(P) (0.75≤Eₕ≤1.26 eV) layers on InP. This curve shows that the In/Ga ratio in the solid is neither affected by the PH₃/AsH₃ ratio in the vapour phase nor by the strain up to 2%. The latter was determined by us for ternary InₓGa₁₋ₓAs/InP structures. In the case of compressive strain, one, four, eight and sixteen InGaAsP QWs in the active layer were embedded within InGaAsP barrier- and separate confinement layers were grown. These latter InGaAsP (Eₕ=1.15 eV) layers were grown lattice-matched to InP for the single, four and sixteen QW structure, and under 0.2% tensile strain, which increases the bandgap slightly to 1.17 eV, in an eight and sixteen QW structure to compensate for the strain in the QWs, thus using a similar approach as reported by Miller et al. [30] for 1.5 μm wavelength compressively strained MQW lasers. The well widths were adjusted to about 55 Å thickness to obtain room temperature photoluminescence emission near 1.3 μm wavelength.
In structures applying tensile strain, four 120-Å-wide InGaAsP QWs were grown, embedded within either unstrained InGaAsP ($E_g=1.15$ eV) or 0.3% compressively strained InGaAsP ($E_g=1.14$ eV) barrier- and separate confinement layers. The details of the strained-layer InGaAsP/InP active layer structures employed in this study are summarized in Table 6.1.

**Table 6.1.** Listing of active layer structures (number of 1% strained InGaAsP quantum wells followed by the abbreviations CS, CCS, TS, and CTS, meaning Compressive Strain, Compensated Compressive Strain, Tensile Strain, and Compensated Tensile Strain, respectively), lasertype (FP or DFB), quantum well width ($L_x$), thickness of separate confinement heterostructure (SCH) and barrier (B) layers, bandgap of SCH/B, and strain in SCH/B of the devices studied in this work.

<table>
<thead>
<tr>
<th>Structure</th>
<th>Lasertype</th>
<th>$L_x$ (Å)</th>
<th>Thickness SCH/B (Å)</th>
<th>$E_g$(SCH/B) (eV)/Strain in SCH/B (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1CS</td>
<td>FP</td>
<td>55</td>
<td>700/-</td>
<td>1.15/0</td>
</tr>
<tr>
<td>4CS</td>
<td>FP + DFB</td>
<td>55</td>
<td>550/150</td>
<td>1.15/0</td>
</tr>
<tr>
<td>8CCS</td>
<td>FP</td>
<td>55</td>
<td>550/150</td>
<td>1.17/0.2%TS</td>
</tr>
<tr>
<td>16CCS</td>
<td>FP</td>
<td>55</td>
<td>550/70</td>
<td>1.17/0.2%TS</td>
</tr>
<tr>
<td>16CS</td>
<td>FP</td>
<td>55</td>
<td>550/70</td>
<td>1.15/0</td>
</tr>
<tr>
<td>4TS</td>
<td>FP + DFB</td>
<td>120</td>
<td>550/150</td>
<td>1.15/0</td>
</tr>
<tr>
<td>4CTS</td>
<td>FP + DFB</td>
<td>120</td>
<td>550/150</td>
<td>1.14/0.3%CS</td>
</tr>
</tbody>
</table>

Furtheron in this paper, the structures will be identified by a number, indicating the number of quantum wells, followed by CS (compressive strain), CCS (compensated compressive strain), TS (tensile strain), and CTS (compensated tensile strain). The structures were used for subsequent fabrication of Semi-Insulating Planar Buried Heterostructure (SIPBH) Fabry-Perot lasers having about 1.5-2.0 μm wide mesas which were wet-chemically etched using an SiO₂ mask. An Fe-doped InP layer of about 2 μm thickness and an n-InP "zinc-diffusion-blocking-layer" [31] were then grown by selective LP-OMVPE. Ferrocene (Fe(C₅H₇)₂) was used as the iron-source. After removal of the SiO₂ mask, a p-InP and a p⁺-InGaAs contact layer were grown on the entire wafer to
complete the SIPBH structure.

Distributed feedback (DFB) lasers at 1.3 \( \mu m \) wavelength using either the 1% compressively or the 1% tensile strained InGaAsP/InP MQW active layers were also fabricated. The active layer structures used are given in table 6.1. In the DFB lasers, an InGaAsP buried grating [32] positioned above the active layer in the p-InP cladding layer, was applied. The buried grating has two major advantages. First, the coupling constant can be accurately adjusted which is of importance for linear DFB lasers for analog applications, and second high-quality quantum well structures are more easily obtained for planar substrates than for corrugated InP substrates, independently of the coupling constant. The first-order grating was fabricated into a quaternary layer, separated from the active layer by about 2000 Å p-InP, by holographic definition of the mask and subsequent wet-chemical etching. The grating was overgrown with 0.8 \( \mu m \) p-InP by LP-OMVPE. DFB lasers with the double channel planar buried heterostructure (DCPBH) lateral current confinement were fabricated by etching the channels into the wafer and subsequent liquid phase epitaxial (LPE) regrowth of p-n-p InP current blocking layers and a p+ InGaAsP contact layer. SIPBH-FP and DCPBH-DFB lasers with various cavity lengths and facet reflectivities, obtained by facet coating, were mounted p-side down on Cu heatsinks. Their characteristics are reported in section 6.5.

### 6.4. Materials Characterization

The strained InGaAsP/InP quantum well structures were characterized using photoluminescence (PL) at 300 and 4K, and cross section transmission electron microscopy (TEM). The PL measurements at 300 and 4K were performed using 647 nm light from a krypton-ion laser and 514.5 nm light from an argon-ion laser, respectively. The emission was dispersed with a 0.75 m grating monochromator and detected by a Ge detector, which was cooled to 77K in the low temperature measurements. The spectra were recorded using phase-sensitive detection.

The PL measurements are summarized in table 6.2, showing subsequently the structure of the active layer, the PL-peak wavelength, the PL-linewidth (full width at half maximum) at standard excitation density, and at reduced (by a factor of five) excitation intensity within brackets, and the PL-intensity relative to bulk InGaAsP layers. All data were measured at 300K, except in the last column, PL-linewidths at 4K are given.
Table 6.2. Summary of the room temperature PL-wavelength, PL-full width at half maximum (FWHM) linewidth under standard and reduced excitation conditions within brackets, PL-intensity relative to bulk InGaAsP and the FWHM at 4K for the different structures used in this study.

