Microstructure and Resistivity of Laser-Annealed Au-Ge Ohmic Contacts on GaAs

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ABSTRACT

The results of a study of laser-annealed Au-Ge ohmic contacts to GaAs are presented. The specific contact resistivity was observed to decrease with increasing laser energy density while the grain size of the polycrystalline microstructure (as observed by transmission electron microscopy) increased. At higher energy densities, both parameters were found to remain constant within the experimental conditions used. Transmission electron micrographs, and sputtering Auger electron spectroscopic data showing Ga, As, and Ge redistribution within the Au-Ge film are also presented.

Recent research has focused on using laser irradiation for making ohmic contacts to GaAs because the localized, short duration heating of a laser beam was expected to result in negligible GaAs decomposition and therefore improve contact quality. Results obtained for laser-annealed ohmic contacts (1, 2) have so far shown no significant improvement in contact properties over contacts made by conventional furnace alloying. A more detailed study of laser-annealed ohmic contacts, focusing on the materials properties, therefore seems necessary.

Transmission electron microscopy (TEM) is useful not only for microstructural study of materials but also for identification of material types and phases. It has been used to study the alloying of thin films of Au (3) and Au-Ge-In (4) on GaAs. However, only the microstructural properties of these contacts were investigated.

We report here a study of laser-annealed Au-Ge/GaAs ohmic contacts using TEM to correlate the microstructure and the electrical properties as obtained from specific contact resistivity measurements. Specifically, the variation of the specific contact resistivity as a function of the laser energy density was measured and is presented along with variations in the microstructure and the compositional depth profile as measured by Auger electron spectroscopy (AES).

Sample Preparation and Experimental Procedure

Te-doped (100) n-GaAs substrates of carrier concentration, $2 \times 10^{18}$ cm$^{-3}$ were used. They were first cleaned in trichlorethylene, acetone, and methanol. An oxide was then grown in deionized water and etched off with HCl. An Au-Ge layer of 1000Å thickness was deposited in vacuum ($2 \times 10^{-6}$ Torr) from an alloy of eutectic composition [Au-Ge(12%)] which was evaporated to completion in a resistance heated tungsten boat. This deposition was done through a metal evaporation mask of parallel slits that resulted in the contact structure shown in Fig. 1.

Ohmic contacts were formed by single pulse irradiation of the AuGe film with a pulsed Nd-YAG laser of 1.06 micron wavelength, 30 nsec pulse duration, and beam spot diameter of 0.1 cm. The pulse energy density was varied from 0.6 to 3 J/cm$^2$.

The ohmicity of the contacts was determined from curve tracer I-V plots while the specific contact resistivity was determined by a four-probe measurement technique as shown in Fig. 1. Current (10 mA) is passed through two of the contact pads while the voltage between one of the current-carrying pads and

![Fig. 1](image-url)

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Fig. 1. (a) Four-point probe geometry for ohmic contact measurements; (b) resistance measured as a function of distance.
each of the inner pad is measured. The resistance variation as shown in Fig. 1b has been derived by Schuldt (5) by solving Laplace's equation numerically with the ohmic contact plane as a boundary and is given by

\[ R(x) = \left( \frac{\rho_B}{H} \right) x + \left( \frac{\rho_B}{nt} \right) \ln \left( 1 - \exp \left( -\frac{D}{H} \right) \right) + f \left( \rho_c, \frac{D}{H} \right) \]  

[1]

where \( \rho_B \) is the bulk resistivity, \( H \) is the sample thickness, \( W \) sample width, \( D \) contact length, and \( f(\rho_c, D/H) \) is a resistance parameter which is a function of the specific contact resistivity \( \rho_c \) and the ohmic contact geometry. Using \( R_o \), the intercept from the four-probe measurements, the specific contact resistivity \( \rho_c \) was obtained from numerical solutions for \( f(\rho_c, D/H) \) (5). Prior to the measurements, the sample was scribed into rectangular strips in a direction perpendicular to the Au-Ge stripes. This was to ensure current flow paths only in the planes perpendicular to the contact plane and along the length of the GaAs strip. Equation [1] is valid only for this geometry.

The samples were prepared for TEM examination by chemical jet polishing with bromine methanol from the back side of the substrate so that only the top surface was observed.

