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Selective molecular beam epitaxy (MBE) growth of GaAs/AlAs ridge structures containing 10 nm scale wires and side quantum wells (QWs) and their stimulated emission characteristics

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Abstract

We report on the formation and optical characteristics of ridge waveguide structures containing quantum wires (QWRs). To form a waveguide, we adopted digital alloys, i.e. short-period superlattices (SLs), rapid changes of As fluxes using valved cells, growth temperatures (T_s), and introduction of growth interruptions. Stimulated emission from these ridge QWR laser structures was observed at temperatures 4.7–290 K by optical pumping.

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1. Introduction

Faceted structures by selective MBE growth on a patterned substrate have been explored with great interest as a promising method to organize nanometer size structures in a single continuous growth [1,2]. We have earlier reported the successful growth of ridge QWRs by MBE [3–8].

QWR semiconductor lasers have attracted a great deal of attention with a prediction of improved performances due to the peaked structure of the one-dimensional (1D) density of states [9]. Stimulated emissions from the QWRs on V-groove substrates [1] and the T-shaped QWRs by cleavage and over growth [10] are reported. Since the ridge QWR structures have some advantages in producing large lateral confinement energy and high crystal quality, they are promising to realize room-temperature QWR lasers with novel properties of the 1D system.

2. Experimental procedure

The ridge structures are formed on a GaAs (0 0 1) wafer with mesa stripes at various T_s and the As_4

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flux ranging from 5×10^{-6} to 3×10^{-5} Torr. The growth rates of GaAs and AlAs were 0.25 and 0.1 μm/h, respectively.

The optical waveguide structure is a simple separate confined heterostructure, which consists of two cladding layers (A) and (E) sandwiching two core layers (B) and (D) and an active layer (C).

Ridge shape width (W) and morphologies are strongly dependent on the various growth parameters such as T_s, growth rates, III–V ratio, As flux, growth interruption (GI) and Al composition. The active layer, the core layers and the cladding layers have different requirements, hence we modify the growth conditions according to each layer to keep their morphology. We chose growth parameters under the following considerations. As flux influences both W and uniformity of the ridge structure, when the As flux was less than 3×10^{-7} Torr, the ridge structure tended to have defects missing the top section. When the As flux was very high and exceeded 3×10^{-5} Torr, random faceted structures were formed on the (1 1 0) side plane, and roughened the (1 1 1)B planes. For the cladding layer where the Al content is high, a relatively wide ridge of 100 nm is favorable for light confinement. On the other hand the W of the GaAs active layer should be as narrow as 10 nm. We must grow the cladding layers at higher T_s, and the active layer (C) at lower T_s. The GaAs ridge gets wider during GI due to frequent re-arrangement motion of Ga atoms [7]. However, long GI on the ridge surface may introduce uncontrollable change of shape and oxidation and contamination especially in the AlAs surface which may cause degradation of performance. So we limited the introduction of GI to 5 s only on the GaAs surface. So far the AlGaAs alloy growth on a faceted structure has been difficult, as it often leads to irregular faceted structures and fluctuation in Al local contents, because of different diffusion behaviors of Ga and Al adatoms. We adopted AlGaAs digital alloys which are short-period SLs, because the growth of GaAs and thin AlAs layers of good uniformity had already been established, and it also enables a rapid change of the Al content of the core layer ($x = 0.2$) from cladding layers ($x = 0.4$) during the growth.

The lower cladding layer (A) was grown at 600°C. This layer (A) was an SL $(AlAs_{2.8}/GaAs_{4.2})_{360}$, the averaged Al content of which is 0.4. We then grew the lower core layer (B) at 580°C which was an SL $(AlAs_{1.4}/GaAs_{5.6})_{45}$ followed by an SL of shorter period $(AlAs_{0.7}/GaAs_{2.8})_{10}$. We introduced GI of 5 s after each GaAs layer of this shorter SL layer. Then the T_s was rapidly cooled down to 530°C, and an active GaAs layer (C) of 5 nm was grown. During this growth, we rapidly increased the As flux to 3×10^{-5} Torr, by opening the As valved cell, to obtain a smaller W and a stronger lateral confinement. After the growth of this layer the As flux was decreased to 5×10^{-7} Torr by closing the As valve to maintain the uniformity. The active layer was covered by the core SL $(GaAs_{5.6}/AlAs_{1.4})_{50}$ layer (D) at 580°C. The optical waveguide was completed by the growth of the cladding SL $(AlAs_{2.8}/GaAs_{4.2})_{300}$ layer (E) and a GaAs cover layer (F) of 30 nm thick at 600°C.

