Studies on GaAs-Based Light-Emitting Devices on Si Substrates for Optoelectronic Integrated Circuits

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1996

Yoshiaki Hasegawa
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Chapter 1. Introduction

1.1 Background

A combination of GaAs-based compound semiconductors with highly developed Si has a great potential for creation of novel semiconductor materials incorporating photonic functions such as light emission and high speed operation into highly integrated Si-based electronic devices. Therefore, GaAs-based semiconductors on Si substrates (GaAs/Si) have attracted much interest as promising materials for future optoelectronic integrated circuits (OEICs). In the conventional Si-based devices, metallic wires and bonds have been used as electrical connections. On the other hand, for example, a Si-based ultra large scale integrated circuit (ULSI) chip with GaAs-based optical input/output (I/O) devices can hopefully promise the increased I/O speed and increased resistance to electromagnetic interference without time delays by optical chip-to-chip interconnects and communications. The communications can be carried out by high-speed optical links coupled through fibers or free-space propagation.

The merits and demerits of heteroepitaxy and direct bonding (epitaxial lift-off) are tabulated in Table I.I. For the realization of such a monolithic integration of GaAs and Si, many researches have been carried out by heteroepitaxy or direct bonding. In the latter, a homoepitaxially grown GaAs layer is transferred and bonded to Si substrate by adhesive metals and intermediate layers after detaching it from GaAs substrate. Direct bonding of GaAs/Si will provide high-quality GaAs film and possible uses of commercially processed Si chips. There are, however, some serious problems of troublesome fabrication, small area uniformity,
Table I.I. Merits and demerits of heteroepitaxy and direct bonding for GaAs on Si.

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<th>Heteroepitaxy</th>
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<td><strong>Merits</strong></td>
<td>Easy fabrication</td>
<td>High-quality GaAs film on Si</td>
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<td>Large area uniformity</td>
<td>Possible uses of commercially processed Si chips</td>
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<td>as a substrate for other semiconductor devices</td>
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<td>for solar cell applications</td>
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<td>Low cost</td>
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<td><strong>Demerits</strong></td>
<td>Induced thermal stress</td>
<td>Troublesome fabrication</td>
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<td>High density dislocations</td>
<td>Troublesome fabrication</td>
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<td>Induced thermal stress</td>
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<td>High resistance at the bonding interface</td>
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induced thermal stress and high resistance at the bonding interface. The heteroepitaxial growth of GaAs/Si has been actively studied over the past decade by molecular beam epitaxy (MBE) and metalorganic chemical vapor deposition (MOCVD). This heteroepitaxial growth technique has some attractive merits such as easy fabrication, large area uniformity and low cost. However, GaAs/Si material grown by heteroepitaxy suffers from three major problems:

(1) the anti-phase domain (APD) generation due to the polar (GaAs)-on-nonpolar (Si) system;

(2) the generation of high density of threading dislocations (>10^6 cm^-2) due to the ~4 % mismatch between the lattice constants of GaAs (0.5653 nm) and Si (0.5431 nm); and

(3) the large biaxial tensile stress (~10^9 dyn/cm^2) generated during cooling after the growth due to the difference in the thermal expansion coefficients of GaAs (6.4×10^-6 /K) and Si (2.3×10^-6 /K).

The relationship between lattice constants and thermal expansion coefficients at 300 K for various semiconductor materials is shown in Fig. 1.1. The first problem can lead to a low minority carrier lifetime^{28} and poor morphology^{29}. The latter two problems can contribute to the reduced minority carrier lifetimes^{30} and the severe bowing of the GaAs/Si wafer which introduces microcracks^{31-34} of the GaAs layer on Si. Thermally induced stress also enhances the migration of dislocations^{35}. Considering the widespread applications of GaAs/Si, it is believed that the heteroepitaxy has more attractive advantages rather than the direct bonding. Therefore, for the growth of high-quality GaAs/Si, the above-
Fig. 1.1. Relationship between lattice constants and thermal expansion coefficients at 300 K for various semiconductor materials.
mentioned problems must be overcome.

1.2 Heteroepitaxial Growth of GaAs on Si

The above-mentioned first problem (APD generation) has been already solved by use of nominally (100)-oriented Si substrates with a slight misorientation between 2 and 4° towards the [011] direction\[^{11,29,36-38}\]. This effect is accomplished through the formation of a (100) Si surface with a large density of double-atomic-layer steps on the Si surface as a result of the heat treatment of the misoriented Si substrate. Therefore, the Si substrates oriented 2° off (100) towards the [011] direction were used in this study. Although a single-domain GaAs layer was obtained by use of the misoriented Si substrate, the fundamental problems associated with the differences of lattice constant and thermal expansion coefficient still remain.

Most of the dislocations are confined at the GaAs/Si interface when grown by conventional MOCVD two-step growth technique\[^{39}\] with an initial nucleation layer growth at relatively low temperature (~400 °C) and a subsequent growth at normal growth temperature. Growth at low temperature results in a mode approaching two-dimensional GaAs nucleation layer growth\[^{40}\]. The number of threading dislocations can be decreased by inserting strained-layer superlattices (SLSs)\[^{11}\] such as InGaAs/GaAs\[^{41,42}\] and GaAsP/GaAs\[^{9,43}\]. Nozawa and co-workers\[^{44}\] have reported that the etch pit density (EPD) was drastically reduced to $7 \times 10^4$ cm$^{-2}$ by a combination of InGaAs/GaAs SLSs inserted into GaAs layers on Si grown by migration enhanced epitaxy (MEE) at 300 °C. Recently, other new approaches such as (GaAs)$_m$(GaP)$_n$ strained short-period superlattices (SSPSs)\[^{45}\], ZnSe\[^{46,47}\] and GaSe\[^{48}\] buffer layers, thin
Si intermediate layer\textsuperscript{13,49,50} and so on, have been proposed. In addition, in-situ\textsuperscript{41,42,51,52} and ex-situ\textsuperscript{11,53-55} thermal annealings are also effective in reducing the dislocation density due to the increase of dislocation velocity and interaction. The growth on the pre-patterned Si substrates is also useful for reducing the dislocation density\textsuperscript{12,55-62}. For example, Ismail et al.\textsuperscript{57} have demonstrated low dislocation density of \( \sim 10^5 \) cm\(^{-2} \) in the GaAs layer grown on a sawtooth patterned Si substrate. This effect seems to be caused by the presence of a thin SiO\(_2\) layer at the GaAs/Si interface, resulting in the relaxation of lattice mismatch strain.

The reduction of thermal stress has been demonstrated by uses of selective-area growth\textsuperscript{63-66} and post-growth patterning\textsuperscript{42,54,67,68}. The residual stress and the defect density in the GaAs/Si decrease with reductions of the growth area. This is due to the fact that the stress near the edge is released, because the edge facet is free to move\textsuperscript{67}. Another approach to reduce the thermal stress is proposed by using a selective etching AlGaAs release layer to undercut GaAs mesa structures\textsuperscript{12,43,69-76}. The completely undercut mesas are restrained in their original positions by photoresist positioners, and the residual stress is completely removed by this mesa release and deposition (MRD) technique\textsuperscript{69-71}. Sakai et al.\textsuperscript{43,72-76} have proposed the undercut GaAs/Si (UCGAS) where only part of the GaAs layer is undercut. This UCGAS is useful for both the reduction of dislocation density and thermal stress after post-growth annealing. This effect is believed to be caused by the absence of the GaAs/Si interface which acts as a source for dislocation supply.

In spite of these efforts, the threading dislocation density below \( 10^4 \) cm\(^{-2} \), which is allowed to use for practical devices
(especially light-emitting devices), has not been realized yet.

1.3 Device Applications: Light-Emitting Devices on Si

In spite of the above-mentioned problems, numerous GaAs-based electrical and optical devices, including field-effect transistors (FETs)\(^{77}\), high-electron-mobility transistors (HEMTs)\(^{78,79}\), modulators\(^{80}\), waveguides\(^{81}\), photodetectors\(^{82}\), solar cells\(^{10,83}\), LEDs\(^{74,76}\) and lasers\(^{84-91}\), have been fabricated on Si substrates by heteroepitaxy.

The characteristics of GaAs/Si electrical devices have been improved to the same level as that of GaAs/GaAs, because the majority carriers are hardly influenced by the dislocation density.

On the other hand, the characteristics of the light-emitting devices operated by minority carriers are easily affected by dislocations\(^{10}\). However, most of the interest has been focused on the realization of reliable GaAs-based LEDs and lasers on Si because of needs for optical interconnects in future OEICs. In 1984, Windhorn et al.\(^{84}\) were the first to report the pulsed operation at 77 K for an AlGaAs/GaAs double-heterostructure (DH) laser on Si grown by MBE (threshold current density: \(J_{th} = \sim 10\, \text{kA/cm}^2\)). They also reported that MBE-grown AlGaAs/GaAs single quantum well (SQW) laser was operated under pulsed condition at room-temperature in 1986 (\(J_{th} = \sim 4\, \text{kA/cm}^2\))\(^{85}\). In the same year, the first AlGaAs/GaAs DH laser on Si grown by MOCVD was reported by Sakai et al.\(^{86}\). This laser was operated under pulsed condition at room-temperature (\(J_{th} = \sim 1.5\, \text{kA/cm}^2\)). In 1987, Deppe et al.\(^{87}\) succeeded in the first room-temperature continuous-wave (cw) operation for an AlGaAs/GaAs SQW laser grown by MOCVD on MBE-grown GaAs-coated Si substrate (\(J_{th} = \sim 2\, \text{kA/cm}^2\)). This laser was operated for \sim 4.5\, \text{h} under cw...
condition at room-temperature. The cw lifetime was increased to ~16 h by use of microcracks for an AlGaAs/GaAs SQW laser on Si\textsuperscript{33}) in 1988. In 1989, room-temperature cw operation for an AlGaAs/GaAs laser on Si grown by only MOCVD was achieved by Egawa et al.\textsuperscript{88}). In 1990, Deppe et al.\textsuperscript{89,90}) reported the first demonstration of vertical-cavity surface-emitting laser (VCSEL) on Si. The cw lifetime of ~56.5 h was reported by Choi et al.\textsuperscript{91}) for an AlGaAs/InGaAs SQW laser on Si in 1991. For the LEDs on Si, Wada et al.\textsuperscript{74}) reported the lifetime over 2500 h for an undercut AlGaAs/GaAs LED under direct current (dc) condition at room-temperature in 1992. Furthermore, in 1993, an AlGaAs/GaAs SQW laser on a sawtooth Si substrate was operated under room-temperature cw condition for ~100 h\textsuperscript{12}). The summary of recent advances in the GaAs-based light-emitting devices on Si is shown in Table I.II. In this table, those of the InP-based light-emitting devices on Si\textsuperscript{92-95}) are also shown as references. The InP-based lasers on Si have advantages of reliability over the GaAs-based lasers on Si, because the dark-line defect (DLD) growth velocity in the active layer of the InP-based lasers is about two order lower than that of the GaAs-based lasers\textsuperscript{96,97}). Therefore, a long-wavelength (1.5 \textmu m) InGaAsP/InGaAs multi quantum well (MQW) laser on Si\textsuperscript{94,98}) has exhibited no degradation after more than 8000 h under cw condition at room-temperature, despite the high dislocation density of ~10\textsuperscript{7} cm\textsuperscript{-2}.

1.4 Purpose and Organization of Dissertation

GaAs-based light-emitting devices on Si suffer from rapid degradation, and reliable GaAs-based devices on Si for practical use have not been realized yet. The understanding of the rapid degradation mechanisms is very important in order to obtain highly
Table I.II. Summary of recent advances in GaAs- and InP-based light-emitting devices on Si.

<table>
<thead>
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<th>Year</th>
<th>GaAs-based light-emitting devices</th>
<th>InP-based light-emitting devices</th>
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<td>1995</td>
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| 1993 | AlGaAs/GaAs SQW laser on sawtooth Si (MOCVD) 
      | lifetime ~100 h (300 K, CW) [IMEC] |
| 1992 | Undercut (UC) AlGaAs/GaAs LED on Si (MOCVD) 
      | lifetime > 2500 h (300 K, DC) [Tokushima Univ.] |
      | Stress-free mesa release and deposition (MRD) AlGaAs/GaAs DH laser on Si (MOCVD) [MIT] |
| 1991 | AlGaAs/GaAs SQW laser on Si with microcracks (MBE+MOCVD) 
      | lifetime ~500 h (200 K, CW) [Illinois Univ.] |
      | AlGaAs/InGaAs SQW laser on Si (MOCVD) 
      | lifetime ~56.5 h (300 K, CW) [MIT] |
      | AlGaAs/GaAs SQW laser on Si using selective growth (MOCVD) 
      | lifetime ~10 h (300 K, CW) [MIT] |
| 1990 | AlGaAs/GaAs VCSEL on Si (MBE) 
      | I₀ = 125 mA (300 K, Pulsed) [AT&T] |
| 1989 | AlGaAs/GaAs SQW laser on Si (MOCVD) (300 K, CW) [NIT] |
| 1988 | AlGaAs/GaAs SQW laser on Si with microcracks (MBE+MOCVD) 
      | lifetime ~16 h (300 K, CW) [Illinois Univ.] |
| 1987 | AlGaAs/GaAs SQW laser on Si (MOCVD) (375 K, Pulsed) 
      | (< 160 K, CW) [AT&T] |
| 1986 | AlGaAs/GaAs DH laser on Si (MOCVD) (290 K, Pulsed) [NIT] |
      | AlGaAs/GaAs SQW laser on Si (MBE) (300 K, Pulsed) [MIT] |
| 1984 | AlGaAs/GaAs DH laser on Si (MBE) (77 K, Pulsed) [MIT] |
|      | InP/InGaAsP DH laser on Si (MOCVD+VME) 
      | lifetime > 800 h (300 K, CW) (λ = 1.3 μm) [NTT] |
|      | InGaAsP/InGaAs MQW laser on Si (MOCVD+VME) lifetime > 8000 h (300 K, CW) (λ = 1.5 μm) [NTT] |
      | *VME; vapor mixing epitaxy |
reliable GaAs-based LEDs and lasers on Si. Novel approaches must be proposed for suppression of the degradation.

The dissertation is composed of six chapters, and each of them is summarized below.

In chapter 2, degradation mechanisms related to DLDs for AlGaAs/GaAs quantum well lasers on Si are described. Using electroluminescence (EL) topography, the detailed observation of rapid degradation of the lasers on Si is carried out under cw aging operation at room-temperature. The generation and rapid expansion of the <100> DLDs is clearly observed, resulting in the deterioration of lasing characteristic parameters such as internal differential quantum efficiency, differential gain coefficient and internal loss. Additionally, it is also shown that the DLD growth velocity depends on the injected current density.

