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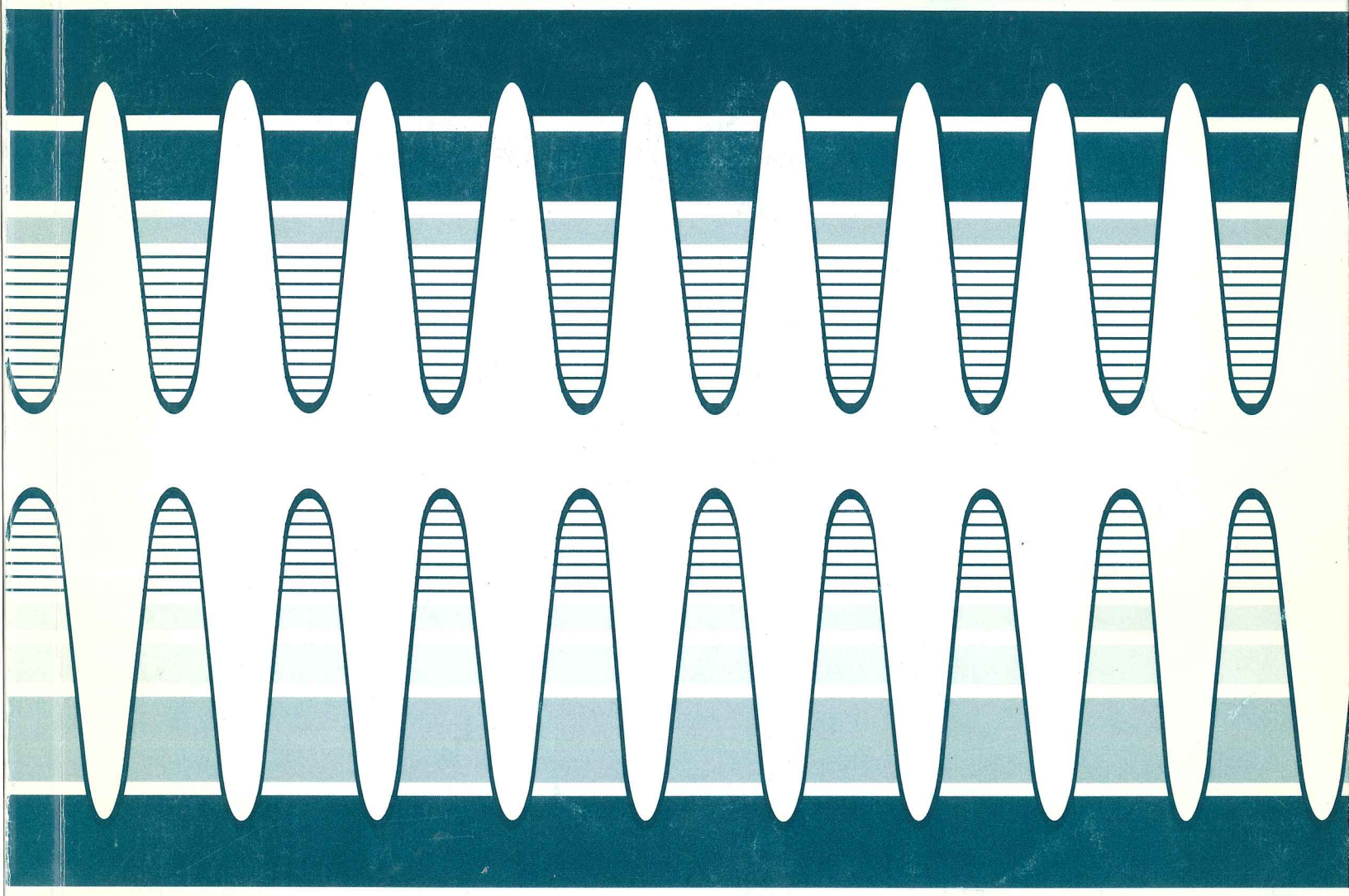

1995 ELECTRONIC MATERIALS CONFERENCE

Technical Program with Abstracts

recombination enhanced motion results in their spreading throughout the active region, killing the device luminescence. Since they originate from pre-existing defects, we believe that eliminating the "grown in" stacking faults and threading dislocations will result in arresting the device degradation.

