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Optical and electrical quality of InGaP grown on GaAs with low pressure metalorganic chemical vapour deposition

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The growth of InGaP has been investigated in a low pressure MOCVD reactor. Experiments have been performed in order to study the influence of several parameters on the growth process, such as the influence of the growth temperature, V/III ratio and total gas flow on the solid composition and on the growth rate. The optical and electrical quality of the grown InGaP layers is discussed in relation to the amount of lattice mismatch of the epilayers. Finally the morphology of the epitaxial InGaP layers is described. A defect, typical for the growth of InGaP on GaAs, is presented. A hypothesis of the origin of this defect is given which results in a method to avoid its formation.

1. Introduction

The last few years there has been a great interest in the ternary III/V semiconductor $\text{In}_{1-x}\text{Ga}_x\text{P}$, especially when lattice matched to GaAs due to its potential in optoelectronic devices as lasers and solar cells [1–4]. It can be employed as active layer in laser diodes, in particular in combination with AlInGaP acting as a barrier [2]. In this way it is possible to obtain lasers operating near $0.67 \mu\text{m}$. Even shorter wavelengths can be realized if quantum well heterostructures consisting of these two semiconductors are used [2,5–7]. A second field of interest for this material is that of high efficiency solar cells. In order to obtain these high efficiencies, it is inevitable to use multi junction solar cells. Recently a tandem solar cell consisting of a GaAs bottom and an InGaP top cell, interconnected with a GaAs tunnel diode, demonstrated an efficiency of 27.3% [3].

The great interest in this material can be explained when its material qualities are compared with the more often used ternary alloy AlGaAs. When lattice matched to GaAs, InGaP has a direct bandgap of around 1.9 eV [1,8,9] which is as large as AlGaAs containing around 40% alu-

minum. This is about the maximal bandgap before AlGaAs becomes indirect. However, InGaP suffers none of the problems which are connected with AlGaAs. When growing AlGaAs with MOCVD one has to be extremely careful to avoid traces of oxygen and humidity in the reactor [10,11]. These contaminants will lead to high resistivity and bad optical quality of the samples. Contrary to AlGaAs, InGaP is relatively insensitive toward traces of oxygen and water [10]. This will ease the growth of InGaP as compared to AlGaAs. As a result of this the GaAs/InGaP interface employs a remarkable low recombination velocity, as low as 1.5 cm/s [12], which is about 30 times lower than the best values reported for the AlGaAs/GaAs interface (53 cm/s [13]). Another advantage of InGaP, lattice matched to GaAs, compared with AlGaAs is the fact that it possesses no so called DX centers. So it is no problem to grow highly doped n-type InGaP in contrast to AlGaAs in which a decrease of the free carrier concentration is observed with increasing aluminum content while the dopant input concentration is kept constant [14].

Although it seems that InGaP has only advantages over AlGaAs and no disadvantages, still there are several drawbacks. One of the main

problems of this material is the composition control. Independent of the growth method used (LPE, VPE, MOCVD, MBE and GSMBE) control of the alloy composition must be within 0.15% absolute, because of the large differences in lattice constants of InP, GaP and GaAs [15]. In growing AlGaAs this problem does not occur because of the small difference in lattice constant of AlAs and GaAs. Furthermore the difference in thermal expansion coefficients between GaAs and InGaP [15] can cause additional stress in the grown epilayers upon cooling down from growth to room temperature.

Additionally, when InGaP is grown with MOCVD techniques, the material exhibits a typical behaviour. Depending on the growth conditions like the growth temperature, growth rate, V/III ratio and substrate orientation, the bandgap can differ between high (≈ 1.9 eV) and low (≈ 1.8 eV) values for a fixed composition, lattice matched to GaAs [16]. The shift in bandgap energy can be as large as 100 meV at room temperature [17,18]. This behaviour of the bandgap energy is thought to originate from a CuPt type of ordering on the group III sublattice. Ordered material shows low values of the band gap energies and completely disordered exhibits the highest values [16–21].

In this paper we will present the growth and characterization of high quality InGaP material grown in a low pressure MOCVD reactor. The influence of the growth temperature, V/III ratio, total gas flow and gas phase composition on the quality of the InGaP epilayers is investigated. Electrical and optical qualities of the grown material are examined with $C-V$, Hall and photoluminescence measurements. Our samples exhibit, to our knowledge, the highest mobilities reported in literature so far.

