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Continuous growth of heavily doped p^+-n^+ Si epitaxial layer using low-temperature photoepitaxy

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Heavily doped p^+ and n^+ silicon epitaxial layers were continuously grown at 600 °C using photoenhanced epitaxy. The heavily phosphorus-doped photoepitaxial layer with a carrier concentration above 1×10^{17} cm^{-3} grown on the p^- substrate shows very high density surface pits due to phosphorus precipitation, suggesting poor crystal quality. Unexpectedly, when this n^+ photoepitaxial layer is continuously grown on a heavily boron-doped p^+ photoepitaxial layer at a boron concentration above 1×10^{19} cm^{-3}, surface pits completely disappear and crystal quality is greatly improved. The very low growth temperature enabled an extremely abrupt impurity profile to be achieved for the p^+-n^+ layer.

Low-temperature epitaxy with high crystal quality is required to fabricate ultralarge-scale integrated circuits. Impurity doping of epitaxial films is necessary for many device applications. Heavy impurity doping above the concentration of 1×10^{18} cm^{-3} with the abrupt impurity profile is often essential. Conventional silicon epitaxy with chemical vapor deposition (CVD) technique uses processing temperatures exceeding 1000 °C and causes impurity redistribution and thermal stress. Attempts to obtain abrupt impurity profiles with heavy phosphorus and boron doping are not successful because the crystal quality is degraded due to impurity atom precipitation and diffusion of doped impurity atoms results from high deposition temperature. Low-temperature epitaxial techniques, including ultrahigh vacuum CVD method,[1] plasma-enhanced CVD,[2] mercury-photosensitized CVD,[3] and low kinetic energy particle process epitaxy,[4] have been reported, but still do not produce the required device-level crystal quality in the heavy impurity doping region.

We have developed a low-temperature epitaxial technique using direct ultraviolet (UV) photosensitization of disilane gas (Si_2H_6).[5,6] UV irradiation enhances the disilane's vapor phase reaction, increasing the growth rate, and activates the surface reaction and migration of adsorbed species, lowering the growth temperature. UV irradiation also enhances the diborane (B_2H_6) photochemical reaction in the vapor phase and the incorporation of adsorbed impurity atoms into substitutional positions in the Si crystal, increasing boron doping efficiency and enabling heavy boron doping with high crystal quality.[7] This letter describes p^+-n^+ continuous growth using this photoepitaxy, the resulting crystal quality, and the impurity profile.

Low-temperature photoepitaxy was done in a specially designed quartz chamber having a base pressure of 1×10^{-4} Pa. A (100) oriented Si wafer was heated by infrared lamps from behind. As the UV light source, we used a high-pressure mercury lamp which emitted UV light: a sharp line of 185 nm wavelength and a continuous wave longer than 195 nm wavelength. Three of these lamps were located on the quartz chamber, and vertically irradiated the wafer surface with a UV intensity of 1.2 W/cm^{-2} (integrated value of wavelengths >300 nm).[6,7] Disilane of 1.5 cc/min was used as the Si source, phosphine (PH_3) as the n-type doping source, diborane as the p-type doping source, and hydrogen of 20 ℓ/min as the carrier gas. Deposition pressure was 27 kPa. The Si wafer was cleaned by conventional wet methods before loading to the growth chamber. To reduce the native oxide, the wafer was then heated at 900 °C for 10 min in hydrogen ambient before deposition. The temperature was then reduced to the deposition temperature at 600 °C.

In the first place, we describe the phosphorus doping characteristics in low-temperature photoepitaxy. At a phosphine flow rate below 10^{-2} cc/min, which produces an electron concentration of 1×10^{18} cm^{-3}, the growth rate is independent of the phosphine flow rate at 6 nm/min. The growth rate of the photoepitaxy was four times that of thermal growth. The increase of the growth rate due to UV irradiation is caused by the photochemical dissociation of the disilane in the vapor phase, which has been investigated using the Fourier transform infrared (FTI) spectroscopy.[8] In phosphine flow rates above 10^{-2} cc/min, the growth rate decreased linearly to the increase in phosphine flow rate. A decreased growth rate, due to the addition of high-volume phosphine, has been observed in conventional CVD epitaxy and is attributed to homogeneous nucleation in the vapor phase.[9] In low-temperature photoepitaxy, this nucleation was not observed. At low temperature, however, a high electron density on crystal surface increases the hydrogen surface coverage, decreasing the adsorption probability of photochemically produced silylene species and the dehydrogenation of adsorped species.[10,11] The growth rate of the photoepitaxial layer then decreased at a phosphine flow rate above 10^{-2} cc/min.