In chapter 3, increased lifetimes of strained AlGaAs/InGaAs quantum well lasers on Si with InGaAs intermediate layers (InGaAs ILs) are presented. The increased lifetimes are achieved by both introducing InGaAs ILs and replacing the conventional GaAs active layers with the InGaAs layers. EL topography shows the suppression of <100> DLD growth velocity with the increase of In content in the InGaAs active layer. Furthermore, a remarkable improvement in reliability of a post-growth annealed AlGaAs/In_{0.07}Ga_{0.93}As laser on Si with InGaAs IL is also given.

In chapter 4, improved characteristics of AlGaAs/GaAs light-emitting devices on Si by reducing the active regions are described. The structures of vertically-stacked quasi GaAs quantum wires grown on a V-grooved GaAs/Si substrate are confirmed by a high-resolution scanning electron microscopy (SEM). The extremely low-threshold current AlGaAs/GaAs quantum wire-like laser on Si is also demonstrated. Furthermore, an AlGaAs/GaAs LED on Si with self-
formed GaAs islands active region is presented. The shape and size of GaAs islands formed by droplet-epitaxy are characterized by an atomic force microscopy (AFM). The reliability of this LED is also demonstrated.

In chapter 5, the fabrication of AlGaAs/GaAs VCSELs on Si is presented. Room-temperature pulsed operation for a VCSEL on Si with bottom mirror of a 20-pair of AlAs/GaAs quarter-wave multilayer distributed Bragg reflector (DBR), top mirror of nonalloyed AuZn/Au and a SQW active layer is described. Furthermore, it is demonstrated that a VCSEL on Si with bottom mirror of a 23-pair of AlAs/Al_{0.1}Ga_{0.9}As DBR, top mirror of transparent Au and ten quantum wells succeeded in the cw operation at low-temperature up to 150 K.

Chapter 6 is the summary of this dissertation.
References


Chapter 2. Degradation Mechanisms of AlGaAs/GaAs Quantum Well Lasers on Si Substrates

2.1 Introduction

Observation of the degradation of GaAs-based lasers on Si is very important for understanding of the rapid degradation mechanism and development of long-life reliable lasers. Van der Ziel and co-workers\textsuperscript{1,2} have reported that growth of nonradiative dark regions including DLDs was observed in an AlGaAs/GaAs SQW laser on Si using pulsed current excitation. Martins et al.\textsuperscript{3} have reported that the initial degradation of the AlGaAs/GaAs SQW laser on Si occurred in the vicinity of the p-n junction. Recently, Egawa et al.\textsuperscript{4} have observed optical and electrical degradation of the AlGaAs/GaAs SQW laser on Si. However, rapid degradation mechanisms related to DLDs under cw aging operation have not been clarified.

In this chapter, the influences of DLDs on lasing characteristics and the dependence of DLD growth velocity on the injected current density for the AlGaAs/GaAs quantum well laser on Si are described. This chapter is organized as follows: In section 2.2, epitaxial growth and fabrication process of the AlGaAs/GaAs quantum well laser on Si are described. The influences of DLDs on lasing characteristics under cw aging operation are presented in section 2.3. In section 2.4, the dependence of $<100>$ and $<110>$ DLD growth velocities on the injected current density under direct current (dc) operation is described. This chapter is concluded in section 2.5.

2.2 Epitaxial Growth and Fabrication Process
The samples used in this study were grown on n⁺-Si substrates (Sb-doped, 0.02 Ω·cm) oriented 2° off (100) towards the [011] direction by MOCVD, using an rf-heated horizontal reactor operated at atmospheric pressure. Trimethylgallium (TMG) and trimethylaluminum (TMA) were used as group-III sources, and arsine (AsH₃) was used as a group-V source. Hydrogen selenide (H₂Se) and diethylzinc (DEZ) were used as n- and p-dopants, respectively. The carrier gas was hydrogen (H₂). The substrates were rinsed in organic solvents and then cleaned in solutions of H₂SO₄:H₂O₂ (4:1 by volume) and HF:H₂O (1:1 by volume). Prior to the growth, the substrates were heated in H₂ atmosphere at 1000 °C for 10 min in order to remove the native oxide from the substrate surface. The growth temperature was 750 °C except for the initial GaAs nucleation layer which was grown on Si at 400 °C in the two-step growth technique. An initial 10-nm-thick undoped GaAs nucleation layer and a 2.1-μm-thick n⁺-GaAs buffer layer (Se-doped, 2×10¹⁸ cm⁻³) were grown. During growth of the n⁺-GaAs layer, the substrate temperature was cycled five times from 350 to 850 °C in an AsH₃ atmosphere. This thermal cycle annealing is effective in reducing the threading dislocations in the active layers of the lasers⁵,⁶). Figure 2.1 shows a schematic cross-section of an AlGaAs/GaAs triple quantum well (TQW) laser on Si. The laser structure comprised a 1.0-μm-thick n-Al₀.₇Ga₀.₃As lower cladding layer (Se-doped, 1×10¹⁸ cm⁻³), a 60-nm-thick undoped Al₀.₇Ga₀.₃As lower optical confining layer, three 9-nm-thick undoped GaAs quantum wells separated by 5.5-nm-thick undoped Al₀.₇Ga₀.₃As barrier layers, a 60-nm-thick undoped Al₀.₇Ga₀.₃As upper optical confining layer, a 1.0-μm-thick p-Al₀.₇Ga₀.₃As upper cladding layer (Zn-doped, 1×10¹⁸ cm⁻³) and an 90-nm-thick p⁺-GaAs contact layer (Zn-doped, 1×10¹⁹ cm⁻³). After the growth, laser devices were fabricated as follows. A 0.1-μm-
Fig. 2.1. Schematic cross-section of AlGaAs/GaAs TQW laser on Si.
thick SiO₂ insulating layer was deposited on the p⁺-GaAs contact layer and 10-μm-wide stripe contact windows were opened with a 300 μm pitch along the <110> direction by chemical etching of SiO₂. Then, 50-nm-thick Ti/150-nm-thick Au were evaporated on the p⁺-GaAs layer as the p-side electrode. After thinning the n⁺-Si substrate to a thickness of 100 μm, 50-nm-thick AuSb/150-nm-thick Au were evaporated on the Si substrate as the n-side electrode. The contacts were annealed in a N₂ atmosphere at 380 °C for 30 s to reduce the series resistance. The samples were cleaved into chips and mounted on In-soldered Cu heatsinks in the p-side-up configuration. In order to observe the electroluminescence (EL) topograph of the active region with a Si vidicon TV camera (Hamamatsu C2741-03), part of the p-side electrode of the lasers was thinned by chemical etching in a solution of KI : I₂ : H₂O (113 g : 65 g : 100 cm³)⁷.

The AlGaAs/GaAs TQW lasers on Si with 120-800 μm cavity length were prepared. The cw aging test was carried out at a constant output power of 1 mW/facet under automatic power control (APC) operation at a heatsink and ambient temperature of 20 °C. The degree of degradation was studied through the change of the increase ratio of cw driving current (ΔI), lasing characteristic parameters such as internal differential quantum efficiency (ηᵢ), internal loss (αᵢ) and differential gain coefficient (β), and EL topograph of the active region⁸). ΔI was determined using

\[ ΔI = \left( \frac{I - I₀}{I₀} \right) \times 100, \quad (2.1) \]

where I and I₀ are driving current and its initial value, respectively.
2.3 Influences of DLDs on Lasing Characteristics

All AlGaAs/GaAs TQW lasers on Si degraded within a few minutes. No relationship between the lifetime and the cavity length of the lasers was observed in this experiment. Figure 2.2 shows the log-log relationship between the increase ratio of driving current (ΔI) and aging time for the lasers on Si. The driving current increases almost linearly with operating time at the initial slow degradation stage of ΔI < 60-80 %. However, at the subsequent rapid degradation stage of ΔI > 60-80 %, the driving current increases approximately as the eighth power of operating time. These results indicate that the degradation mechanisms differ in the initial slow degradation stage and the subsequent rapid one. In order to investigate this difference, the EL topographs of the laser on Si were observed at each degradation stage. Typical EL topographs at each degradation stage of ΔI = 0, 40, 100, 200 % are shown in Fig. 2.3. At the initial slow degradation stage of ΔI = 40 %, generation of a few <100> DLDs was observed. Subsequently, expansion of <100> DLDs was clearly observed at the rapid degradation stages of ΔI = 100, 200 %. Detailed observations of DLD growth are described in section 2.4.

In order to observe the change of lasing characteristics for the lasers on Si, light-versus-current (L-I) characteristics were measured at 20 °C at each degradation stage by interrupting the aging test. Figure 2.4 shows the typical L-I characteristics of the 425-μm-long laser on Si under the pulsed condition at each degradation stage. Current pulses 0.2 μs long were used at 5 kHz repetition rate. It can be seen that the threshold current increases and the external differential quantum efficiency (η_d) decreases. For lasers on Si with various cavity lengths, the DLD densities observed by EL topography were the same at each
Fig. 2.2. Log-log relationship between increase ratio of driving current (ΔI) and aging time for AlGaAs/GaAs lasers on Si.
Fig. 2.3. Typical EL topographs of the laser on Si at each degradation stage of ΔI=(a) 0, (b) 40, (c) 100 and (d) 200 %. 
Fig. 2.4. Typical L-I characteristics of the laser on Si under pulsed condition at each degradation stage.
degradation stage. Therefore, the pulsed measurement was carried out for the lasers in order to estimate the change of the lasing characteristics. The lasing characteristic parameters such as $\eta_i$, $\alpha_i$ and $\beta$ were determined using $^9$, $^{10}$

$$
1/\eta_d = (1/\eta_i) \left[ 1 + \alpha_i L/\ln(1/R) \right],
$$

(2.2)

$$
J_{th} = J_0/\eta_i + (\eta_i \beta \Gamma)^{-1} \left[ \alpha_i + L^{-1}\ln(1/R) \right],
$$

(2.3)

where $L$ and $R$ are cavity length and reflectivity of the mirror facet, respectively. The $J_{th}$ is related to $J_0$ and $\Gamma$ which are transparency current density and the optical confinement factor, respectively. Using the fact that the value of $R$ is generally 0.32 for an uncoated facet, the values of $\eta_i$ and $\alpha_i$ can be obtained from the relationship of $1/\eta_d$ versus $L$ in eq. (2.2). This relationship at each degradation stage of $\Delta I = 0, 40, 100$ % is shown in Fig. 2.5. It is found that $\eta_i$ gradually decreases as the degree of degradation progresses. Assuming that $\Gamma$ is 0.081 for the TQW laser $^{11}$, the values of $b$ and $J_0$ can be estimated from eq. (2.3). Figure 2.6 shows the relationship of $J_{th}$ versus $L^{-1}\ln(1/R)$ at each degradation stage. $\beta$ is found to decrease gradually. Typical changes of lasing characteristics such as $\eta_i$, $\alpha_i$ and $\beta$, which are normalized by their initial values before degradation, are shown in Fig. 2.7. These initial values, $\eta_i(0)$, $\alpha_i(0)$ and $\beta(0)$, were 0.77, 25 cm$^{-1}$ and 0.85 cm/A, respectively. At the initial slow degradation stage of $\Delta I < 60-80$ %, $\eta_i$ decreases markedly. The decrease in $\eta_i$ seems to be caused by the increased number of nonradiative recombination centers. In order to study the increase of nonradiative recombination centers, the current-versus-voltage (I-V) characteristic of the laser on Si was measured at each degradation.
Fig. 2.5. Relationship of $1/\eta_d$ versus $L$ for the laser on Si at each degradation stage.
Fig. 2.6. Relationship of $J_{th}$ versus $L^{-1}\ln(1/R)$ for the laser on Si at each degradation stage.
Fig. 2.7. Typical changes of $\eta_1/\eta_1(0)$, $\alpha_i/\alpha_i(0)$ and $\beta/\beta(0)$, normalized by initial values before degradation, as a function of increase ratio of driving current ($\Delta I$) for the laser on Si.
stage. Figure 2.8 shows the typical change of recombination current ($I_r$) which has the ideality factor of 2. $I_r$ was normalized by the initial value [$I_r(0)$] of 2.5 nA at 0.6 V, and the change of $I_r$ was evaluated at this voltage. During the cw aging operation, the ideality factor was increased from 2.06 to 2.20. As shown in Fig. 2.8, $I_r$ drastically increased at the initial slow degradation stage. Thus, the slow degradation related to the generation of <100> DLDs is probably caused by the increased nonradiative recombination current due to the increase of nonradiative recombination centers. Furthermore, the degradation is also caused by the decrease in $\beta$, as shown in Fig. 2.7. Although it can be seen that $\alpha_i$ slightly decreases at this degradation stage, the reason for this is not yet understood. At the subsequent rapid degradation stage of $\Delta I > 60-80 \%$, $\eta_i$ negligibly changes, whereas $\alpha_i$ drastically increases and $\beta$ markedly decreases. These results suggest that the rapid degradation due to the expansion of <100> DLDs is caused by increased absorption of emitted light and decreased gain.

2.4 Dependence of DLD Growth Velocity on Injected Current Density

2.4.1 <100> DLD Growth Velocity

It has been reported that the <100> DLD growth velocity for an AlGaAs/GaAs laser grown on the GaAs substrate generally depends on the injected current density\textsuperscript{12}). The <100> DLDs are caused by the formation of complicated dislocation networks due to the recombination-enhanced dislocation climb (REDC) motion in the active region\textsuperscript{13,14}). The injected carriers trapped by nonradiative recombination centers at DLDs emit the phonon energy which causes the lattice vibrations, resulting in enhancement of the REDC motion. For an AlGaAs/GaAs laser on Si, the dependence of the <100> DLD...
Fig. 2.8. Typical change of recombination current ($I_R$) as a function of increase ratio of driving current ($\Delta I$) for the laser on Si.
growth velocity on the injected current density was studied for the first time. Figure 2.9 shows typical magnified EL topographs of an AlGaAs/GaAs TQW laser on Si under dc operation at a constant current density of 0.5 kA/cm\(^2\) and temperature of 20 °C. No dark defects were observed before degradation [Fig. 2.9(a)]. At the first degradation stage [Fig. 2.9(b)], a few <100> DLDs were generated. In the subsequent stage [Fig. 2.9(c)], these DLDs extended toward the center of the active region.