<table>
<thead>
<tr>
<th>Structure</th>
<th>$\lambda_{PL}$ (nm)</th>
<th>FWHM (meV)</th>
<th>Normalized Intensity</th>
<th>FWHM$^{4K}$ (meV)</th>
</tr>
</thead>
<tbody>
<tr>
<td>1CS</td>
<td>1320</td>
<td>46.2(33.4)</td>
<td>0.72</td>
<td>10.2</td>
</tr>
<tr>
<td>4CS</td>
<td>1324</td>
<td>39.4</td>
<td>1.7</td>
<td>10.7</td>
</tr>
<tr>
<td>8CCS</td>
<td>1322</td>
<td>39.7(32.0)</td>
<td>2.4</td>
<td>--</td>
</tr>
<tr>
<td>16CCS</td>
<td>1334</td>
<td>36.5(36.0)</td>
<td>1.8</td>
<td>9.5</td>
</tr>
<tr>
<td>16CS</td>
<td>1328</td>
<td>43.0(36.0)</td>
<td>1.7</td>
<td>--</td>
</tr>
<tr>
<td>4TS</td>
<td>1304</td>
<td>91.8</td>
<td>0.28</td>
<td>--</td>
</tr>
<tr>
<td>4CTS</td>
<td>1335</td>
<td>86.9</td>
<td>0.33</td>
<td>--</td>
</tr>
<tr>
<td>Bulk</td>
<td>1310</td>
<td>55.8</td>
<td>1</td>
<td>7.8</td>
</tr>
</tbody>
</table>

The compressively strained InGaAsP QW structures all show very intense and narrow-linewidth PL emissions. Under reduced excitation intensity, the room temperature PL-linewidths become as narrow as 32-33 meV (=1.3 kT) even for the 8CCS and 16CCS structures. This indicates the reproducibility in thickness and composition of the quantum wells grown in this study. Table 6.2 also shows that the tensile strained InGaAsP QW structures have much wider PL-linewidths and lower intensities than the compressively strained structures which is unavoidable as band structure calculations [33] show that in these wide (120 Å) tensile strained quantum wells several energy levels are lying relatively close to each other. These levels become populated and consequently this results in a broadened shoulder at the short-wavelength side of the PL-peak. The low intensity may be explained from the small conduction band offset between the well and the barrier caused by the tensile strain [23]. In addition, the effects of the strain compensation can be observed in table 6.2. The structures with the strain compensation show more intense emissions with narrower linewidths than the structures without strain-compensation.

Figure 6.2 shows, as an example, the cross-sectional transmission electron micrograph (TEM) of the as-grown 16CS InGaAsP QW structure. The
layer thicknesses are uniform and reproducible, with very sharp interfaces between the layers. No defects are observed despite the strain.

6.5. Device Characteristics

6.5.1. Tensile Strained InGaAsP/InP QW Lasers

As-cleaved, 250 μm cavity length tensile strained InGaAsP/InP MQW lasers (4TS) show at 20°C CW threshold currents of 16 mA and differential efficiencies of 67%, whereas the strain-compensated InGaAsP MQW (4CTS) lasers show threshold currents of only 12 mA and differential efficiencies of 68% (fig. 6.3). The smaller threshold current for the strain-compensated structures was observed for two different wafers grown in different runs and is therefore believed to be real. However, the 4TS and the 4CTS lasers both show very similar $T_0$-values of 47K as deduced from the CW threshold currents at 20°C.
Figure 6.3. CW light-current characteristics of 250 μm cavity length as-cleaved tensile strained InGaAsP 4QW laser, with (4CTS) and without (4TS) strain-compensation.

and 60°C heatsink temperatures, and also the maximum operating temperatures of 85-90°C are very similar. Both structures show TM polarized light emission, demonstrating that lasing indeed occurs from electron to light hole recombinations, as expected from the strain-induced modification of the valence subband structure. A significantly improved performance is obtained by reducing the mirror loss of these short-cavity-length devices. By applying one high-reflectivity (HR) coating (R=98%), the average threshold current of the 4CTS InGaAsP-InP MQW lasers reduced from 12 to 6.4 mA at 20°C and from 30 to 12 mA at 60°C, respectively. This implies an increase in T₀-value to 63K. Figure 6.4 shows the CW light-current characteristics up to 100°C of a single facet HR coated 4CTS InGaAsP/InP MQW laser. At 20°C, about 23 mW output power is obtained at only 50 mA drive current, whereas at 100°C, the threshold current remains low at 31 mA with still over 20 mW output power obtainable at 200 mA drive current. Further reduction of the mirror loss by applying a high-reflectivity coating to the front facet (R=92%), and reduction of the cavity length to 150 μm, results in a CW threshold current of 2.0 mA at 20°C. At 100°C, a threshold current as low as 6.8 mA was measured as shown in figure 6.5. This makes these devices well-suited for high-temperature applications.
Figure 6.4. CW light-current characteristics up to 100°C of a strain-compensated tensile strained InGaAsP MQW (4CTS) laser (L=250 μm, front facet as-cleaved, rear facet coated to 98% reflectivity).

6.5.2. Compressively Strained InGaAsP/InP QW Lasers

The structural details of the lasers employing 1, 4, 8 and 16 compressively strained InGaAsP quantum wells in the active layer, including the bandgaps and thicknesses of their separate confinement- and barrier layers are given in table 6.1. InGaAsP/InP SQW (1CS) lasers with 250 μm cavity length and HR/HR’ (R=92%/98%) coated facets show a CW threshold current of 2.6 mA at 20°C, and a T₀-value of 50-55K between 20 and 60°C heatsink temperature. Figure 6.6 shows the light-current characteristics up to 100°C of a 150 μm cavity length HR/HR’ coated SQW (1CS) laser. At 10°C, the threshold current is as low as 1.3 mA, and the differential quantum efficiency is 22%.
Figure 6.5. CW light-current characteristics up to 100°C of a 150 μm cavity length strain-compensated tensile strained InGaAsP MQW (4CTS) laser applying HR-coated facets. At 100°C, a threshold current as low as 6.8 mA was measured.

