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# The effect of nonstoichiometry and polarity of the (111) plane on microtwin formation in ion-implanted GaAs

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To study the effect of nonstoichiometry and polarity of the (111) plane on recrystallization of an implantation-induced amorphous layer, arsenic-implanted (100) and (111) *A* and *B* GaAs wafers were annealed under vacuum and specified arsenic vapor pressures. It has been found that the formation of microtwins is suppressed by annealing at the specified arsenic pressure. This suggests that microtwin formation is closely related to the arsenic deficiency or arsenic vacancies. Moreover, the result of direct implantation into (111) *A* (Ga) and *B* (As) surfaces indicates that the twins would preferentially grow on the (111) *A* plane, which has a smaller solid-phase regrowth rate than that of the (111) *B* plane.

## I. INTRODUCTION

There have been many reports concerning the solid-phase epitaxial regrowth in the implantation-induced amorphous layer in (100) GaAs,<sup>1-4</sup> which is characterized by imperfect regrowth in contrast to the Si (100) surface. This imperfect regrowth is mainly caused by the generation of extended defects, such as microtwins and dislocation loops. However, the detailed mechanism of microtwin formation in GaAs has not been satisfactorily elucidated so far, in particular with respect to the effect of nonstoichiometry produced by implantation<sup>5,6</sup> and post-annealing as well as the growth plane. As for the latter problem it has been reported<sup>2,4</sup> that the twins grow on the (111) plane, which has a smaller regrowth rate than that of (100) and (110) planes like the case of Si and Ge. In these investigations, however, no attention was paid to the polarity effects of the (111) planes, *A* and *B*, on twin growth in GaAs, although their regrowth rates are possibly different.

The purpose of this paper is thus to clarify the effect of nonstoichiometry on the formation of microtwins and the polarity dependence of solid-phase regrowth in implanted GaAs. Arsenic-implanted (100) GaAs wafers and (111) *A* and *B* GaAs wafers were annealed under vacuum and different arsenic vapor pressures, respectively, and the resultant defects were analyzed by means of a Rutherford-backscattering-channeling (RBSC) technique.

## II. EXPERIMENT

As<sup>+</sup> ions were implanted at an energy of 80 keV into liquid-encapsulated Czochralski (LEC) grown (100) GaAs with a dose of  $6 \times 10^{13}$  cm<sup>-2</sup>. The thickness of the amorphous layer produced was dependent on the beam current or the wafer temperature during the implantation. In this work, wafers with an approximately 350-Å buried amorphous layer were used. For purposes of comparison, a wafer with a

preamorphous but heavily damaged layer was also provided (hereafter called the reference sample). To study directly the polarity dependence of solid-phase regrowth, (111) *A* and *B* wafers grown by the horizontal Bridgman (HB) method under controlled vapor pressure were also implanted under the same conditions as those for the LEC (100) wafers.

These wafers were post-annealed in a vacuum of  $10^{-6}$  Torr. The implanted side was placed upward so as to enhance arsenic evaporation. To study the effect of nonstoichiometry on defect restoration, they were also annealed under various applied arsenic vapor pressures ( $P_{As}$ ). The values of  $P_{As}$  used were  $6.8 \times 10^{-3}$ , 0.09, 0.8, and 5.4 Torr for the annealing temperatures of 300, 350, 400, and 450 °C, respectively. These values were considerably higher than the optimum arsenic vapor pressures  $P_{As}^{opt}$  at each temperature. Annealing at  $P_{As}^{opt}$  produced the lowest density of lattice defects in GaAs.<sup>7,8</sup>  $P_{As}^{opt}$  is given by the following equation<sup>7</sup>:

$$P_{As}^{opt} = 2.6 \times 10^6 \exp\left(\frac{-1.05 \text{ eV}}{kT_H}\right) \text{ Torr},$$

where  $T_H$  is the annealing temperature. The corresponding values of  $P_{As}^{opt}$  are calculated as  $1.5 \times 10^{-3}$ ,  $8.3 \times 10^{-3}$ , 0.036, and 0.13 Torr for the respective annealing temperatures mentioned above.