4:10PM, O6

Defect Structure in ZnSe Based II-VI Semiconductor Blue-green Laser Structures and Their Degradation: G.C. Hua, D.C. Grillo, M.D. Ringle, J. Han and R.L. Gunshor, School of Electrical Engineering, Purdue University, West Lafayette, Indiana 47907; M. Hovinen[a)] and A.V. Nurmikko, Division of Engineering and Department of Physics, Brown University, Providence, RI 02912; N. Otsuka, Department of Materials Science, Japan Advanced Institute of Science and Technology, Hokuriku, Nomigun, Ishikawa 923-12, Japan; a) Now at Tampere University of Technology, P.O. Box 589, Tampere, Finland

Pre-existing crystal defects have been reported to be the major limiting factor in the development of current ZnSe based II-VI semiconductor blue-green light-emitting devices. Using transmission electron microscopy (TEM), we have studied the defect structure in ZnCdSe/ZnSSe/ZnMgSSe laser structures and in ZnMgSSe epilayers grown on GaAs substrates with GaAs buffer layer. Paired stacking faults have been found in these films with a density typically in the 10^5 cm^{-2} range. TEM stereomicroscopy indicated that the paired stacking faults are nucleated near (or at) the II-VI/GaAs interface. The nucleation level of stacking faults was further investigated by examining wedge-shaped plan-view TEM samples prepared from those structures. Our results show that stacking faults are not necessarily all nucleated right at the II-VI/GaAs interface, but are nucleated within a certain thickness (up to 100 nm) above and including the II-VI/GaAs heterovalent interface. Nucleation of stacking faults is extrinsic and likely associated with the nucleation and/or initial growth stage of the II-VI layers. The TEM study of degraded laser diodes has shown that paired stacking faults create threading dislocations which pass through the ZnCdSe quantum well region. During laser operation, the threading dislocations become the "seed" of nonluminescent dark defects, which are patches of dislocation networks developed at the quantum well region.

A single triangular type of defect has been observed in ZnSSe epilayers with a thickness of about 1 m grown on GaAs. This type of defect has been found to be extremely unstable under TEM electron beam excitation.

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4:30PM, O7

Structural Study of Degraded ZnMgSSe Blue Light Emitters: K. Nakano, S. Tomiya, M. Ukita, H. Yoshida, S. Itoh, E. Morita, M. Ikeda, and A. Ishibashi, Sony Corporation Research Center, 174, Fujitsuka-cho, Hodogaya-ku, Yokohama 240, Japan

Using Electroluminescence (EL) topography and transmission electron microscopy (TEM), we investigated the nonluminescent regions which form while current is being injected to ZnMgSSe/ZnSSe/ZnCdSe-based blue light emitters. Small dark spots were observed just after turn-on and spread out forming a rough triangle in the <100> directions in the EL image of the active region. TEM studies showed that the small dark spots are pre-existing stacking faults originating at the substrate/epitaxial layer interface. The non-luminescence triangles were found to be associated with a high density of dislocation dipoles and dislocation loops. Each dipole was aligned along two <100> directions in the {111} planes. The Burgers vectors were of the type a/2<101> 45 inclined to the (001) junction plane.

This result shows that the Burgers vectors are the same as that of dislocation networks in degraded AlGaAs light emitters. The orientation of the dipoles is different, however, from that in III-V devices. This suggests that the mechanism of defect formation in II-VI optical devices are different from that in III-V devices. The defect growth mechanism in degraded blue emitting devices will be discussed.