Figure 2.10 shows the dependence of the <100> DLD growth velocity on the injected current density for an AlGaAs/GaAs TQW laser on Si at 20 °C. It can be seen that the DLD growth velocity strongly depends on the current density. The estimated DLD growth velocity was 14–25 μm/h at a current density of 1 kA/cm\(^2\). On the other hand, it was reported that the <100> DLD growth velocity was estimated to be 2–10 μm/h at about 1 kA/cm\(^2\) for an AlGaAs/GaAs SQW laser on GaAs\(^1\). The faster DLD growth velocity of the laser on Si is thought to be caused by the larger number of nonradiative recombination centers such as point defects in the active region than that of the laser on GaAs. It is known that the DLD growth is also enhanced by the rise of junction temperature of the laser\(^1\)\(^\text{4,16}\). The dependence of the <100> DLD growth velocity on the junction temperature was also studied for the laser on Si. The junction temperature is influenced by the injected current density and the ambient temperature under cw operation. The rise of junction temperature was estimated by measurement of the wavelength in the LED mode, because the wavelength shifts continuously by the rise of junction temperature. For example, the peak wavelength at 2.5 kA/cm\(^2\) was about 5.4 nm longer than that at 0.5 kA/cm\(^2\). Thus, it can be estimated that the junction temperature at 2.5 kA/cm\(^2\) is about 30 °C higher than that at 0.5 kA/cm\(^2\), because it is observed

-33-
Fig. 2.9. Typical magnified EL topographs of the laser on Si after (a) 0, (b) 7, (c) 21 and (d) 45 min under dc operation at a constant current density of 0.5 kA/cm² and temperature of 20 °C.
Fig. 2.10. Dependence of the $<100>$ DLD growth velocity on the injected current density for the AlGaAs/GaAs laser on GaAs:

- $2 - 10 \, \mu m/h$ [15]
that the wavelength of the laser on Si is shifted by about 0.18 nm/°C with variation in ambient temperature. For the current density of 0.5 kA/cm² at 50 °C, the DLD growth velocity was estimated to be 15-20 μm/h. This growth velocity is much slower than that of 2.5 kA/cm² at 20 °C. This result indicates that the DLD growth velocity depends more strongly on the injected current density than on the junction temperature. For suppressing the rapid degradation of the AlGaAs/GaAs laser on Si, the reduction of operating current density is very effective because this reduces the <100> DLD growth velocity.

2.4.2 <110> DLD Growth Velocity

The <100> DLDs have effective width in the <110> direction, and thus are regarded as <110> DLDs. The cause of <110> DLDs is thought to be the growth of dislocations due to the recombination-enhanced dislocation glide (REDG) motion. Matsui et al. have also demonstrated that the <100> DLD was propagated by a combination of REDC and REDG motions. Therefore, the <110> DLDs in this study is thought to be caused by the REDG motion. The <110> DLDs for an AlGaAs/GaAs laser on GaAs are often observed by introducing high stress (~10⁹ dyn/cm²). Figure 2.11 shows the dependence of the <110> DLD growth velocity on the injected current density for an AlGaAs/GaAs TQW laser on Si at 20 °C. This estimated DLD growth velocity was 5-8 μm/h at a current density of 1 kA/cm². For the laser on Si, the <110> DLD growth velocity was estimated to be about 1/3 of the <100> DLD growth velocity. In contrast, it was reported that the <110> DLD growth velocity was about 1/10 of the <100> DLD growth velocity for an AlGaAs/GaAs laser on GaAs in the absence of high stress. It appears that the faster <110> DLD growth velocity for the laser on Si is caused by its much higher...
Fig. 2.11. Dependence of the $<110>$ DLD growth velocity on the injected current density for the laser on Si at 20 °C.
residual tensile stress than that of the laser on GaAs.

Finally, a rapid degradation mechanism related to increased internal loss (absorption loss) due to both <100> and <110> DLDs is considered for an AlGaAs/GaAs laser on Si, as shown in Fig. 2.12. The <100> DLD is represented as an effective absorption region with effective width $W_a$ and effective absorption coefficient $\alpha_{d1}^{19}$. For an AlGaAs/GaAs laser on GaAs with $W_a$ of ~5 $\mu$m and $\alpha_{d1}$ of ~200 cm$^{-1}$, it was reported that the degradation due to the increase of $\alpha_i$ was mainly caused by the increase in the number of <100> DLDs ($N_{d1}$)$^{19}$. In contrast, the rapid degradation related to the increase of $\alpha_i$ for an AlGaAs/GaAs laser on Si seems to be caused by the increase of both $N_{d1}$ and $W_a$. The increase of $W_a$ is probably caused by the growth of <110> DLDs which result from the high residual stress for the laser on Si. The reduction of this stress seems to be also very important for the improvement of reliability of the laser on Si.

2.5 Conclusions

The influences of DLDs on lasing characteristics and the dependence of the DLD growth velocity on the injected current density for an AlGaAs/GaAs quantum well laser on Si have been studied. The generation of <100> DLDs caused the decrease of $n_i$ and the slow increase of driving current. At the subsequent rapid degradation stage, expansion of the DLDs caused the remarkable increase of $\alpha_i$ and the decrease of $\beta$, and finally resulted in rapid increase of driving current. Furthermore, it was also found that the <100> and <110> DLDs growth velocities strongly depended on the injected current density. The <100> DLDs seem to have been caused by the formation of complicated dislocation networks due to the REDC motion in the vicinity of the threading dislocations in the active...
Fig. 2.12. Rapid degradation mechanism related to increased internal loss for the laser on Si.
region. On the other hand, the <110> DLDs are thought to have originated from the growth of dislocations due to the REDG motion by the large residual thermal stress.
References

Chapter 3. AlGaAs/InGaAs Quantum Well Lasers on Si Substrates with InGaAs Intermediate Layers

3.1 Introduction

It has been recently reported that strained AlGaAs/InGaAs quantum well lasers grown on GaAs substrates, which are promising for new applications such as pumping Er\textsuperscript{3+}-doped optical fiber amplifiers at 980 nm\textsuperscript{1-3)}, show very slow degradation rates and long lifetimes\textsuperscript{4-12}). Waters and co-workers\textsuperscript{7,9}) have observed that strained AlGaAs/InGaAs SQW lasers on GaAs are significantly more resistant to <100> DLD propagation than AlGaAs/GaAs lasers on GaAs. Fukagai et al.\textsuperscript{10}) have reported that no <100> DLD was observed and DLDs grew in the <110> direction for the strained AlGaAs/InGaAs double quantum well (DQW) laser on GaAs. In addition, they have demonstrated that the <110> DLD growth velocity in the AlGaAs/InGaAs laser on GaAs was estimated to be ~1/100 of that for the <100> DLD growth velocity in the AlGaAs/GaAs laser on GaAs\textsuperscript{10}). This suppressed <100> DLD growth is believed to result from the dislocation pinning due to the introduction of In in the GaAs active layer. Kirkby\textsuperscript{13}) has shown that dislocation motion can be drastically reduced by the incorporation of In (and other group-III or V impurities) into GaAs, which is similar to alloy hardening of metals\textsuperscript{14}). To date, there has been only one demonstration of the increased lifetime for GaAs-based lasers on Si by replacing the conventional GaAs quantum well active layer with the InGaAs layer\textsuperscript{15}). However, differences of degradation mechanism between AlGaAs/GaAs and AlGaAs/InGaAs quantum well lasers on Si have not been clarified. Furthermore, for the AlGaAs/InGaAs laser on Si, the relationship between reliability and residual stress in the InGaAs
active layer has not been also made clear.

In this chapter, the reliable AlGaAs/InGaAs SQW lasers on Si with InGaAs dislocation suppressing intermediate layers (InGaAs ILs) are demonstrated and the differences of degradation mechanism between AlGaAs/GaAs and AlGaAs/InGaAs SQW lasers on Si are also studied. This chapter is organized as follows: The epitaxial growth and fabrication process of the AlGaAs/InGaAs lasers on Si are described in section 3.2. In section 3.3, lasing characteristics of these lasers are presented. The reliability and degradation mechanism of the lasers are described in section 3.4. In section 3.5, increasing lifetime of the AlGaAs/InGaAs laser on Si by post-growth annealing is presented. This chapter is concluded in section 3.6.

3.2 Epitaxial Growth and Fabrication Process

Figure 3.1 shows a schematic cross-section of the AlGaAs/InGaAs SQW laser on Si with InGaAs IL. The 20-nm-thick n\textsuperscript{+}-In\textsubscript{0.08}Ga\textsubscript{0.92}As IL was inserted between a 2.0-\textmu m-thick n\textsuperscript{+}-GaAs buffer layer and a 1.0-\textmu m-thick n-Al\textsubscript{0.7}Ga\textsubscript{0.3}As lower cladding layer. A 70-nm-thick undoped Al\textsubscript{0.3}Ga\textsubscript{0.7}As lower confining layer, various 9-nm-thick undoped In\textsubscript{x}Ga\textsubscript{1-x}As (x=0, 0.02, 0.07, 0.10) SQW active layer, a 70-nm-thick undoped Al\textsubscript{0.3}Ga\textsubscript{0.7}As upper confining layer, a 1.0-\textmu m-thick p-Al\textsubscript{0.7}Ga\textsubscript{0.3}As upper cladding layer, and an 80-nm-thick p\textsuperscript{+}-GaAs contact layer were subsequently grown on the n-Al\textsubscript{0.7}Ga\textsubscript{0.3}As lower cladding layer. Thermal cycle annealing was performed five times by varying the substrate temperature between 350 and 850 °C during the n\textsuperscript{+}-GaAs buffer layer growth. After the growth, the lasers were fabricated by defined 10-\textmu m-wide oxide stripe opening to contact the p\textsuperscript{+}-GaAs layer. In order to compare the lasing characteristics, an
Fig. 3.1. Schematic cross-section of AlGaAs/InGaAs SQW laser on Si with InGaAs IL.
AlGaAs/GaAs SQW laser on GaAs and AlGaAs/In\textsubscript{x}Ga\textsubscript{1-x}As (x=0, 0.01, 0.07) SQW lasers on Si without InGaAs IL, which were similar laser structures described above, were also prepared.

Figure 3.2 shows variation of lattice constants of Si, GaAs, GaAs/Si, In\textsubscript{0.02}Ga\textsubscript{0.98}As and In\textsubscript{0.07}Ga\textsubscript{0.93}As during cooling-down from the growth temperature to room-temperature\textsuperscript{16}). The linear thermal expansion coefficients for these materials in the temperature range between 300 and 1000 K can be obtained from the slope of each line. During cooling from the growth temperature (~1000 K) to room-temperature (~300 K), the large thermal stress of ~2\times10\textsuperscript{9} dyn/cm\textsuperscript{2} is generated in GaAs/Si by the difference in the thermal expansion coefficients between GaAs and Si. Therefore, the lattice constant of GaAs/Si increases from point A to B in Fig. 3.2 at the room-temperature. If the In content (x=0.02) in the In\textsubscript{x}Ga\textsubscript{1-x}As which is lattice-matched for the GaAs/Si at the middle temperature between 300 and 1000 K is chosen, the difference in the lattice constants between In\textsubscript{0.02}Ga\textsubscript{0.98}As and GaAs/Si is ~0.09 % at 300 K and +0.14 % at 1000 K, as shown in Fig. 3.2. Therefore, the thermally induced stress in the In\textsubscript{0.02}Ga\textsubscript{0.98}As layer grown on GaAs/Si can be relieved without introduction of misfit dislocations. Even AlGaAs/GaAs laser structure on GaAs also has the lattice-mismatch of ~+0.14 %. On the other hand, the difference in the lattice constants between In\textsubscript{0.07}Ga\textsubscript{0.93}As and GaAs/Si is +0.27 % at 300 K and +0.49 % at 1000 K. Thus, the misfit dislocations can be introduced in the In\textsubscript{0.07}Ga\textsubscript{0.93}As layer grown on GaAs/Si if the In\textsubscript{0.07}Ga\textsubscript{0.93}As layer is thicker than the critical thickness (hc).

Figure 3.3 shows the hc calculated for a GaAs/In\textsubscript{x}Ga\textsubscript{1-x}As single layer from the model of Matthews and Blakeslee\textsuperscript{17}) and the model of People and Bean\textsuperscript{18}). The circles corresponding to the 9-nm-thick In\textsubscript{x}Ga\textsubscript{1-x}As (x=0, 0.01, 0.02, 0.07, 0.10) SQW active layers used in...
Fig. 3.2. Variation of lattice constants of Si, GaAs, GaAs/Si, In$_{0.02}$Ga$_{0.98}$As and In$_{0.07}$Ga$_{0.93}$As during cooling-down from the growth temperature to room-temperature.
Fig. 3.3. Critical thickness calculated for a GaAs/In$_x$Ga$_{1-x}$As single layer from the model of Matthews and Blakeslee and the model of People and Bean.
this study are also shown. For all the active layers, the thickness is enough below the hc. Therefore, the lattice-mismatched strain in the active layers can be accommodated elastically without formation of misfit dislocations.

Under cw condition at room-temperature, the lasing characteristics such as polarization, threshold current density, emission wavelength, lasing characteristic parameters and reliability were studied\(^{19-22}\). In addition, the crystallinity of the active layers in the lasers was characterized by cathodoluminescence (CL) observations using a SEM (Topcon ABT-55).

3.3 Lasing Characteristics

3.3.1 Polarization

It is well known that the stress in the active layers of the lasers affects the polarization of the output light\(^{23-28}\). Therefore, the residual stress in the active layers can be relatively estimated from the degree of polarization (\(\rho\)). The degree of \(\rho\) is given by\(^{23}\)

\[
\rho = \frac{(L_{TE} - L_{TM})}{(L_{TE} + L_{TM})},
\]

where \(L_{TE}\) and \(L_{TM}\) are the intensities of the transverse electric (TE) and transverse magnetic (TM) polarized output light, respectively. Therefore, \(\rho\) varies from +1 for completely TE polarized emission, to 0 for equal portions of TE and TM polarized light, and to -1 for completely TM polarized emission. Predominantly, TE and TM polarized emissions from the quantum well active layers are associated with electron-to-heavy hole and electron-to-light hole quantum state transitions, respectively.
Figure 3.4 shows a typical polarization of the AlGaAs/GaAs SQW laser on Si under cw condition at room-temperature. It is found that the lasing occurs in only the TE mode. In order to estimate the stress in the active layers, it is necessary to observe the spontaneous emission below threshold current ($I_{th}$) because of elimination of the dependence on mode reflectivity affecting polarization\(^{23}\). The dependence of $\rho$ on the normalized current ($I/I_{th}$) for five kinds of lasers with $\sim$400-\(\mu\)m-long cavity length (L) is shown in Fig. 3.5. The degree of $\rho$ of a conventional AlGaAs/GaAs laser on Si is lower than that of an AlGaAs/GaAs laser on GaAs. This result indicates that the tensile stress in the active layer of the AlGaAs/GaAs laser on Si causes the light- and heavy-hole valence band quantum states to move closer together with increasing the TM polarized emission\(^{29-33}\). It is noticeable in Fig. 3.5 that the stress in the active layer can be changed from tensile to compressive stress with increasing the In content in the InGaAs active layer. The data indicates that the stress in the active layer of the AlGaAs/In\(_{0.02}\)Ga\(_{0.98}\)As laser on Si with InGaAs IL can be relieved, which is equivalent to that of the AlGaAs/GaAs laser on GaAs. The AlGaAs/In\(_{0.01}\)Ga\(_{0.99}\)As laser on Si without InGaAs IL has also the strain-relieved active layer. On the other hand, it was confirmed that the AlGaAs/In\(_{0.07}\)Ga\(_{0.93}\)As laser on Si with InGaAs IL has a compressive stress in the active layer.

