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Nanoscale InP islands embedded in InGaP

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We have prepared small InP islands which are embedded in $In_{0.485}Ga_{0.515}P$ grown on a (100) GaAs substrate by solid-source molecular beam epitaxy. The InP islands form due to the 3.7% lattice mismatch between $In_{0.485}Ga_{0.515}P$ and InP. Atomic force microscopy measurements show that the island size is typically ~50 nm in diameter and ~5 nm in height for nominally two monolayers of InP deposited on $In_{0.485}Ga_{0.515}P$. The energy of the photoluminescence (PL) peak shifts from 1.85 to 1.53 eV as the nominal InP thickness increases from 2 to 10 monolayers. A minimum PL linewidth of 21.6 meV and the maximum intensity of the PL originating from the InP islands is observed for 7.3 monolayers InP. © *1995 American Institute of Physics.*

The formation of small islands due to epitaxial growth of lattice mismatched heterostructures has been investigated for many different material combinations. It has been demonstrated that up to a certain thickness the three-dimensional island growth occurs dislocation-free despite a strain relaxation.[1–3] The measured width of such coherent islands is typically ~100 nm for Ge on Si and ~25 nm in case of $In_{0.5}Ga_{0.5}As/GaAs$.[1,3] Therefore it has recently attracted increasing interest as a technique to fabricate self-assembling nanoscale quantum boxes.[4–8] The island formation provides a self-organizing *in situ* technique for making quantum boxes with the advantage that no lithography and etching processes are needed.

The size of the epitaxial islands mainly depends on the lattice mismatch, growth temperature, surface structure, and other growth conditions such as the III/V ratio. Molecular beam epitaxy (MBE) is well known to allow precise submonolayer deposition which is necessary in order to achieve reproducible size control of the islands. The main challenge is to find a growth mode which minimizes the size fluctuations of the islands in order to obtain sharp luminescence peaks.

So far this concept has mainly been applied to $In_xGa_{1-x}As/GaAs$.[4–7] Another interesting material system is $In_{0.485}Ga_{0.515}P/InP$. The lattice mismatch between InP and $In_{0.485}Ga_{0.515}P$ is 3.7%, while $In_{0.485}Ga_{0.515}P$ has the same lattice constant as GaAs. Both materials have a direct bandgap E_g, which is 1.945 eV for $In_{0.485}Ga_{0.515}P$ and 1.42 eV for InP at 4.2 K.[9] Due to the high energy gap of $In_{0.485}Ga_{0.515}P$, InP islands grown on InGaP should have different optical properties compared to the $In_xGa_{1-x}As/GaAs$ family. Specifically, we expect to observe a luminescence at higher energies.

In this letter, we report on the formation of InP islands embedded in a 400 nm thick $In_{0.485}Ga_{0.515}P$ layer grown by solid source MBE on (100) GaAs substrates. To our knowledge, this has only been done previously by vapor phase epitaxy (VPE).[8] In order to investigate the optical properties of the InP islands, the samples are characterized by photoluminescence (PL) spectroscopy. We discuss PL measurements for various nominally grown InP thicknesses while keeping the growth conditions constant. The island size distribution is estimated by atomic force microscopy (AFM).

All samples are grown on semi-insulating (100) GaAs substrates in a conventional MBE system using GaP as P_2 source. For further details about the P_2 source see Ref. 10. After removing the oxide, a 200 nm thick GaAs buffer is grown under a beam equivalent As_4 pressure of 5×10^{-6} Torr at a substrate temperature of 580 °C. Then the deposition is stopped to decrease the substrate temperature to 350 °C and change from As to P rich growth conditions. The deposition sequence is 200 nm $In_{0.485}Ga_{0.515}P$, followed by 2–14.5 monolayers (MLs) InP. The InP islands are finally overgrown by a 200 nm thick $In_{0.485}Ga_{0.515}P$ layer. This layer sequence is grown at 470 °C under a P_2 pressure of 4×10^{-6} Torr. The surface of the $In_{0.485}Ga_{0.515}P$ layer shows a 2×1 reconstruction in reflection high energy electron diffraction (RHEED). We use a growth rate of 1 ML per second for $In_{0.485}Ga_{0.515}P$ and 0.485 ML/s for InP. The composition of the $In_{0.485}Ga_{0.515}P$ layers is determined by double-crystal x-ray diffraction (DCXR).

Figure 1 shows an AFM picture of a sample with nominally two MLs of InP deposited on InGaP without the 200 nm $In_{0.485}Ga_{0.515}P$ cap layer. The sample was cooled down rapidly (50 °C/min) after the InP deposition to limit reordering of the InP islands due to surface diffusion processes.

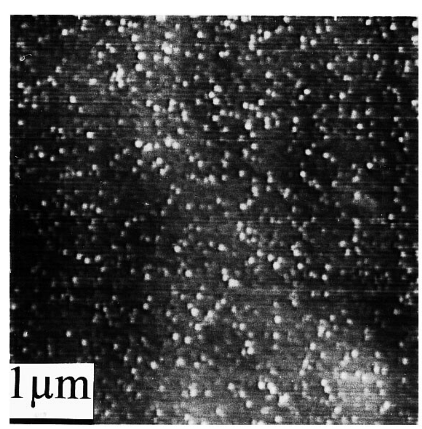

FIG. 1. Atomic force micrograph of a sample containing 2 ML InP on InGaP.