In the electron doping concentration below 1×10^{18} cm^{-3}, the doped electron concentration was proportional to the phosphine flow rate, and it saturated at about $n=1\times10^{19}$ cm^{-3}. Adding more phosphine degraded the crystal quality, eventually resulting in the growth of a phosphorus-silicon compound. The electron Hall mobility in the n^+ photoepitaxial layer was smaller than that of bulk Si, indicating poor crystal quality. The heavily phosphorus-doped n^+ photoepitaxial layer with a phosphorus concentration above 1×10^{17} cm^{-3} had many surface pits (Fig. 1). These surface pits increased as the doped phosphorus concentration increased, and the maximum pit density was at 10^7 cm^{-2} for a doped electron concentration of 5×10^{18}

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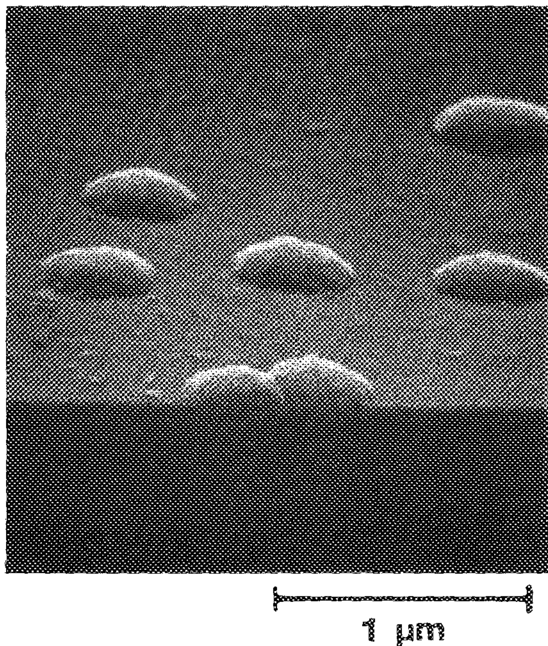

FIG. 1. Scanning electron microscopy (SEM) micrograph of surface pits on the n^+ photoepitaxial layer with a phosphorus concentration of 5×10^{18} cm^{-3} grown on the p^- substrate ($\rho = 10\ \Omega$ m).

cm^{-3}. The origin of these surface pits may be phosphorus precipitation, but details are not clear, and their origin requires further study.

We attemped continuous growth of a heavily phosphorus-doped n^+ layer on the heavily boron-doped p^+ layer using photoepitaxy. Conventional CVD epitaxy cannot grow a heavily boron-doped single crystal at boron concentration above 1×10^{18} cm^{-3} at a growth temperature below 900 °C because of the severe degradation of crystal quality due to boron atom precipitation. Photoepitaxy can grow a single crystal at a maximum boron doping concentration of 1.5×10^{20} cm^{-3} without any crystal defects at 600 °C.[7] Using

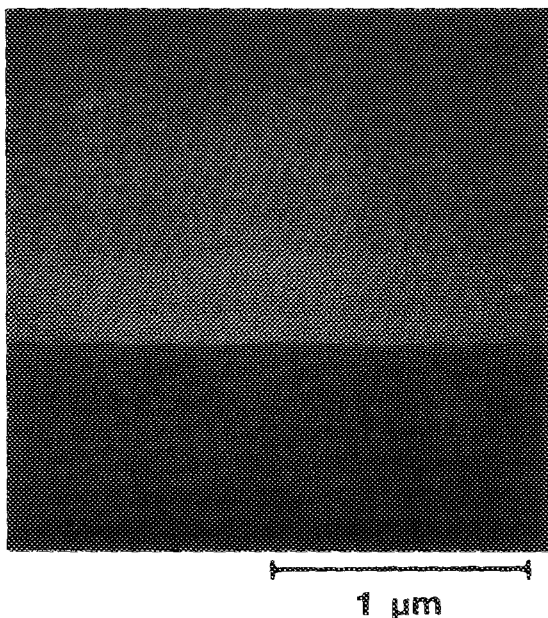

FIG. 2. SEM micrograph of surface of the n^+ photoepitaxial layer with a phosphorus concentration of 5×10^{18} cm^{-3} grown on a p^+ photoepitaxial layer with a boron concentration of 2×10^{19} cm^{-3}.

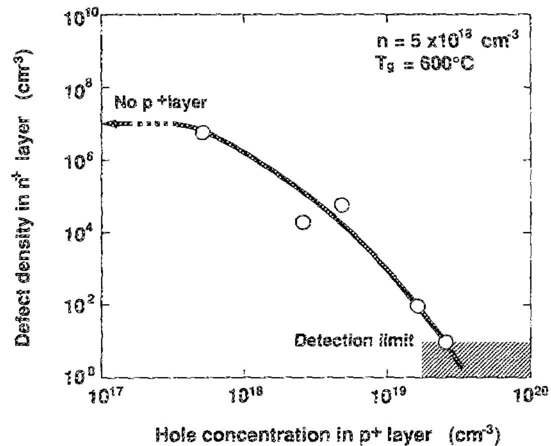

FIG. 3. Dependence of hole concentration in the p^+ photoepitaxial layer on surface-pit density at the n^+ photoepitaxial layer.

such low-temperature photoepitaxy, we could grow a heavily phosphorus-doped n^+ layer with a 5×10^{18} cm^{-3} phosphorus concentration on a p^+ photoepitaxial layer with a 2×10^{19} cm^{-3} boron concentration by simply changing the doping gas from diborane to phosphine and without removing the wafer from the growth chamber. The growth temperature for both layers was 600 °C. The n^+ photoepitaxial layer was 150 nm thick and the p^+ photoepitaxial layer was 50 nm thick.

