strates. As can be seen in the figure, the growth fronts of L-SPE and defects in the grown layers are evident by the difference of the etching rates. In samples A, the film grows laterally from the both sides of the bare Si regions onto the SiO_2 patterns. After 8-h annealing, formation of polycrystalline islands was observed in the remaining a-Si film and the L-SPE was stopped by these polycrystalline grains. The maximum growth length in this sample was about $6 \mu m$. On the other hand, it can be seen from the roughly etched surface that the a-Si film in sample B changed to polycrystalline state at least after 4-h annealing. In this sample, however, some L-SPE was also observed at the left-hand sides of the Si/Si structures. This fact suggests that the internal stress is somewhat reduced even in a room-temperature-deposited film by reducing the step height of the SiO_2 patterns. It was also found that the L-SPE was not enhanced in the room-temperature-deposited films even if they were implanted with Si ions. We conclude from these results that the use of dense a-Si films is essentially important for L-SPE, though partial L-SPE may occur even in porous a-Si films when the thickness of the underlying SiO_2 film is thin enough.

Finally, the L-SPE growth rate along the $\langle 010 \rangle$ direction in sample A was derived from the micrographs in Fig. 4 and the related ones. The result shown in Fig. 5 has revealed that the L-SPE rate at 600 °C is about 1.2×10^{-8} cm/s in the steady state, though the rate is faster in the initial stage shorter than 2 h. The steady growth rate is $\frac{1}{4}$ to $\frac{1}{8}$ of that of V-SPE on the (100) plane.[7] The origin of the slower rate is not clear, but the rate is probably lowered due to formation of some kinds of facets and/or defects. The maximum SPE length after 10-h annealing was about $6 \mu m$.

In summary, we showed that dense a-Si films formed by evaporation on heated substrates and subsequent amorphization by Si^+ ion implantation grew laterally onto SiO_2 patterns by SPE. The lateral SPE growth rate along the $\langle 010 \rangle$ direction was $\frac{1}{4}$ to $\frac{1}{8}$ of that of vertical SPE, though the rate was enhanced in the initial stage. The maximum SPE length along this direction was about $6 \mu m$ after 10-h annealing at 600 °C.

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Electrical properties of undoped and Si-doped $Al_{0.48}In_{0.52}As$ grown by liquid phase epitaxy

Toshiyuki Tanahashi, Kazuo Nakajima, Akio Yamaguchi, and Itsuo Umebu

Fujitsu Laboratories Limited, Atsugi Semiconductor Materials Laboratory, 1677 Dannoki, Ono, Atsugi, 243-01 Japan

Electrical properties of undoped $Al_{0.48}In_{0.52}As$ layers grown by liquid phase epitaxy (LPE) were studied for the first time. The carrier concentration $n(cm^{-3})$ and the mobility $\mu(cm^2/Vs)$ were $\mu = 4600$ cm^2/Vs at $n = 4.7 \times 10^{15}$ cm^{-3} and $\mu = 4500$ cm^2/Vs at $n = 5.9 \times 10^{15}$ cm^{-3} at room temperature. The doping data of Si in LPE-grown $Al_{0.48}In_{0.52}As$ and the resulting mobility and carrier concentration were also studied. The distribution coefficient of Si was determined to be 0.019. The mobility of $Al_{0.48}In_{0.52}As$ is comparable to that of InP when compared at the same carrier concentration.

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$In_{1-x}Ga_xAs_{1-y}P_y$-InP lasers have been playing an important role in long range fiber-optical communication systems. But the energy gap difference in $In_{1-x}Ga_xAs_{1-y}P_y$-InP double heterostructures (DH) is not sufficient, and so the carrier leakage[1] through the barrier is concerned in relation to the poor temperature characteristics of the lasers.[2]

$Al_{0.48}In_{0.52}As$ lattice matches to InP, and has a larger energy gap than InP.[3] Therefore, more effective carrier confinement effects can be expected in $In_{1-x}Ga_xAs_{1-y}P_y$ DH lasers with $Al_{0.48}In_{0.52}As$ confining layer than with InP one. $Al_{0.48}In_{0.52}As$ can also be used as thin intermediary layers to increase Schottky barrier height for $In_{0.52}Ga_{0.48}As$ field-effect transistors.[4] Up to now, this material has been grown by
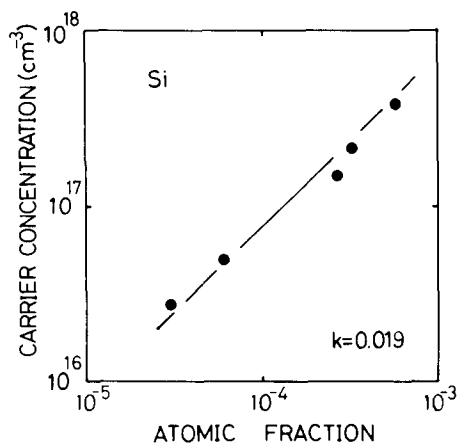
FIG. 1. Carrier concentration as a function of the atomic fraction of Si in the growth solution.

