Epitaxial Growth-Nitride

Sublimation Growth of Bulk AlN Crystals: Materials Compatibility and Crystal Quality
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Crystal Growth of Aluminum Nitride by Sublimation Close Space Technique
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Heteroepitaxial Growth of Insulating AlN on 6H-SiC by MBE
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Structural and Electronic Characterisation of Heteroepitaxially Grown AlN on Si(111) Using Surface-Reconstruction Induced Epitaxy
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RHEED Studies of In Effect on the N-polarity GaN Surface Kinetics Modulation in Plasma-assisted Molecular Beam Epitaxy
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The Growth of (SiC)x (AlN)12x Epitaxial Thin Films on 6H-SiC, by Ion-assisted Dual Magnetron Sputter Deposition
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Growth and Characterization of GaGdN and AlGdN on SiC by RF-MBE
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Heteroepitaxial Growth of Defect-Free 3C-SiC on (Late News) Step-Free Hexagonal (0001) SiC Mesas
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Sublimation growth of bulk AlN crystals: materials compatibility and crystal quality

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Perspectives of bulk AlN crystal growth are attracting much interest, as AlN wafers would be the nearly ideal substrate for nitride based electronic and optoelectronic devices having small lattice mismatch with GaN, similar coefficient of thermal expansion and high resistivity. Since the pioneering work of Slack and McNelly [1] the problem of chemical compatibility of materials used in the growth cell was recognized as the key issue in bulk AlN growth. To date various crucible materials such as graphite and SiC coated graphite, nitrides (TiN, Ta_2N) and high-melting metals (W, Re and W-Re alloys) have been employed, but AlN samples are polycrystalline, still limited in size and contain much impurities (the latter is evidenced by reported crystal coloration). In this work crystal growth results obtained with the use of different materials in the reactor system are discussed in terms of process stability and crystal quality.

Crystal growth experiments were conducted in a restively heated furnace using graphite or tungsten heating elements in the temperature range 1900-2200°C. The source material was AlN powder (99%, Chempur, Germany, main residual impurity - oxygen). Seed plates 10x10 mm^2 were cut from on axis 6H-SiC crystals grown in our laboratory. The growth process was performed in an atmosphere of pure nitrogen and N_2+H_2 mixtures under the pressures ranging from 30 to 500 mbar. The following material combinations were tested:

N	Crucible	Heater	Isolation
1	Dense graphite	Graphite	Porous graphite
1a	SiC coated graphite	Graphite	Porous graphite
2	Tungsten	Tungsten	Tungsten or/and
2a	Tungsten	Tungsten	AlN and Al_2O_3 ceramics
3	Tungsten	Graphite	Porous graphite

First two arrangements 1 and 1a represent 'standard' high-temperature material combinations: (1): The main challenges with a 'pure graphite ambient' is crucible permeability for aluminum vapor and seed surface graphitization at T>2000°C. However carbon contamination of the crystal can be kept at a relatively low level, if oxygen is carefully removed from the system

prior to growth. In the initial process stage the carbothermic reduction of aluminum oxide film according to the reaction $Al_2O_3+3C+N_2=2AlN+3CO$ is very useful for preparation of an oxygen-free AlN powder charge.
(1a): Contrary to the results of Balkas et al. [2] SiC coating of graphite crucible was found to be unstable at least at T>2000°C. According to our estimations, at elevated temperatures, vapor pressures of SiC species are only slightly lower, than that of AlN, leading to growth of mixed AlN-SiC crystals. AlN-SiC alloys are interesting as substrates for MBE growth of AlN, and also exhibit useful semiconductor properties. However, the combination of 1a can not be considered as a promising way to grow semiconductor grade AlN.