3.3.2 Threshold Current Density

In order to study the lasing characteristics, these lasers with various L were operated under cw condition at room-temperature. Figure 3.6 shows the relationship of $J_{th}$ versus L for the AlGaAs/GaAs lasers on Si and AlGaAs/In\(_x\)Ga\(_{1-x}\)As (x=0.02, 0.07) lasers on Si with InGaAs ILs. For L=200-600 \(\mu\)m, the value of $J_{th}$ of the
Fig. 3.4. Typical polarization of AlGaAs/GaAs SQW laser on Si under cw condition at room-temperature.
Fig. 3.5. Dependence of degree of polarization, ρ, on the normalized current (I/I_th).
Fig. 3.6. Relationship of threshold current density versus cavity length for lasers on Si.
AlGaAs/In_{0.02}Ga_{0.98}As lasers is lower than that of the AlGaAs/GaAs lasers. This result is the same as the dependence of $J_{th}$ on the In content which was observed for AlGaAs/InGaAs lasers on GaAs\textsuperscript{34-36}). It is also found that the value of $J_{th}$ of the AlGaAs/In_{0.07}Ga_{0.93}As lasers tends to be higher than that of the AlGaAs/In_{0.02}Ga_{0.98}As lasers.

In order to characterize the crystallinity of the active layer in the AlGaAs/In_{0.07}Ga_{0.93}As laser on Si with InGaAs IL, the CL observation was carried out using a SEM at low temperature (~130 K). The accelerated beam voltages were 20 and 30 kV. This CL observation was also carried out for the conventional AlGaAs/GaAs laser on Si without InGaAs IL and AlGaAs/In\textsubscript{x}Ga\textsubscript{1-x}As (x=0, 0.02, 0.10) lasers on Si with InGaAs ILs as references. Figure 3.7 shows the top-viewed CL images of the four kinds of lasers on Si. These images were taken at the wavelength (energy) of the each active layer. The nonradiative recombination regions at dislocations and the other defects can be seen as dark spots (DSs). The dark spot density (DSD) of the conventional AlGaAs/GaAs laser on Si is estimated to be 0.8-1.0$\times10^7$ cm\textsuperscript{-2}. However, the value of this DSD may be lower than that of EPD by molten KOH etching, because it is possible that some of the DSs overlap with each other, making DSD lower than EPD\textsuperscript{37}). The DSD is effectively reduced to 4-6$\times10^6$ cm\textsuperscript{-2} for the AlGaAs/GaAs laser on Si with InGaAs IL due to the suppression of threading dislocations by introduction of InGaAs IL. For the AlGaAs/In_{0.02}Ga_{0.98}As laser, the DSD is estimated to be 3-4$\times10^6$ cm\textsuperscript{-2}. It is particularly noteworthy that the DLDs oriented towards the <110> direction can be clearly seen in the AlGaAs/In_{0.07}Ga_{0.93}As and AlGaAs/In_{0.10}Ga_{0.90}As lasers. Furthermore, it is found that the number of the <110> DLDs increased with increasing the In content in the InGaAs active layer. This
Fig. 3.7. CL images of the (a) AlGaAs/GaAs laser on Si without InGaAs IL, (b) AlGaAs/In$_{0.02}$Ga$_{0.98}$As, (c) AlGaAs/In$_{0.07}$Ga$_{0.93}$As and (d) AlGaAs/In$_{0.10}$Ga$_{0.90}$As lasers on Si with InGaAs ILs.
result clearly suggests that the higher strain associated with the increase of the In content enhances the bend of threading dislocations in the active layer\textsuperscript{29,38,39}, resulting in the raise in the $J_{th}$ due to the increase of nonradiative recombination centers\textsuperscript{15}, as shown in Fig. 3.6.

3.3.3 Emission Wavelength

The emission wavelengths of the AlGaAs/In$_x$Ga$_{1-x}$As ($x=0, 0.02, 0.07$) lasers on Si with InGaAs IL were measured. In this measurement, the ~400-μm-long cavities were used for these lasers, because short cavities ($L<300$ μm) show a sharp decrease in the wavelength due to the emission shifts from $n=1$ to $n=2$ quantum state transition in the quantum well. This shift in the emission wavelength with decreasing $L$ are the result of saturation of the available quantum well gain\textsuperscript{36,40}. It was confirmed that the emission wavelength monotonously increased from ~845 nm for the AlGaAs/GaAs laser, to ~865 nm for the AlGaAs/In$_{0.02}$Ga$_{0.98}$As laser, and to ~885 nm for the AlGaAs/In$_{0.07}$Ga$_{0.93}$As laser. This increasing wavelength is associated with the decrease of InGaAs band-gap energy with increasing In content\textsuperscript{34,35}. These wavelengths are from $n=1$ transition. For the AlGaAs/In$_{0.10}$Ga$_{0.90}$As laser on Si with InGaAs IL, however, the emission wavelength from $n=1$ transition could not be measured and the laser was operated at ~855 nm from $n=2$ transition. The reason is thought to be that the increase of $J_{th}$ due to the very poor crystallinity of the active layer [as shown in Fig. 3.7(d)] resulted in the difficulty of the $n=1$ transition.

3.3.4 Lasing-Characteristic Parameters

For the AlGaAs/GaAs lasers on Si with and without InGaAs IL, and AlGaAs/In$_x$Ga$_{1-x}$As ($x=0.02, 0.07$) lasers on Si with InGaAs ILs,
the lasing-characteristic parameters such as \( \eta_i \), \( \alpha_i \), \( \beta \) and \( J_0 \) were determined by eqs. (2.1) and (2.2), and are summarized in Table III.I. It was assumed that \( \Gamma \) is 0.027 for the SQW laser \(^{41} \). An AlGaAs/GaAs SQW laser on GaAs was also characterized as a reference.

The value of \( \eta_i \) seems to be considerably associated with the crystallinity of the active layer. That of \( \eta_i \) increased from 61 % for the conventional AlGaAs/GaAs laser without InGaAs IL, to 77 % for the AlGaAs/GaAs laser with InGaAs IL, and to 81 % for the AlGaAs/In\(_{0.02}\)Ga\(_{0.98}\)As laser. These improvements are certainly caused by InGaAs IL which improves the crystallinity of the active layer, as shown in Fig. 3.7(b). However, it decreased to 72 % for the AlGaAs/In\(_{0.07}\)Ga\(_{0.93}\)As laser. This is probably due to the increased nonradiative recombination centers related to the generation of bent threading dislocations \(<110>\) DLDs) for the AlGaAs/In\(_{0.07}\)Ga\(_{0.93}\)As laser, as shown in Fig. 3.7(c).

The value of \( \alpha_i \) was reduced from 26 cm\(^{-1}\) for the conventional AlGaAs/GaAs laser, to 15 cm\(^{-1}\) for the AlGaAs/GaAs laser with InGaAs IL, and to 17 cm\(^{-1}\) for the AlGaAs/In\(_{0.02}\)Ga\(_{0.98}\)As laser. For the AlGaAs/In\(_{0.07}\)Ga\(_{0.93}\)As laser, it decreased to 19 cm\(^{-1}\). However, this reason is not understood yet.

Additionally, it was found that the \( \beta \) was associated with the In content in the InGaAs active layer. The value of \( \beta \) was 1.4 cm/A and 1.7 cm/A for the AlGaAs/GaAs lasers on Si with and without InGaAs IL, respectively. It increased from 1.7 cm/A for the AlGaAs/In\(_{0.02}\)Ga\(_{0.98}\)As laser to 2.7 cm/A for the AlGaAs/In\(_{0.07}\)Ga\(_{0.93}\)As laser. In particular, it is noticeable that the \( \beta \) of 2.7 cm/A is larger than that of 2.0 cm/A for the AlGaAs/GaAs laser on GaAs. This improvement seems to have been caused by the compressive stress in the active layer, as shown in Fig. 3.5. The similar improvement of \( \beta \) has been observed for the
Table III.I. Summary of lasing-characteristic parameters for lasers.

<table>
<thead>
<tr>
<th>Device Type</th>
<th>$\eta_i$</th>
<th>$\alpha_i$</th>
<th>$\beta$</th>
<th>$J_0$</th>
</tr>
</thead>
<tbody>
<tr>
<td>AlGaAs/GaAs laser on GaAs</td>
<td>87</td>
<td>21</td>
<td>2.0</td>
<td>100</td>
</tr>
<tr>
<td>AlGaAs/GaAs laser on Si</td>
<td>61</td>
<td>26</td>
<td>1.7</td>
<td>270</td>
</tr>
<tr>
<td>AlGaAs/GaAs laser on Si with InGaAs IL</td>
<td>77</td>
<td>15</td>
<td>1.4</td>
<td>280</td>
</tr>
<tr>
<td>AlGaAs/In$<em>{0.02}$Ga$</em>{0.98}$As laser on Si with InGaAs IL</td>
<td>81</td>
<td>17</td>
<td>1.7</td>
<td>250</td>
</tr>
<tr>
<td>AlGaAs/In$<em>{0.07}$Ga$</em>{0.93}$As laser on Si with InGaAs IL</td>
<td>72</td>
<td>19</td>
<td>2.7</td>
<td>410</td>
</tr>
</tbody>
</table>
AlGaAs/InGaAs lasers on GaAs\textsuperscript{36}). This improvement has been also theoretically explained as a result of the additional strain-induced separation of energies of the the various valence band quantum state transitions\textsuperscript{42-45}).

In addition, the value of $J_0$ is much larger for the AlGaAs/In\textsubscript{0.07}Ga\textsubscript{0.93}As laser than for the AlGaAs/In\textsubscript{0.02}Ga\textsubscript{0.98}As laser. Although this is the opposite tendency observed for the AlGaAs/InGaAs lasers on GaAs\textsuperscript{36,46,47}), the reason is thought to be the increase of the nonradiative recombination current at the bent threading dislocations for the AlGaAs/In\textsubscript{0.07}Ga\textsubscript{0.93}As laser.

3.4 Reliability and Degradation Mechanism

3.4.1 Reliability

In order to study the reliability of these lasers, the cw aging tests were carried out at a constant output power of 1 mW/facet under APC operation at room-temperature. The conventional AlGaAs/GaAs laser on Si showed very rapid degradation only in a few minutes (as shown in Fig. 2.2) because the laser has a high DSD of 0.8-1.0\times10\textsuperscript{7} cm\textsuperscript{-2} and a large tensile stress in the active layer\textsuperscript{48,49}). The AlGaAs/GaAs laser on Si with InGaAs IL also degraded very rapidly as well as the the AlGaAs/GaAs laser on Si without InGaAs IL. This result indicates that the reduction of DSD from 0.8-1.0\times10\textsuperscript{7} to 4-6\times10\textsuperscript{6} cm\textsuperscript{-2} in the GaAs active layers, which was caused by inserting the InGaAs IL, is hardly effective in increasing the lifetime of lasers on Si. In addition, the AlGaAs/In\textsubscript{0.07}Ga\textsubscript{0.93}As and AlGaAs/In\textsubscript{0.07}Ga\textsubscript{0.93}As lasers on Si without InGaAs ILs also showed rapid degradation within a few minutes. In particular, the rapid degradation of the stress-relieved AlGaAs/In\textsubscript{0.07}Ga\textsubscript{0.93}As laser seems to show that the relief of stress
in the active layer with high DSD (~10^7 cm^-2) is scarcely effective in improving the reliability.

Figure 3.8 shows the results from the aging tests at room-temperature for the several AlGaAs/In_{0.02}Ga_{0.98}As and AlGaAs/In_{0.07}Ga_{0.93}As lasers on Si with InGaAs ILs. For the stress-relieved AlGaAs/In_{0.02}Ga_{0.98}As lasers, the lifetimes reaching 500 mA were ~40, 50 and 85 min. For the longest lifetime of ~85 min, the current increased at a rate of ~65 mA/h in the first 60 min, by ~230 mA/h in the next 15 min and then the laser showed a sharp increase in the degradation rate with time, which presumably resulted from the propagation of DLDs. On the other hand, despite the higher J_{th} of the AlGaAs/In_{0.07}Ga_{0.93}As laser than that of the AlGaAs/In_{0.02}Ga_{0.98}As laser (as shown in Fig. 3.6), the lifetimes of the AlGaAs/In_{0.07}Ga_{0.93}As lasers were ~70, 80 and 105 min in which the latter two lasers showed gradual degradation without the sharp increase of current. These results suggest that the reduction of DSD by the InGaAs IL and the dislocation pinning by the InGaAs active layer are required to realize the reliable GaAs-based lasers on Si.

3.4.2 Degradation Mechanism by EL Observation

Using EL topography, the degradation mechanisms of the AlGaAs/In_{0.02}Ga_{0.98}As and AlGaAs/In_{0.07}Ga_{0.93}As lasers on Si with InGaAs ILs were investigated.

Figure 3.9 shows typical EL topographs of the AlGaAs/In_{0.02}Ga_{0.98}As laser under dc operation at a constant current density of 1.0 kA/cm^2 and temperature of 20 °C. A DS originating from a pre-existing defect was observed before degradation [Fig. 3.9(a)]. At the first degradation stage after 35 min [Fig. 3.9(b)], one <100> DLD and many DSs were generated. In the next stage [Fig.
Fig. 3.8. Results from the APC aging tests under cw condition at room-temperature for the AlGaAs/In0.02Ga0.98As and AlGaAs/In0.07Ga0.93As lasers on Si with InGaAs ILs.
Fig. 3.9. Typical magnified EL topographs of the AlGaAs/In$_{0.02}$Ga$_{0.98}$As laser on Si after (a) 0, (b) 35, (c) 45 and (d) 120 min under dc operation at a constant current density of 1.0 kA/cm$^2$ and temperature of 20 °C.
the DLD extended towards both <100> and <110> directions in the active region. After 120 min [Fig. 3.9(d)], the DLD growth was enhanced towards <110> direction, and each DSs were broad in the vicinity of its originating defects which resulted from the formation of dislocation loops and dipoles at the threading dislocations\(^{50,51}\). Compared with the degradation of an AlGaAs/GaAs laser on Si (as shown in Fig. 2.9), the generation and growth of DSs are particularly remarkable. This seems to indicate that the DLD formation is somewhat suppressed by the dislocation pinning in the InGaAs active layer\(^{7,9,10,12,13}\).