FIG. 2. Mobility as a function of the carrier concentrations for undoped and Si-doped $Al_{0.48}In_{0.52}As$ layers.

molecular beam epitaxy (MBE).[4–9] Recently, however, we developed the liquid phase epitaxial (LPE) growth of lattice-matched $Al_{0.48}In_{0.52}As$ on InP.[10]

In this letter the mobility and the carrier concentration of undoped $Al_{0.48}In_{0.52}As$ are shown. The doping of Si in LPE-grown $Al_{0.48}In_{0.52}As$ and the resulting mobility and carrier concentration are studied.

$Al_{0.48}In_{0.52}As$ epitaxial layers of this work were grown onto Fe-doped (100) InP substrates. The experimental apparatus consisted of a horizontal furnace system and a conventional sliding graphite boat. Pd purified hydrogen was used as ambient gas. Materials used were semiconductor-grade Al, In, InAs, and Si. An AlIn mother alloy, which contained 0.1 at. % Al, was used as the Al source.[10]

The composition of the growth solution was $X^l_{Al} = 0.00066$, $X^l_{As} = 0.142$, and $X^l_{In} = 0.85734$. In doping experiments, the atomic fraction of Si in the solution was between 3×10^{-5} and 6×10^{-4}. The growth solution was kept at 810 °C for 30 min and then cooled at the rate of 0.3 °C/min. The InP substrate was covered with the InP polycrystal and in situ etched by undersaturated In and InP solution. $Al_{0.48}In_{0.52}As$ layers were grown from 778 to 768 °C.

The Hall coefficient and the resistivity were measured on samples in which grown layers were mesa etched into bridge shapes. The Ohmic contacts were made by alloying indium dots onto the epilayer at 400 °C for 3 min in N_2 ambient gas. The measurements were performed under the conditions of 0.1 mA and 2 kOe at room temperature.

The layer thickness of $Al_{0.48}In_{0.52}As$ was about 0.5 μm. The lattice mismatch between the epilayer and the substrate was less than 1×10^{-3}, which was measured by a double crystal x-ray diffraction technique. The growth morphology and the growth rate of Si-doped $Al_{0.48}In_{0.52}As$ were similar to those of undoped layers. No traces of oxidization of Si were observed at all.[11]

The carrier concentration $n(cm^{-3})$ and the mobility $\mu(cm^2/Vs)$ of two runs of undoped $Al_{0.48}In_{0.52}As$ layers were $\mu = 4600$ cm²/Vs at $n = 4.7 \times 10^{15}$ cm⁻³ and $\mu = 4500$ cm²/Vs at $n = 5.9 \times 10^{15}$ cm⁻³ at room temperature. These mobilities are as high as those of InP with the similar carrier concentrations. Ohno et al. reported that undoped

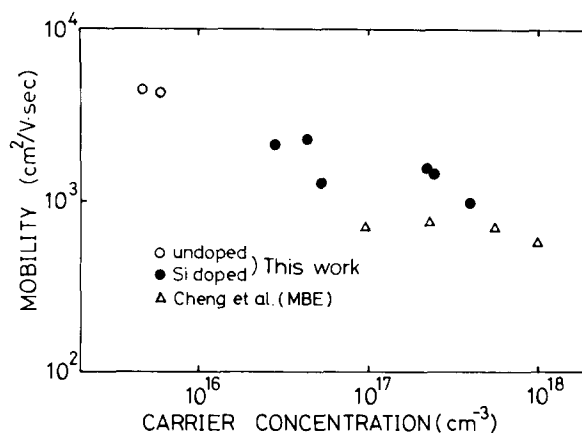$Al_{0.48}In_{0.52}As$ layers grown by MBE showed high resistivity and they attributed it to the high density of deep levels.[3] In our case high purity LPE-grown $Al_{0.48}In_{0.52}As$ layers were obtained without special purification of solution such as the long time baking.[12]

Figure 1 shows a double log plot of the carrier concentration versus the atomic fraction of Si in the growth solution. The gradient of the line in Fig. 1 is approximately unity over the doping range studied here from 3×10^{16} to 4×10^{17} cm⁻³. From these data a distribution coefficient k of Si is calculated to be 0.019, where it is defined as the ratio of the atomic fraction of the impurity species in the solid to that in the liquid. Astles et al. reported Si doping in InP. Using their data, the distribution coefficient of Si in InP is estimated to be 0.13. It is approximately seven times as high as that in $Al_{0.48}In_{0.52}As$ though their growth temperatures were 640–680 °C.[11]