The experimental arrangement for RBSC analysis is schematically shown in the inset in Fig. 2, where a low exit angle of 5° and He<sup>+</sup> beam energy of 1.5 MeV were employed to improve the depth resolution (~45 Å). Aligned spectra were obtained from the <100> direction and <111> *A* and *B* directions for the (100) wafer and (111) *A* and *B* wafers, respectively. The effective acceptance angle of a solid-state detector was limited to within  $\pm 0.5^\circ$  by using an aperture. The exposed total beam charge during data accumulation amounted to 20 μC under the beam current of 10 nA. The areas of defect peaks in the aligned spectra were obtained

with the assumption of a linear background. The detailed experimental conditions have been described elsewhere.<sup>9,10</sup>

### III. RESULTS AND DISCUSSION

Figure 1 shows the annealing behavior of the (100) GaAs wafers with a completely amorphous layer 350 Å thick when annealed under vacuum and  $P_{As}$ . The results of the reference sample with a preamorphous layer under vacuum anneal only are also shown in Fig. 1 for comparison. The temperature dependence of the aligned spectra for the sample with a complete amorphous layer of 350 Å is shown in Figs. 2(a) and 2(b), each corresponding to annealing in vacuum and the applied arsenic vapor pressure  $P_{As}$ . Two main features are clearly seen from Figs. 1 and 2: (1) The GaAs with a complete amorphous layer behaves quite differently from the reference sample with a preamorphous layer; (2) there is a remarkable effect of applied arsenic pressure on the reduction of the defects. The behavior of the reference sample subjected to annealing above 150 °C has already been discussed in terms of point defects including extended forms such as clusters and complexes.<sup>9-11</sup> For the sample with the completely amorphous layer, near-perfect restoration was observed by annealing at 200 °C for 15 min in vacuum. Prolonged annealing at 200 °C did not bring any increase in peak area and in minimum yield  $\chi_{min}$ . On further annealing above 200 °C in vacuum, the density of the defects increased and attained a maximum at 300 °C, but the defects almost completely disappeared at 400 °C. The yield peaks observed at 300 and 350 °C shifted slightly toward the surface as shown in Fig. 2(a). Such behavior cannot be explained on the basis of the point defects stated above. As the present amorphous layer is almost completely recrystallized by annealing at 200 °C, the defects generated above 200 °C may be associated with those generated during the regrowth. It is most plausible to attribute those defects to the microtwins that were frequently observed in implanted GaAs,<sup>1,2,4,12,13</sup> in view of

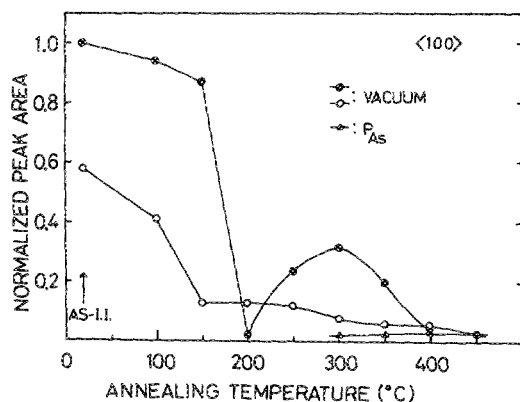


FIG. 1. Normalized peak areas of the defects after isochronal (15-min) annealing of the sample with an amorphous layer of 350 Å: Annealing in vacuum (●) and applied arsenic vapor pressure  $P_{As}$  (▲). The reference data (○) were taken from the vacuum-annealed sample with a preamorphous implanted layer.