2. Experimental procedure

All samples were grown in a commercial available low pressure MOCVD reactor with a horizontal reactor cell [22]. Pure arsine (AsH_3) and phosphine (PH_3) were used as group V species. For the group III species trimethyl gallium (TMG)

and trimethyl indium (TMI) were used as precursors. All growth experiments were performed on (100) 2° off towards [110] oriented GaAs wafers. The growth runs were performed at a reactor pressure of 20 mbar varying the growth temperature between 580 and 700°C. The total gas flow through the reactor was chosen to be 5 or 7 SLM, respectively, while the V/III ratio was varied between 50 and 400. The mole fractions of TMG and TMI varied between 0.0043% and 0.0065% each. The resulting growth rate varied, depending on the growth temperature and total gas flow and amount of group III precursor, between 1 and 2.2 $\mu\text{m/h}$.

The solid composition of the layers was determined using electron microprobe analysis (EPXMA) and X-ray diffraction. Transmission electron microscopy (TEM) and high resolution transmission electron microscopy (HRTEM) were used to investigate dislocations in the InGaP epilayers and to determine the amount of ordering. All samples were electrically characterized by Hall–Van der Pauw measurements performed at room temperature. Temperature dependent Hall–Van der Pauw measurements down to 4.2 K were performed at some selected samples.

The optical material quality of the grown $\text{In}_{1-x}\text{Ga}_x\text{P}$ epilayers was investigated by photoluminescence measurements. These measurements were performed at room temperature and 4.2 K. At the latter temperature, the sample was mounted in a cryostat with the sample in He exchange gas. Optical excitation was provided by the 2.41 eV (514.5 nm) line from an Ar^+ laser with an excitation density of 5.3 W/cm^2 . The luminescence was dispersed by a double monochromator fitted to a cooled photomultiplier tube with a S1 response.

3. Results and discussion

3.1. Influence of the growth temperature on the solid composition and growth rate of $\text{In}_{1-x}\text{Ga}_x\text{P}$

Experiments were done in order to study the influence of the growth temperature on the composition of the $\text{In}_{1-x}\text{Ga}_x\text{P}$ epilayers and on the

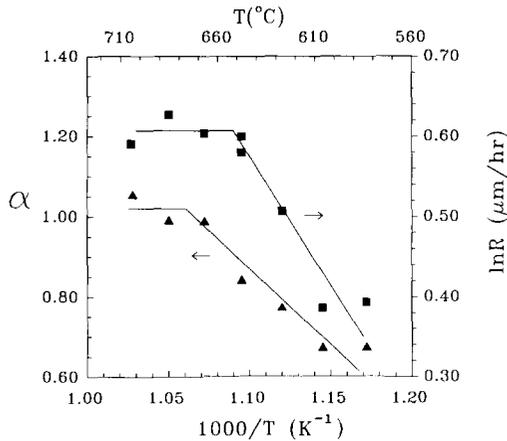


Fig. 1. Plot of ratio of rates of incorporation of gallium and indium growth species (α) and growth rate (R) versus reciprocal temperature ($1000/T$). α is defined as $\bar{r}_{\text{TMG}}/\bar{r}_{\text{TMI}}$, where \bar{r}_{TMG} is the concentration independent growth rate of GaP and is \bar{r}_{TMI} the concentration independent growth rate of InP as defined in section 3.1.

growth rate of this material. In these experiments the growth temperature was varied between 580 and 700°C keeping the gas phase composition constant. This implies that the fraction trimethylgallium in the gas phase X_G ($= [\text{TMG}]/([\text{TMG}] + [\text{TMI}])$) and the V/III ratio, i.e. the ratio between the amount of phosphine and the total amount of group III growth components (TMG + TMI), were unchanged. The total amount of group III species in the gas phase was 0.018%, the V/III ratio was 250.

The growth rate is plotted versus the reciprocal temperature (K^{-1}) in fig. 1. From this figure it is clear that starting from a growth temperature of about 650°C and higher the growth rate of $\text{In}_{1-x}\text{Ga}_x\text{P}$ is temperature independent, implying that the growth rate in this temperature region is gas phase diffusion limited. Because of the high V/III ratio used (250), the group III growth species are the growth rate limiting species. The diffusion of these growth components through the gas phase toward the growing surface is the limiting factor in the growth rate of $\text{In}_{1-x}\text{Ga}_x\text{P}$ at these temperatures. At growth temperatures below 650°C, the growth is temperature dependent with an apparent activation energy $E_{\text{act}} = 5$ kcal/mol, as can be deduced from fig. 1. This

value for E_{act} indicates that the growth is determined by gas phase diffusion coupled to gas phase decomposition of, at least one of the group III growth components, i.e. TMG or TMI. However, in growing GaAs in the same reactor at the same temperature and pressure conditions, the same apparent activation energy (4.2 kcal/mol) has been found [22]. This means that in this temperature region probably the decomposition of TMG in the gas phase coupled to the diffusion through the gas phase of Ga growth species is the limiting factor in the growth of $\text{In}_{1-x}\text{Ga}_x\text{P}$. This is consistent with the fact known from literature that the decomposition of TMG is completed at higher temperatures than that of TMI [23–28]. When gas phase decomposition kinetics would solely determine this process an apparent activation energy of about 20 kcal/mol would have been found [29].

