CHARACTERISATION OF BF$_2$+ AND B+ IMPLANTED SILICON AFTER RAPID THERMAL ANNEALING.

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A detailed characterisation of rapid thermally annealed BF$_2$+ and B+ implanted silicon has been carried out using the complementary techniques of in line 4 point probe and Van der Pauw electrical measurements, transmission electron microscopy, Rutherford backscattering/channeling and secondary ion mass spectrometry. Our results after rapid thermal annealing indicate that defect trapping and release of boron strongly influence both the diffusion behaviour and the electrical properties of the samples.

1. Introduction

Rapid thermal annealing (RTA) of low energy B+ and BF$_2$+ ion implanted silicon has been widely studied in recent years [1–7] as an attractive approach to obtaining shallow p+-n junctions. Several issues have been addressed: i) investigation of “anomalous” transient diffusion of boron [1,2]; ii) suppression of boron diffusion during activation using BF$_2$+ or preamorphised wafers to minimise, in particular, rapid diffusion of the “channeled” boron tail [3–6]; iii) optimisation of RTA conditions to remove remnant defects and provide maximum activation whilst minimising diffusion and maintaining good p–n junction properties (i.e. low leakage currents) [5–7]; iv) determination of the role of fluorine and defect–fluorine interactions during RTA of BF$_2$+ implants [6,7]. Although the general RTA behaviour has been reasonably well characterised and is largely understood, details of defect–dopant (and defect–fluorine) interactions and their influence on diffusion and activation, particularly in the high dose (> 10$^{15}$ cm$^{-2}$) regime, have not been well characterised. In this paper we report on some recent results of a detailed electrical and physical characterisation of B+ and BF$_2$+ implanted silicon subjected to RTA in the temperature range 600 to 1100°C.

2. Experimental

Silicon wafers of (100) orientation and nominally 5 Ω cm n-type were implanted with either B+ (at 20 keV) or BF$_2$ (at 60 and 100 keV). The implants were carried out at room temperature to a dose of 3 × 10$^{15}$ cm$^{-2}$ in each case. Small 7 × 7 mm$^2$ samples were then subjected to RTA at temperatures in the range 600–1100°C. The RTA equipment used in these experiments was built in-house and incorporates quartz halogen lamps and parabolic reflectors, as described previously [8]. The temperature under well defined RTA conditions was measured by inserting a control sample (of identical dimensions) into the system with a well bonded thermocouple attached. A total RTA on-time of 15 s was employed and the measured temperature refers to the maximum temperature attained during this time. One set of samples was given a 3-stage furnace anneal (550°C, 2 h; 850°C, 15 min; and 550°C, 2 h) for comparison with the RTA samples.

All annealed samples were examined by 4 point probe and Van der Pauw techniques to provide sheet resistance, average mobility and carrier concentration. Selected samples were thinned and examined by plan-view transmission electron microscopy (TEM) to determine the nature of residual defects. Rutherford backscattering and channeling (RBS–C) was employed to measure the depth distribution of gross damage and secondary ion mass spectrometry (SIMS) was used to measure the boron and/or fluorine depth distributions.

3. Results and discussion

3.1. Electrical measurements

Fig. 1 summarises the electrical data for 20 keV B+ and 100 keV BF$_2$+ implants. The differences between the two implanted species are to be expected in view of the initial damage. The B+ implanted layer is amorphous and the incomplete activation of this species at RTA temperatures of < 900°C is consistent with previous work [3–5]. For BF$_2$+, almost complete activation follows solid phase epitaxial (SPE) growth, which is complete by 700°C (15 s). However, there are two features of the electrical measurements which are somewhat unexpected. Firstly, the furnace-anneal results indicate significantly lower electrical activity than RTA
treatments, particularly for the \( B^+ \) implants. As indicated later, we interpret this difference in terms of dopant–defect interactions which are particularly sensitive to thermal history. Secondly, average Hall mobilities (of holes) measured for both \( B^+ \) and \( BF_2^+ \) implants, following \( > 800^\circ C \) RTA, were anomalously low \((< 25 \text{ cm}^2/\text{Vs})\) in view of the expected maximum carrier concentration from \( 3 \times 10^{15} \text{ cm}^{-2} \) implants. Indeed, calculations of active dose (or average free carrier concentration), using the measured sheet resistance and the average Hall mobility, in some cases gave values which exceeded the implanted dose. Furthermore, even when the boron profiles had exhibited appreciable diffusion, thus lowering the average carrier concentration (see SIMS profiles in fig. 4), the measured mobility remained constant. Although more careful differential Hall measurements are necessary to verify this anomalous electrical behaviour, we speculate that the presence of active defects in the original implanted layer may be playing some role in determining the electrical properties of the layers.

3.2. Residual disorder

In fig. 2a we show selected RBS–C spectra for \( B^+ \) implanted layers. The spectrum corresponding to the as-implanted case indicates a broad damage distribution centred at about 700 Å, with considerable damage extending out to the surface. (Note the significantly larger surface peak compared with the annealed cases.) The deep damage profile correlates well with the LSS profile [9] for \( 20 \text{ keV B}^+ \) in silicon \((R_p = 662 \text{ Å and } \Delta R_p = 282 \text{ Å})\).