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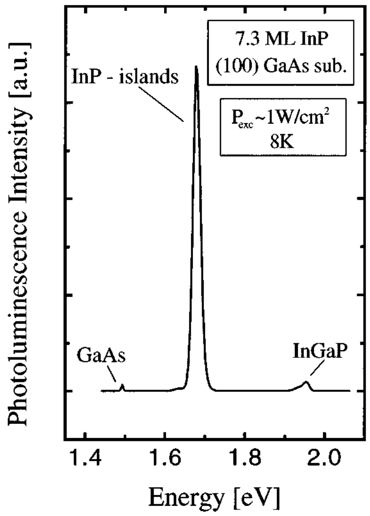

FIG. 2. Low-temperature (8 K) photoluminescence spectrum for InP islands embedded in 400 nm InGaP. The amount of nominally deposited InP is 7.3 ML.

Figure 1 demonstrates that the island formation has started already. This is also confirmed in the RHEED pattern, which becomes spotty after depositing about two MLs of InP. The diameter of the islands is typically 30–50 nm with an average height of about 5 nm. The density is in the order of $10^9/cm^2$.

We assume there are few misfit dislocations because the InP islands have the same order in size like the coherent islands in the Ge/Si and $In_{0.5}Ga_{0.5}As/GaAs$ system as reported in Refs. 1 and 3. It cannot be excluded that there is some rearrangement of the islands and further nucleation of misfit dislocations when the islands are overgrown and finally embedded in InGaP. The overgrowth is crucial to improve the optical properties of the samples. Without such a cap layer we measured only a very weak and broad PL which could be attributed to the islands. For nominally 2 ML InP, the island PL peak has a width of more than 100 meV and the intensity is smaller than the InGaP PL.

To study the optical properties of embedded islands in detail, we varied the InP deposition between 2 and 14.5 MLs. Figure 2 shows the PL spectrum for the sample with 7.3 MLs InP in 400 nm InGaP at 8 K. The 514.5 nm line of an Ar^+ laser is used for excitation. The excitation power is 1 W/cm^2 in this case. The spectrum shows three peaks at 1.95, 1.68, and 1.49 eV. They can be attributed to the InGaP barrier, the InP islands and the GaAs buffer layer, respectively. The assignment of the peak at 1.68 eV to the InP islands is confirmed by measurements on a reference sample, which was grown without InP deposition. This sample does not show a PL peak between the InGaP and GaAs bulk lines. The full width at half maximum of the InP island peak shown in Fig. 2 is only 21.6 meV. In addition, the intensity is very high as compared to the InGaP and GaAs bulk peaks. The ratio of the integrated intensities between the InP and InGaP peak is more than 30, which is quite remarkable keeping in mind that the PL originates from only 7.3 MLs (about 2 nm) InP compared to the 400 nm thick InGaP layer.

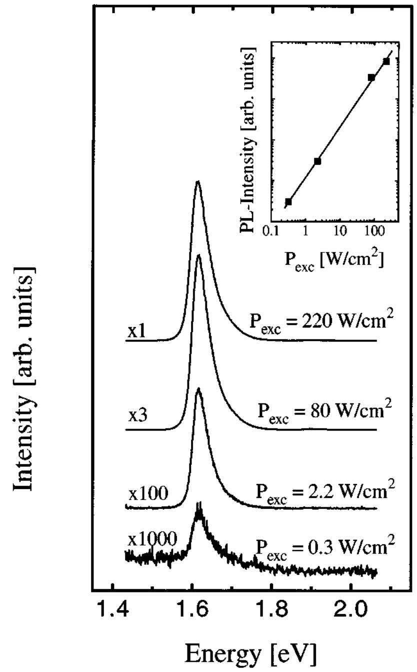

FIG. 3. Room-temperature spectra of the 7.3 ML InP sample measured at various excitation power densities P_{exc}.

Figure 3 shows four PL spectra of the 7.3 ML InP sample measured at room temperature (RT) with various excitation power densities. The essential result is strong PL from the InP islands even at RT. Additionally the intensity ratio between the islands and the InGaP barrier signal is increased. Even when the excitation power is varied between 0.15 and 220 W/cm^2 in Fig. 3, there is no saturation of the PL intensity, which indicates that the island-attributed PL is not due to defects or impurities. Comparable results are obtained at 8 K.