**Results**

Figure 2 shows the \( I-V \) characteristics of contacts formed under various conditions. The as-deposited Au-Ge film is a rectifying contact as Fig. 2a shows, but after laser irradiation at 0.6 and 1.1 J/cm\(^2\), the contact becomes ohmic (Fig. 2b, c). The change in the specific contact resistivity as the laser energy density is varied is shown in Fig. 3. It can be seen that the electrical properties of the contacts are improved at higher laser energy densities. The corresponding microstructural changes are shown in Fig. 4 in which the microstructure of the contact irradiated at 0.6 J/cm\(^2\) appears similar to that of the as-deposited Au-Ge film. The grain size of the contacts irradiated at 1.1 J/cm\(^2\) has increased significantly while the grains show features which were not observed in the other contacts fabricated at lower energy densities. The results obtained are summarized in Fig. 5 which compares the measured residual resistance \( R_o \) of the contacts and the corresponding grain size. At energy densities higher than 1.1 J/cm\(^2\), both the specific contact resistivity and the grain size seem to remain constant.

The TEM diffraction patterns also showed changes corresponding to the formation of ohmic contacts by laser irradiation. For example, three rings can be observed in the diffraction patterns of Fig. 4b and c which do not exist in the diffraction pattern of the as-deposited AuGe film shown in Fig. 4a. By matching lattice spacings, these extra rings have been identified with germanium. The dark field image shown in Fig. 6 was obtained with the objective aperture of the TEM on the Ge(111) ring. It indicates segregation of Ge at the grain boundary. The white spots within the grains were also found to be germanium.

Using Auger analysis combined with argon ion sputtering, measurements were made on the as-deposited samples as well as the laser irradiated samples in order to observe the changes in the composition depth profile of the contacts corresponding to the observed electrical and microstructural changes (as a function of laser energy density). The elemental depth distributions of oxygen (503 eV), gallium (1070 eV), germanium (1147 eV), arsenic (1228 eV), and gold (2024 eV) are shown in Fig. 7 for the as-deposited AuGe film. The essential features of this profile (such as the sharp interface) do not change with laser irradiation. There appears, however, to be an increasing redistribution of all the elements present with increasing laser energy density. In Fig. 8, the profiles of Ga and As show a significant redistribution of these two elements.
Within the energy density range used, this redistribution does not reach the surface and thus no gallium segregation at the surface was observed as was the case with furnace alloyed Au-Ge (and Au-Ge-Ni) ohmic contacts (6). Figure 9 shows a corresponding decrease in the Ge profile which indicates a redistribution of Ge into GaAs. This could not be confirmed by the present AES data because of the coincidence of the arsenic and germanium Auger electron transitions at 1147 eV.

**Discussion**

It has been observed that pulsed laser annealing of implanted semiconductors results in melting, the duration of which is longer than the pulse width of the laser beam (7, 8). Specifically, it was found that the duration of melting increased with the laser energy density and that for Te-implanted GaAs, the melt duration was approximately 1 μsec at an energy density of 1 J/cm² (7). In view of the short time periods involved in laser annealing, it seems that the drastic change in microstructure observed in this work...
could not be due to solid-state diffusion (the solid-state diffusion length is much less than the grain sizes observed). This kind of microstructural change is therefore due to melting of the Au-Ge film and an adjacent layer of the GaAs substrate. By analogy with observations of laser annealing of aluminum on silicon (9), this melting is followed by rapid regrowth of Ge-rich GaAs with a top layer of alloys or compounds of Ga-Ge, Au-Ga, and Au-Ge. This kind of structure, consisting of successive layers of GaAs heavily doped with Ge; Au-Ge, Ni-Ga, and Au-Ga alloys and compounds has been observed in furnace alloyed Au-Ge-Ni ohmic contacts (10).

The grain size of a polycrystalline structure which has regrown from a melt is related to a growth parameter which is known to vary inversely as the regrowth velocity (11). The regrowth velocity however, has been shown to also vary inversely as the energy density of the laser beam (7). Thus, the increase in grain size observed in this work may be attributed to an increase in the growth parameter as a result of a decrease in the regrowth velocity with energy density. The decrease of the specific contact resistivity on the other hand may be due to an increased Ge doping of the Ge-rich GaAs layer (12). In GaAs, Ge may go into either Ga or As vacancies depending on the temperature and the vapor pressures (13) of As and Ga. It appears that the constancy of the specific contact resistivity at the higher laser energy densities is due to an increasing concentration of Ge in As sites, instead of in Ga sites as is required for n-type doping.