Figure 3.10 shows typical EL topographs of the AlGaAs/In\(_{0.07}\)Ga\(_{0.93}\)As laser under dc operation at a constant current density of 1.0 kA/cm\(^2\) and temperature of 20 °C. As the EL observation was just started [Fig. 3.10(a)], several pre-existing DSs and <110> DLDs which are bent dislocations [as shown in Fig. 3.7(c)] were observed. In the following stage [Fig. 3.10(b)], one <100> DLD was generated from one DS in the center of the active region. In the subsequent stages [Figs. 3.10(c) and 3.10(d)], the DLD extended toward the edge of the active region. The growth of pre-existing <110> DLDs was not clearly observed after 76 min.

Figure 3.11 shows the dependence of the <100> DLD growth velocity on the injected current density for the AlGaAs/In\(_{0.02}\)Ga\(_{0.98}\)As and AlGaAs/In\(_{0.07}\)Ga\(_{0.93}\)As lasers at 20 °C. The data of AlGaAs/GaAs laser on Si (as shown in Fig. 2.10) was replotted in Fig. 3.11 as a reference. It was found that <100> DLD growth velocity (\(V\)) for the AlGaAs/GaAs laser was most strongly dependent on the injected current density (\(J\)), and the relationship between \(V\) and \(J\) for this laser was estimated as \(V \propto J^{1.55}\). In contrast, the DLD growth velocity was reduced to \(V \propto J^{1.51}\) for the AlGaAs/In\(_{0.02}\)Ga\(_{0.98}\)As laser. For the AlGaAs/In\(_{0.07}\)Ga\(_{0.93}\)As laser,
Fig. 3.10. Typical magnified EL topographs of the AlGaAs/In$_{0.07}$Ga$_{0.93}$As laser on Si after (a) 1, (b) 16, (c) 26 and (d) 76 min under dc operation at a constant current density of 1.0 kA/cm$^2$ and temperature of 20 °C.
Fig. 3.11. Dependence of <100> DLD growth velocity on the injected current density for the lasers on Si.
the growth velocity was drastically reduced to $V \propto J^{1.28}$. For example, the value of $V$ at $J=2.0$ kA/cm$^2$ was reduced from $\sim 65$ µm/h for the AlGaAs/GaAs laser, to $\sim 45$ µm/h for the AlGaAs/In$_{0.02}$Ga$_{0.98}$As laser, and to $\sim 20$ µm/h for the AlGaAs/In$_{0.07}$Ga$_{0.93}$As laser. These results indicate that the degradation mechanism related to $<100>$ DLDs of the AlGaAs/InGaAs lasers on Si with InGaAs ILs is basically the same as that of the AlGaAs/GaAs laser on Si, but the increasing lifetimes of the AlGaAs/InGaAs lasers were caused by the reduction of the DLD growth velocity due to dislocation pinning in the InGaAs active layer$^7,10)$. After all, for the AlGaAs/InGaAs lasers on Si with InGaAs ILs, it seems that the dislocation pinning by higher In content ($x=0.07$) in the InGaAs active layer has a more attractive potential of improving the reliability rather than low (stress-relieved) In content ($x=0.02$).

3.5 Increased Lifetime of AlGaAs/InGaAs Laser on Si by Post-Growth Annealing

It has been reported that post-growth annealing is effective in improving the crystallinity of the GaAs/Si$^{52-54})$. Therefore, the post-growth annealing was performed for the conventional AlGaAs/GaAs laser on Si and AlGaAs/In$_x$Ga$_{1-x}$As ($x=0.02, 0.07$) lasers on Si with InGaAs ILs. The processes of the post-growth annealing were as follows: A 0.1-µm-thick SiO$_2$ cap layer was deposited on the surface of each samples in order to prevent As atom detachment from the surface. Next, the samples were loaded into MOCVD reactor and annealed at 750 °C for 1 h in an AsH$_3$ atmosphere.

In order to study the crystallinity of the active layers in the post-growth annealed lasers, the CL observation was carried out at low-temperature. For the AlGaAs/GaAs laser, the improvement of
crystallinity was hardly observed. On the other hand, the remarkable improvement was particularly observed for the AlGaAs/In$_{0.07}$Ga$_{0.93}$As laser. Figure 3.12 shows a CL image of the AlGaAs/In$_{0.07}$Ga$_{0.93}$As laser after post-growth annealing. Compared with that of the laser before annealing as shown in Fig. 3.7(c), it can be seen that all <110> DLDs successfully vanished. This improvement is thought to be a result of the reduced bent threading dislocations, which resulted from the migration and interaction of the dislocations due to post-growth annealing.

The conventional 10-μm-wide oxide stripe contact lasers were fabricated from the annealed samples in order to investigate the lasing characteristics. All three kinds of lasers were operated under cw condition at room-temperature. Figure 3.13 shows a typical L-I characteristic of a post-growth annealed AlGaAs/In$_{0.07}$Ga$_{0.93}$As laser on Si with InGaAs IL under cw condition at room-temperature. This laser with 600-μm-long cavity exhibited the $I_{th}$ of 73 mA and the $\eta_d$ of 21%. The emission wavelength was also ~875 nm. For the annealed AlGaAs/In$_{0.02}$Ga$_{0.98}$As and AlGaAs/In$_{0.07}$Ga$_{0.93}$As lasers, the $\eta_i$ was estimated by eq. (2.2). The value of $\eta_i$ for the AlGaAs/In$_{0.02}$Ga$_{0.98}$As laser was increased from 81 to 85% by post-growth annealing. For the AlGaAs/In$_{0.07}$Ga$_{0.93}$As laser, that of $\eta_i$ was remarkably increased from 72 to 82% by the annealing. However, the value of $\eta_i$ was not changed for the AlGaAs/GaAs laser. These results correspond to the results of crystallinity of the active layers observed by CL.

The AlGaAs/GaAs lasers with and without the annealing degraded very rapidly within a few minutes. On the other hand, several annealed AlGaAs/In$_{0.02}$Ga$_{0.98}$As and AlGaAs/In$_{0.07}$Ga$_{0.93}$As lasers showed slow degradations and long lifetimes, as shown in Fig. 3.14. For the annealed AlGaAs/In$_{0.02}$Ga$_{0.98}$As laser, the longest lifetime
Fig. 3.12. CL image of the AlGaAs/In$_{0.07}$Ga$_{0.93}$As laser on Si after post-growth annealing.
RT: CW

After post-growth annealing
(750 °C, 1 hour)

Cavity length: 600 µm

$I_{th} = 73$ mA ($J_{th} = 1.2$ kA/cm²)

$\eta_d = 21\%$

$\lambda = 875$ nm

Fig. 3.13. Typical L-I characteristic of a post-growth annealed AlGaAs/In$_{0.07}$Ga$_{0.93}$As laser on Si with InGaAs IL under cw condition at room-temperature.
Fig. 3.14. Results from the APC aging tests under cw condition at room-temperature for the post-growth annealed AlGaAs/In0.02Ga0.98As and AlGaAs/In0.07Ga0.93As lasers on Si with InGaAs ILs.
as ~2.3 h which was ~1.6 times as long as the longest lifetime of 87 min obtained from this laser before annealing. Furthermore, the longest lifetime obtained in this study was ~24 h for the annealed AlGaAs/In$_{0.07}$Ga$_{0.93}$As laser (as shown in Fig. 3.13). In this case, the current increased at a rate of ~10 mA/h in the first 3 h, by ~6 mA/h after 15 h, by ~14 mA/h after 20 h, and then the laser showed a sharp increase in the degradation rate with time. This lifetime of 24 h was ~14 times as long as the longest lifetime of ~105 min recorded from this laser before annealing. These results are believed to have been arisen from the improvement of crystallinity in the InGaAs active layers by post-growth annealing.

In order to investigate the degradation mechanism of the long-life AlGaAs/In$_{0.07}$Ga$_{0.93}$As laser, the top-viewed EL observation was carried out under dc condition at a constant current density of 1.5 A/cm$^2$ and temperature of 20 °C. Figure 3.15 shows typical EL opographs of an annealed AlGaAs/In$_{0.07}$Ga$_{0.93}$As laser. A dark region extending towards the <110> direction (perpendicular to the stripe direction) was observed in the first degradation stage [Fig. 3.15(a)]. In the next stage after 1 h [Fig. 3.15(b)], several <110> DLDs were generated. In the subsequent stage [Figs. 3.15(c) and 3.15(d)], the <110> DLDs were grown, and many DSs were generated and some of them were transformed to <110> DLDs. This <110> DLD growth velocity was estimated to be ~6 µm/h at J=1.5 kA/cm$^2$, which is lower than the <100> DLD growth velocity of ~15 µm/h at the same value of J for this laser before annealing (as shown in Fig. 3.12). This degradation mechanism related to <110> DLDs is completely different from that mainly related to <100> DLDs observed in the other lasers on Si as shown in Figs. 2.9, 3.9 and 3.10. The result obtained from this EL observation indicates that the drastically increased lifetime of the annealed AlGaAs/In$_{0.07}$Ga$_{0.93}$As laser was
Fig. 3.15. Typical magnified EL topographs of the post-growth annealed AlGaAs/In$_{0.07}$Ga$_{0.93}$As laser on Si after (a) 4 min, (b) 1, (c) 6 and (d) 18 h under dc operation at a constant current density of 1.5 kA/cm$^2$ and temperature of 20 °C.
caused by the complete suppression of <100> DLD growth due to the dislocation pinning and the slow <110> DLD growth associated with REDG motion by stress\(^9,51\) (as shown in Fig. 3.5). In particular, the absence of <100> DLDs suggests that the REDC motion\(^9,51\) which causes rapid degradation is effectively suppressed in the annealed In\(_{0.07}\)Ga\(_{0.93}\)As active layer. The more increasing lifetime of the AlGaAs/InGaAs laser on Si can be expected to be achieved by suppression of the <110> DLD growth. This will be realized by use of the stress-compensated structure\(^55-60\) such as InGaAs compressive-stress active layer sandwiched by GaAsP tensile-stress confining layers.

3.6 Conclusions

The characteristics of AlGaAs/InGaAs SQW lasers on Si with InGaAs ILs have been studied. The residual stress in the active layer was relatively estimated by use of the degree of polarization. An AlGaAs/In\(_{0.02}\)Ga\(_{0.98}\)As laser on Si with InGaAs IL has a stress-relieved active layer, while an AlGaAs/In\(_{0.07}\)Ga\(_{0.93}\)As laser on Si with InGaAs IL has a compressive-stress induced active layer. Compared with a conventional AlGaAs/GaAs laser on Si, the value of \(J_{th}\) was reduced for an AlGaAs/In\(_{0.02}\)Ga\(_{0.98}\)As laser due to the increase of \(\eta_i\) and \(\beta\). The AlGaAs/In\(_{0.07}\)Ga\(_{0.93}\)As laser exhibited higher \(J_{th}\) than that of the AlGaAs/In\(_{0.02}\)Ga\(_{0.98}\)As laser because of the reduction of \(\eta_i\) by the increased bending of threading dislocation in the active layer. The lifetimes under APC operation at room-temperature were remarkably increased for these lasers, which resulted from the reduction of DSD by the introduction of InGaAs IL and the reduction of <100> DLD growth velocity by the dislocation pinning in the InGaAs active layer. Furthermore, for
these post-growth annealed lasers, the drastically increased lifetimes were realized. In particular, the longest lifetime was \(~24\) h for the annealed AlGaAs/In\(_{0.07}\)Ga\(_{0.93}\)As laser, which was caused by completely suppressed \(<100>\) DLD growth and slow \(<110>\) DLD growth.
References


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Chapter 4. AlGaAs/GaAs Light-Emitting Devices on Si Substrates with Small Active Regions

4.1 Introduction

As described in chapters 2 and 3, it has been demonstrated that DLDs originating from the threading dislocations grew very rapidly with increasing injected current density for the GaAs-based lasers on Si\(^1\). It is believed that the failure of the lasers on Si is caused by the rapid DLD growth. Although many previous efforts have been mainly focused on the reduction of the dislocation density, it has not been realized that the dislocation density is reduced to a level at least below \(\sim 10^4\ \text{cm}^{-2}\) observed in the GaAs layer on a GaAs substrate.

If the sizes of active regions are drastically reduced, the number of dislocations in the active regions can be reduced\(^2\). Therefore, the reliability can be expected to be improved for the GaAs-based LEDs and lasers on Si with the small active regions.

Recently, Kapon and co-workers\(^3-5\) have demonstrated the AlGaAs/GaAs lasers with GaAs quantum wire-like active regions by use of the regrowth technique of MOCVD on nonplanar GaAs substrates. The two-dimensional quantum confinement of injected carriers has also been observed from the optical properties of the quantum wire-like lasers\(^6-15\). Furthermore, the high-quality self-formed nanometer-scale GaAs and InGaAs quantum dot-like islands have been currently realized on GaAs substrates using the several growth techniques such as droplet-epitaxy\(^16,17\), Stranski-Krastanow (SK) growth mode\(^18-26\) and self-organizing growth mode on GaAs (311)B substrates\(^27-31\). The lasers with these islands in the active regions have been also fabricated\(^32-34\).
In this chapter, the successful fabrications of vertically-stacked quasi GaAs quantum wires grown on a V-grooved GaAs/Si substrate\textsuperscript{35}) and self-formed GaAs islands grown by droplet-epitaxy on a GaAs/Si substrate\textsuperscript{36}) are described. These shapes and sizes were observed using a high-resolution SEM (HITACHI S-5000) and an AFM (SPI 3700/SPA-300). These GaAs structures were also applied to the active regions of an AlGaAs/GaAs laser and an LED on Si. This chapter is organized as follows: In section 4.2, fabrication of an AlGaAs/GaAs quantum wire-like laser on a V-grooved GaAs/Si substrate is described. Fabrication of an AlGaAs/GaAs LED on Si with self-formed GaAs islands active region is presented in section 4.3. This chapter is concluded in section 4.4.