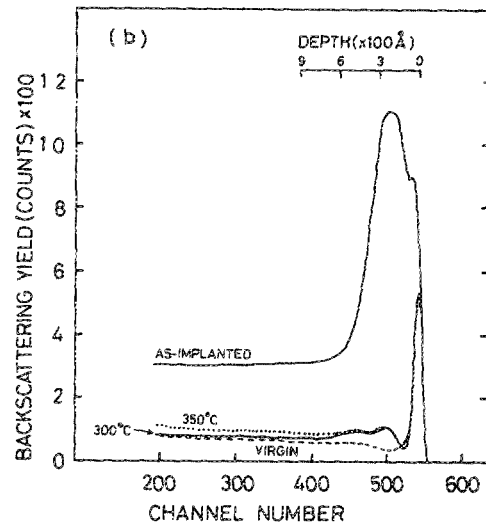
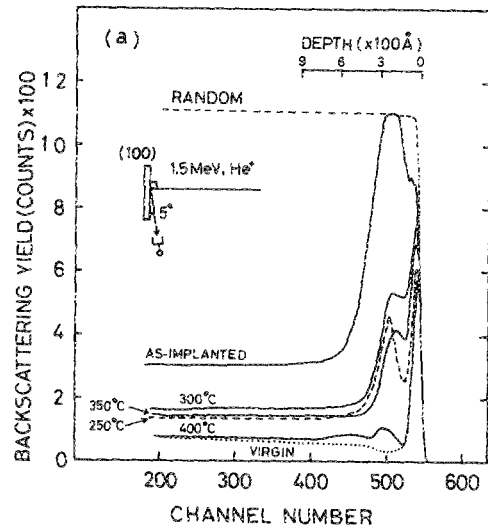


FIG. 2. (100) aligned spectra obtained from the (100) sample with an amorphous layer of 350 Å. Annealed for 15 min in (a) vacuum and (b) applied arsenic vapor pressure  $P_{As}$ . The measurement geometry is schematically shown in inset.

the fact that the microtwins are generated at ~200–250 °C and annihilated at 400 °C for a thin amorphous layer.<sup>3,4</sup> It should be noted here that small-sized microtwins contribute mainly to direct scattering rather than dechanneling.<sup>2</sup> The observed spectra shown in Fig. 2 can be explained quite well by direct scattering, because the size of the present twins must be less than the thickness of the amorphous layer. Besides, the observed large effect of the applied arsenic vapor pressure on the reduction of the defects at relatively low temperatures would exclude the idea that the defects are composed of arsenic-vacancy-related point defects. This conclusion could be supported further by the fact that the defects disappeared upon vacuum annealing at 400 °C, where arsenic vacancies should be generated more easily.

Thus, the observed annealing behavior may result from the inherent thermal property of the defects or microtwins. The possibility of the occurrence of other extended defects, e.g., dislocation loops, would also be excluded because they are not annihilated even above 400 °C. The above considerations will lead to the important implication that the formation of microtwins is suppressed by annealing under applied arsenic overpressures. This means that microtwin formation is closely related to the arsenic deficiency or arsenic vacancies. This can be regarded as the first experiment that enables one to clarify the nonstoichiometric conditions for twin formation in solid-phase epitaxial regrowth of GaAs, although it was indicated<sup>5,6,14</sup> that microtwin formation might be caused by implantation-induced deviation from the stoichiometric composition. From the present experiment, it becomes evident that annealing-induced nonstoichiometry will play a dominant role in twin formation over the implantation-induced nonstoichiometry in the case of a thin amorphous layer. The latter nonstoichiometry may become more important for annealing of a thicker amorphous layer. It is to be noted here that the equivalent nonstoichiometric condition for twin formation was reported<sup>15</sup> in the LEC growth of GaAs where the twins tend to be formed in the Ga-rich melt.

As stated above, the yield peaks observed at 300 and 350 °C shifted toward the surface, but the peaks observed below 250 °C were found at the same position as those of the as-implanted sample. It is well known for a thin buried amorphous layer<sup>3</sup> that microtwins are located at the position where bidirectional regrowth fronts meet each other. Therefore, the peak shift toward the surface can be explained by the slower regrowth of the outer amorphous layer than that of the inner one. Since more arsenic vacancies are generated in the outer region of the amorphous layer, the slower regrowth may be attributed to the deviation from stoichiometric composition of GaAs.