A 'pure tungsten ambient' (2 and 2a) includes two other challenges. First of all W can be strongly attacked by aluminum vapor not only at high-temperatures, but also at a temperatures below 1000°C by aluminum melt droplets, if partial decomposition of AlN occurs during pre-heating of the furnace in vacuum. The life time of tungsten heaters is relatively short. We have found that W-heater stability can be sufficiently improved by a precise furnace design, but the need in further corrections undoubtedly remains. Secondly, in a tungsten furnace it is very difficult to get rid of inevitable oxygen contamination of AlN charge. Purging with N_2+H_2 mixture at 1000-1200°C is really effective only for day-long reaction times. Seeded growth is often complicated by formation of oxynitrides in the initial stage of growth.
The combination of a W-crucible and a graphite heater (3) offers important advantages, such as long heater life time, effective use of heating power and flexible cell design. A critical issue arising here is the probability of crucible damage because of WC formation on the outer surface. However it was found, that in the absence of direct contact of tungsten with graphite parts, the crucible erosion is mainly determined by the action of AlN and not by carbon. Formation of oxynitrides was not observed.

Finally, results on morphology and purity of the grown AlN crystals and layers are presented and discussed. The growth of AlN-SiC mixed crystals is described in terms of process conditions and crystallization parameters.

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Crystal growth of aluminum nitride by sublimation close space technique

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Introduction

Silicon carbide (SiC) and III-nitride like gallium nitride (GaN) and aluminum nitride (AlN) is the most promising semiconductor materials with applications to high frequency, power and temperature devises, because these semiconductors have a wide gap, a high carrier mobility, etc. In order to create such devises, high quality semiconductor substrates and layers are needed. Besides, there are many polytypes in SiC and it is known that 2H-SiC has the widest gap in SiC polytypes. In order to obtain 2H-SiC crystal, substrates that have 2H polytype and a small lattice miss match are needed. 2H-AlN is the most suitable substrate in order to grow 2H-SiC. In this background, we tried to grow AlN layers.

Heteroepitaxial growth of 2H-AlN on 6H-SiC was carried out by sublimation close space technique (SCST) [1,2] in order to obtain high quality AlN layers with a high growth rate. SCST was used in the crystal growth of SiC epitaxial layer with a high growth rate and high quality [1,2]. Besides, in the crystal growth of III-nitride compound, ammonia (NH_3) is generally used as a source of nitrogen (N). In this study, nitrogen gas (N_2) was used as a source of N.

Experiment

Crystal growth was carried out by SCST. In order to grow AlN, aluminum carbide (Al_4C_3) powder was used as a source material of Al and crystal growth was carried out in N_2 atmosphere. The growth temperature was 1900°C and the growth pressure was 1 atm. The distance between the source and the substrate was 1.0mm. 6H-SiC (0001) Si-face and (0001) 5° off oriented toward $<11\bar{2}0>$ were used as substrates. These substrates were prepared in our laboratory. The substrate and the source material were set in a graphite crucible. In order to absorb carbon from Al_4C_3 and the crucible, tantalum (Ta) plate was used. This crucible was heated by RF-generator at a frequency of 36.2 kHz. At these conditions, the growth rate was about 10μm/h.

Result and discussion

Fig.1 shows the Raman spectrum of the grown layer. The substrate which had the off axis-angle was used and the film thickness of this layer was 10μm. Two peaks, SiC and AlN, are observed. This is due to the fact of that this layer is not enough thick. However, it is confirmative that this grown layer is AlN [3].

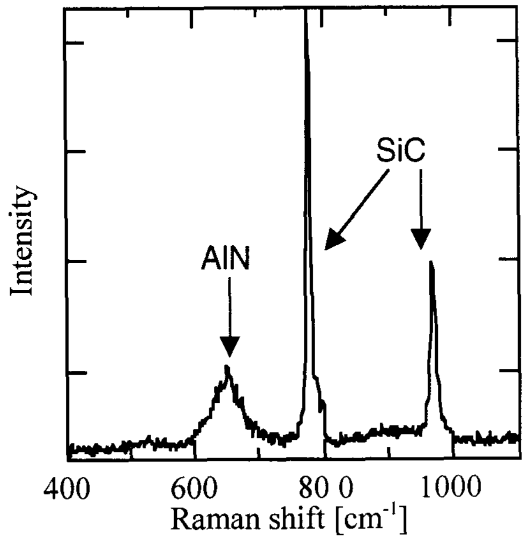

Fig. 1 Raman spectrum of grown layer

Fig. 2 shows a surface morphology of the grown layers by an optical microscope. This layer was grown on the substrate that had the off-axis angle. Because step like morphology is observed on the surface, it is considerable that crystal growth proceeds in the 2 dimensional growth, in other wards, step-flow growth [4]. When the substrate which has no off axis-angle was used, different surface morphology, hexagonal pattern, was observed. In this case, it is considerable that crystal growth proceeds in the 3 dimensional growth. Additionally, in this case, polycrystalline AlN layers were sometimes grown. Therefore, in order to obtain single crystal AlN layers, crystal growth must proceed in the 2 dimensional growth, in other wards, substrates must have the off-axis angle.