4.2 AlGaAs/GaAs Quantum Wire-Like Lasers on V-Grooved GaAs/Si Substrates

4.2.1 Vertically-Stacked Quasi GaAs Quantum Wires

Figure 4.1 shows the flowchart of the fabrication process for an AlGaAs/GaAs quantum wire-like laser on a V-grooved GaAs/Si substrate. In the first growth, an initial 10-nm-thick undoped GaAs nucleation layer and a 2.5-µm-thick n\textsuperscript{+}-GaAs buffer layer were grown on an n\textsuperscript{+}-Si substrate [Fig. 4.1(a)]. During growth of the n\textsuperscript{+}-GaAs layer, thermal cycle annealing was performed five times by varying the substrate temperature between 350 and 850 °C in an AsH\textsubscript{3} atmosphere. Next, V-shaped grooves with depth of ~1.2 µm and width of ~2.7 µm were formed with a 300 µm pitch along the [011] direction on a (100) GaAs/Si substrate using photolithography and wet chemical etching techniques [Fig. 4.1(b)]. As a reference, a GaAs/Si substrate with U-shaped grooves with depth of ~0.7 µm, top-width of ~2.7 µm and bottom-width of ~1.0 µm was also prepared. The chemical
Fig. 4.1. Flowchart of the fabrication process for an AlGaAs/GaAs quantum wire-like laser on a V-grooved GaAs/Si substrate.
etching in a solution of $\text{H}_2\text{SO}_4 : \text{H}_2\text{O}_2 : \text{H}_2\text{O}$ (1 : 8 : 40 by volume) forms the V and U-grooves with (111)A side-walls\textsuperscript{37}. The V-grooved substrate was then etched in a solution of $\text{H}_2\text{SO}_4 : \text{H}_2\text{O}_2 : \text{H}_2\text{O}$ (4 : 1 : 1 by volume) in order to create the round corner at the bottom of the V-groove. After the photoresist mask was removed, the substrates were stored in methyl alcohol. The grooved substrates were then cleaned in HCl and loaded into the reactor for the second growth. For the second growth, the AlGaAs/GaAs laser structure was grown on both V- and U-grooved GaAs/Si substrates by MOCVD. The structure consisted of a 10-nm-thick $n^+$-GaAs buffer layer, five periods of a 10-nm-thick $n^+$-GaAs/10-nm-thick $n^+$-Al$_{0.3}$Ga$_{0.7}$As buffer superlattice, a 0.51-μm-thick $n$-Al$_{0.7}$Ga$_{0.3}$As lower cladding layer, a 60-nm-thick undoped Al$_{0.3}$Ga$_{0.7}$As lower confining layer, three 3-nm-thick undoped GaAs quantum wells separated by 5.5-nm-thick undoped Al$_{0.3}$Ga$_{0.7}$As barrier layers, a 60-nm-thick undoped Al$_{0.3}$Ga$_{0.7}$As upper confining layer, a 0.51-μm-thick $p$-Al$_{0.7}$Ga$_{0.3}$As upper cladding layer and a 0.1-μm-thick $p^+$-GaAs contact layer [Figs. 4.1(c) and 4.1(d)]. Thickness of each layer was determined from GaAs and AlGaAs growth rates on a planar substrate. After the growth, the samples were cleaved perpendicular to the direction of the grooves for the high-resolution SEM observation.

Figure 4.2 shows a cross-sectional SEM micrograph of the overall AlGaAs/GaAs quantum wire-like laser grown on a V-grooved GaAs/Si substrate. The active layers of this laser are sharply defined along the V-groove with (111)A side-walls. Near the center of the V-groove, a (100) bottom surface is bounded between the two (111)A side-walls. No cracks were observed at the center of the V-grooves. This result indicates that relaxation of the large residual thermal stress in the GaAs/Si did not occur after forming the V-grooves or during cooling from the growth temperature (750 °C).
Fig. 4.2. Cross-sectional SEM micrograph of the overall AlGaAs/GaAs quantum wire-like laser on a V-grooved GaAs/Si substrate.
to room-temperature in the second growth. Figures 4.3(a) and 4.3(b) show a cross-sectional magnified SEM micrograph and a schematic diagram of the vertically-stacked quasi GaAs quantum wires near the center of the V-groove in Fig. 4.2, respectively. As shown in Fig. 4.3(a), the crescent-shaped quasi GaAs quantum wires of size (11-15 nm thick at its center) × (83-127 nm wide) are successfully grown on the V-grooved GaAs/Si substrate. They are realized by Ga atom accumulation from (111)A side-walls to the (100) bottom surface of the V-groove. However, nonuniformity among each quantum wire can also be seen. In going from the first (bottom) to the third (top) quantum wire, the lateral width of each quantum wire is increased and the vertical thickness is decreased. The reason may be that the 5.5-nm-thick Al0.3Ga0.7As barrier layers are too thin to recover the larger curvature of the center of the subsequently grown quantum wires10,13,15). Another significant result is that each quantum wire seems to be independent of the quantum well grown on the (111)A side-walls, which results in the formation of quantum wires of a small size. This indicates that this quantum wire-like laser has a much attractive advantage of low dislocation number arising from the reduction in the size of the active region. For example, assuming that the dislocation density is 1×10^7 cm^-2, the dislocation number of this quantum wire-like laser with 300-μm-long cavity is estimated to be ~3, while that of the conventional quantum well laser on Si with 10-μm-wide stripe contact window and 300-μm-long cavity (shown in Fig. 2.1) are ~300. This reduction of dislocations contributes to the suppression of rapid degradation for GaAs-based lasers on Si.

Additionally, the growth of quantum wells on a U-grooved GaAs/Si substrate have been also studied. The overall AlGaAs/GaAs laser structure grown on the U-grooved substrate is shown in Fig. 4.4(a). The active layers of this laser do not seem to be sharply
Fig. 4.3. Cross-sectional image of the vertically-stacked quasi GaAs quantum wires of size (11-15 nm)×(83-127 nm) near the center of the V-groove in Fig. 4.2: (a) magnified SEM micrograph and (b) schematic diagram.
Fig. 4.4. Cross-sectional SEM micrographs: (a) overall AlGaAs/GaAs patterned quantum well laser on a U-grooved GaAs/Si substrate and (b) vertically-stacked GaAs patterned quantum wells near the center of the U-groove.
formed along the U-groove which has a (100) bottom surface and (111)A side-walls. Figure 4.4(b) shows the vertically-stacked GaAs patterned quantum wells near the center of the U-groove in Fig. 4.4(a). The crescent-shaped GaAs quantum wires seen in Fig. 4.3(a) are not formed on the U-grooved substrate because another new facet in addition to the (111)A appears near the center of the U-groove. This new facet is thought to be (311)A by the measurement of the angle from the SEM micrograph of Fig. 4.4(b). This results in the formation of GaAs patterned quantum wells on the U-grooved substrate and indicates that the sharp V-groove with (111)A side-walls plays an important role in the successful formation of GaAs quantum wires.

4.2.2 Lasing Characteristics

The extremely low-threshold characteristics of an AlGaAs/GaAs laser with the vertically-stacked quasi GaAs quantum wires of size (20-21 nm) x (100-260 nm) grown on a V-grooved GaAs/Si substrate were also demonstrated.38-40 This quantum wire-like laser has been fabricated by the growth of three 9-nm-thick GaAs quantum wells and a p+-GaAs current-blocking layer. Laser devices were fabricated as follows: A 0.1-μm-thick SiO₂ insulating layer was deposited on the p+-GaAs contact layer and 2-μm-wide stripe contact windows were opened just on the grooves by wet chemical etching of SiO₂ in order to confine the injected current. Ti/Au was then evaporated on the p+-GaAs layer as the p-contact. After thinning the n+-Si substrate to a thickness of ~100 μm, AuSb/Au was evaporated on the Si substrate to form the n-contact. In order to reduce the series resistance, the contacts were annealed in a N₂ atmosphere at 380 °C. The samples were then cleaved into chips and mounted on In soldered Cu heatsinks in the p-side-up configuration. This quantum wire-like laser was tested at room-temperature under pulsed (0.2 μs pulses at
5 kHz repetition rate) and cw conditions.

Figures 4.5(a) and 4.5(b) show a cross-sectional SEM micrograph of the overall AlGaAs/GaAs quantum wire-like laser on Si with the p⁺-GaAs current-blocking layer and a schematic diagram of the injected current distribution, respectively. As shown in Fig. 4.5(a), a 1.6-μm-thick n⁺-GaAs and a 1.0-μm-thick p⁺-GaAs (Zn-doped, 1×10¹⁹ cm⁻³) layers were grown on a Si as a GaAs buffer layer. Figure 4.5(b) shows that the p⁺-GaAs layer acts as an internal current-blocking layer and prevents the current leakage path outside the V-groove. This internal current blocking is provided by the p-n-p-n junction structure outside the V-groove. However, the process of SiO₂ opening just on the V-groove must be carefully carried out, because the SiO₂ on both edges of the groove are easily removed and the light can be emitted at these edges due to the current leakage path.

Figure 4.6 shows the I-V characteristics of this quantum wire-like laser outside the V-groove with a p⁺-GaAs current-blocking layer and inside the V-groove without the layer. The turn-on voltage was 1.2 V without the current-blocking layer. On the other hand, no current flow was observed with the layer up to 5 V. This result indicates that the p⁺-GaAs layer is useful enough to confine the injected current into the quantum wires near the center of the V-groove.

Figure 4.7 shows the L-I characteristic and the emission spectrum of this quantum wire-like laser under pulsed condition at room-temperature. A pulsed Iₚₜ as low as 9.8 mA was obtained for the 123-μm-long laser. The value of ηₖ was 21 %. The peak emission wavelength was 858 nm at 2.0×Iₚₜ. In addition, the laser beam was mainly polarized parallel to the (100) substrate plane (TE-like mode). In order to compare the lasing characteristics, a
Fig. 4.5. Cross-sectional image of the overall AlGaAs/GaAs quantum wire-like laser on Si with p$^+$/GaAs current-blocking layer: (a) SEM micrograph and (b) schematic diagram of the injected current distribution.
Fig. 4.6. I-V characteristics of the quantum wire-like laser on Si outside the V-groove with $p^+$-GaAs current-blocking layer and inside the V-groove without the layer.
Fig. 4.7. L-I characteristic and emission spectrum of the quantum wire-like laser on Si under pulsed condition at room-temperature.
conventional AlGaAs/GaAs quantum well laser grown on a planar GaAs/Si substrate with three 9-nm-thick GaAs quantum wells and a 2-μm-wide stripe contact window was prepared. Under cw operation, the distribution of $I_{th}$ against cavity length for the quantum wire-like lasers and the conventional quantum well lasers on Si were measured. The minimum $I_{th}$ was 24-64 mA for the quantum well laser cavities of length 100-380 μm. For the quantum wire-like lasers, $I_{th}$ was as low as 16-20 mA for cavities of length 150-380 μm. In particular, to date, a pulsed $I_{th}$ of 9.8 mA and a cw $I_{th}$ of 16 mA for the quantum wire-like lasers are the lowest values so far reported for conventional quantum well lasers on Si$^{41-47)}$. The improvement in $I_{th}$ is thought to have been achieved by employing quantum wire structures with a reduction in the size of the active region.

4.3 AlGaAs/GaAs LEDs on Si with Self-Formed GaAs Islands Active Regions

4.3.1 Self-Formed GaAs Islands on GaAs/Si Substrates

A 2.5-μm-thick thermal-cycle annealed $n^+$-GaAs buffer layer was grown on an $n^+$-Si substrate by the two-step growth technique. The growth temperature was 750 °C except for the initial 10-nm-thick undoped GaAs nucleation layer which was grown on Si at 400 °C. In the preliminary experiment, the GaAs islands were grown on the $n^+$-GaAs/Si substrate at 700 °C using droplet-epitaxy$^{17)}$ by the supply of TMG and successive AsH$_3$ supply. At the same time, the GaAs islands were also grown side-by-side on an $n^+$-GaAs (100) substrate as a reference. Figure 4.8 shows the typical supply sequence of TMG and AsH$_3$ for droplet-epitaxy. The details about growth sequence of this droplet-epitaxy is as follows: After the substrate temperature reached the growth temperature at 700 °C, the AsH$_3$ flow was stopped.
Fig. 4.8. Typical supply sequence of TMG and AsH$_3$ for droplet-epitaxy.
5 s later, the TMG was supplied for 6 s. Ga droplets were formed at this stage. The AsH₃ was supplied again with 5 s interval after stopping the TMG supply. At this stage, the Ga droplets were transformed to GaAs islands. This growth mechanism is believed to be similar to the vapor-liquid-solid (VLS) equilibrium phase growth model [48-50]. Using an AFM, the self-formed GaAs islands on the GaAs/Si substrate were observed as shown in Fig. 4.9. In this experiment, the TMG flow rate and time were 10 sccm and 6 s, respectively. This growth condition corresponds to a 5-nm-thick GaAs layer by the simultaneous supply of TMG and AsH₃. It can be seen that the GaAs islands on the GaAs/Si are randomly formed as well as the islands on the GaAs substrate in spite of the surface roughness (∼20 nm) of the GaAs layer on Si [51]. The islands exhibited the conical shape with the heights of 90-170 nm, the effective diameters of 600-750 nm and the density of 1-2×10⁷ cm⁻². It was also observed that the islands aligned on the cracks of GaAs/Si. This self-alignment of the islands seems to be caused by the Ga droplet accumulation on the bottom of the cracks. In addition, the size and density of self-formed GaAs islands could be controlled by the growth condition such as the TMG flow rate. Figures 4.10(a) and 4.10(b) show the dependence of island size and density on the TMG flow rate. The TMG flow time was kept constant at 6 s. As shown in Fig. 4.10(a), the island height decreases with the increase of the TMG flow rate, while the diameter increases with increasing the TMG flow rate. Figure 4.10(b) shows that the island density can be controlled between 10⁶ and 10⁷ cm⁻² by varying the TMG flow rate, which results from increasing density of the Ga dimer of critical nucleus with the increase of TMG flow rate.