Next, we describe the growth plane of microtwins. As discussed earlier,<sup>2,4,12,13</sup> the twins formed in (100) GaAs wafers are known to grow from the patched (111) planes, which are revealed at the rough interface between the amorphous layer and its underlying crystalline region. Bhattacharya, Pronko, and Lee indicated from their transmission electron microscope (TEM) observations<sup>3</sup> that the twins were on at least two of the equivalent four (111) planes in the (100) GaAs wafer. But no further attention was paid to the possibility for the microtwins to grow preferentially on either the (111) *A* and *B* plane. Four (111) planes revealed in the (100) wafer are divided into two *A* and *B* planes with possibly different regrowth rates. It is known that twins show a preference to grow on the crystalline plane with a slower regrowth rate. Therefore in order to know the polarity dependence of the regrowth rate on (111) plane, direct implantation into the (111) *A* and *B* wafers is required. Implantation into the (111) wafer has already been performed, but no attention was paid to the polarity of the (111) plane in the literature.<sup>14</sup> The plane used in the previous implantation seems to be a (111) *B* plane, judging from the present results. We have attempted ion implantation into (111) *A* and *B* wafers. Figure 3 shows the annealing behavior of (111) *A* and *B* samples that were observed directly at the (111) *A* and

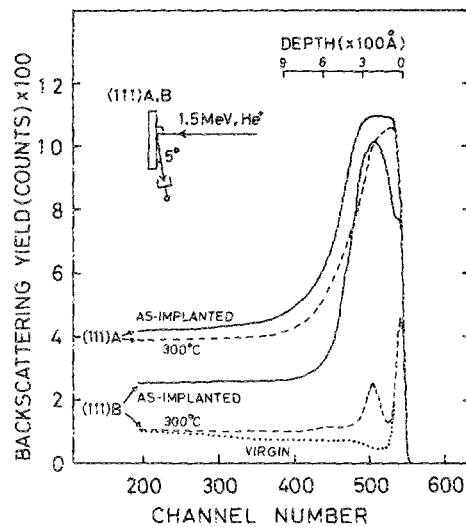


FIG. 3. (111) *A* and *B* aligned spectra directly obtained from (111) *A* and *B* wafers, respectively. Spectra obtained after annealing (300 °C, 15 min) in vacuum (broken curves) and from as-implanted samples (solid curves) are shown.

*B* directions, respectively. The implantation conditions were the same as those described above. Post-annealing was done at 300 °C for 15 min in a vacuum. In the case of the (111) *B* wafer, a slight regrowth already occurred during the implantation due to a temperature rise, and after annealing considerable restoration was observed. In the case of the (111) *A* wafer, an amorphous layer of about 460 Å was formed by implantation and only a slight regrowth (~100 Å) occurred from the inner interface even after annealing at 300 °C. It is thus apparent that the (111) *A* plane has a smaller solid-phase regrowth rate than the (111) *B* plane. It is expected, therefore, that microtwins preferentially grow on the (111) *A* plane and are scarcely observable on the (111) *B* plane.

To confirm this result, we performed another experiment to study the polarity dependence of dechanneling using a (100) wafer. Two each of the (111) *A* and *B* planes in the (100) wafer were selected by rotating the wafer by 90° intervals while keeping the wafer at an inclination of 54.75° to the He<sup>+</sup> analyzing beam in the (110) plane, which is placed perpendicular to the (100) wafer surface. If the twins lie preferentially on the (111) *A* plane and are exactly oriented to the (111) *B* twin axis, reduced dechanneling will be observed in the direction of the (111) *B* twin axis. This is due to the fundamental property of twins, that the twinned region is seen as the (111) direction only from the twin axis.<sup>16</sup> Assuming isotropic twins, the (111) *B* direction should have about half the dechanneling rate of that from the (111) *A* direction, since (111) *B* twins have a negligibly small concentration and the (111) *B* dechanneling rate is more than 1 order of magnitude larger than that of (111). Anisotropic

twin growth along the  $\langle 111 \rangle$  direction with a dendritic shape<sup>17</sup> will further increase the difference in the dechanneling rate between the  $\langle 111 \rangle A$  and  $B$  directions and also between  $\langle 111 \rangle$  and other axes such as  $\langle 100 \rangle$  and  $\langle 110 \rangle$ . However, we observed no noticeable change in the dechanneling rate between the two polar directions even for the sample annealed at 300 °C as shown in Fig. 1. This would suggest that the present twins are misoriented by at least a critical angle of the  $\langle 111 \rangle$  axis ( $\sim 0.5^\circ$ ). This is thought to be quite reasonable in view of the presence of strain and the rough nature at the initial growth front. Small-sized twins may also be responsible for the negligible dechanneling. It seems impossible that anisotropic twin growth occurs in the case of a thin amorphous layer, but this was observed in the case of a considerably thick ( $\sim 3.9 \mu\text{m}$ ) amorphous layer.<sup>17</sup>