The mobilities of both undoped and Si-doped $Al_{0.48}In_{0.52}As$ layers as a function of the carrier concentration are shown in Fig. 2, together with the data of MBE-grown Sn-doped $Al_{0.48}In_{0.52}As$ reported by Cheng et al.[7] The mobility increases with lowering the carrier concentration. It is 2100 and 1600 cm²/Vs at $n = 3 \times 10^{16}$ cm⁻³ and $n = 2 \times 10^{17}$ cm⁻³, respectively. The mobilities obtained here are comparable to those of InP. For MBE-grown $Al_{0.48}In_{0.52}As$ the value of the mobility known up to now is about 800 cm²/Vs at 1×10^{17} cm⁻³ (Refs. 7 and 9). The mobility of LPE-grown layers is twice as high as that of MBE-grown ones.

In summary, high purity $Al_{0.48}In_{0.52}As$ layers with the carrier concentration 5×10^{15} cm⁻³ were obtained by LPE growth without the special purification. Si doping to that material was studied and the distribution coefficient was determined to be 0.019. The mobility of $Al_{0.48}In_{0.52}As$ is comparable to that of InP when compared at the same carrier concentration. By this work, the promising aspect of this material as the confining layer of $In_{1-x}Ga_xAs_{1-y}P_y$ lasers has been fully shown.

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Anion inclusions in III-V semiconductors

H. Gant,[a] L. Koenders, F. Bartels, and W. Mönch

Laboratorium für Festkörperphysik, Universität Duisburg, D-4100 Duisburg, Germany

On all surfaces cleaved from InP, GaP, GaAs, and GaSb a local excess of anion atoms was detected as bright spots in the image of the electron current absorbed by the samples and identified by using a scanning Auger electron spectrometer exhibiting a spatial resolution of 20 μm. Arsenic zones were detected on horizontal Bridgman- as well as liquid encapsulation Czochralski grown GaAs indicating their existence to be independent of the crystal growth technique. The densities of anion zones varied between 10 and 500 per square centimeter on a cleave-to-cleave basis. Their diameters typically measured approximately 50 μm. The investigations indicate that the anion zones originate from inclusions in the bulk becoming exposed by cleavage.

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Presently available single crystals of III-V compounds locally deviate from ideal stoichiometry. Already in 1974, Arthur[1] postulated the existence of a diffusion process in GaAs supplying As from the bulk to the surface. This conclusion has been strongly supported by studies with GaAs layers prepared by molecular beam epitaxy.[2,3] Ga-rich GaAs(001) surfaces were prepared by annealing the samples above 800 K. During cooling to below 770 K the surfaces converted to As-rich ones but heating to above 800 K restored the Ga-rich structure again. This cycle could only be repeated a few times indicating that the substrate contained a limited supply of arsenic.

The first direct evidence of a local excess of anions within GaAs samples grown by the horizontal Bridgman (HBM) and the liquid encapsulation Czochralski (LEC) growth methods was reported by Cullis et al.[4] They detected As precipitates with diameters up to 1000 Å attached to dislocations by using transmission electron microscopy. By utilizing locally resolved Auger electron spectroscopy (AES) Bartels et al.[5] recently observed segregates of pure arsenic at surfaces cleaved from GaAs samples grown by the HBM and the traveling heater methods (THM). Since the arsenic zones were detected immediatedly after cleavage they are inclu-sions within the bulk becoming uncovered by cleavage.

In the present study investigations with locally resolved AES were extended to other III-V compounds and to GaAs samples grown by LEC. Since anion zones have been observed at all the InP, GaP, GaAs, and GaSb surfaces investigated anion inclusions seem to be a general problem in III-V semiconductors.

Clean surfaces were prepared by cleavage from single-crystal bars in an UHV chamber at a base pressure of 4×10^{-9} Pa. The bars initially measured $4.2 \times 7 \times 35$ mm with the (110) direction parallel to the long axis. The LEC-grown InP, GaP, and GaAs as well as HBM-grown GaSb samples were n doped with carrier concentrations of $\approx 2 \times 10^{17}$ cm^{-3} at room temperature. The AES studies were performed with a cylindrical mirror analyzer (Varian) equipped with an integral, scanning electron gun. The primary energy was set to 2 kV yielding a beam diameter and thus a spatial resolution of approximately 20 μm at a beam current of 2 μA. In another mode, the electron beam was scanned across the cleaved surface, and the electron current absorbed by the sample was displayed on a video screen. Local variations of this current and thus of the brightness of the screen are due to changes of the secondary electron emission.

Already immediately after cleavage the pictures of the current absorbed by the samples were found to be nonuniform and to exhibit a few bright spots, i.e., the electron emission was locally decreased. For the chemical identification of