These characteristics make the devices very attractive for bias-free modulation and for applications in optical interconnects. When coupled into a multimode fiber (graded index, 50 μm diameter, NA=0.2), an output power of typically 0.8-1.0 mW is obtained at 10 mA drive current. The longitudinal mode spectrum is typically multimode with about four to six modes. Figure 6.7 shows the CW threshold currents (left) and differential efficiencies (right), both at 20 and 60°C, respectively, of 250 μm cavity length lasers with as-cleaved facets for all structures fabricated. Included in figure 6.7 are also data from 1.3 μm lasers with bulk InGaAsP, as well as data on lattice-matched InGaAsP/InP 8QW (8LM) active layers, reported recently by Tsang et al. [34]. All lasers of
Figure 6.6. CW light-current characteristics up to 100°C of a 150 μm cavity length R=92%/98% coated compressively strained InGaAsP SQW (1CS) SIPBH laser. At 10°C the threshold current is as low as 1.3 mA.

Figure 6.7 have the SIPBH lateral current confinement structure.

At 20°C, the compressively strained MQW lasers employing the 4CS InGaAsP QW active layer show higher threshold currents than the compressively strained devices with larger well number. This indicates the gain saturation effects of these short cavity length lasers having a mirror loss of about 50 cm⁻¹. This is well illustrated by increasing the cavity length to 500 μm (mirror loss reduces to about 25 cm⁻¹) which results in a very small increase in threshold current from 15 to 16 mA. By increasing the number of quantum wells to eight (8CCS), the threshold current at 20°C reduces to 8.5 mA, which is similar to the threshold current reported for the lattice-matched InGaAsP MQW (8LM) lasers. For the sixteen QW lasers, the threshold currents for the strain-compensated
structure are lower than for the lasers without strain-compensation; 9.5 mA for the 16CCS versus 12 mA for the 16CS lasers, respectively. This is a similar trend as observed for the strain compensation if applied in tensile strained InGaAsP MQW lasers (4CTS versus 4TS).

At 60°C, the threshold currents of the compressively strained and lattice-matched InGaAsP/InP MQW lasers with the well number ≥8 are significantly lower than for the tensile strained, the compressively strained InGaAsP 4QW (4CS), and for the bulk InGaAsP lasers, as shown in figure 6.7. The compressively strained InGaAsP MQW lasers (8CCS, 16CS and 16CCS) show

Figure 6.7. CW threshold currents (left) and differential efficiencies (right) both at 20 and 60°C, respectively, of 250 μm cavity-length as-cleaved SIPBH lasers employing different active layer structures. The numbers indicate the number of quantum wells and the abbreviations TS, CTS, CS, CCS and LM stand for Tensile Strained, Compensated Tensile Strained, Compressively Strained, Compensated Compressively Strained, and Lattice Matched, respectively. The data of the lattice-matched MQW lasers are taken from ref. 34.
between 20 and 60°C heatsink temperature a smaller increase in threshold current, i.e. a larger $T_0$-value than the lattice-matched InGaAsP 8QW (8LM) lasers. The ratios of the threshold currents correspond to $T_0$-values of 52, 60, 60 and 42K for the 8CCS, 16CCS, 16CS and 8LM InGaAsP MQW lasers, respectively. Apart from the improved high-temperature performance of the compressively strained InGaAsP/InP MQW lasers, figure 6.7 shows another distinct improvement obtained by the application of either compressive or tensile strain, namely, the differential efficiency being significantly higher for the strained InGaAsP MQW structures than for the lattice-matched InGaAsP/InP 8QW structures [34]. From the cavity length dependence of the inverse differential efficiency, the internal efficiency and the internal loss of the 4CS InGaAsP QW lasers are estimated to be 78% and 5.6 cm$^{-1}$, respectively, 76% and 4.3 cm$^{-1}$ for two different wafers. This results in differential efficiencies as high as 70% for 250 μm cavity length as-cleaved lasers. The low internal loss is confirmed by the strong effect of the mirror loss on the threshold current. The application of one HR (R=98%) coating reduces the threshold current of 250 μm cavity length lasers almost by a factor of 2, and significantly enhances the front facet differential efficiency. These lasers emit 125 mW CW output power at 20°C, 80 mW at 60°C, and 24 mW at 100°C, respectively. The maximum output powers of these short-cavity-length lasers are limited by the thermal heating, and therefore, significantly higher output powers are expected by increasing the cavity length. For lasers with 1 mm cavity length and AR/HR (R$_e$=4%, R$_e$=98%) coatings, the differential efficiency is as high as 60% (0.565 mW/μA). As shown in figure 6.8, this results in extremely high output powers as high as 380 mW at 20°C, which is the highest ever reported output power for InGaAs(P) buried heterostructure lasers [27].

For the 8CCS InGaAsP QW lasers, from the cavity length dependence of the inverse differential efficiency, an internal efficiency of 79% and an internal loss as low as 4.3 cm$^{-1}$ were deduced. These figures are very similar to data obtained from compressively strained InGaAsP/InP 4QW devices, and this independence of the internal loss on the number of quantum wells is indicative for the reduction of intervalence band absorption in these compressively strained InGaAsP/InP MQW lasers. For lattice-matched InGaAsP QW (8LM) lasers, an internal efficiency of 67% and an internal loss of 11.5 cm$^{-1}$ were reported [34]. Consequently, a significantly lower differential efficiency was obtained, 47% versus 72% for the 8CCS InGaAsP QW lasers, as shown in figure 6.7. Furthermore, the 8CCS InGaAsP/InP MQW lasers show other very favourable characteristics. Figure 6.9 shows that by applying one HR coating a low threshold (5 mA) and over 25 mW CW output power can be obtained at 100°C.
heatsink temperature. By applying also a HR coating to the front facet and reducing the cavity length to 150 μm, the threshold current reduces to 2.5 mA at 20°C and only 6.7 and 10.2 mA at 60 and 100°C, respectively, as shown in figure 6.10. This makes these devices well-suited for applications at high temperature.