Following the 3-step furnace anneal, the RBS–C spectrum indicates, rather surprisingly, a significant increase in yield from the deep damage, but almost complete removal of near-surface damage. We illustrate differences in the nature of the residual \( B^+ \) disorder in the TEM micrographs of figs. 2b and c. Fig. 2b shows the typical “salt and pepper” defect structure characteristic of room temperature \( B^+ \) implanted silicon [10]. This structure essentially remains following a RTA at \( 600^\circ C \). Note also the spot streaking and splitting in the selected area diffraction pattern. After the 3 step furnace anneal (fig. 2c) the deep damage has coalesced into visible rods and loops. Analysis of the near-surface region of this sample (in the thin region to the right of fig. 2c) indicates an absence of such defects, consistent with the RBS–C data. Indeed, taken together, the RBS–C and TEM data suggest that the point defects and clusters which are present following implantation migrate and coalesce into extended defects of interstitial character distributed about the peak of the boron distribution.

The final RBS–C spectrum in fig. 2a shows the \( 1000^\circ C \) 15 s RTA result: clearly the deep damage peak is much reduced and the spectrum closely resembles that of a virgin sample, suggesting that considerable annealing out of the defects has taken place.

In fig. 3 we illustrate typical RBS–C and TEM results for the 60 and 100 keV \( BF_2^+ \) implants. In these cases, the as-implanted amorphous layer is removed on
Fig. 2. a) RBS-C spectra for RTA (1000°C, 15 s) and 3-step furnace annealing of 20 keV B⁺ implanted silicon. b) and c) TEM micrographs and SAD's for 600°C RTA and 3-step furnace annealing, respectively, of 20 keV B⁺ implanted silicon.

Fig. 3. a) RBS-C spectra for RTA (1000°C, 15 s) and 3-step furnace annealing of 60 keV BF₂⁺ implanted silicon. b) TEM micrograph and diffraction pattern for 100 keV BF₂⁺ implanted silicon after RTA at 900°C for 15 s.

annealing by SPE leaving behind both a well-defined damage peak close to the original amorphous-crystal interface (~1000 Å in fig. 3a) and a smaller, less well defined peak corresponding to the range of fluorine (~600 Å for 60 keV BF₂⁺ in fig. 3a). The TEM micrograph from a wedge sample in fig. 3b indicates that the residual disorder following SPE commences at a sharply defined depth and consists of large perfect loops of interstitial type, residing on (100) planes parallel to the surface. The left hand side of the micrograph corresponds to thicker crystal: the thinner right hand side is closer to the surface and is free of large loops.

IV. PHASES/RTA
Fig. 4. a) Boron SIMS profiles: as-implanted (1), after RTA 1000°C, 15 s (2), and 3-step furnace annealing (3) of 20 keV B⁺ implanted silicon. b) Boron SIMS profiles after RTA (1000°C, 15 s) and 3-step furnace annealing of 100 keV BF₂⁺ implanted silicon; curves as in a). c) Fluorine SIMS profiles of b). (Courtesy of Charles Evans and Associates).

3.3. Boron and fluorine profiles

Fig. 4 shows typical SIMS profiles for 20 keV B⁺ and 100 keV BF₂⁺ implants. The boron profiles in fig. 4a indicate a marked boron “channeling” tail, which has substantially diffused for the 3-stage furnace anneal, consistent with previous results [1, 2, 6]. It is interesting that the majority of boron near the profile peak has not diffused during this 850°C, 15 min furnace anneal sequence, although some diffusion should have been observed at 850°C based upon expected concentration-dependent diffusion data for B in silicon [11]. We suggest that the boron is predominantly trapped in complexes and at extended defects surrounding the boron profile, identified in fig. 2. The low activity for the furnace anneal of B⁺ implanted silicon is attributed to this process. After a 1000°C, 15 s RTA the boron appears to be released from these defects and can then diffuse into the silicon (fig. 4a) to become fully active (fig. 1b).

Fig. 4b shows selected boron profiles for BF₂⁺ (100 keV) implants. In this case the boron near the peak does diffuse (in addition to the “channeling” tail) during the 850°C furnace anneal. However, a considerable fraction is clearly trapped at the deep damage (corresponding to the original amorphous-crystalline interface as measured by RBS-C) and possibly also at the depth corresponding to the fluorine projected range. Following 1000°C, 15 s RTA the boron is again released from defects to diffuse into the silicon and attain full activation.

Fig. 4c shows fluorine profiles corresponding to the boron profiles in fig. 4b. After furnace annealing, and RTA up to 1000°C (15 s) fluorine appears to be trapped at both the original amorphous-crystalline interface disorder and at the original fluorine range. This defect gettering of fluorine is consistent with previous observations [6, 7, 12]. TEM also provided some evidence that the fluorine may exist, in part, in the form of bubbles in these regions. Finally, the SIMS data clearly indicate the important role of defects in gettering and trapping both the boron dopant and fluorine.
4. Conclusions

1) Isochronal RTA results indicate that activation processes are decidedly different for B⁺ and BF₂⁺ implants and are strongly influenced by dopant-defect interactions. The 3-stage (550, 850 and 550°C) furnace annealing schedule used in this study produces low electrical activity as a result of boron trapping at defects.

2) The diffusion of high dose boron (and BF₂) in silicon is strongly influenced by defects and defect-trapping. Care should be exercised when measuring diffusion parameters in such cases to ensure that dopant-defect interactions are properly accounted for.

3) We suggest that the anomalously low mobilities and anomalously high activities obtained for high temperature RTA of both B⁺ and BF₂⁺ may result from (electrically active) defect complexes.

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References