Figure 2 shows the surface morphology of the n^+ photoepitaxial layer grown on a p^+ photoepitaxial layer. None of the surface pits in Fig. 1 were observed, and a very specular surface was obtained. Figure 3 shows the characteristics of the hole concentration in the p^+ photoepitaxial layer and the surface-pit density at the n^+ photoepitaxial layer surface. The hole concentration was controlled by the diborane flow rate, and growth conditions for the n^+ photoepitaxial layer were fixed. The surface-pit density decreased markedly with increasing hole concentration. Eventually, surface pits completely disappeared at hole concentration above 1×10^{19} cm^{-3}. This n^+-p^+ continuous growth also greatly improved the crystal quality and electrical properties of the n^+ photoepitaxial layer. Table I shows the activation ratio of phosphorus atoms and the electron mobility in the n^+ photoepitaxial layer grown on a p^+ photoepitaxial layer, compared to that in an n^+ photoepitaxial layer grown directly on the p^- substrate. The phosphorus activation ratio determined by Hall measurement and secondary-ion mass spectrometry (SIMS) analysis was improved from 67 to 100%,

TABLE I. Activation ratio of phosphorus atoms and electron Hall mobility in the n^+ photoepitaxial layer grown on the p^- substrate or p^+ photoepitaxial layer. The p^- substrate has a resistivity of 10 Ω cm and the p^+ photoepitaxial layer has a boron concentration of 2×10^{19} cm^{-3}.

	Photo-n^+/photo-p^+/p^- sub.	Photo-n^+/p^- sub.
Phosphorus atom activation ratio	100%	67%
Electron mobility (μ/μ_0)[a]	1.0	0.6

[a] μ_0: bulk Si electron mobility.

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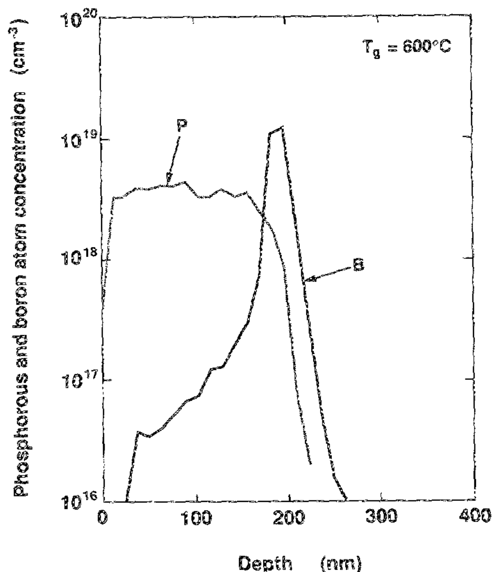

FIG. 4. Phosphorus and boron profiles in the p^+-n^+ continuously grown layer measured by SIMS analysis.

and the electron Hall mobility became the same as that of bulk Si in n^+-p^+ continuous growth. The growth rate and doped phosphorus atom concentration in the n^+ photoepitaxial layer increased slightly due to n^+-p^+ continuous growth.

Even though lattice mismatch and stresses at the interface of n^+-p^+ photoepitaxial layers are larger than those at the interface of the n^+ photoepitaxial layer and p^- substrate, the crystal quality of the n^+ photoepitaxial layer grown on the p^+ photoepitaxial layer was much better than that grown on the p^- substrate. The high hole density on the surface apparently decreases the hydrogen surface coverage, increasing the adsorption probability of the reactive species, and activates dehydrogenation and surface migration of adsorbed species such as silylene or phosphorus-hydrogen compounds.[10,11] Phosphorus atoms on the crystal surface are then less likely to be precipitated, and fit to the crystal growth site corresponding to the substitutional position in the Si crystal. As a result, the poor crystal quality and phosphorus activation ratio of the heavily phosphorus-doped photoepitaxial layer was greatly improved.

The low growth temperature of 600 °C enabled us to obtain extremely abrupt impurity profiles. Figure 4 shows the boron and phosphorus profiles of continuously grown p^+-n^+ photoepitaxial layers which were measured by SIMS analysis. Boron and phosphorus profiles were sharp at the edges and flat throughout the doped region, and no phosphorus atoms were precipitated at the n^+-p^+ interface.

To summarize, a heavily phosphorus-doped photoepitaxial layer with a 5×10^{19} cm^{-3} electron concentration was continuously grown on a heavily boron-doped photoepitaxial layer with a 2×10^{19} cm^{-3} hole concentration at 600 °C. Surface pits on the n^+ photoepitaxial layer that were caused by the precipitation of phosphorus atoms were greatly decreased by increasing the hole concentration in the p^+ photoepitaxial layer; these pits completely disappeared at hole concentrations above 1×10^{19} cm^{-3}. This n^+-p^+ continuous growth technique also greatly improved the electrical properties and phosphorus atom activation ratio in the n^+ photoepitaxial layer. The impurity profiles of the phosphorus and boron atoms were extremely abrupt.

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