

# Optimised two layer overgrowth of a lateral strain-modulated nanostructure

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## Abstract

Recently it has been shown that lateral carrier confinement in an InGaAs quantum well (QW) embedded in GaAs can be achieved by using a laterally patterned InGaP stressor layer on top of the heterostructure. To exploit this effect in a device the structure has to be planarized by a second epitaxial step. It has been shown that the lateral strain modulation almost vanishes after overgrowth with GaAs, whereas overgrowth with a single ternary layer of opposite strain compared to the stressor layer suffers from strain induced decomposition. Here we show that the lateral carrier confinement of the initially free standing nanostructure can almost be maintained using a two step process for overgrowth, where a strained thin ternary layer is grown first followed by GaAs up to complete planarization of the patterned structure. Thickness and composition of the ternary layer are adjusted on the basis of finite element calculations of the strain distribution (FEM). The strain field achieved after overgrowth is probed by X-ray grazing-incidence diffraction (GID).

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## 1. Introduction

Strain can be employed as an additional degree of freedom for device design. It has been reported [1] that lateral carrier confinement can be achieved in an InGaAs quantum well (QW) via lateral patterning of a highly strained InGaP stressor layer grown on top of the QW. The lateral strain modulation is caused by laterally periodic lattice relaxation of the stressor layer causing lateral band gap variation in the underlying InGaAs-QW. The resulting dimensional reduction of carrier movement due to the lateral carrier confinement also can result in a higher modal gain of respective laser devices [2]. For a free-standing lateral nanostructure the lateral carrier confinement could be optimised [3]. For device applications, however, the effect must be maintained after planarization by a second epitaxial growth step. The overgrowth has to preserve the lateral strain modulation and the overgrown layer should be almost homogeneous in composition and free of

structural defects. For overgrowth of etched GaAs gratings with ternary layers, i.e.  $\text{Al}_x\text{Ga}_{1-x}\text{As}$ , it has been shown that self-organization processes may occur [4]. In a previous paper we have shown that strain driven migration of Ga and Al has to be taken into account [5]. This indicates that for overgrowth of semiconductor lateral nanostructures with ternary layers strain related phenomena have to be considered.

In this paper we show that strain induced carrier confinement effectively can be maintained without defect creation using a two step process for overgrowth and planarization of the initial laterally patterned nanostructure. Finite element calculations (FEM) in terms of linear elasticity theory have been applied for the calculation of the strain distribution in the overgrown structures. The created strain distribution is verified by X-ray grazing incidence diffraction (GID).

## 2. Experimental

For both the first and the second epitaxial step we used metalorganic vapour phase epitaxy (MOVPE). In a first step

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a vertical layer structure was grown. It is composed of a 10 nm thick  $\text{In}_{0.16}\text{Ga}_{0.84}\text{As}$ -QW with a strain of  $\varepsilon = -1.14\%$  grown on a (0 0 1) GaAs substrate followed by a 10 nm thick GaAs layer (barrier and etch stop) and a 120 nm thick tensily strained InGaP layer with  $\varepsilon = 0.3\%$  and a 10 nm thick GaAs cap layer. The free standing lateral grating with a period of 500 nm was prepared by holographic photolithography and subsequent selective wet chemical etching of the GaAs cap and the InGaP stressor layer resulting in  $\{1\ 1\ 1\}$  facets as valley sidewalls [1]. The sample was cleaved and one part was overgrown with GaAs whereas the other one was overgrown with compressively strained InGaP. The gratings were heated up to  $600^\circ\text{C}$  under phosphine overpressure to prevent desorption of phosphorus from the grating sidewalls. It has been shown before that this is a crucial step to preserve the shape of the grating and to prevent the growth of InGaAs quantum wires which occurs if arsine is used during heating-up [6]. The overgrowth of GaAs took place at  $580^\circ\text{C}$  whereas InGaP was grown at a temperature of  $600^\circ\text{C}$ .