3. TEM observation

Fig. 1 shows a dark field image of a cross-sectional TEM micrograph of the ridge QWR laser structure. The clear TEM image supports the high crystalline quality of MBE grown ridge waveguide structures. We can see each period of SL and trace the time evolution of the ridge shape during the growth.

The ridge top region shown as (a)–(e) in Fig. 1, seems brighter than the region grown on (1 1 1)B surface, shown as (A)–(E). A closer investigation reveals that the AlAs layer at the top region is thicker, on the other hand the GaAs layer of the ridge top is thinner than that of the (1 1 1)B region. This makes the Al content at the top region higher than that of the (1 1 1)B region. This is very diffrent from the MOCVD grown wire laser on a V-groove [1].

If W is constant as seen in the growth of layers in (a), (b), and (d2), the amount of atoms moving out of the top region equals that migrating into it. As Al atoms of this region increased, the same amount of Ga atoms left. It is not easy to attribute the cause of loss of Ga atoms at the ridge top region, to either desorbtion or surface diffusion [7]. When W decreases as seen in (d1), Ga atoms migrate more than

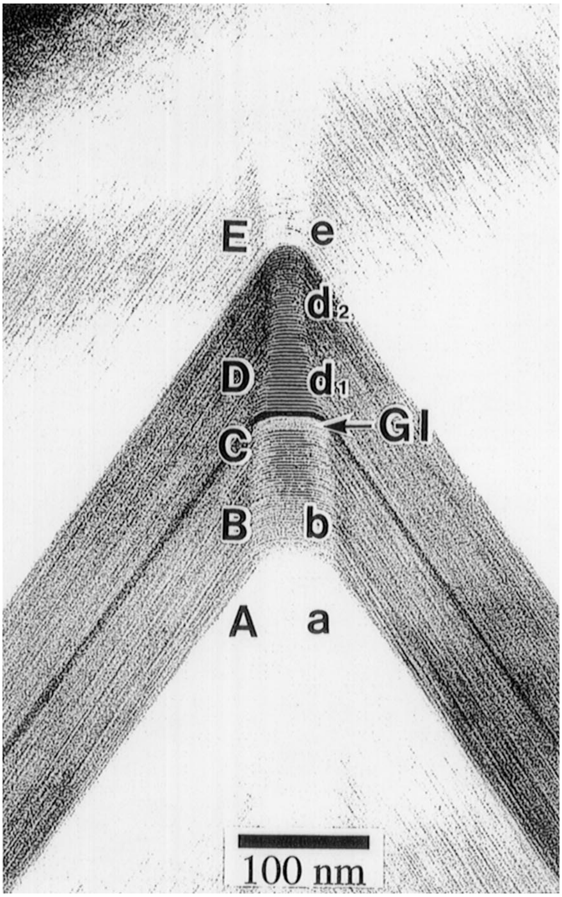

Fig. 1. A TEM cross-sectional dark field image of the sample. Characters A, B, C, D, and E are representive of lower cladding SL layer, lower core SL layer, active layer, upper core SL layer, and upper cladding SL layer, respectively. Lowercase characters indicate ridge top regions of the corresponding layers. GI indicates the SL layer in which growth interruption was introduced.

Al atoms, which enhances the growth rate of ridge top and lowers the local Al content.

Introduction of GI causes intermixing of Ga and Al atoms, as no periodic structure was observed with layers in GI in Fig. 1, nor was there any apparent increase in W, which was observed in a GaAs ridge.

Fig. 2 shows time evolution of the W_S and T_s for several locations of the sample. The W_S of layers have some deviations, the W_S of the end of the cladding layer (A) deviate from 40 to 62 nm. The W_S of the end of the core layer (B) are 21–56 nm.