4.3.2 L-I Characteristics
Fig. 4.9. AFM image of the self-formed GaAs islands on GaAs/Si substrate. The TMG flow rate and time were 10 sccm and 6 s, respectively.
Fig. 4.10. Dependence of (a) island size and (b) island density on the TMG flow rate. The TMG flow time was kept constant at 6 s.
Figures 4.11(a) and 4.11(b) show cross-sectional SEM micrographs of the overall LED on Si and the magnified GaAs islands active region, respectively. The LED structure comprised a 2.5-μm-thick thermal-cycle annealed n⁺-GaAs buffer layer, a 0.74-μm-thick n-Al₀.₇Ga₀.₃As lower cladding layer, a 70-nm-thick undoped Al₀.₃Ga₀.₇As lower confining layer, undoped GaAs islands active region formed by droplet-epitaxy, a 70-nm-thick undoped Al₀.₃Ga₀.₇As upper confining layer and a 0.74-μm-thick p-Al₀.₇Ga₀.₃As upper cladding layer. An 80-nm-thick p⁺-GaAs contact layer was finally grown. After the growth, surface-emitting LED was fabricated from this wafer by depositing Ti/Au contact ring as the p-side electrode. AuSb/Au were evaporated on the n⁺-Si substrate as the n-side electrode. During the droplet-epitaxy in the LED growth, the TMG flow rate and time were 10 sccm and 6 s, respectively. This growth condition is the same as that shown in Fig. 4.9. As shown in Fig. 4.11(a), it is found that two bright areas in Al₀.₃Ga₀.₇As/GaAs region are GaAs islands. The island is connected to some extent with a very thin GaAs layer at the interface of the Al₀.₃Ga₀.₇As layers. However, it can not be seen that the islands are connected each other.

Figure 4.12 shows a typical L-I characteristic of this LED under dc condition at room-temperature. The insets in Fig. 4.12 show a top-viewed electroluminescence (EL) topograph and an emission spectrum of this LED. It is found that the light is emitted from each GaAs island. This LED was operated up to 27 μW at 190 mA, and the light output power increased linearly with increasing current up to 21 μW at 130 mA. At higher current, however, it was saturated because of thermal heating. This tendency was confirmed by the fact that the output power increased linearly up to higher current without saturation under pulsed (0.2 μs pulses at 5 kHz repetition
Fig. 4.11. Cross-sectional SEM micrographs: (a) overall AlGaAs/GaAs LED on Si with self-formed GaAs islands active region and (b) magnified GaAs island active region.
Fig. 4.12. Typical L-I characteristic of the LED on Si under dc condition at room-temperature. The insets show a top-viewed EL topograph and an emission spectrum of this LED.
rate) operation. The peak emission wavelength at 80 mA was 868 nm and the full width at half maximum (FWHM) was 49 nm. An AlGaAs/GaAs LED on Si with a 9-nm-thick GaAs quantum well active region was also prepared as a reference. The structure of this LED was the same as that of an LED with the islands active region except for the active region structure. The output power of the LED with the islands active region decreased by ~1/10 of that of the LED with a quantum well active region due to the reduction in the size of the active region. From this result, however, it can be estimated that the output power density in the active region of the LED with the islands is about twice as high as that of the LED with a quantum well at the same injected current. This indicates that the high-quality GaAs islands could be successfully formed on Si. In addition, considering the island size and density, the dislocation number of the LED with the islands active region is estimated to be reduced to ~1/100 of that of the LED with a quantum well active region.

In order to investigate the reliability of the LED on Si with GaAs islands active region, the automatic current control (ACC) aging test was carried out under room-temperature dc condition at a constant current of 60 mA (0.5 kA/cm²). This test was also carried out for an LED on Si with a quantum well active region. Figure 4.13 shows a typical degradation data under ACC aging tests of these LEDs. As shown in Fig. 4.13, the LED with a quantum well showed a rapid degradation in which the output power decreased rapidly to a half of the initial value only in a few minutes. In contrast, the output power of LED with the islands decreased very slowly and reached to a half of the initial value after ~14 hours in spite of the higher output power density in the active region. In particular, it should be noted that the decrease in the output power
Fig. 4.13. Typical degradation data under ACC aging tests at room-temperature of the LEDs on Si with the islands active region and with a conventional quantum well active region.
of the LED with the islands is much more suppressed than that of the LED with a quantum well at the initial degradation stage within few minutes.

In order to study in detail the differences of degradation in these LEDs, EL observations were carried out on 10-μm-wide stripe contact laser structures fabricated from the same each LED wafer. The fabrication of these laser structures has been described in section 2.2. Figure 4.14 shows top-viewed EL topographs of progressive degradation stages of a laser structure with the islands active region at 20 °C. The injected current density during this observation was 0.5 kA/cm². For a laser structure with a quantum well, the EL topographs have been shown in Fig. 2.9. In the laser structure with a quantum well, a few <100> DLDs originating from the threading dislocations were generated and rapidly extended toward the center of the stripe. On the other hand, it was observed that the DLDs growth was suppressed and only two bright spots (arrows shown in Fig. 4.14(a)) emitted from the islands vanished after 30 min in the laser structure with the islands [Fig. 4.14(b)]. At the subsequent degradation stage after 60 min, the degradation seems not to progress so fast because no change was observed on the EL topograph [Fig. 4.14(c)]. In the more progressive degradation stage, it was observed that almost bright spots gradually darkened. Figure 4.14 probably indicates that most of the islands (except these islands marked by arrows as shown in Fig. 4.14(a)) could avoid the threading dislocations and nonradiative recombination at the dislocations was drastically reduced by the confinement of the injected carriers in the islands. This resulted in improving the initial degradation as shown in Fig. 4.13. The subsequent slow degradation as shown in Fig. 4.13 is thought to be caused by the darkened islands probably due to gradual condensation of point
Fig. 4.14. Typical EL topographs of the laser structure on Si with the islands active region after (a) 0, (b) 30 and (c) 60 min under dc operation at a constant current density of 0.5 kA/cm² and temperature of 20 °C.
defects in the islands. After all, this improved reliability is believed to have been caused by low dislocation numbers and suppressed <100> DLD growth in the active region due to reduction in the size of the active region.

4.4 Conclusions

The successful fabrication of an AlGaAs/GaAs quantum wire-like laser on Si with vertically-stacked quasi GaAs quantum wires was described. It was confirmed that the GaAs quantum wires of size (11-15 nm) × (83-127 nm) were grown on a V-grooved GaAs/Si substrate. Furthermore, the extremely low-threshold current (9.8 mA) AlGaAs/GaAs quantum wire-like laser on Si, which has the vertically-stacked quantum wires of size (20-21 nm) × (100-260 nm), was also demonstrated.

The fabrication of an AlGaAs/GaAs LED on Si with self-formed GaAs islands active region was also presented. The GaAs island size and density could be controlled by the TMG flow rate. At the TMG flow rate of 10 sccm and time of 6 s, the islands showed the conical shape with the heights of 90-170 nm, the diameters of 600-750 nm and the density of 1-2×10⁷ cm⁻². This LED could be operated under dc condition at room-temperature. The reliability of this LED was much superior to that of an LED with a quantum well active region because of low dislocation numbers and suppressed DLD growth due to reduction of the active region by use of the small GaAs island structure.
References

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Chapter 5. AlGaAs/GaAs Vertical-Cavity Surface-Emitting Lasers on Si Substrates

5.1 Introduction

The vertical-cavity surface-emitting lasers (VCSELs) on Si are the most promising devices for optical interconnections in future OEICs\(^{1-9}\). In addition, the residual thermal stress and dislocations in the active region and the threshold current can be reduced in the VCSELs on Si with the small active regions\(^{1}\). Therefore, the VCSEL structure also has the potential for the realization of reliable GaAs-based lasers on Si. There has been only one demonstration\(^{1}\) of the VCSEL on Si fabricated by heteroepitaxial growth. For this VCSEL on Si, the pulsed \(I_{th}\) was \(-125\) mA at room-temperature, corresponding to the \(J_{th}\) of \(-71\) kA/cm\(^2\). To date, however, cw operation has not been realized yet.

In this chapter, the low-threshold pulsed operation of AlGaAs/GaAs VCSEL on Si at room-temperature\(^{10,11}\) was described. Furthermore, cw operation at low-temperature was also presented. This chapter is organized as follows: In section 5.2, room-temperature pulsed operation is described. Low-temperature cw operation is presented in section 5.3. This chapter is concluded in section 5.4.

5.2 Room-Temperature Pulsed Operation

5.2.1 Epitaxial Growth and Fabrication Process

The VCSEL structure was grown on Si substrate at 750 °C by the conventional two-step growth technique. Figure 5.1 shows the cross-sectional SEM micrograph of the AlGaAs/GaAs VCSEL on Si. The
Fig. 5.1. Cross-sectional SEM micrograph of the AlGaAs/GaAs VCSEL on Si.
structure consisted of a 0.85-μm-thick thermal-cycle annealed n⁺-GaAs buffer layer, a 20-pair of n⁺-AlAs/n⁺-GaAs (71 nm/59 nm) quarter-wave multilayer DBR, a 0.46-μm-thick n-Al0.7Ga0.3As lower cladding layer, a 70-nm-thick undoped Al0.3Ga0.7As lower confining layer, an undoped 9-nm-thick GaAs SQW active layer, a 70-nm-thick undoped Al0.3Ga0.7As upper confining layer, a 0.34-μm-thick p⁺-Al0.7Ga0.3As upper cladding layer, and an 80-nm-thick p⁺-GaAs contact layer. 50-nm-thick AuSb/150-nm-thick Au were used for the contact on the n⁺-Si substrate. Nonalloyed 50-nm-thick AuZn/150-nm-thick Au of 40×40 μm², which was used for the top mirror and the electrical contact, was formed by the photoresist patterning and the lift-off technique. The devices were mounted, junction up, on a sample holder and tested under pulsed condition (0.1 μs pulses at 1MHz repetition rate) at room-temperature. Output power from the VCSEL was measured by detecting the light emitted around the edges of the AuZn/Au metallization. The DSD was also measured by electron-beam-induced current (EBIC) method using a SEM (Topcon ABT-55).

5.2.2 Characteristics of Reflectors

Figures 5.2(a) and 5.2(b) show the cross-sectional transmission electron microscopy (TEM) micrographs of the overall VCSEL on Si and near the GaAs/Si interface, respectively. Some threading dislocations, which originate at the GaAs/Si interface, can be effectively confined into the thermal-cycle annealed n⁺-GaAs layer [Fig. 5.2(b)]. However, dislocation bending was not observed in the AlAs/GaAs DBR [Fig. 5.2(a)]. Although some previous studies¹,¹²,¹³ have indicated that the AlAs/GaAs DBR is effective in bending the dislocations due to interdiffusion-assisted dislocation motion in AlAs/GaAs¹²) or crystal hardening of AlAs¹⁴), Fig. 5.2(a) shows that
Fig. 5.2. Cross-sectional TEM micrographs: (a) overall VCSEL on Si and (b) near the GaAs/Si interface.
many dislocations originating at the GaAs/Si interface propagate into the active layer. This VCSEL on Si, which was grown under the thermal cycle annealing and the AlAs/GaAs DBR, also showed the DSD of $2-3 \times 10^7$ cm$^{-2}$. This value is almost the same value for the sample without the AlAs/GaAs DBR. These results indicate that the AlAs/GaAs DBR is not effective in bending the threading dislocations because there is less mismatch (~0.14 %) in the lattice constant for the AlAs/GaAs layers.

The calculated and measured reflectivities of the 20-pair of AlAs/GaAs DBR on Si are shown in Fig. 5.3. The reflectivity of the DBR was measured with a monochromator and the reference reflectivity was supplied by a deposited aluminum mirror. The measured reflectivity was above 90 % at the wavelength region between 820 and 870 nm. The recorded maximum reflectivity was 93 % at 860 nm, which was slightly longer than the designed wavelength of 850 nm, presumably due to a small nonuniformity of the thickness during growth. As shown in Fig. 5.3, it is found that the measured peak reflectivity of 93 % is lower than the calculated reflectivity of 99.3 % at 860 nm. Figures 5.4(a) and 5.4(b) show the magnified cross-sectional SEM micrographs of the AlAs/GaAs DBRs grown on GaAs and Si substrates, respectively. The interfacial sharpness with uniform and smooth heterointerfaces is clearly seen in the DBR on GaAs. However, the DBR on Si shows quasiperiodic zigzag roughness and nonuniformity in the AlAs/GaAs layers$^{15}$. Compared with Fig. 5.4(a), the thickness variations are estimated to be from -8 to +6 % for the AlAs layer and from -3 to +14 % for the GaAs layer in the DBR on Si as shown in Fig. 5.4(b). Additionally, Figure 5.5 shows the AFM image of the surface morphology for the VCSEL on Si. The microroughnesses consisting of hillocks and depressions with a different size were observed in the surface morphology. The surface
Fig. 5.3. Calculated and measured reflectivities of the 20-pair of the AlAs/GaAs DBR on Si.
Fig. 5.4. Magnified cross-sectional SEM micrographs of the AlAs/GaAs DBR on (a) GaAs and (b) Si substrates.
Fig. 5.5. AFM image of the surface morphology for the VCSEL on Si.
and interfacial roughnesses are probably due to the three-dimensional GaAs growth on Si at the initial growth stage. It has been theoretically demonstrated that the reflectivity spectrum and its peak value are very insensitive to small thickness variation (a few monolayers) in the DBR\textsuperscript{16-18}. However, the large zigzag roughness and interfacial roughness (as shown in Figs. 5.4(b) and 5.5) introduce a large optical diffraction and scattering, resulting in the lower reflectivity of the DBR on Si.

In contrast, the top mirror has the reflectivity of 60% for the alloyed and 96% for the nonalloyed AuZn/Au. The higher reflectivity of the nonalloyed AuZn/Au seems to have been caused by a smooth morphology due to the lack of thermal annealing\textsuperscript{19}.

5.2.3 Lasing Characteristics

The turn-on voltage and the series resistance were 1.3 V and 26 Ω, respectively. These values are comparable to those of the conventional edge-emitting lasers on Si. These results indicate that the 20-pair of n\textsuperscript{+}-AlAs/GaAs DBR on Si hardly makes the series resistance high. Figure 5.6 shows an L-I characteristic of the VCSEL on Si under pulsed condition at room-temperature. This VCSEL exhibited the I\textsubscript{th} of 79 mA, corresponding to the J\textsubscript{th} of 4.9 kA/cm\textsuperscript{2}. Additionally, the peak emission wavelength was 840.03 nm and the FWHM was 0.28 nm. Compared with the previously reported results\textsuperscript{1}), this VCSEL on Si exhibits remarkable improvements such as a lower threshold and a narrower emission spectrum, which are probably due to relatively lower dislocation density and laser structure. However, this pulsed J\textsubscript{th} of 4.9 kA/cm\textsuperscript{2} is much higher than many reported results\textsuperscript{2,20-24} of the VCSELS on GaAs substrates. A cw operation could not be achieved in this VCSEL because of a lower optical gain of the SQW active layer and a lower reflectivities of
Fig. 5.6. L-I characteristic of the VCSEL on Si under pulsed condition at room-temperature.
reflectors.