In fig. 1, also the ratio of the incorporation rates of gallium and indium species, α , is plotted versus the reciprocal temperature. This ratio α is defined as $\bar{r}_{\text{TMG}}/\bar{r}_{\text{TMI}}$. In this expression \bar{r}_{TMG} is the concentration independent growth rate of GaP and \bar{r}_{TMI} is the concentration independent growth rate of InP [30]. The observed growth rate is given by

$$r = C_0 \bar{r},$$

where C_0 is the input concentration of the group III component. Explicit expressions for \bar{r} are not really needed here. For the regimes where transport through the gas phase is rate determining or where the chemical kinetics dominate, it is given that [30] for the diffusionally limited process:

$$\bar{r}(D_0, v_0, h, z, T) = A \frac{D_0}{h} \exp\left(-b \frac{D_0}{v_0 h} \frac{z}{h}\right),$$

and for the kinetically limited process:

$$\bar{r}(k_0, T) = k_0 \exp\left(-\frac{E_a}{RT}\right),$$

where D_0 is the binary diffusion coefficient of the group III component at room temperature, v_0 is the mean horizontal gas flow velocity, h is the free height above the susceptor, z is the coordinate along the susceptor (heating starts at $z = 0$),

A and B are temperature-dependent dimensionless numbers, which are a weak function of T and the thermal diffusion factor α_T only, k_0 is the pre-exponential factor of the rate constant, E_a is the activation energy for chemical reaction or reaction enthalpy for equilibria, T is the growth temperature and R is the gas constant. A detailed discussion of these equations is given by Van Sark et al. [30].

At temperatures lower than 660°C, the factor α decreases with decreasing temperatures. The decreasing factor α indicates decreasing (concentration independent) growth rate of GaP. This supports our view that decomposition of TMG is most probably the rate limiting factor of the growth rate at these temperatures.

At higher temperatures, above 660°C, α levels off to unity. This result again points to a gas phase diffusion limited growth process at these temperatures. The difference in diffusion coefficient between Ga and In growth species is small enough ($\approx 15\%$) to ensure that it has hardly any noticeable effect on the solid state composition.

In fig. 2, the ratio of the fraction of gallium in the solid state (X_{Ga}) and the fraction indium in the solid state ($1 - X_{Ga}$) is plotted versus the ratio of the input partial pressure TMG and the input partial pressure TMI for three different growth

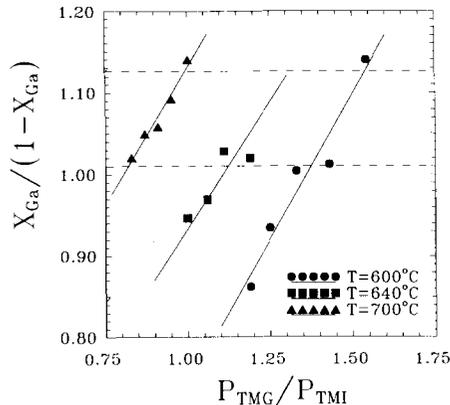


Fig. 2. Plot of ratio of the fraction of gallium in the solid (X_{Ga}) to the solid phase fraction of indium ($1-X_{Ga}$) in the $In_{1-x}Ga_xP$ epilayer as a function of the ratio of input mol-fractions TMG (P_{TMG}) and TMI (P_{TMI}) at three different growth temperatures. The dashed lines represent the lattice matched region around $X_{Ga}/(1 - X_{Ga})$.

temperatures. In these experiments the input partial pressure TMI is kept constant and the input partial pressure TMG is varied. At all three growth temperatures this ratio ($X_{Ga}/(1 - X_{Ga})$) varies linearly with the ratio of the input partial pressures (P_{TMG}/P_{TMI}). In all these situations the slope is about 0.7. In cases of gas phase diffusion limited growth of a III/V semiconductor with mixing on the group III sub-lattice like InGaP, the amounts of indium and gallium in the solid state should correspond to the partial pressures of TMI and TMG in the gas phase, i.e. the distribution coefficient between the gas phase and the solid state should be about one. The deviation we obtained in our experiments from this theoretical distribution coefficient must be explained by a nonlinear, but reproducible, reaction of the electronic TMG massflow controller.