The increased differential efficiency observed for the compressively strained InGaAsP MQW structures, is very similar to observations for 1.5 μm wavelength strained InGaAs/InGaAsP MQW structures which was attributed to elimination of the intervalence band absorption. Figure 6.7 shows the lower temperature sensitivity of the threshold current and the differential efficiency of 250 μm cavity length compressively strained InGaAsP MQW lasers compared to tensile strained InGaAsP MQW and bulk InGaAsP lasers. However, the $T_0$-value is not a fixed figure but rather depends on the threshold carrier density
which may be reduced by reducing the mirror loss, i.e. by increasing the mirror reflectivity or by increasing the cavity length. This is indicated in figure 6.11 showing that the $T_0$-values (deduced from the CW threshold currents between 20 and 60°C heatsink temperature) vary from 44 to 66K for different-cavity-length 8CCS InGaAsP QW lasers with various facet coatings. Similarly, as reported in section 6.5.1, for 250 μm cavity length tensile strained InGaAsP MQW lasers the $T_0$-value increased from 47K to 63K by applying a HR coating to one facet.
Figure 6.10. CW light-current characteristics of HR/HR' coated 150 µm cavity length InGaAsP/InP compressively strained 8QW SIPBH laser. The threshold currents are as low as 2.5 and 10.2 mA at 20 and 100°C, respectively.
Figure 6.11. Dependence of the characteristic temperature on cavity length of strain-compensated compressively strained 8QW (8CCS) lasers applying various facet reflectivities.

6.5.3. Linewidth Enhancement Factor

It was predicted theoretically [35] and recently demonstrated experimentally [13, 14] for 1.5 μm wavelength strained-layer InGaAs/InP MQW lasers that the linewidth enhancement factor, $\alpha_H$, is reduced by the application of grown-in compressive as well as tensile strain. The linewidth enhancement factor is the ratio of the variation of the refractive index, $n$, with carrier density, $N$, to the variation of optical gain, $g$, with the carrier density. This can be expressed as

$$\alpha_H = -\frac{4\pi}{\lambda} \frac{dn/dN}{dg/dN} \quad (6.1)$$

where $\lambda$ is the wavelength of the light. The linewidth enhancement factor is the key parameter that determines the spectral linewidth, $\Delta v$ [36], and the wavelength chirp [37]. The changes in refractive index and the mode gain were deduced from the wavelength shift and the change in the modulation depth of the Fabry-Perot fringes as a function of carrier density. The measurements were performed at different injection currents below threshold. The linewidth enhancement factors of the 8CCS and 4TS InGaAsP QW SIBPH lasers calculated from eq. (6.1) are plotted versus the wavelength in figure 6.12. The linewidth enhancement factor strongly depends on the emission wavelength and
Figure 6.12. The linewidth enhancement factor versus the wavelength of strain-compensated compressively strained (8CCS) and tensile strained (4TS) InGaAsP MQW SIBH lasers. The gain-peak wavelengths are indicated by the arrows.

becomes small with decreasing wavelength. At the lasing wavelengths, which are indicated by the arrows, linewidth enhancement factors around 2 were measured for both structures. These data are in excellent agreement with data reported by Kano et al. [28] for compressively strained InGaAsP/InP MQW lasers at 1.3 μm. The αₚ-parameters reported here are a marked improvement compared to αₚ=5.7 and αₚ=3.5 measured at the gain peak wavelength of λ=1.3 μm bulk InGaAsP and lattice-matched InGaAsP/InP MQW lasers, respectively [38]. From figure 6.12, it is expected that the linewidth enhancement factor becomes smaller by proper wavelength detuning, as possible in a DFB laser. The reduction of the linewidth enhancement factor in both the compressively and the tensile strained MQW lasers indicates the advantage of the near symmetrical band structure caused by the biaxial strain.

6.5.4. Reliability

The compressively and tensile strained InGaAsP/InP single and multiple quantum well SIBH lasers were subjected to a stress test in which 250 μm cavity length as-cleaved lasers were operated at 150 mA drive current at 100°C heatsink temperature for 20 hours. The threshold currents were measured at 70°C
with an accuracy of 0.1 mA before and after this test. The 1CS, 4CS and 8CCS SIPBH lasers showed averaged threshold current increases of less than 0.5%. For the strain-compensated 16QW (16CCS) lasers the increase in threshold current was 1.7%; about a factor of four larger increase was observed for similar structures without strain compensation (16CS).

The tensile InGaAsP/InP MQW lasers showed somewhat higher increase in threshold current. Also here, the strain-compensated MQW lasers showed significantly less increase in threshold current than the lasers without strain compensation (6.9% for the 4CTS versus 33% for the 4TS devices). None of the lasers failed to lase after this test. For the tensile strained MQW structures more extensive studies are required, but the reliability data for 1% compressively strained MQW lasers clearly indicate long-life operation of these lasers.

6.5.5. InGaAsP/InP MQW Distributed Feedback Lasers

Distributed feedback (DFB) lasers employing compressively as well as tensile strained InGaAsP/InP 4QW active layers with the DCPBH lateral current confinement (4CS, 4TS, and 4CTS) were fabricated. Lasers with coated facets and 500 μm cavity length were evaluated. From the widths of the stop band, the coupling constant was estimated to be 15-30 cm⁻¹ (κ.L = 0.75-1.5).

Compressively strained DCPBH-DFB lasers with the rear facet coated to 80% reflectivity and the front facet as-cleaved showed threshold currents of 16 mA at 25 °C and differential efficiencies of 34%. The single mode yield, defined as the fraction of lasers showing a side mode suppression ratio (SMSR) > 30 dB at 5 mW output power, was 50%. Generally, the SMSR was larger than 40 dB. On some DFB lasers an additional AR coating (R=5%) was applied to the front facet. Figure 6.13 shows the CW front facet output power at 25°C versus the drive current of such a device as an example. Single mode output powers as high as 90 mW were obtained as shown in the inset.

Tensile strained InGaAsP/InP MQW DCPBH-DFB lasers (4TS) having the rear facet coated to 80% reflectivity and the front facet as-cleaved showed at 25°C threshold current of 18 mA and differential efficiencies of 33%. The single mode yield of these lasers was as high as 85% most probably due to the small difference between gain maximum and Bragg wavelength (≤6nm). However, strain-compensated tensile strained InGaAsP MQW DCPBH-DFB (4CTS) lasers with similar facet reflectivities showed threshold currents of only 11.5 mA at 25°C. This low threshold current was measured despite the larger wavelength detuning of about 20 nm resulting in only 10% single mode yield. The single mode yield was improved to 60% by coating the front facet to 5% reflectivity. This resulted in an increase of both the threshold current to 16 mA
Figure 6.13. CW light-current characteristic of compressively strained InGaAsP 4QW DCPBH-DFB laser with 500 μm cavity length and R=5%/80% coated facets. The insets show the single mode operation (SMSR > 40 dB) at 50 and 90 mW, respectively.

and the differential efficiency to 37%. The tensile strained InGaAsP MQW DFB lasers lased in the TM mode. AR/HR (5%/80%) coated tensile strained InGaAsP
Figure 6.14. CW light-current characteristic of tensile strained InGaAsP 4QW DCPBH-DFB laser with 500 μm cavity length and R=5%/80% coated facets. The insets show the single mode TM operation (SMSR > 40 dB) at 25, 50 and 80mW CW output power, respectively.
MQW DFB lasers also showed high single mode output powers up to over 80 mW as shown in figure 6.14. The small linewidth enhancement factors were confirmed by linewidth measurements using heterodyne detection. The 4CS InGaAsP DFB lasers (L=500 μm, R_t=5%, R_r=80%) showed linewidths of 1 MHz at 25 mW, and 500 kHz at 50 mW output power, respectively.