The sample geometry was examined by scanning electron microscopy (SEM) in the backscattered electron (BSE) mode on cleaved edges whereas the structural properties were studied using grazing incidence X-ray diffraction (GID). This method allows for the determination of the in-depth evolution of the lateral lattice mismatch. Thus, the prediction from the FEM calculations can be checked. The measurements were carried out at the European synchrotron radiation facility (ESRF) in Grenoble at beamline ID10B using the vertical diffractometer. For different angles of incidence  $\alpha_i$  ranging from  $0.08^\circ$  to  $1.7^\circ$  the reflected diffracted intensity from the sample was monitored as a function of reciprocal lattice units (rlu), i.e., in terms of H, K, L. For the measurements a parallel monochromatic X-ray beam with a wavelength  $\lambda = 0.154\text{ nm}$

was used. To achieve a very high resolution in reciprocal space a Ge analyser was mounted in front of the detector. The coordinate system used in our experiments is rotated by  $45^\circ$  with respect to the usual cubic coordinate system. Thus, the (2 0 0) reflection in our notation is equivalent to the (2 2 0) reflection in cubic coordinates. We recorded longitudinal scans at the (2 0 0) reflection which is in our geometry sensitive to strain (see [3,6]). Since the ridges and valleys of the grating run parallel to  $[1\ -1\ 0]$  we detect the lateral lattice mismatch in  $[1\ 1\ 0]$  direction where a change of the lateral lattice parameter is expected.

### 3. Results of single layer overgrowth

SEM micrographs of the samples with free standing lateral structure and after overgrowth with GaAs are shown in Fig. 1a and b, respectively. After overgrowth with GaAs the sample is planarized and no irregularities are found in the overgrown material. The other sample (see Fig. 1c), where compressively strained InGaP (100 nm) was used for filling-in, exhibited the well-known behaviour for ternary material grown over a grating, namely self-organization. This behaviour is caused by the different migration of In and Ga and leads to an In-rich vertical InGaP-layer (indicated by arrow in Fig. 1c). From earlier experiments it is known that the high strain in this region causes defect formation above the valley (Fig. 1d) thus leading to plastic relaxation [6].

In Fig. 2a, the GID intensity curves are shown for the free standing grating as a function of the angle between the incoming X-ray beam and the sample surface  $\alpha_i$ . Increasing  $\alpha_i$  corresponds to an increasing penetration depth into the sample [8]. It is seen that for small  $\alpha_i$  the intensity maximum

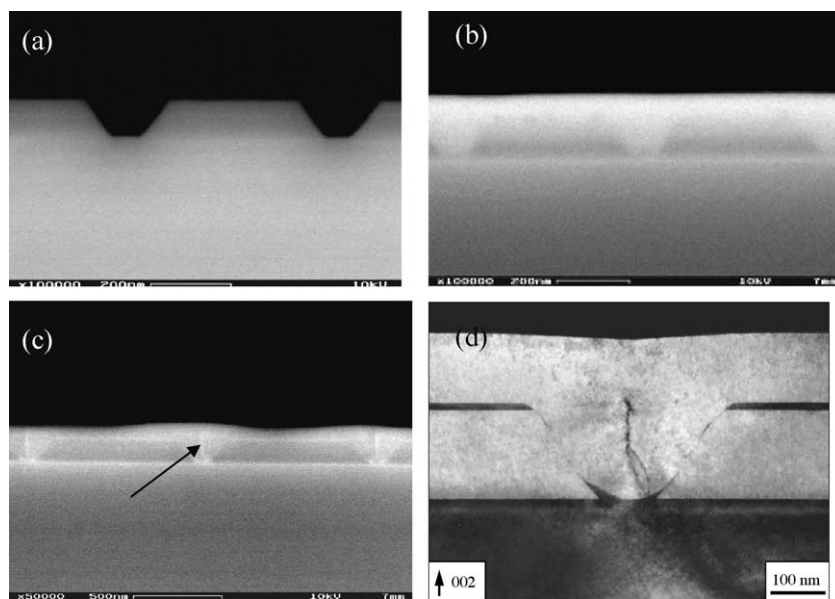


Fig. 1. SEM micrographs of the free standing grating (a), the grating overgrown with GaAs at  $580^\circ\text{C}$  (b) and with InGaP (c), (d) shows a transmission electron micrograph from the area indicated by arrow in (c) with the defect structure resulting from self organization.