At higher temperatures, more TMI (relative to TMG) is needed in the gas phase in order to

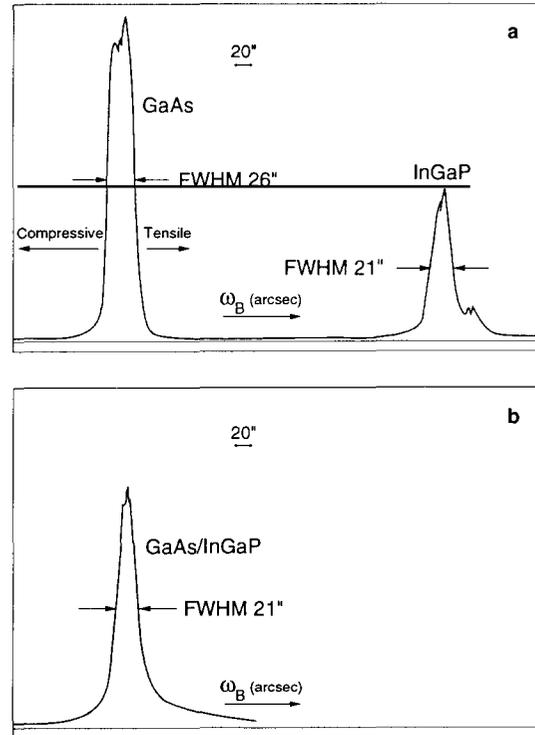


Fig. 3. X-ray rocking curves for (a) a slightly mismatched epilayer and for (b) a perfectly lattice matched epilayer of InGaP on GaAs at $T = 300$ K.

grow lattice matched material at a fixed indium content. This most probably is caused by a larger desorption rate of indium growth species than that of gallium growth species from the surface at higher temperatures. This is consistent with the lower value of the InP bond strength as compared to that of GaP.

The dashed lines in fig. 2 represent the boundaries of the region in which the $\text{In}_{1-x}\text{Ga}_x\text{P}$ is still lattice matched to the GaAs substrate. In this paper, we define lattice matching by a lattice mismatch of smaller than 1×10^{-3} around the exactly lattice matched composition $X_{\text{Ga}} = 0.516$ (at room temperature) [15]. It can be seen from this figure that at all three temperatures lattice matched epilayers can be grown by using different gas phase compositions of TMG and TMI. An example of this is given in fig. 3. In this figure, two rocking curves are plotted of InGaP epilayers on GaAs substrates. In fig. 3a, the epilayer is slightly (tensile) mismatched ($\Delta a/a_0 = -1 \times 10^{-3}$), and in fig. 3b, a perfectly lattice matched epilayer is given. The FWHM of the rocking curve of the perfectly matched epilayer is only 21 arc sec or less, a good value for this material.

3.2. The effect of the V/III ratio and the total gas flow on the growth rate and solid composition

Experiments have been performed in order to investigate the effect of the V/III ratio on the growth rate. In these experiments the amount of III component was kept constant and the concentration of phosphine in the gas phase was varied. The ratio between the concentration of TMG and TMI was chosen in such a way that lattice matched epilayers could be grown at the temperatures used.

For a total gas flow of 7 SLM (i.e. the sum of all the precursor flows and carrier gas flow) and at three different growth temperatures, the results are plotted in fig. 4. From this figure it is obvious that the V/III ratio within the region investigated has no influence at all on the growth rate. From fig. 4 one can conclude that at the two higher temperatures used (640 and 700°C) the growth rate indeed is purely gas phase diffusion limited because both growth rates are exactly

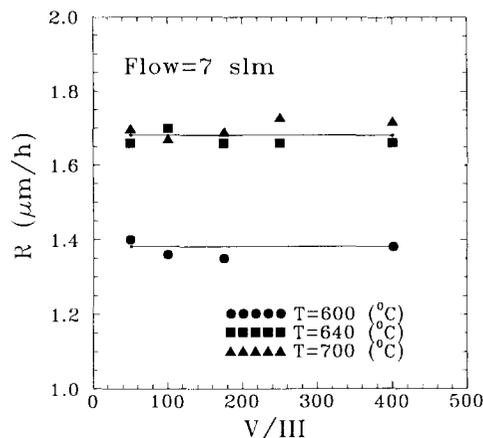


Fig. 4. Growth rate of the InGaP epilayers as a function of the input V/III ratio for 3 different growth temperatures and for a total flow of 7 SLM.

equal while at 600°C the growth rate is smaller because gas phase kinetics still plays a role.