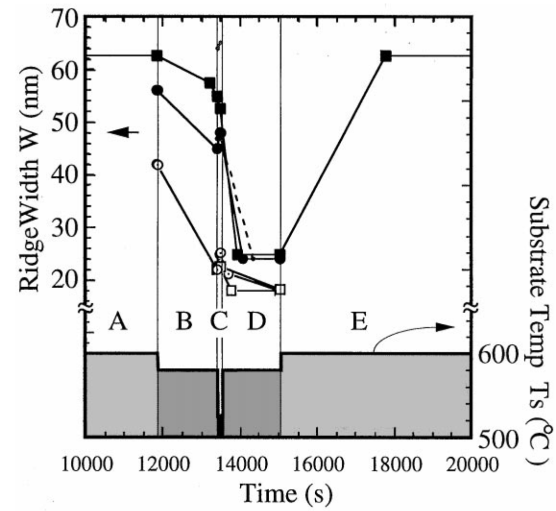

Fig. 2. Time evolution of measured ridge widths (W) and setting substrate temperatures (T_s). The characters A–E correspond to the layers of the sample.

The variation of the W_S of the active layer (C) explains the rather wide PL peak from QWR. Each line indicates the time evolution of W_S for specific sample positions. Where the W of the cladding layer (A) is wide as shown by closed square the W of the active layer (C) tends to be wide, the W of the active layer (C) is strongly affected by the W of the cladding layer (A).

The W of the upper core layer (D) is narrower than active the layer, though it was grown at higher T_s. This delayed reponse of W_S is attributed to two origins; the first is due to the limited heat conductance of the GaAs subtrate and heater, the temperature of the ridge top surface does not change as quickly as indicated by the thermo-couple or the pyrometer. The second is more essential, the change of W is caused by diffusion due to the difference of growth rates on the (0 0 1) top region and that on (1 1 1)B. The difference of MBE growth rates under usual conditions is not as large as that of MOCVD, which causes delay in W_S during the MBE growth.

4. Optical measurement

The samples were cleaved to form optical cavities of various lengths $L = 0.3$–1 mm in the wire

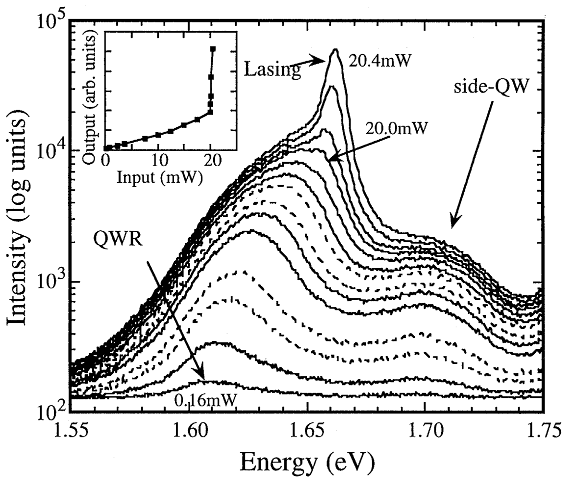

Fig. 3. PL spectra at 8.9 K under various excitation powers. The inset shows the power dependence of the peak at 1.662 eV.

direction. Photoluminescence (PL) spectra were measured at various excitation intensities. The Ti sapphire pulsed laser with an energy of 1.766 eV was used. The pulse duration and repetition rate are 150 fs and 76 MHz.

Fig. 3 shows PL spectra at 8.9 K under various excitation powers. The peak observed at 1.609 eV at weak excitation was PL from ridge QWR. When the excitation power increases, this peak shifts to a higher energy. This blue shift indicates the saturation of carriers in the ground and lower-order excited states in QWRs and the occupation of higher-order excited states. Finally at a pump power of 20 mW, stimulated emission was observed at 1.662 eV. The broad shape of this lasing peak indicates lasing in multi-mode or the contribution of neighboring QWRs. This lasing energy is higher than the ground-state PL energy of QWR by 53 meV, where the energy separation between the ground state of QWR and that of the side-QW was 90–100 meV. Lasing was observed at all temperatures between 4.7 and 290 K [11].

5. Conclusions

A ridge QWR laser structure was formed by using digital alloys, and rapid changes of T_s and As fluxes using valved cells, and the introduction of GI. A close TEM observation reveals plentiful information about the unique nature of MBE ridge growth mechanisms, and some deficiencies of the ridge waveguide structure to be improved.

Stimulated emission was observed in ridge QWR laser structures up to 290 K. The origin of stimulated emission was the transitions between higher-order excited states of QWRs.

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