6.6. Conclusions

We reported 1.3 μm wavelength tensile- and compressively strained InGaAsP/InP quantum well lasers with superior performance over lattice-matched MQW and bulk InGaAsP lasers. Low threshold currents, high differential efficiencies, and high output powers even at elevated temperatures make these devices of high interest for applications in optical telecommunication systems. Similar to 1.55 μm wavelength strained-layer MQW lasers, for compressively strained 1.3 μm InGaAsP/InP MQW lasers the enhanced differential efficiency and the independency of the internal loss on the quantum well number indicates the reduction of the intervalence band absorption. This facilitates high CW power operation up to 380 mW for Fabry-Perot, and about 90 mW for distributed feedback lasers, respectively. Linewidth enhancement factors below 2 indicate the advantage of the near symmetrical band structure induced by the strain. The reduction of the total strain in the strained-layer MQW active layer by applying oppositely strained barrier- and separate confinement layers is effective to enhance the performance and reliability of those structures with a total strained-layer QW thickness near the critical thickness.

6.7. Acknowledgements

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6.8. References


Chapter 7

Summary

In this thesis, work on 1.3, 1.48, and 1.55 μm wavelength InGaAsP/InP quantum well (QW) semiconductor lasers for applications in optical fiber communication systems is reported. The 1.3 and 1.55 μm wavelength lasers are of interest because they match the minima in dispersion and attenuation of the optical fiber, respectively, while lasers emitting at 1.48 μm wavelength are of interest due to their ability to pump Er3+-doped fibers for all-optical amplifiers (so called EDFA’s).

We have demonstrated that the performance of semiconductor lasers emitting at these wavelengths can be significantly improved by modifying the composition of the quantum wells such that the QW-lattice is elastically deformed as a result of the grown-in strain. The materials for this study have been grown by low-pressure organometallic vapour phase epitaxy (LP-OMVPE).

In chapter 1, a brief historical review of the most important group III-V compound semiconductor lasers is given.

In chapter 2, it is shown that the performance of conventional InGaAsP long wavelength lasers employing bulk, i.e., about 0.2 μm thick active layers is limited due to intrinsic loss-mechanisms resulting from the valence subband structure. It can be shown theoretically that a more optimal band structure for semiconductor lasers is obtained by employing strained-layer quantum well active layers. The quantum well lattice can be grown under compressive and tensile strain by varying the composition. As a result of the elastic tetragonal deformation of the lattice, the in-plane hole effective mass and consequently the loss-mechanisms are strongly reduced resulting in improved performance lasers. The composition of the quantum wells and the interface quality in particular determine the semiconductor laser characteristics.

In chapter 3, available growth techniques for layered structures are reviewed. Their strengths and weaknesses with respect to the growth of InGaAs(P)/InP structures are emphasised. LP-OMVPE, used in this work, is discussed in greater detail.

In chapter 4, (strained-layer) InGaAs(P)/In(GaAs)P interfaces are studied. High resolution X-ray diffraction measurements show that during growth pauses group V elements at the crystal surface are exchanged by group V elements
from the vapour phase. This does not affect the flatness of InP surfaces, whereas InGaAs(P) surfaces become undulated. Using optimized gas phase switching procedures up to 1.8% compressively strained InGaAs(P)/In(GaAs)P quantum wells with abrupt and flat interfaces are grown. This is indicated by Shubnikov-de Haas measurements revealing the existence of very high mobility 2-dimensional electron gases at InP-to-InGaAs(P) ($\lambda_{sh}=1.65, 1.55, \text{ and } 1.3 \text{ \mu m}$) heterojunctions. Direct comparison of low-temperature photoluminescence spectra of quantum well structures grown on InP substrates with small ($\leq 0.2^\circ$) and larger ($\geq 2^\circ$) misorientations from the (001) plane is a powerful characterization technique. As a result of variations of the terrace length compared to the excitonic diameter, the characteristic differences in the spectra can unambiguously be ascribed to integer number atomic layer inter-well width variations. Extremely narrow photoluminescence linewidths, e.g. 8.8 meV for a 6 Å thick well of InGaAsP, confirm the quality of the layer and the abruptness of the interfaces. Maximum quantum shifts corresponding to 90% bandgap difference between well and barrier layers were measured. Flat, abrupt interfaces and quantum wells of thicknesses down to several monolayers are demonstrated by transmission electron microscopy. The abruptness and flatness of unstrained quantum wells show no dependency on the substrate misorientation up to 25° from the (001) plane. However, strained-layer (e.g. 1.8%) quantum wells show island growth at small misorientations (e.g. 2°) from the (001) plane.

In chapter 5, device structures, the fabrication, and the performance of semiconductor lasers in the 1.48-1.55 \mu m wavelength band employing In$_x$Ga$_{1-x}$As/InGaAsP quantum wells grown under 2.2% biaxial tensile ($x=0.22$) to 1.8% biaxial compressive strain ($x=0.80$) are reported.

Due to the strain-induced band structure modifications, for compressive strain the in-plane heavy hole effective mass is reduced, and the intervalence band absorption is eliminated. In addition, the Auger recombination is reduced, though less than predicted. Consequently, enhanced transverse electric (TE) polarised gain is obtained, leading to a significantly reduced threshold current density from 500 to about 100 A/cm$^2$. This enables the fabrication of buried heterostructure (BH) lasers showing sub-milliampere (0.8 mA) threshold currents, differential efficiencies up to 80%, record high CW output powers (325 mW), and narrow linewidths due to the reduced linewidth enhancement factors ($\alpha_{th}=0.8$).

