featuring pure and volatile reagents. The SC-1 solution for the first processing step is a mixture of NH_4OH (ammonium hydroxide), H_2O_2 (peroxide) and H2O; also known as "APM" for "ammonia/peroxide mixture." The SC-2 solution for the second processing step is a mixture of HCl (hydrochloric acid), H_2O_2 and H_2O; also known as "HPM" for "hydrochloric/peroxide mixture." A typical composition for the SC-1 solution ranges from 5:1:1 to 7:2:1 parts by volume of: H_2O:H_2O_2:NH_4OH. A typical composition for the SC-2 solution ranges from 6:1:1 to 8:2:1 parts by volume of H_2O:H2O2:HCl. The exact compositions for both solutions are not critical for proper performance; the recommended proportions are reliable and simple to prepare and use. Cleaning in either mixture is carried out at about 75-85° C. (e.g., about 80° C.) for about 10-20 minutes followed by a quench and overflow rinse in running DI H_2O.

The same N, Ga and Al sources were used. There was no additional high temperature thermal treatment [18] of the Si substrates before the growth. To achieve a uniform temperature distribution, the wafers were supported only by thin Mo wires to avoid any hard thermal contact between the substrate and the Mo holder used for affixing the sample during growth to the heater assembly. The AlN buffer layer was grown at the same temperature followed by growth of GaN at a temperature 20° C. higher than in the previous MBE growth. The GaN layers were grown under a Ga rich condition and the growth rate was limited by the nitrogen flux. During the growth, reflection high-energy electron diffraction (RHEED) was used to monitor the surface condition.

TEM Studies

Characterization of GaN grown on implanted and un-implanted Si substrates was performed by TEM using a JEOL 3010 transmission electron microscope. Since the RBS results shown in FIGS. **2**-**3** suggest highest dechanneling at 600° C., and decreasing (but still observable) at 750° C., we used the Molecular Beam Epitaxy (MBE) growth method instead of the more conventional MOCVD technique, since the lower growth temperature can be more easily applied. We were aware that the quality of the layer can suffer due to utilization of a low growth temperature, but a compromise was attempted to meet the objective of redirecting misfit dislocations from the interface to the substrate. AlN layers were grown slightly above 600° C. with a thickness of about 30 nm followed by a 1 μm thick GaN layer grown at a 50-70° C. higher temperature than used for the buffer layer. Si wafers were cleaned by the same procedure I. Annealing of the substrates prior to growth was not used. Samples with two different implantation energies (15 keV and 30 keV) and same fluence (1×10^{16} He cm^{-2}) were studied. In addition, the same structures were grown on un-implanted Si using the same growth and substrate cleaning procedure.

The first GaN layer about 1030 nm thick was grown on 30-40 nm thick AlN on top of an implanted (15 keV and fluence 1×10^{16} cm^{-2}) Si substrate (FIG. **4**). The measured distance from the interface to the He bubbles, based on TEM micrographs, was 120 nm, close to the nominal value of 130 nm determined by RBS. In this sample, misfit dislocations interacted with the strain field created in the vicinity of many He bubbles and these dislocations propagated further into the substrate. End-of range defects in the form of dislocation loops formed due to implantation were also clearly observed. There are also some areas of the samples where large voids are formed at the interface (FIG. **5**). Any interaction with dislocation loops in the vicinity of such a void was not observed.

A higher implantation energy (30 keV) and fluence (1×10^{16} cm^{-2}) (FIG. **6**) was also studied. The GaN layer (about 1050 nm thick) was grown on 40 nm thick AlN on top of an implanted Si substrate. The distance from the interface to the He bubbles in this case was about 240 nm, close to the value of 270 nm determined by RBS. Only a narrow band with He bubbles was observed in this area of the Si substrate. In this sample some misfit dislocations interacted with the strain field around the bubbles and propagated deep to the substrate where end of range defects are formed and beyond. However, the distance between the dislocations that propagated from the interface to the Si substrate was much larger (200-400 nm) than in the sample with a lower fluence and shorter distance from the Si surface. It is most likely that the He bubbles are formed too deep in the substrate and the strain field around them is not large enough to interact with misfit dislocations formed at the interface with the AlN buffer layer. Dislocation density in the GaN layer was rather large (9×10^{11} cm^{-2}).

Samples that were implanted (15 keV and fluence of 1×10^{16} cm^{-2}) through the SiO_2 cap layers (removed before the growth), where He bubbles are formed at the depth of 70 nm did not show strong interaction with the misfit dislocations since many elongated voids were present at the interface with the AlN buffer layer. From these studies it was clear that there is only a narrow range of distances for which interaction with misfit dislocations is possible. It was also observed that despite many misfit dislocations being "pushed" back to the Si substrate the GaN layer quality did not improve. It was clear that this low growth temperature was one of the parameters determining the GaN quality. The second parameter, that we also considered was the cleaning procedure, since the traditionally used annealing procedure for oxide removal could be not applied in this case.

To check the reproducibility of the results a new set of GaN/AlN/Si samples was grown in the same MBE system using the cleaning procedure II, in accordance with the present invention. Cross-section and plan-view samples were studied by TEM. Diffraction patterns and bright field micrographs prepared from plan-view samples show a drastic difference in the structural quality of these samples. Diffraction patterns (FIGS. **7***a* and **7***b*) show a much higher arcing of the diffraction spots from the samples grown on un-implanted Si, suggesting smaller grains and misorientation between them (FIG. **7***a*). This arcing is practically negligible in the samples grown on implanted substrates and single crystal pattern is observed (FIG. **7***b*). This substantial improvement in the structural quality of the GaN layer is confirmed by images taken from these layers grown on the implanted and un-implanted substrates (FIGS. **8** and **9**). The size of the grains for the layers grown on implanted Si (FIG. **8**) was many times larger (with diameter larger than 1000 nm) than for the growth on un-implanted Si (FIG. **9**). The size of the GaN grains for the samples grown on un-implanted Si was in the range of 40-150 nm. A large misorientation between these grains was also observed.

To assess the structural quality of the larger areas of these samples 0002 x-ray rocking curves were measured on the GaN samples grown on Si implanted with 15 keV and 30 keV with the fluence of 1×10^{16} cm^{-2}. For comparison the measurements were also performed on the GaN layer grown on un-implanted Si, where Si surface cleaning II and growth procedure was identical to the previous two samples. The intensity of x-ray rocking curves for both samples grown on implanted Si is 4 times higher and FWHM (full with at half maximum) is about 3 times narrower in comparison to the samples grown on un-implanted Si. One can notice that the intensity of the x-ray rocking curve obtained for the sample implanted with 15 keV is still slightly higher in comparison