4:50PM, O8

Observation of [100] and [010] Dark Line Defects in Optically Degraded ZnSSe-based LEDs by Transmission Electron Microscopy: L. Salamanca-Riba, and L.H. Kuo, Materials and Nuclear Engineering Department, University of Maryland, College Park, MD 20742-2115

We have used transmission electron microscopy to study the [100] and [010] dark line defects (DLDs) produced after photodegradation of a ZnSSe-based/GaAs heterostructure. Our results show that the DLDs are networks of elongated dislocation loops or half loops that originate in the quantum well region during device operation. The loops start as very small dislocation loops that form along a trace left by a mobile defect, the so-called "bug", observed during photodegradation. (G.M. Haugen, et al., submitted to Appl. Phys. Lett.) Upon further illumination the dislocation loops elongate and become dislocation half loops when they reach the film surface. In this fashion, networks of dislocations form primarily along the [100] and [010] directions.

In addition to DLDs the degraded regions show that the Frank-type stacking faults become tangles of dislocations. In contrast, the Shockley-type stacking faults remain unchanged for the photodegradation conditions studied indicating that they are more resistant to photodegradation than the Frank-type stacking faults. We believe that the Frank-type faults are sources of the degradation and generation of DLDs. The bugs, which are very small clusters of vacancies, are emitted from the Frank-type stacking faults. These vacancies produce lattice distortions as they move through the lattice. Along these distortions dislocation loops can easily form upon further irradiation.

Thursday, June 22, 1995, PM

Session P: Growth and Characterization of GeSi and GeSiC Alloys

Session Chairman: Ya-Hong Xie, AT&T Bell Laboratories, Room 1E342, 600 Mountain Ave, Murray Hill, NJ 07974
Co-Chairman: Jonathan P. Pelz, Department of Physics, the Ohio-State University, Columbus, OH 43210

1:30PM, P1+

Band-Edge Photoluminescence in Pseudomorphically Strained $Si_{1-x-y}Ge_xC_y$ Layers on Si (100) Substrates: A. St. Amour[1], J. C. Sturm[2], K. Brunner[3], and J. Weber[3], [1]Dept. of Electrical Engineering, Princeton University, Princeton, NJ 08544, [2]Institut fur Halbleitertechnik, University of Stuttgart, Stuttgart, Germany (present address: Dept. of Electrical Engineering, Princeton University, Princeton, NJ 08544), [3]Max-Planck Institute for Solid-State Physics, Stuttgart, Germany

add: L. Lefroix, M.L.W. Thewalt

Recently there has been strong interest in $Si_{1-x-y}Ge_xC_y$ alloy layers for extending the $Si_{1-x}Ge_x$/Si heterojunction system. To date, several groups have well established that the addition of substitutional carbon up to 2% can be accomplished without SiC precipitation, and that one C atom can compensate for the macroscopic strain induced by 8 Ge atoms. However, there has been only one previous report of photoluminescence and the bandgap in this material [1]. In that work, defect lines were also present under conditions for which band-edge luminescence was observed. In this work, band-edge luminescence without dislocation luminescence is reported, and the dependence of the bandgap on C concentration is established.

The SiGeC layers were grown at 625∞C by rapid thermal chemical vapor deposition using dichlorosilane, germane, and methylsilane in a hydrogen carrier. They were between 30 and 50 nm thick, and were capped with a thin Si layer. The carbon concentration was measured by the shift of the SiGeC peak in X-ray rocking curves towards the Si substrate peak as C was added. The shift was found to be linear with the methylsilane flow. Assuming that one carbon atom compensated the strain of 8 germanium atoms, and assuming that the methylsilane flow did not alter the germanium fraction in the samples, which was measured by X-ray diffraction to be 18% in the control sample without C, C fractions of 0.1%, 0.4%, and 0.7% were estimated for the three samples.

Photoluminescence measurements were performed with an Ar^+ pump

with the samples immersed in liquid nitrogen. At high pump power densities, band edge luminescence was seen in the 0.1% and 0.4% C samples similar to that seen in the strained control $Si_{1-x}Ge_x$ layer (a no-phonon (NP) peak and TO phonon replica). The dependence of the bandgap, as measured by the position of the no-phonon peak, was surprisingly small (05 meV/% C), less than in the previous work with dislocation lines present. Since adding C to strained SiGe layers also reduces the biaxial strain, which would be expected to increase the bandgap, this suggests that the addition of the initial amounts of C actually reduces the bandgap of fully relaxed alloys. This would lead to a substantial downward bowing of the bandgap vs. lattice constant curve for relaxed layers, as is also commonly observed in III-V's when there is a big lattice mismatch between the constituents.