As a result of the present work, it has been found in the implanted (100) GaAs wafer that the formation of microtwins is closely related to the presence of an arsenic-deficient phase or arsenic vacancies. It has been made clear that twins form preferentially on the (111)  $A$  (Ga) plane because this plane has a smaller solid-phase regrowth rate than that of the (111)  $B$  plane.

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- <sup>1</sup>S. S. Kular, B. J. Sealy, K. G. Stephens, D. Sadana, and G. R. Booker, *Solid State Electron.* **23**, 831 (1980).
- <sup>2</sup>M. G. Grimaldi, B. M. Paine, M.-A. Nicolet, and D. K. Sadana, *J. Appl. Phys.* **52**, 4038 (1981).
- <sup>3</sup>R. S. Bhattacharya, P. P. Pronko, and S. C. Ling, *J. Appl. Phys.* **53**, 1804 (1982).
- <sup>4</sup>R. S. Bhattacharya, A. K. Rai, P. P. Pronko, J. Narayan, S. C. Ling, and S. R. Wilson, *J. Phys. Chem. Solids* **44**, 61 (1983).
- <sup>5</sup>L. A. Christel and J. F. Gibbons, *J. Appl. Phys.* **52**, 5050 (1981).
- <sup>6</sup>Y. I. Nissim, L. A. Christel, T. W. Sigmon, J. F. Gibbons, T. J. Magee, and R. Ormond, *Appl. Phys. Lett.* **39**, 598 (1981).
- <sup>7</sup>J. Nishizawa, H. Otsuka, S. Yamakoshi, and K. Ishida, *Jpn. J. Appl. Phys.* **13**, 46 (1974).
- <sup>8</sup>H. Otsuka, K. Ishida, and J. Nishizawa, *Jpn. J. Appl. Phys.* **8**, 632 (1969).
- <sup>9</sup>J. Nishizawa, I. Shiota, and Y. Oyama, in *Proceedings of the 3rd Symposium on Ion Beam Technology, Hosei University, Tokyo, 1984*, edited by T. Sebe (Hosei University, Tokyo, 1985), pp. 21–31.
- <sup>10</sup>J. Nishizawa, I. Shiota, and Y. Oyama, *J. Phys. C* **19**, 1 (1986).
- <sup>11</sup>J. Nishizawa, I. Shiota, and Y. Oyama, *J. Phys. D* **19**, 1073 (1986).
- <sup>12</sup>A. K. Rai, R. S. Bhattacharya, P. P. Pronko, and S. C. Ling, *J. Appl. Phys.* **54**, 2307 (1983).
- <sup>13</sup>D. K. Sadana, T. Sandis, and J. Washburn, *Appl. Phys. Lett.* **44**, 623 (1984).
- <sup>14</sup>K. Gamo, T. Inada, J. W. Mayer, F. H. Eisen, and C. G. Rhodes, *Radiat. Eff.* **33**, 85 (1977).
- <sup>15</sup>C. G. Kirkpatrick, R. T. Chen, D. E. Holmes, P. M. Asbeck, K. R. Elliott, R. D. Fairman, and J. R. Oliver, in *Semiconductors and Semimetals*, edited by R. K. Willardson and A. C. Beer (Academic, New York, 1984), Vol. 20, Chap. 3.
- <sup>16</sup>G. Foti, L. Csepregi, E. F. Kennedy, J. W. Mayer, P. P. Pronko, and M. D. Reichtin, *Philos. Mag. A* **37**, 591 (1978).
- <sup>17</sup>D. K. Sadana, H. Choksi, J. Washburn, P. F. Byrne, and N. W. Cheung, *Appl. Phys. Lett.* **44**, 301 (1984).