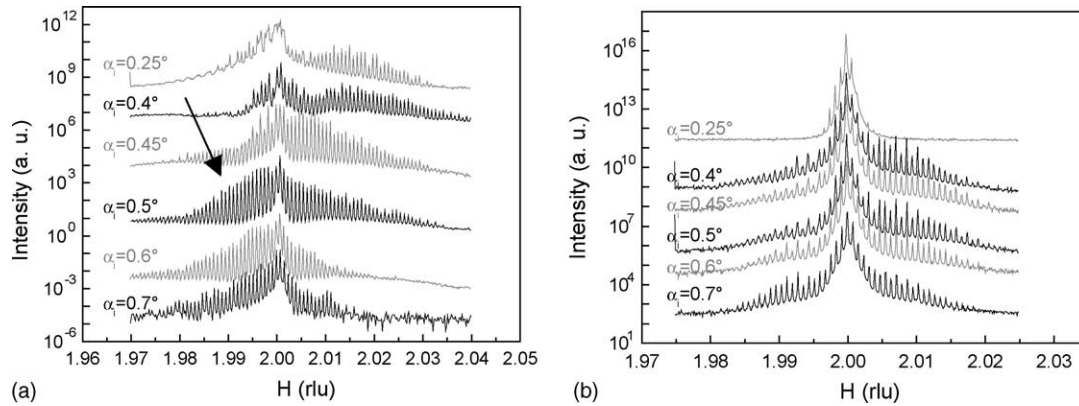


Fig. 2. Longitudinal GID scans at the strain sensitive (200) reflection for the free standing nanostructure (a) and after overgrowth with GaAs (b).

is slightly shifted to  $H > 2$  indicating a smaller lateral lattice parameter compared to GaAs.

This is an indication of strain relaxation in the near surface region of the InGaP stressor layer, which was initially tensily strained. After patterning the lateral lattice parameter shifts back to its own smaller value compared to GaAs. For  $\alpha_i > 0.45^\circ$  a local intensity maximum emerges at  $H < 2$  which is a measure for a region with a larger lateral lattice parameter than GaAs (see arrow in Fig. 2a). This signal originates from the depth where the InGaAs-QW is located and stems from the region below the valleys of the grating. The corresponding tensile lateral strain amounts to  $\varepsilon_{||} = 0.4\%$ .

In Fig. 2b, the GID scans for the sample overgrown with GaAs are presented. Independently from  $\alpha_i$  we always observe a symmetrical GID curve with the intensity maximum located at  $H = 2$ , i.e. there is no indication for strain. The initial lateral strain modulation vanishes after overgrowth with GaAs.

We also measured GID scans of the sample after overgrowth with a compressively strained single InGaP layer (see Fig. 3). For all  $\alpha_i$  we observe a shoulder at  $H > 2$  in the GID curves. These features are related to a region near the sample surface and are obviously caused by the high compressive strain due to the self-organization of the grown-in InGaP

layer. On the other hand, peaks at  $H < 2$  are not visible, i.e. the tensile strain for  $\alpha_i > 0.4^\circ$  seen for the free-standing sample is not maintained. The sharp peak at  $H = 2$  observed for the scans with  $\alpha_i > 0.7^\circ$  corresponds to the GaAs substrate. From GID measurements we conclude that single layer overgrowth is not suited to maintain the strain modulation of the initial patterned nanostructure.