For small tensile strains ($\leq 1\%$) the valence band maximum becomes (nearly) degenerate. This results in lasers with very unfavourable characteristics, e.g., very high threshold currents. Significantly improved semiconductor lasers are obtained by increasing the QW-width and/or the strain. The transverse
magnetic (TM) polarised emission demonstrates that lasing is due to electron-light hole recombinations. Extremely low threshold current densities (92 A/cm²), low threshold currents (0.62 mA), high power operation up to 220 mW for narrow stripe BH lasers, and small linewidth enhancement factors (αₚ=1.5) were demonstrated. Analysis of laser characteristics under externally applied hydrostatic pressure indicates that both the Auger recombination and intervalence band absorption are reduced by tensile strain. As a result of the small conduction band offset, an important remaining loss mechanism still concerns the carrier spillover into the barrier layers. However, this spillover is far less than expected initially due to the strong confinement of the holes, inducing an electrostatic field reducing the electron leakage.

Due to the strain-induced reduction of the in-plane effective hole mass for both signs of the strain, the temperature sensitivity of the Auger recombination is increased. As a consequence, T₀-values, a measure of the temperature dependence of the threshold current, have increased less than theoretically predicted. Typical T₀-values for unstrained, tensile strained, and compressively strained multiple quantum well lasers range from 40-60K, 70-80K, and 90-100K, respectively, for long cavity length lasers. Despite the smaller T₀-values for tensile strained quantum well devices, they benefit from their larger well widths which reduce the operating carrier density, and thus also give less Auger recombination. This is reflected by the record CW operating temperature of 140°C compared to about 120°C maximum operating temperature for compressively strained QW devices both with as-cleaved facets.

Reduction of the non-radiative-loss mechanisms leading to enhanced differential gain, together with the large output powers achieved, were expected to result in enhanced modulation bandwidths. Although strained-layer quantum well lasers have broken the existing record, the remaining discrepancy with theory is now explained by the presence of different factors which cause gain saturation, e.g. due to carrier transport times to the wells.

Reliability studies demonstrate that compressively strained MQW lasers remain stable under severe operating conditions where conventional bulk InGaAsP lasers show limited device performance. At 70°C heatsink temperature, 500 mA drive current, the devices deliver up to 90 mW CW output power. A median lifetime as large as 37 years at 70°C was deduced from extrapolation of our experimental data. In total 233,000 hours of testing at 70°C without any device were recorded.

In chapter 6, compressively and tensile strained quaternary InGaAsP/InP quantum well lasers emitting at 1.3 μm wavelength are reported. For both signs of the strain, improved device performance over bulk InGaAsP and lattice matched InGaAsP/InP quantum well lasers was observed. Compressively strained
InGaAsP/InP MQW lasers show improved differential efficiencies, CW threshold currents as low as 1.3 mA for high-reflectivity coated single quantum well active layers, and record CW output powers as high as 380 mW for HR-AR coated devices. The excellent high temperature characteristics of tensile strained quantum well lasers are confirmed. At 100°C minimum CW threshold currents as low as 6.8 mA versus 10 mA for compressively strained quantum well lasers were observed.

For both types of strained-layer quantum wells, strain-compensation applied by lattice-mismatching barrier- and separate confinement layers oppositely to the quantum wells, results in higher intensity, narrower linewidth photoluminescence emissions, and reduced threshold currents as compared to uncompensated devices. Furthermore, strain-compensation is shown to be effective in improving the reliability of strained-layer MQW structures with the total quantum well thickness grown near the critical value.

The results from this work show that properly designed and grown strained-layer quantum well lasers outperform existing long wavelength lasers. Although the work in this area is rather new, several benefits shown in this work are already explored "in the field", such as in high power 1480 nm pump sources for EDFA’s, in high power single mode and linear DFB lasers, in low threshold devices for optical interconnects, in low chirp, high speed DFB lasers, and in semiconductor optical amplifiers.
Samenvatting

In dit proefschrift worden de resultaten gerapporteerd van het onderzoek aan InGaAsP/InP quantum well (QW) halfgeleider lasers die licht uitzenden met een golflengte van 1,3, 1,48, of 1,55 μm en toegepast worden in glasvezel communicatie systemen. De golflengte van 1,3 μm is interessant omdat deze samenvalt met het minimum in de dispersie en die van 1,55 μm vanwege een minimum in de demping in de glasvezels. Lasers die licht genereren met een golflengte van 1,48 μm zijn van belang als pomplaser voor Er³⁺-gedoteerde optische fiber versterkers (de zgn. EDFA’s).

Wij hebben experimenteel aangetoond dat de eigenschappen van halfgeleider lasers bij genoemde golflengtes aanmerkelijk verbeterd kunnen worden door de samenstelling van de QW-laagjes te wijzigen zodanig dat het kristalrooster van de QW elastisch vervormd wordt als gevolg van de ingegroeide mechanische spanning. De materialen voor dit onderzoek zijn gegroeid m.b.v. organometal gasfase epitaxie by verlaagde druk; een werkwijze die in het vakgebied bekend is als "Low-Pressure Organometallic Vapour Phase Epitaxy" (LP-OMVPE).

In hoofdstuk 1 wordt een historisch overzicht gegeven van de belangrijkste halfgeleider lasers opgebouwd uit groep III-V verbindingen.

In hoofdstuk 2 blijkt dat de eigenschappen van halfgeleider lasers met "klassieke" ongeveer 0,2 μm dikke InGaAsP actieve laag beperkt worden door intrinsieke verliesmechanismen als gevolg van de structuur van de valentiebanden. Theoretisch kan worden aangetoond dat een sterk verbeterde bandenstructuur voor halfgeleider lasers wordt verkregen door in de actieve laag juist QW-laagjes met ingegroeide mechanische spanning toe te passen. Door variatie van de samenstelling kan het kristalrooster van de QW zowel onder druk- als trekspanning worden gegroeid waarbij, als gevolg van de elastische tetragonale roostervervorming, de effectieve gatenmassa en daarmee ook de verliesmechanismen afnemen, met als resultaat een sterke verbetering van de lasereigenschappen. De eigenschappen van QW lasers worden niet alleen bepaald door de samenstelling van het kristal maar vooral door de kwaliteit van de overgangen (interfaces) tussen verschillende lagen.