5.3 Low-Temperature CW Operation

5.3.1 Epitaxial Growth and Fabrication Process

In order to achieve the cw operation, it is very important for the VCSEL on Si to increase the optical gain and the reflectivities of reflectors, because the VCSEL has a much shorter cavity length than that of the conventional edge-emitting laser and the mirror loss \((1/L\ln(1/R))\) is very high. The higher reflectivity of the DBR on Si can be obtained by reducing the absorption losses of the \(n^+\)-GaAs layers in the DBR\(^1\). Therefore, the conventional AlAs/GaAs DBR was replaced with the AlAs/Al\(_{0.1}\)Ga\(_{0.9}\)As DBR. Theoretically, the reflectivity increases by \(~1\%\) for the AlAs/Al\(_{0.1}\)Ga\(_{0.9}\)As DBR\(^1\). In contrast, the increasing optical gain is expected to be realized by use of MQW active layers structure due to the improvement of the gain flattening (saturation)\(^{25-31}\).

The optimization of MQW active layers in the VCSEL, especially the number of quantum wells (\(N_w\)), has been discussed to achieve low threshold operation from the viewpoint of matched gain effect\(^{29,31}\). The threshold gain \(g_{th}\) of the VCSEL with internal loss and matched gain effect is given by\(^{29,31}\)

\[
\Gamma R N_w L_w g_{th} = 1/2 \ln[1/(R_{r} R_{r})] + \sum_{w,b} (\Gamma R N_i L_i \alpha_{ai}) \\
+ (L - \sum_{w,b} N_i L_i) \alpha_C, \quad (5.1)
\]

\[
\Gamma R = 1 + \sin(2\pi\sigma) / (2\pi\sigma), \quad \sigma = 1/\lambda \sum_{w,b} (N_i L_i), \quad (5.2)
\]

where \(L_w\) and \(L_b\) are the well thickness and barrier thickness, respectively. The \(L_b\) of 5.5 nm was used for the Al\(_{0.3}\)Ga\(_{0.7}\)As
barrier layer. $R_f$ and $R_r$ are the reflectivities of the front (top) and rear (bottom) side reflectors, respectively. $\Gamma_r$ is the gain matching factor. $\alpha_a$ and $\alpha_c$ are the absorption losses in the active and cladding layers, respectively. According to eq. (5.1), a large $N_w$ in the cavity is desirable to increase the gain. However, from eq. (5.2), in order to achieve the matched gain ($\Gamma_r > 1$), the total MQW thickness must be thinner than the $\lambda/2n$ ($\lambda$:wavelength, $n$:refractive index), which requires a reduction of $N_w$. For example, the $N_w$ should be less than 14 if a 9-nm-thick GaAs quantum well is used as an active layer. Finally, from eq. (5.1), an optimum $N_w$ is mainly determined by the reflectivities of reflectors.

From these viewpoints, an AlGaAs/GaAs VCSEL on Si with 10 GaAs quantum wells and a 23-pair of AlAs/Al$_{0.1}$Ga$_{0.9}$As DBR was fabricated. Figure 5.7 shows a schematic cross-section of this VCSEL on Si. The laser structure comprised a 2.1-μm-thick thermal-cycle annealed n$^+$-GaAs buffer layer, a 23-pair of n$^+$-AlAs/n$^+$-Al$_{0.1}$Ga$_{0.9}$As (71 nm/60 nm) quarter-wave multilayer DBR, a 0.29-μm-thick n-Al$_{0.7}$Ga$_{0.3}$As lower cladding layer, a 51-nm-thick undoped Al$_{0.3}$Ga$_{0.7}$As lower confining layer, ten undoped 9-nm-thick GaAs MQW ($N_w=10$) active layers separated by 5.5-nm-thick undoped Al$_{0.3}$Ga$_{0.7}$As barrier layers, a 51-nm-thick undoped Al$_{0.3}$Ga$_{0.7}$As upper confining layer, a 0.22-μm-thick p-Al$_{0.7}$Ga$_{0.3}$As upper cladding layer, and a 34-nm-thick p$^+$-GaAs contact layer. After the growth, a laser device was fabricated as follows: A 0.1-μm-thick SiO$_2$ insulating layer was deposited on the p$^+$-GaAs contact layer and 50 μmφ contact windows were opened by wet chemical etching of SiO$_2$. Next, 50-nm-thick Ti/60-nm-thick Au were evaporated on the p$^+$-GaAs layer as the p-side electrode except for the 20 μmφ at the center of the 50 μmφ contact window. A 70-nm-thick semitransparent Au was then evaporated on the p-side surface. A 50-nm-thick AuSb/150-nm-thick Au was used for the
Fig. 5.7. Schematic cross-section of the VCSEL on Si with 23-pair of AlAs/Al_{0.1}Ga_{0.9}As DBR and multi quantum wells active layers.
n-side electrode on the n⁺-Si substrate. In this VCSEL on Si, the light output was measured by detecting the light emitted through the 20 μmφ semitransparent Au.

The measured peak reflectivity of the 23-pair of AlAs/Al₀.₁Ga₀.₉As DBR on Si was 94 % at 840 nm. Although a little increase of reflectivity was observed in comparison with the 20-pair of AlAs/GaAs DBR on Si, the reflectivity of 94 % still remains to be much lower than the calculated value of 99.4 % probably due to the large zigzag roughness and the interfacial roughness in the DBR on Si. The top mirror had the reflectivity of 93 % and transparency of 0.6 % for the semitransparent Au.

5.3.2 Lasing Characteristics

Figure 5.8 shows an L-I characteristic and emission spectrum of this VCSEL under cw operation at 100 K. The value of $I_{th}$ was 73 mA, which corresponded to $J_{th}$ of 3.7 kA/cm². The $\eta_d$ was 0.8 %, which is much lower than that of the edge-emitting lasers on Si (for example, Fig. 3.13). This seems to have been caused by the large absorption in the semitransparent Au and the lateral current spreading due to the absence of current confinement structure such as mesa etching²₀,³₂) and ion implantation³₃,³₄). The FWHM of this spectrum was 39.5 nm at 0.96×$I_{th}$ and 2.2 nm at 1.2×$I_{th}$. The relatively broad FWHM of 2.2 nm may be due to the generation of multi transverse-modes near the single fundamental mode²²). The peak wavelength at 1.2×$I_{th}$ was 844 nm, which was near the wavelength achieving the peak reflectivity of the DBR. Figure 5.9 shows the polarization of this VCSEL for two perpendicular directions under cw operation at 100 K. It can be seen that the output light was scarcely polarized below $I_{th}$ and was clearly polarized above $I_{th}$. It has been reported that there is no fundamental polarization...
Fig. 5.8. L-I characteristic and emission spectrums of the VCSEL on Si under cw condition at 100 K.
Fig. 5.9. Polarization of the VCSEL on Si for two perpendicular directions under cw condition at 100 K.
selection mechanism for the VCSELs on GaAs substrates with quantum well active layers in contrast to the edge-emitting lasers with quantum wells, and the polarization is probably due to some anisotropy introduced by the processing\textsuperscript{35}). Additionally, it has been also reported that the polarization can be controlled by introducing stress\textsuperscript{36,37}). These influences for the polarization characteristic observed in the VCSEL on Si are not yet understood.

The L-I characteristics operated between 30 and 150 K were also shown in Fig. 5.10. It is found that the $I_{th}$ increased with the increase of temperature. Up to 150 K, this VCSEL on Si could be operated under cw condition. The value of $I_{th}$ was 82 mA ($J_{th}$=4.2 kA/cm$^2$) at 150 K. The successful cw operation up to 150 K was believed to have been mainly achieved by the increased gain due to increased $N_w$. The characteristic of the cw operation is expected to be more improved by the optimum $N_w$ with increasing $R_f$ and $R_r$ from eq. (5.1).

However, this VCSEL on Si showed rapid degradation under cw condition at low-temperature and pulsed condition at room-temperature. The VCSEL degraded very rapidly only in ~1 min under cw condition at 100 K. Figure 5.11 shows a result from the APC aging test of the VCSEL on Si under room-temperature pulsed condition (0.2 µs pulses at 5 kHz repetition rate) at a constant output power of 210 µW. The inset in this figure shows an L-I characteristic of this VCSEL. The values of $I_{th}$ and $\eta_d$ were 104 mA and 1.8 %, respectively. The $I_{th}$ of 104 mA corresponds to the $J_{th}$ of 5.3 kA/cm$^2$. As shown in Fig. 5.11, the current increased gradually at a rate of ~30 mA/h in the first 170 min and then the VCSEL showed a rapid increase in current by ~114 mA/h. This rapid degradation under pulsed operation is believed to have been caused by rapid DLD growth due to high injected current density, which has
Fig. 5.10. L-I characteristics of the VCSEL on Si under cw condition between 30 and 150 K.
Fig. 5.11. Result from the APC aging test of the VCSEL on Si under pulsed condition at room-temperature. The inset shows an L-I characteristic of this VCSEL.
been shown in Figs. 2.10 and 2.11.

5.4 Conclusions

The room-temperature pulsed operation of the AlGaAs/GaAs VCSEL on Si with the 20-pair of AlAs/GaAs DBR and a SQW active layer was demonstrated. The measured peak reflectivity of the 20-pair of the AlAs/GaAs DBR was 93% at 860 nm. A TEM observation showed that the AlAs/GaAs DBR is not effective in bending the threading dislocations because there is less mismatch in the lattice constant for the AlAs/GaAs layers. The $I_{th}$ and $J_{th}$ of this VCSEL were 79 mA and 4.9 kA/cm$^2$, respectively. Furthermore, low-temperature cw operation of the VCSEL on Si with the 23-pair of AlAs/Al$_{0.1}$Ga$_{0.9}$As DBR and MQW ($N_w=10$) active layers was also demonstrated for the first time. Up to 150 K, the VCSEL could be operated under cw condition.
References


Chapter 6. Summary

This dissertation has described the studies on GaAs-based light-emitting devices on Si substrates for future OEICs. At first, the observation and understanding of the rapid degradation mechanism of the AlGaAs/GaAs quantum well lasers grown on Si have been carried out. Then, the increasing lifetimes of these lasers have been achieved by both introducing InGaAs intermediate layers and replacing the conventional GaAs active layers with the InGaAs layers. Finally, the novel approaches for improving characteristics of the AlGaAs/GaAs light-emitting devices on Si have been presented. The demonstration of AlGaAs/GaAs VCSELs on Si for OEICs have been also described.

In chapter 1, the background and purpose of this research, and the current status of heteroepitaxial growth of GaAs/Si and device applications were presented.

The results obtained in this research are summarized as follows:

In chapter 2, the influences of DLDs on lasing characteristics and the dependence of the DLD growth velocity on the injected current density for an AlGaAs/GaAs quantum well laser on Si were described. The generation of <100> DLDs caused the decrease of $\eta_i$ and the slow increase of driving current. At the subsequent rapid degradation stage, the expansion of DLDs caused the remarkable increase of $\alpha_i$ and the decrease of $\beta$, and finally resulted in the rapid increase of driving current. Additionally, it was also found that the $<100>$ and $<110>$ DLDs growth velocities strongly depended on the injected current density.

In chapter 3, the increased lifetimes of AlGaAs/InGaAs SQW lasers on Si with InGaAs ILs were presented. Furthermore, the
differences of degradation mechanism between AlGaAs/GaAs and AlGaAs/InGaAs lasers on Si were also described. The observation of degree of polarization indicated that the AlGaAs/In$_{0.02}$Ga$_{0.98}$As laser on Si with InGaAs IL had a stress-relieved active layer, while the AlGaAs/In$_{0.07}$Ga$_{0.93}$As laser on Si with InGaAs IL had a compressive-stress induced active layer. The lifetime of the AlGaAs/In$_{0.02}$Ga$_{0.98}$As laser increased to ~85 min under cw condition at room-temperature. The AlGaAs/In$_{0.07}$Ga$_{0.93}$As laser also exhibited the increased lifetime of ~105 min. These improvements were caused by the reduction of threading dislocations by use of the InGaAs IL and the reduction of $<$100$>$ DLD growth velocity due to the dislocation pinning in the InGaAs active layer. Furthermore, a drastic increased lifetime of ~24 h was realized for the post-growth annealed AlGaAs/In$_{0.07}$Ga$_{0.93}$As laser, which resulted from the completely suppressed $<$100$>$ DLD growth and slow $<$110$>$ DLD growth.

In chapter 4, the improved characteristics of AlGaAs/GaAs light-emitting devices on Si with small active regions were described. The AlGaAs/GaAs quantum wire-like laser on Si with vertically-stacked quasi GaAs quantum wires was successfully fabricated. It was confirmed that the GaAs quantum wires of size $(11-15 \ \text{nm}) \times (83-127 \ \text{nm})$ were grown on a V-grooved GaAs/Si substrate. Furthermore, the AlGaAs/GaAs quantum wire-like laser on Si with a p$^+$-GaAs current-blocking layer and quasi GaAs quantum wires of $(20-21 \ \text{nm}) \times (100-260 \ \text{nm})$ exhibited extremely low-threshold current of 9.8 mA under pulsed condition at room-temperature. The AlGaAs/GaAs LED on Si with the self-formed GaAs islands active region was also fabricated. The GaAs island size and density could be controlled by the TMG flow rate. At the TMG flow rate of 10 sccm and the time of 6 s, for example, the islands showed the conical shape with the heights of 90-170 nm, the diameters of 600-750 nm and the density of
1-2×10^7 cm^{-2}. The reliability of this LED was much superior to that of an LED on Si with a conventional GaAs quantum well active region, which resulted from the reduction of dislocation numbers and the suppressed DLD growth due to the reduction in the size of the active region. The dislocation numbers in these light-emitting devices on Si with the small active regions could be estimated to be reduced to ~1/100 of that of the devices on Si with the quantum well active regions.

In chapter 5, the AlGaAs/GaAs VCSELs on Si, which are the most promising devices for future OEICs, were presented. The room-temperature pulsed operation was realized for a VCSEL on Si with a 20-pair of AlAs/GaAs DBR and a SQW active layer. The I_{th} and J_{th} of this VCSEL were 79 mA and 4.9 kA/cm^2, respectively. Additionally, 150 K cw operation was achieved for a VCSEL on Si with a 23-pair of AlAs/Al_{0.1}Ga_{0.9}As DBR and MQW (10 quantum wells) active layers.

With further improvements in crystallinity of GaAs/Si grown by heteroepitaxy and device structures, the characteristics such as reliability of the GaAs-based light-emitting devices on Si will be improved significantly. Consequently, the monolithic integration of GaAs-based optical I/O devices with Si-based ULSI chips will be realized in the near future. The results in this study are hopefully expected to give a helpful guidance for the realization of practical GaAs-based light-emitting devices on Si for future OEICs.
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Papers and Letters


(invited paper).


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