#### 4. Finite element calculations of strain distribution after overgrowth

A linear elasticity model in a finite element method (FEM) approach has successfully been applied to optimise the geometrical parameters of the free standing nanostructure. In particular we optimised the thickness and composition of the InGaP stressor layer and the shape of the etched trench in order to maximize the lateral strain modulation in the underlying InGaAs-SQW [3]. In this paper we apply the FEM method to calculate the strain distribution in gratings after overgrowth. For the calculations we used the initial vertical layer structure as described above. A grating period of 500 nm and a valley width of 90 nm were chosen. As indicated above the ridges and valleys run along [1–10] direction ( $z$ -direction). In the valleys the GaAs cap layer and the InGaP layer are removed down to the GaAs etch stop layer. The valley sidewalls consist of  $\{111\}$  facets. The direction of grating replication is denoted as the  $x$ -direction and the  $y$ -direction is defined to be parallel to the growth direction. The strain distribution of a half-period of the free standing nanostructure and the structure after overgrowth with GaAs has been already shown in [7]. The calculations revealed that the lateral lattice misfit vanishes after filling-in. This is in accordance with the findings of the GID measurements and can be explained by the fact that the relaxation of the tensily strained InGaP at the free grating surfaces is turned back due to the re-growth with GaAs. In this case no or only a very small lateral strain modulation is observed in the InGaAs-QW region. The situation changes after filling-in with compressively strained InGaP ( $\varepsilon = -0.3\%$ ). Here the lateral strain modulation within

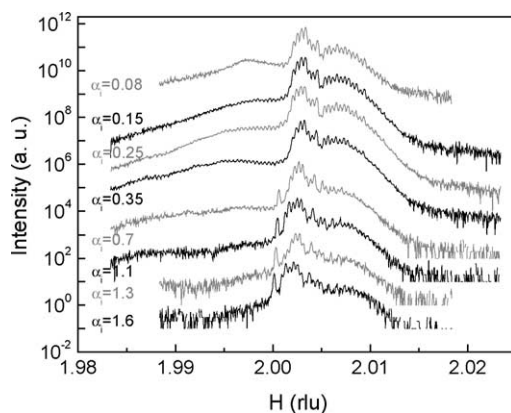


Fig. 3. GID from a nanostructure after single layer InGaP overgrowth.

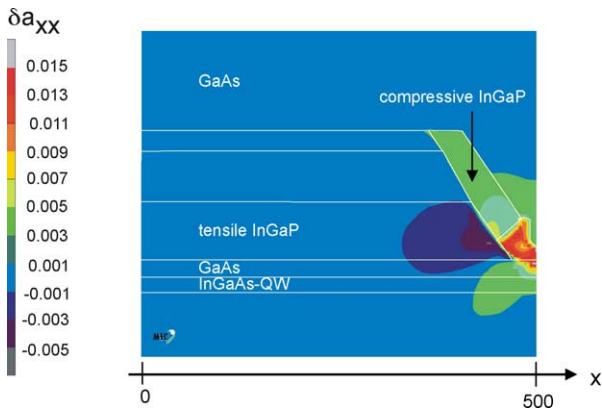


Fig. 4. FEM results of two layer overgrowth.

the InGaAs-QW can be maintained, in general. Unfortunately the effect of strain-driven self-organization, shown in Fig. 1d, cannot be simulated based on the presently used approach. The observed defect formation leads to a plastic relaxation in the highly strained vertical QW, which is undesired. To find a way out of this problem we divided the material used for overgrowth into a strained layer and a lattice matched layer.

Fig. 4 shows the results of FEM calculations using a two layer system for overgrowth, i.e. a thin compressively strained InGaP layer followed by a thicker GaAs layer. The calculated lateral lattice misfit distribution  $\delta a_{xx}(x, y)$  is plotted for a half-period of the lateral nanostructure with  $\delta a_{xx} = (a_{Lx} - a_0)/a_0$ , where  $a_{Lx}$  is the lattice constant in  $x$ -direction and  $a_0$  is the lattice constant of the undisturbed GaAs substrate. For the calculation we considered that the compressively strained InGaP covers the  $\{111\}$  facets of the grating. Since its nominal lattice constant is larger than GaAs this material tends to expand the underlying material. The strain distribution clearly shows a region with a high positive lateral lattice mismatch (gray colour) close to the bottom of valley. In the depth where the InGaAs-QW is located still a lateral lattice misfit with a value of 0.5% is observed (see light gray colour in the QW region). From FEM calculations one can conclude that about 90% of the initial strain before overgrowth can be maintained.