J.C.Sturm and A. St. Amour were supported by the von Humboldt Stiftung, ONR, and USAF Rome Lab, and NSF.

1. P.Boucaud et al, Appl. Phys. Lett., **64**, 875 (1994).

1:50PM, P2+

Optical Properties of $Ge_{1-y}C_y$ Alloys: B.Orner, A.Khan, D.Hits, F.Chen, K.Roe, J.Pickett, X.Shao, R.G.Wilson*, P.Berger, and J.Kolodzey, Electrical Engineering Dept., 140 Evans Hall, University of Delaware, Newark, DE 19716; *Hughes Research Lab, 3011 Malibu Canyon Road, Malibu, CA 90265

Recently, alloys of the group IV elements have come under investigation for heterostructure devices compatible with Si circuit technology. Due to the immiscibility of C, these alloys are metastable but can be synthesized by non-equilibrium growth techniques such as molecular beam epitaxy (MBE). No stable compounds of Ge and C are known, but $Ge_{1-y}C_y$ random alloys are interesting because a prediction based on the linear interpolation of the , X, and L conduction band minima yields an energy gap that is theoretically direct in k-space for C atomic fractions in the range 0.03y0.1. While we do not speculate here on the actual band structure of $Ge_{1-y}C_y$, we stress the importance of measuring its optical properties which are not yet well understood. If sufficient C can be incorporated to reduce the lattice constant to the value a=4.54 Å, $Ge_{1-y}C_y$ may become useful as a lattice-matched substrate for cubic GaN, an important material for blue light emitting diodes.

We have grown $Ge_{1-y}C_y$ alloys by MBE on (100) Si substrates. The substrates were prepared by degreasing, etching and an HF dip. Growth occurred at substrate temperatures between 500 and 600°C. The Ge beam was produced by thermal evaporation from a solid source using a pyrolytic boron nitride crucible. The C beam was produced by sublimation from a resistively heated pyrolytic graphite filament. In-situ electron diffraction indicated the two-dimensional growth of crystals oriented with the substrate. Layer thicknesses ranged from 0.01 to 0.5 mm, measured by optical interference. Compositions were measured by Auger electron spectroscopy and secondary ion mass spectrometry (SIMS). C fractions up to 13 at. % were obtained. Transmission electron microscopy showed that thick layers were relaxed by misfit dislocations at the substrate interface.

Optical absorption in the near-infrared region was measured at room temperature by Fourier Transform Infrared Spectroscopy (FTIR). The absorption increased strongly with photon energy near the bandgap and indicated high quality layers. The absorption edge of thick strain-free layers shifted to higher energies with the addition of C. Thick relaxed layers do not exhibit the decrease in bandgap well known to occur for thin strained layers. From the absorption of thick relaxed layers, we obtained a band gap of 0.732 eV for $Ge_{0.95}C_{0.05}$ (SGC-30), and a band gap of 0.782 eV for $Ge_{0.87}C_{0.13}$ (SGC-31). Photoluminescence (PL) spectra were measured at a temperature of 11 K and indicated broad peaks for thick relaxed layers, and narrower peaks with resolved phonon replicas for thin strained layers. A 10 nm layer of $Ge_{0.94}C_{0.06}$ (SGC-37) showed a PL peak at 0.766 eV with a full width at half maximum of 3 meV. This peak was attributed to no-phonon exciton recombination at the band edge. In comparison, a thicker 0.3 mm layer of similar composition had a PL peak full width at half maximum of 44 meV. We will report on the growth conditions, optical absorption and bandgap energies versus sample composition.

This work was supported by grants from the AFOSR, the ONR, and from W.L.Gore & Associates.