In fig. 5 the gallium composition in the InGaP epilayer is plotted versus the V/III ratio for the three different growth temperatures used (600, 640 and 700°C). At the two highest temperatures, i.e. 640 and 700°C, the solid composition is as expected not dependent of the V/III ratio. However, at 600°C the gallium incorporation increases as the V/III ratio increases. It appears that the decomposition of TMG, and thus the gallium incorporation, is stimulated by increasing amounts of phosphine in the gas phase. It is known in

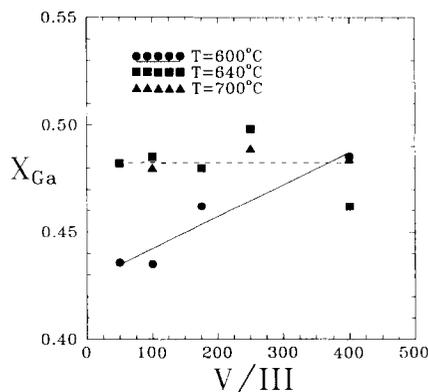


Fig. 5. Gallium content X_{Ga} of the $\text{In}_{1-x}\text{Ga}_x\text{P}$ epilayer as a function of the V/III ratio grown at different temperatures.

literature that the pyrolysis of TMI is enhanced in the presence of phosphine [15,31,32]. Similar observations have been made for the decomposition of TMG in the presence of arsine [26,28]. It now also appears that the decomposition of TMG is enhanced in the presence of phosphine. The influence of total flow rate on the growth of InGaP was also investigated. It appeared that optimal results in our reactor were obtained for a total flow rate of 7 SLM corresponding with a real gas velocity of 228 cm/s in the reactor at the growth temperature.

3.3. Electrical and optical quality of the InGaP layers

The electrical quality of the grown samples was evaluated with Hall–Van der Pauw measurements at room temperature using a cloverleaf configuration. All grown samples were n-type, with carrier concentrations ranging from 1 to $10 \times 10^{15} \text{ cm}^{-3}$. The mobility exhibited a much greater spread, mainly depending on the amount of lattice mismatch. In fig. 6 the mobility is plotted versus the solid composition of the epilayers. The dashed lines represent the boundaries of the lattice matched region. From this figure it can be concluded that the highest values for the mobilities are obtained in the lattice matched area. In

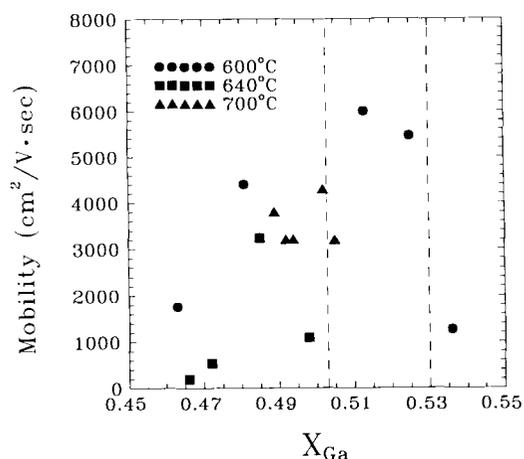


Fig. 6. Mobility ($\text{cm}^2/\text{V}\cdot\text{s}$) versus the solid phase gallium content X_{Ga} of InGaP epilayers grown at different temperatures. The dashed lines represent the lattice matched region.

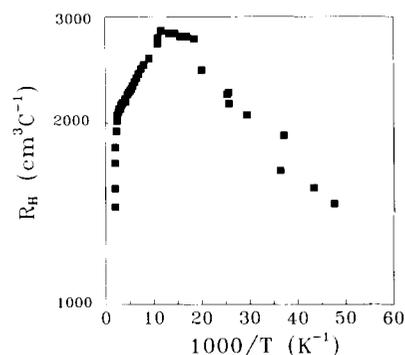


Fig. 7. Plot of the Hall coefficient R_H (cm^3/C) of an InGaP sample versus the reciprocal temperature ($1000/T \text{ (K}^{-1}\text{)}$).

these samples only elastic strain is present, no misfit dislocations were observed. The mobility values around $6000 \text{ cm}^2/\text{V}\cdot\text{s}$ are to our knowledge the highest values reported in literature. However, some care has to be taken in comparing the values of the mobilities of different samples because the degree of ordering of the material can influence these values drastically [33,34].