The energy and the intensity of the InP island luminescence depends strongly on the nominal thickness of the grown InP. As demonstrated in Fig. 4(a), fewer MLs result in a strong blueshift of the InP peak position. For comparison, the direct band gap of InP bulk material is at 1.42 eV (4.2 K). For the sample with nominally 2.4 MLs of InP the PL energy is 1.85 eV, i.e., the difference to InP bulk amounts to 430 meV. The InP PL peak shifts continuously to lower energies and reaches about 1.53 eV for nominally 10 MLs InP. We assume that the luminescence energy is determined by the three-dimensional confinement within the InP islands and by the energy shift due to strain. For a quantitative calculation of the confinement and strain contributions it would be necessary to know the band offset between InGaP and InP, the exact dimensions, and the in-plane as well as the tetragonal lattice contraction within the InP islands. Both the confinement and the compressive strain cause a shift to higher energies. With increasing InP deposition, the islands become bigger. Therefore, we expect a reduced three-dimensional confinement energy. At the same time we expect a partial strain relaxation. Consequently, the PL energy is expected to decrease with increasing InP island size.

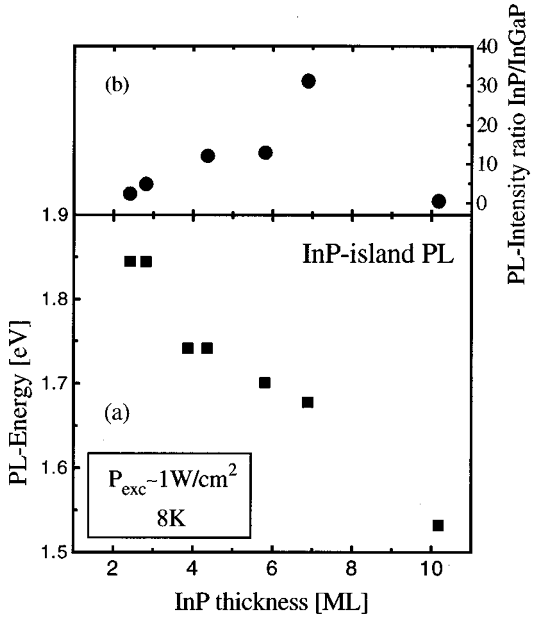

FIG. 4. (a) Energetic position of the island PL as a function of the nominal InP thickness. (b) Ratio of the integrated intensity between the InP island PL and the PL from the $In_{0.485}Ga_{0.515}P$ barrier as a function of the nominal InP thickness.

Figure 4(b) shows the ratio of the integrated intensities between the InP and the InGaP peak as a function of the nominal InP thickness. The maximum intensity is achieved for about 7 MLs. The smaller intensity below 7 MLs of InP can be explained by the smaller volume of InP material and the reduced carrier confinement in the smaller islands. The drastic decrease for 10.2 ML InP is, on the contrary, due to incoherent island growth. Significant strain relaxation within the InP islands is expected with the nucleation of misfit dislocations, which can act as nonradiative recombination centers. This is confirmed by a sample with nominally 14.5 MLs InP, which does not show any island attributed luminescence.

In conclusion, we have shown that the growth of InP on $In_{0.485}Ga_{0.515}P$ by MBE results in the formation of small InP islands. For nominally 2–8 ML they show strong and sharp luminescence even at RT. The luminescence energy can be adjusted precisely due to the submonolayer growth control in MBE. For more than 10 MLs InP, the PL intensity diminishes, which can be explained by increased nucleation of misfit defects due to significant strain relaxation within the InP islands. The shifting of the PL energy with the nominal island thickness as well as the linear PL intensity increase with the excitation power density over four orders of magnitude and the fact that the InP-island related PL is obtained at RT indicate the intrinsic origin of this luminescence.

The surprisingly sharp and very intensive PL emission even at RT promises interesting prospects for device applications in the visible range based on GaAs substrate. To characterize the size distribution, shape, and strain relaxation of the InP islands and for a better understanding of the InP-island PL as a function of the nominal InP thickness a detailed high resolution transmission electron microscopy investigation is necessary.

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[1] D. J. Eaglesham and M. Cerullo, Phys. Rev. Lett. **64**, 1943 (1990).
[2] C. W. Snyder, B. G. Orr, D. Kessler, and L. M. Sander, Phys. Rev. Lett. **66**, 3032 (1991).
[3] S. Guha, A. Madhukar, and K. C. Rajkumar, Appl. Phys. Lett. **57**, 2110 (1990).
[4] J.-Y. Marzin, J.-M. Gérard, A. Izraël, and D. Barrier, Phys. Rev. Lett. **73**, 716 (1994).
[5] D. Leonard, M. Krishnamurthy, C. M. Reves, S. P. Denbaars, and P. M. Petroff, Appl. Phys. Lett. **63**, 3203 (1993).
[6] A. Madhukar, Q. Xie, P. Chen, and A. Konkar, Appl. Phys. Lett. **64**, 2727 (1994).
[7] R. Nötzel, J. Temmyo, and T. Tamamura, Nature (London) **369**, 131 (1994).
[8] J. Ahopelto, A. Yamaguchi, K. Nishi, A. Usui, and H. Sakaki, Jpn. J. Appl. Phys. **32**, L32 (1993).
[9] Landolt-Börnstein, *Numerical Data and Functional Relationships in Science and Technology*, edited by O. Madelung (Springer, Berlin, 1982), Vol. 17a.
[10] T. Shitara and K. Eberl, Appl. Phys. Lett. **65**, 356 (1994).

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