2:10PM, P3

Deposition and Characterization of Delta-doped Layers in Silicon Using Self-saturating Adsorption of PH_3 and B_2H_6 on Si (100): A.Mahajan, R.Gupta, J.Ekerdt, A.Tasch, and S.Banerjee, Microelectronics Research Center, J.J.Pickle Research Campus, University of Texas at Austin, Austin TX 78712

As gate lengths continue to shrink into the deep sub-micron region, short channel effects become more severe and hence the doping concentration in the channel region has to be increased to unreasonable values due to scaling requirements. Introduction of delta-doped layers in the channel region provides high-mobility free carriers due to carrier spill-over effects. Thus the channel doping need not be increased by as much for scaling. These layers can also be used as punchthrough stoppers. Delta-doped layers can be used to generate triangular potential barriers which form the basis for a whole class of devices. So far, in Si, only n-type delta layers using Sb have been demonstrated by a combination of molecular beam epitaxy and solid phase epitaxy. We demonstrate the first process to deposit both n- and p-type delta layers in Si, which is compatible with any Si-hydride based chemical vapor deposition technique. In our process, the Si deposition is paused by stopping the flow of the reactant gas. Following this the dopant gas -PH_3 or B_2H_6- is introduced in the chamber in the temperature range of 400-450°C. We have shown that in this temperature range the dopant gases exhibit self-limiting adsorption on the Si (100) surface. The deposition is then restarted by initiating the flow of reactant gas to grow a Si layer on top of the delta-layer. B_2H_6 adsorption saturates at 0.5 monolayers (ML) with a saturation exposure of 4000 Langmuirs (1L=1x10^{-6} Torr-sec) for an adsorption temperature of 400°C. At 450°C no saturation was observed, instead a B film continues to grow with increasing exposure at a rate of about 1 ML/2000 L exposure. PH_3 adsorption self-limits at 0.1 ML and 1 ML at 450°C and 500°C, respectively. Very little contamination of C and O (<0.1 ML) was observed during the process. The amount of B or P deposited was ascertained using Auger electron spectroscopy (AES) for test layers deposited on the surface, and by secondary ion mass spectroscopy (SIMS) for test layers sandwiched between nominally undoped layers. To estimate the amount of active carriers in the layers we performed Hall measurements on these layers. A high-dose (5x10^{15} cm^{-2}) low energy (30 keV) shallow dopant ion-implant was used to contact the buried delta-doped layers for this purpose. Further we also performed high-frequency C-V measurements to estimate the profiles and carrier concentrations of the carriers. A thin in-situ highly-doped layer of appropriate type was deposited on the sample which contained the buried delta doped layer to form a p-n junction which could be reverse biased for the C-V measurements.

2:30PM, P4

The Effect of Post-Growth Cooling Rate on the Defect Structure in MBE-Grown Buried Layers of $Si_{1-x}Ge_x$ on Si Substrates: M.Fatemi, P.E.Thompson, and M.E.Twigg, Electronics Science and Technology Division, US Naval Research Laboratory, Washington, DC 20375, and J.Chaudhuri, Department of Mechanical Engineering, The Wichita State University, Wichita KS 67208

The thermal stability of buried layers of $Si_{1-x}Ge_x$ depends on the rate with which the sample is cooled after deposition. This finding is based on an extensive series of observations using the high resolution x-ray rocking-curve technique. SiGe layers 200 nm thick, capped with a 50-nm layer of Si and with x(Ge) of approximately 17, 14, and 9 % (at or below the critical thickness) were grown at temperatures of 550, 700, and 800°C. After growth, the samples were subjected to one of two distinct types of cool-down. In the "fast" mode, the heater power was abruptly turned off, whereas in the "slow" mode, the power was reduced in equal steps of 2% of the maximum input power, and the sample reached the equilibrium temperature in about 10 minutes before the power was lowered again. We obtained x-ray rocking curves for the symmetric 004, the glancing-incidence 224, and the glancing-exit 224 reflections, and calculated the strain and composition in each sample for various growth and cool-down conditions. We also examined the breadths of the rocking curves for possible correlation with the degree of strain in each sample. In general, calculated strains showed only a slight reduction (5%) with the increase in the growth temperature, and