To check whether the presence of a 2DEG could have effected our mobilities values we performed some magneto-photoluminescence (MPL) measurements [35] and temperature dependent Hall–Van der Pauw measurements. Both measurements did not give any indication of such a 2DEG in contrast to some effects reported in literature [1,36]. Instead, the temperature dependent Hall–Van der Pauw measurements gave an indication that in a sample at low temperatures a “hopping” conduction mechanism is present instead of the usual carrier freeze out [37]. This is illustrated in fig. 7. In this figure the Hall coefficient (R_H) is plotted versus the reciprocal temperature (K^{-1}). In the lower temperature region suddenly the Hall coefficient drops sharply. This behaviour is typical for a hopping type of conduction [38].

The data points plotted in fig. 6 were obtained from samples with carrier concentrations of 1 to $10 \times 10^{15} \text{ cm}^{-3}$. The corresponding mobilities are relative high, which ensures, in combination with the carrier concentrations, a low compensation ratio for those samples. This implies that the

epilayers, which were all n-type, contain only low concentrations of acceptor atoms. So the effect of the misfit dislocations on the mobility will not be screened by large amounts compensating acceptors [39]. Because in the lattice matched region no misfit dislocations are present, we indeed find an effect of the strain on the electrical properties of the material in this region. Samples grown with a larger mismatch exhibited a dislocation pattern (see section 3.5). The electrical properties of these dislocations are most likely the reason for the lower values of the mobilities. Future experiments with more samples grown lattice matched at different growth temperatures may reveal a significant relation between the growth temperature and mobility.

The optical properties of the samples were measured using photoluminescence techniques at 4.2 K. At this temperature all samples showed luminescence. However, at room temperature some samples did not show any luminescence at all. A typical luminescence spectrum at 4.2 K is given in fig. 8b, for a slightly mis-matched epilayer. In this figure only one peak can be observed, which is common for this material. This peak is generally assigned to a band-to-band transition [1]. For a perfectly lattice matched sample, we found a smallest FWHM at 4.2 K of 9.4 meV for this band-to-band transition peak, as can be

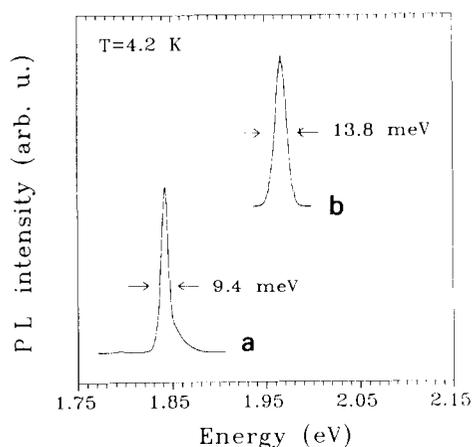


Fig. 8. Photoluminescence spectra recorded at 4.2 K for (a) a slightly mismatched sample and (b) a perfectly lattice matched epilayer InGaP.

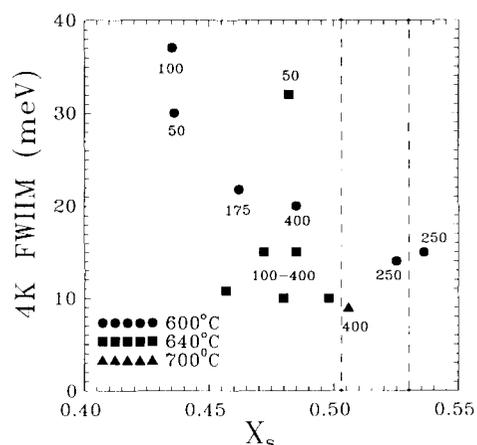


Fig. 9. FWHM of the photoluminescence peaks recorded at 4.2 K versus the solid phase content X_{Ga} of the InGaP epilayers grown at different temperatures. The numbers in this figure correspond with the V/III ratio used for the growth of each sample. The dashed lines represent the lattice matched region.

seen in fig. 8a. In literature sometimes, in less pure material, a second peak in the photoluminescence spectrum is detected which is ascribed to a conduction band to acceptor transition. According to literature zinc is probably this acceptor [5,15]. We did not find any indication of excitonic transitions in our material. Until now, nobody in literature has found luminescence from excitonic transitions in this material. A possible explanation for this absence of excitonic transitions is the local perturbation of the band edges because of the presence of a microstructure in InGaP and the spatial separation of carriers and ionized dopants [40,41].