Our AES results showing limited redistribution of Ga and As are in contrast to previous results obtained for furnace alloyed Au-Ge-Ni ohmic contacts (6) which showed significant Ga and As concentrations at the contact surface. Limited Ga and As segregation as was found here may be due to an incomplete mixing of the melt from which the regrowth takes place. It was found for implanted Si, from numerical calculations, that the melt-front velocity is substantially greater than the velocity of mass diffusion (14). The observed AES profiles for Ga and As are thus probably related to the distribution of Ga and As in the melt which is “frozen” into the ohmic contact layer during solidification.

The observed redistributions of Ga and As can result in Ga and As vacancies in the GaAs substrate. Hence, the substrate concentration of vacancies should be higher in the furnace alloyed contacts than in the laser-annealed ones since there is less redistribution in the latter. One would therefore expect laser-annealed ohmic contacts to have improved contact reliability, since contact reliability has been suggested to depend on vacancy formation in GaAs (15).

Conclusion

We have presented results showing a decrease in the specific contact resistivity with energy density for laser annealed Au-Ge ohmic contacts. We have at the same time shown the correlation between the measured contact resistivity and the microstructure observed by the TEM. In addition, AES results were presented to show the redistribution of Ga and As as a result of the laser annealing.

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REFERENCES

Submicron silicon epitaxial films are becoming increasingly important with the development of VLSI technology. Epitaxial layer thicknesses in the 0.5-1.5 μm range are needed to accommodate the small cell sizes required for VLSI designs. The transients associated with the first few minutes of growth, however, impose severe limitations on the fabrication of such thin films. As is described elsewhere (1), the dopant incorporation process in a silane/arsine epitaxial system requires several minutes to regain steady state when a perturbation such as a step change in dopant gas flow is imposed on the reactor. Similarly, the dopant incorporation process requires several minutes to reach the expected steady-state condition at the beginning of the deposition step (2). Consequently, the epitaxial dopant concentration reaches the expected doping level only for films thicker than that corresponding to this initial transient. Because this transient period corresponds to a transition layer 1-2 μm thick, the expected steady-state doping level may not be achieved in submicron silicon films.

In addition to the dopant incorporation process, the silicon deposition process must also go through a transient period during the initial stages of growth, i.e., some time is required before the expected steady-state epitaxial growth rate is established. A complete understanding of the transients associated with both the dopant incorporation and silicon deposition processes is critical to the fabrication and control of submicron epitaxial films.

The dopant incorporation process has been the subject of previous investigations (1-4). The work presented here deals primarily with the silicon deposition process. The main objective of this work was to investigate the transient associated with the establishment of a steady-state silicon growth rate and the effect of this transient on the epitaxial doping level. The simplest technique that could have been used to determine this transient was to carry out a series of depositions for different growth times and then measure the resulting epitaxial layer thicknesses. However, this method is imprecise because very short initial time intervals are needed in order to obtain information on the initial transient. Moreover, the effect of the initial transient on the film thickness could get compensated by that of the final transient, thereby concealing the information being sought. The most desirable approach would be to measure the film thickness continuously as the film is being deposited. Dumin (5) developed an interferometric technique that measured the thickness of silicon films as they were being deposited on sapphire substrates. This technique utilized the infrared emission from the sapphire substrate and from the growing film. The radiation from the sapphire substrate was partially transmitted through the silicon and partially reflected in the silicon, establishing an interference pattern which was used to determine silicon film thickness. Shaw (6), on the other hand, developed a gravimetric technique in which the weight of the growing gallium arsenide crystal was continuously measured. The use of these techniques, however, poses several difficulties when applied to the epitaxial deposition of silicon. The interferometric technique is best suited to cases in which the index of refraction of the film being deposited is different from that of the substrate, while the gravimetric technique measures not only the weight of the substrate but that of the substrate support as well.

In this paper, a different technique is used to investigate the transient associated with the silicon deposition