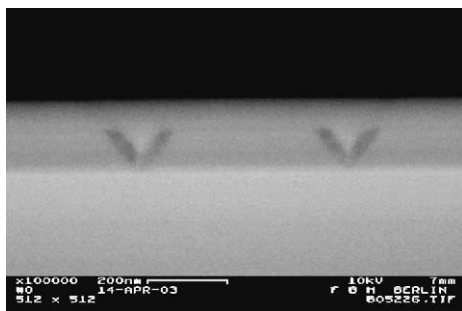


Fig. 5. SEM-BSE image of the cleaved edge of a grating overgrown with 17 nm InGaP with  $\varepsilon = 0.3\%$  and 70 nm GaAs.

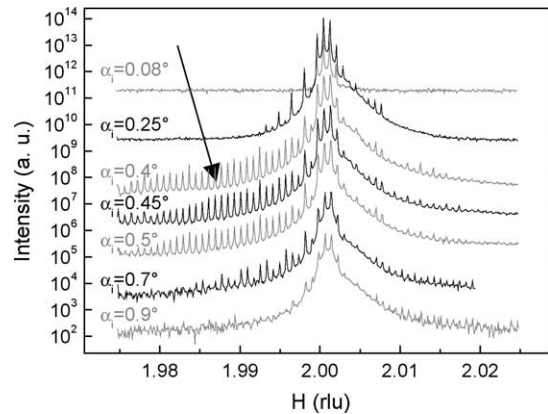


Fig. 6. GID curves of the grating sample after two layer overgrowth.

## 5. Results for two layer overgrowth

Fig. 5 shows a SEM micrograph of the cleavage edge of a nanostructure after two layer overgrowth. Here a nominally 17 nm thick InGaP layer with  $\varepsilon = 0.3\%$  was deposited onto the grating followed by a GaAs layer until planarization was reached. Such a layer sequence was grown over a sample with a triangular grating shape, which is known to yield the best results for lateral carrier confinement. The thin InGaP shows up as a layer with dark contrast.

It is seen, that in the valley the growth starts at the  $\{111\}$  sidewalls. Only in a very small region just above the tip of the valley a different contrast is observed indicating self-organization in the region where the growth fronts of the two facets collide. The sample was planarized using 70 nm GaAs. No growth irregularities are observed.

Fig. 6 shows the GID diffraction curves of the same sample. For small  $\alpha_i$  almost no grating side peaks are observed. Due to planarization there is no lateral periodicity of the lattice parameter in the GaAs region. For increasing penetration depth where the X-ray beam probes the overgrown lateral pattern the lateral periodicity becomes visible. At  $\alpha_i = 0.45^\circ$  the local intensity maximum at  $H \approx 1.99$  (indicated by arrow) shows the highest intensity. The estimate of the penetration depth at  $\alpha_i = 0.45^\circ$  corresponds fairly well to the depth of the InGaAs-QW. Thus, it can be concluded that the lateral strain periodicity can be maintained even after overgrowth. As predicted by the FEM calculations a material with a larger lattice parameter has to be used for filling-in. Since the self-organization has to be taken into account only a moderate compressive strain was used to prevent defect formation and plastic relaxation.

## 6. Conclusions

We have shown that a lateral strain modulation is induced in an InGaAs-QW after lateral patterning of an InGaP stressor layer on top of it. For device applications, however, the structure has to be planarized. If a single GaAs layer is used for the

planarization the lateral strain modulation in the InGaAs-QW vanishes almost completely. In the case of an about 100 nm thick InGaP single layer self-organization leads to an inhomogeneous and not well defined strain distribution which is difficult to predict. Moreover, plastic relaxation occurs due to defect formation. To maintain the lateral strain modulation in the InGaAs-QW we used a two-layer system, namely a combination of a thin compressively strained InGaP layer and a GaAs layer for overgrowth of a free standing triangular grating structure. GID measurements revealed, that in this case the lateral strain modulation in the InGaAs-QW was maintained. It has been demonstrated that FEM calculations can be used to predict the strain distribution within the complete structure after overgrowth. Thus, we have shown that by choosing appropriate MOVPE growth conditions and materials for the second growth step it is possible to preserve the lateral carrier confinement in the InGaAs-QW. Such a planarized structure can be implemented into a device taking advantage of the reduction in free carrier movement.

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