In fig. 9, the FWHM of the band-to-band transition is plotted versus the solid composition of the material. The experiments were performed at different growth temperatures. The experiments performed at 600°C give an indication that the lowest FWHMs are obtained for those samples which are lattice matched. This coincides with results in literature [39,42]. From this figure one can also conclude that in general the two higher growth temperatures (640 and 700°C) give better optical results. Although the results of photoluminescence experiments obtained on our samples are good, the spatial variation of these

results is rather great. These variations of the photoluminescence signal originate from local variations of the material caused by the non-homogeneous ordering in our samples (see next section). This ordering is confined in domains of about 100 Å in diameter which are embedded in a matrix of disordered material. This will be discussed in more detail in the next section.

3.4. Ordering of the InGaP epilayers

We investigated some of the grown samples with TEM on the presence of the well known ordered structure. In accordance with literatures all these samples possessed a CuPt type of ordering on the {111} planes. However, the ordering is only present in two of the four possible {111} planes, most probably the $(\bar{1}11)$ and $(1\bar{1}1)$. One of these two planes shows a more intense diffraction spot than the other indicating that in this plane a higher degree of ordering is present. The diffraction pattern of the samples exhibited also so-called streaks in the $\langle 001 \rangle$ directions. The presence of these streaks suggests that the ordered regions are not perfect, i.e. the ordered domains consist of different type of microdomains [43]. All these TEM observations coincide with results published in literature [16,44–48].

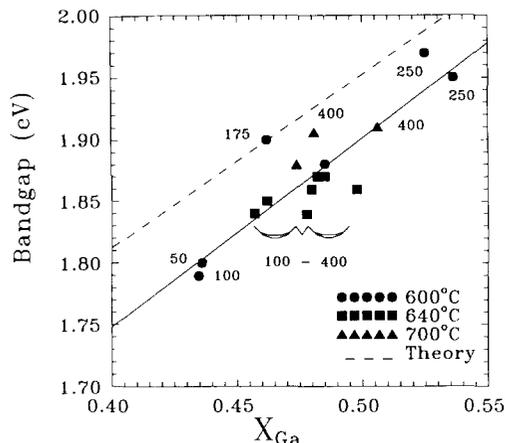


Fig. 10. Bandgap at 4.2 K as measured with PL versus the solid state composition (X_{Ga}) of the InGaP epilayers. The numbers in this figure correspond with the V/III ratio used for the growth of the epilayers. The dashed line shows the theoretical value for the bandgap at 4.2 K without ordering [55].

HREM investigations of an ordered sample revealed that the ordering starts at the InGaP–GaAs interface, so immediately when InGaP starts to grow. The HREM measurements also clearly show that ordering occurs on only two of the four possible [111] planes. In the ordered domains so-called anti phase boundaries (APB) are observed. These are the result of an imperfect variation of the indium and gallium planes in the lattice. So two neighbour indium or gallium planes form an APB. This is conform the literature results [47].

Ordering is known to affect the bandgap of the material, the bandgap shifts to lower values as compared to disordered material. The influence of ordering on the bandgap was investigated using PL techniques. It was found that the peak positions at 4.2 K of the ordered samples were about 60 meV lower than the theoretical values expected for the solid composition of the material [21].

We observed a strong influence of T_{growth} and the V/III ratio on the value of the bandgap. At a growth temperature of 600°C, a maximum was observed in the bandgap at V/III = 175 while at 700°C the samples with the highest V/III ratio (400) showed the highest values for the bandgap. This is shown in fig. 10. These results are comparable with the results published in literature [17,44,49,50].

3.5. Morphology

The four samples grown in the lattice matched area (as defined before) show a specular surface, although some precautions had to be taken to avoid the growth of a typical defect on the surface. Samples with a tensional misfit larger than $\Delta a/a_0 \leq -1.4 \times 10^{-3}$ show a misfit dislocation line pattern only in the $[01\bar{1}]$ direction. The samples grown under compression (larger than $\Delta a/a_0 \geq 4 \times 10^{-3}$) show a dislocation cross-hatched pattern which is oriented in the $[011]$ and in the $[0\bar{1}\bar{1}]$ directions. Ozasa et al. [39] performed double crystal X-ray diffraction measurements on slightly mismatched InGaP epilayers grown on GaAs and also had to conclude from these measurements that the tensile strain in the

epilayers is released more easily than the compressive strain. The difference in the misfit dislocation patterns between the tensile and compressive situations can be explained taking into account the *a*-symmetry of the zincblende structure of III/V compounds [51,52].