In hoofdstuk 3 worden de beschikbare groeitechnieken voor gelaagde structuren behandeld. Hierbij worden de sterkten en zwaktes, speciaal met het oog op de groei van InGaAs(P)/InP structuren belicht. Aan LP-OMVPE wordt extra aandacht geschonken, omdat dit de gebruikte groeitechniek in dit onderzoek is.
In hoofdstuk 4 worden de interfaces tussen InGaAs(P)/In(GaAs)P QW’s zonder en met ingegroeide spanning bestudeerd. Hoge resolutie Röntgendiffractie metingen laten zien dat tijdens een groeistop groep V elementen aanwezig op het kristaloppervlak worden uitgewisseld door groep V elementen aanwezig in de dampfase. Hierbij blijken de InP oppervlakken vlak te blijven terwijl de InGaAs(P) oppervlakken hun vlakheid verliezen. Door optimalisatie van de schakelprocedures voor de gassen zijn InGaAs(P)/In(GaAs)P QW’s tot 1,8% drukspanning gegroeid met abrupte en vlakke interfaces. Dit wordt bevestigd door Shubnikov-de Haas metingen die het bestaan van 2-dimensionale elektronengassen met zeer hoge mobiliteiten op InP-naar-InGaAs(P) ($\lambda_e=1,65, 1,55, \text{ en } 1,30 \mu m$) overgangen aantonen. Directe vergelijking van lage temperatuur fotoluminescentie spectra van QW structuren gegroeid op substraten met kleine ($\leq 0,2^\circ$) en grotere ($\geq 2^\circ$) misorientaties van het (001) kristalvlak blijkt een zeer krachtige karakterisatietechniek. Door de verlopende terrasafstand t.o.v. de excitondiameter, kunnen de karakteristieke verschillen in de spectra eenduidig worden toegeschreven aan laagdiktenvariaties van een geheel aantal atoomlagen binnen een QW. De kwaliteit van de lagen en de abruptheid van de interfaces wordt bevestigd door zeer smalle fotoluminescentieltendes (vb. 8,8 meV voor 6 \AA
dikke InGaAsP laag). Voorts zijn quantum verschuivingen overeenkomend met 90% van het verschil in bandafstand tussen de opsluit- en QW-lagen gemeten. Vlakke, abrupte interfaces en QW lagen van slechts enkele monolagen dik zijn ook aangetoond m.b.v. transmissie elektronenmicroscopie. Voor spanningsvrije QW’s is tot 25° misorientatie van het (001) vlak geen orientatieafhankelijkheid waargenomen op de steilheid en vlakheid van de interfaces. Daarentegen is voor QW’s gegroeid onder spanning eilandgroei waargenomen voor bepaalde combinaties van spanning (vb. 1,8%) en misorientatie (vb. 2°) van het (001) vlak.

In hoofdstuk 5 worden achtereenvolgens besproken: de structuren, het fabricageproces, en de eigenschappen van In$_x$Ga$_{1-x}$As/InGaAsP QW lasers met een emissie golflengte van 1,48-1,55 \mu m. De toegepaste QW’s hadden een ingebouwde spanning variërend van 2,2% trek (x=0,22) tot 1,8% druk (x=0,8).

Als gevolg van een wijziging van de bandenstructuur in de QW’s door het aanbrengen van drukspanning wordt de effectieve gatenmassa kleiner, waarbij ook de "intervalence band absorption" sterk onderdrukt wordt. Bovendien treedt er een vermindering van de Auger recombinatie op, echter kleiner dan voorspeld door de theorie. Door de onderdrukking van deze niet-stralende verliezen neemt de netto versterking toe voor het transversaal elektrisch (TE) gepolariseerde licht, wat o.a. leidt tot een belangrijke verlaging van de drempelstroomdichtheid van 500 tot ongeveer 100 A/cm². Een en ander heeft geresulteerd in lasers met sub-mA drempelstroom (0,8 mA), met differentiële
efficiencies van 80%, met recordwaarden voor het optisch uitgangsvermogen (325 mW), en met smalle lijnbreedtes als gevolg van gereduceerde "linewidth enhancement factors" ($\alpha_{1f}=0.8$).

Bij een geringe trekspanning (<1%) ontstaat een (bijna) ontarda valentiebandenstructuur. Dit heeft ernstige nadelige gevolgen voor de eigenschappen van de lasers, o.a. zeer hoge drempelstroom. Wanneer de QW-dikte en/of de trekspanning worden vergroot, worden de eigenschappen weer aanmerkelijk verbeterd. De transversaal magnetisch (TM) gepolariseerde emissie bewijst dat in dit geval elektronen en lichte gaten recombineren. Bereikte resultaten zijn: een minimale drempelstroomdichtheid van 92 $\text{A/cm}^2$, een drempelstroom van 0.62 mA, een optisch uitgangsvermogen van 220 mW en een kleine "linewidth enhancement factor" $\alpha_{1f}$ van 1.5. Bestudering van deze lasers onder uitwendig opgelegde hydrostatische druk toont aan dat zowel de Auger recombinitie als de intervalentie band absorptie verminderd zijn. I.v.m. de kleine stap in de geleidingsband is nu de lek van elektronen naar de barrièrelagen een belangrijk verliesmechanisme. Dit is echter kleiner dan tot nu toe altijd aangenomen werd doordat de gaten wel zeer goed opgesloten worden en daarmee voor een elektrostatisch veld zorgen waardoor de elektronen minder kunnen weerlekken.

Door de kleinere effectieve gatenmassa voor beide tekens van de spanning is de temperatuursafhankelijkheid van de Augerrecombinitie toegenomen. Dit heeft als gevolg dat de $T_0$-waarden, een maat voor de temperatuursafhankelijkheid van de drempelstroom, minder toenemen dan uit de theorie zou volgen. Typische waarden zijn 40-60K, 70-80K, 90-100K voor relatief lange QW lasers respectievelijk zonder spanning, met trek-, en met drukspanning. Ondanks de lagere $T_0$-waarden voor de lasers met QW's onder trekspanning, blijkt toch het voordeel van de grotere QW-dikte. Hierdoor wordt de ladingsdragersdichtheid kleiner en bij gevolg ook de Auger-recombinitie. Dit wordt geïllustreerd door maximale bedrijfstemperaturen van 140°C voor lasers met de QW's onder trekspanning tegenover 120°C voor lasers met drukspanning in de QW's. Beide resultaten werden behaald voor lasers met gekleefde spiegels, zonder toepassing van spiegelcoatings.

Door de toename van de differentiële versterking a.g.v. de afname van de niet-stralende verliesmechanismen en de hoge optische vermogens, werden grote modulatiebandbreedtes verwacht. Hoewel QW lasers met ingebouwde spanning het bestaande record verbroken hebben, wordt het verschil met de theoretische voorspelling nu verklaard door factoren die verzadiging van de versterking veroorzaken, zoals looptijden van ladingsdragers naar de QW's.