3.6. Defects

In the beginning of this investigation all InGaP epilayers were covered with defects. The defect density increased going from the middle to the edges of the substrates. The defects have all the same orientation, their length axis is oriented along the [011] axis. The density of the defects increased when the material is grown at higher temperatures. The defect density also increased using lower V/III ratios. We observed these defects in both compressive and tensile strained material as in perfectly lattice matched InGaP. So this defect is not caused by stress relaxation as is described in ref. [52]. The defects showed remarkable resemblances to the so-called "oval defects" described in ref. [53].

In order to investigate the origin of these defects GaAs epilayers were grown in the same cell and on the same susceptor as was used for the growth of the InGaP layers and with the same experimental growth procedure. These experiments resulted in GaAs layers covered with the same type of defects. It has to be noticed that GaAs layers grown with the cell and susceptor used for the growth of GaAs exclusively, showed mirror-like surfaces without any oval-like defects. Microprobe analysis of a cleaved and stained defect on the GaAs layer showed that the nucleus [53] of the defect consist of material with a high indium concentration. This nucleus is right on top of the GaAs substrate. So probably some adsorbed indium from the susceptor or cell evaporates onto the GaAs substrate, leading to a defect. In the defects on the InGaP layers the indium content of the nucleus could not be measured due to the masking effect of large amounts of indium in the InGaP epilayers.

Microprobe analysis of the defects on InGaP layers show that these defects are all gallium rich compared to the epilayer. This is especially true

for layers grown at low temperatures (600°C). This difference in composition of defect and epilayer decreases in going to higher temperatures. At 700°C, there is no difference in composition between the defect and the epilayer left although the defects are still present. Differences in enthalpies of formation of GaP and InP (24.4 and 21.2 kcal/mol [54] indicate that the strength of the Ga-P binding is larger than the In-P strength. This is also true for the Ga-As and In-As bindings [54]. So the indium species will diffuse more easily over the surface than the gallium species.

All these observations lead to the conclusion that these defects were caused by an initial adsorption of indium onto the GaAs substrate. This indium originates from previous InGaP growth experiments on the same susceptor. This is supported by the fact that if a susceptor is used on which the previous InGaP layer was covered with an GaAs cap layer the defect density is very low. Desorption of indium from the susceptor and the following adsorption on the GaAs substrate results in the forming of an indium rich-nucleus (InAs?) on the GaAs substrate, which is incorporated into the GaAs buffer layer. An alternative for this GaAs caplayer is a high temperature treatment of the susceptor with HCl.

The experiments described in the former sections of this paper were all carried out with a "clean" susceptor which resulted in defect free epilayers with specular surfaces.

4. Conclusions

InGaP epilayers have been grown on GaAs substrates. The influence of several growth parameters as the growth temperature, V/III ratio and In/Ga ratio in the gas phase has been investigated. It appeared that the growth rate of InGaP is gas phase diffusion determined for growth temperatures of about 650°C and higher. At temperatures lower than 650°C the growth rate is determined by kinetics coupled to the diffusion through the gas phase. The growth process exhibits an apparent activation energy of about 5 kcal/mol. The decomposition of TMG is responsible for this. Also decomposition of TMG is

more effective in the higher temperature region. This, together with the observation that indium growth species will desorb at higher temperatures, leads to the fact that one has to increase the TMI concentration in the gas phase when the growth temperature is raised in order to obtain the same solid composition. At all growth temperatures, perfectly lattice matched InGaP epilayers could be obtained.

The V/III has no influence of the growth rate at the temperatures used, both at a total flow of 5 and of 7 SLM. At the two highest growth temperatures used (640 and 700°C) the solid composition does not depend on the V/III ratio. At 600°C the solid composition becomes more gallium rich when the V/III ratio is increased.

When the total flow rate through the reactor was increased from 5 to 7 SLM, with constant amounts of precursors, the growth rate decreased with about 10%.

The electrical quality of our layers is excellent. The mobilities measured in our samples are high, but are not due to the presence of a 2DEG at the GaAs–InGaP interface. However, care has to be taken in comparing the values for the mobilities with values reported in literature because the ordering present in InGaP can influence this figure. In a sample evidence has been found for a hopping type of conduction at low temperatures. Photoluminescence experiments showed that our samples are of good optical quality. These experiments also point to the fact that ordering is common in our samples. This ordering is a function of the growth conditions. Growing at high temperatures at moderate V/III ratios resulted in disordered material. The morphology is specular whenever lattice matched material was grown but exhibited the typical cross hatched pattern when the samples were not perfectly lattice matched. It turned out that the use of a clean susceptor is necessary in order to avoid defects in the InGaP epilayers.

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