Levensduurtests hebben aangetoond dat lasers met QW's onder drukspanning zeer betrouwbaar zijn, zelfs onder zeer zware bedrijfsscondities.
waarbij "klassieke" InGaAsP lasers niet meer (zo goed) werken. De QW lasers leveren bij een stuurstroom van 500 mA een continu uitgangsvermogen van 90 mW bij 70°C. Bij deze temperatuur is de geëxtrapoleerde mediane levensduur 37 jaar. In totaal hebben we bij 70°C 233.000 laser-testuren geregistreerd zonder één defect.

In hoofdstuk 6 worden lasers die licht genereren met een golflengte van 1,3 μm met quaternaire InGaAsP QW’s zowel onder druk- als trekspanning gerapporteerd. Ook hier zijn voor beide tekens van de spanning in de QW’s verbeterde eigenschappen gemeten t.o.v. de bekende lasers met ongeveer 0,2 μm dikke InGaAsP en spanningsvrije QW actieve lagen. Voor lasers met de QW’s onder drukspanning, is de externe differentiële efficiency verbeterd, is de drempelstroom laag (1,3 mA voor lasers met één enkele QW), en is het uitgangsvermogen zeer hoog (380 mW voor lasers met spiegelcoatings). Het goede hoge temperatuurgehad van lasers met QW’s onder trekspanning wordt ook hier bevestigd. Bij 100°C hebben deze lasers een drempelstroom van 6,8 mA tegen 10 mA voor lasers met QW’s onder drukspanning.

Voor beide typen QW lasers kan de netto mechanische spanning geminimaliseerd worden door spanningsaanpassing, dit is de QW’s en de omliggende barrière lagen met een verschillend teken voor de spanning te groeien. De voordelen hiervan zijn: fotoluminescentiepikken worden hoger en smaller en de drempelstroomen worden lager t.o.v. lasers die deze spanningsaanpassing niet bezitten. Ook is de spanningsaanpassing effectief om de levensduur te verbeteren, en wel in het bijzonder voor lasers met de totale QW-dikte dicht bij de kritische dikte.

De resultaten behaald in dit onderzoek laten zien dat goed ontworpen en gefabriceerde lasers met QW’s onder spanning zich superieur gedragen t.o.v. de nu gangbare lange golflengte laser. Hoewel het werk op dit gebied relatief jong is, worden bepaalde voordelen aangetoond in dit werk al "in het veld" toegepast. Hierbij kunnen worden genoemd: 1480 nm pomplasers voor EDFA’s, lineaire "distributed feedback" (DFB) lasers met groot uitgangsvermogen, lasers met lage drempelstroom voor optische "interconnects", DFB lasers met lage "chirp" voor hoge dichtheid glasvezel telecommunicatie systemen, en optische halfgeleider versterkers.
Curriculum Vitae

Petrus Johannes Adrianus Thijs was born May 20, 1953 in Hooze, The Netherlands. After receiving the diploma from secondary school, in Sept. 1970 he started chemistry studies at the Eindhoven Polytechnic Institute, The Netherlands. The diploma was obtained in June 1974. Then, he did his military service until Sept. 1975, when he joined Philips Research Laboratories, Eindhoven, The Netherlands. Here, he became involved in Liquid Phase Epitaxial growth of AlGaAs/GaAs for semiconductor lasers and the associated characterization techniques, e.g. double crystal X-ray diffractometry and photoluminescence. While at Philips Research Laboratories, in Sept. 1977 part-time studies were started at the Eindhoven University of Technology, department of chemical technology. The academic degree was awarded "cum laude" in March 1983. Since joining Philips Research Laboratories work was carried out under the supervision of Ir. W. Nijman. With the academic degree a research scientist position was obtained. From 1982 until 1985 he worked on Liquid Phase Epitaxial growth of InGaAsP/InP and the associated characterization techniques. Since then, he started-up and studied Low-Pressure Organometallic Vapour Phase Epitaxial growth of InGaAsP/InP and AlGaInAs/InP for long wavelength devices like semiconductor lasers, semiconductor laser amplifiers, detectors, and passive elements like waveguides, couplers and switches. In 1989 work on strained-layer InGaAsP quantum wells was started.

He has authored and co-authored over 60 papers in international journals and conference proceedings, is frequently awarded invited papers, and has served as a program committee member at several international conferences.
List of Publications and Conference Contributions

Parts of this thesis have been published and reported in conference contributions marked by ⋆.


31*. P.J.A. Thijs, J.J.M. Binsma, E.W.A. Young, and W.M.E. van Gils,
"High power and high temperature operation of 1.5 micron wavelength strained-layer InGaAs/InGaAsP SIPBH lasers," Tech. Dig. 3rd Int. Conf. on InP and Related Mat., Cardiff, United Kingdom, April 1991, pp. 184-187.


38*. P.J.A. Thijs, J.J.M. Binsma, L.F. Tiemeijer, P.I. Kuindersma, and T. van Dongen, "Low-pressure organometallic vapour phase epitaxial growth and characterization of strained-layer InGaAs-InGaAsP quantum well


40*. P.J.A. Thijs, J.J.M. Binsma, L.F. Tiemeijer, R.W.M. Slootweg, R. van Roijen, and T. van Dongen, "Sub-mA threshold operation of $\lambda=1.5$ \mu m strained InGaAs MQW lasers grown on (311)B InP substrates," Tech. Dig. 4th Int. Conf. on InP and Related Mat., Newport, USA, April 1992, paper THD4, pp. 461-464.


42*. P.J.A. Thijs, J.J.M. Binsma, L.F. Tiemeijer, and T. van Dongen, "Sub-mA threshold current (0.62 mA) and high output power (220 mW) 1.5 \mu m tensile strained InGaAs single quantum well lasers," Electron. Lett., 28, 829-830 (1992).


54. L.F. Tiemeijer, J.J.M. Binsma, P.J.A. Thijs, T. van Dongen, R.W.M. Slootweg, and J.M.M. van der Heijden, "Polarization insensitive 1300 nm amplifiers employing both compressively and tensile strained
quantum wells in a single active layer," Tech. Dig. 5th Int. Conf. on InP and Related Mat., Paris, France, April 1993, pp. 91-94.


List of publications and conference contributions