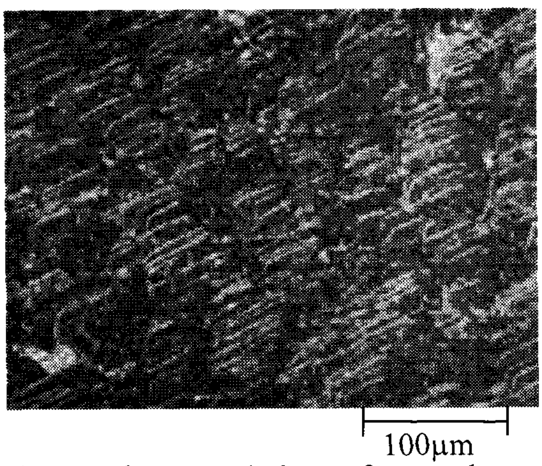

Fig.2 Surface morphology of grown layer

In this study, heteroepitaxial growth of AlN on SiC was carried out by SCST. The growth rate was only 10μm/h. In order to obtain thick layers and bulk crystal, higher growth rate must be obtained. Therefore, the growth condition must be optimized.

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Heteroepitaxial Growth of Insulating AlN on 6H-SiC by MBE

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AlN is expected as an insulator of SiC metal-insulator-semiconductor field effect transistors (MISFETs) due to its high relative dielectric constant (8.5) compared to SiO_2 (3.9). Owing to the small lattice mismatch between AlN and SiC, an AlN layer with low defect density and good structural properties can be obtained. However, there are few reports on high-performance AlN/SiC MIS devices because of large gate leakage current.[1] In order to obtain a high-quality AlN layer on a SiC substrate, a pretreatment of substrate surface is very crucial. We proposed HCl gas etching, which could remove surface polishing scratches and realize an atomically flat terrace structure,[2] as a new pretreatment method of SiC substrate for molecular beam epitaxy (MBE) of III-Ns.[3] As a result, a very flat AlN layer was obtained on an HCl gas etched SiC substrate. In this study, insulating properties of AlN on a 6H-SiC substrate was investigated by a current-voltage (I - V) measurement of Al/AlN/SiC MIS diode structures.

Substrates used in this study are commercially available n-type on-axis 6H-SiC $(0001)_{Si}$ face wafers (off angle < 0.2°). The doping level of substrates is around 1×10^{18} cm^{-3}. Substrates were pretreated by HCl gas etching in a chemical vapor deposition (CVD) system at 1300℃ for 10 min, and transferred through the air into an MBE system. AlN layers were grown at 900℃ by MBE using elemental Al and radio frequency (rf) plasma-excited active nitrogen. Al electrodes (ϕ300 μm) were formed on an AlN surface by vacuum evaporation.

Figure 1 shows insulating properties of 35 nm-thick AlN layers grown on 6H-SiC substrates and AFM images of them. The surface pretreatment of SiC substrate by HCl gas etching strongly influenced the insulating properties of AlN. The AlN layer grown on an HCl gas etched substrate had a very flat surface and exhibited excellent insulating properties. The resistivity of this AlN layer was 6.8×10^{13} $\Omega\cdot$cm. The leakage current was as small as 10^{-8} A/cm^2 below 2.5 MV/cm. This is a considerably hopeful result to apply this AlN for a gate insulator of SiC MISFETs.

Insulating properties of AlN layers grown on HCl gas etched SiC substrates were related to those layer thicknesses as shown in Fig. 2. Thinner AlN layers had more superior insulating properties. A 24.9 nm-thick AlN layer had excellent insulating properties with a small leakage current of 10^{-9} A/cm^2 and a relatively high breakdown field of 4.5 MV/cm. From the result of XRD measurements, thin AlN layers were well-oriented, while beyond 50 nm, the misorientation of AlN layer gradually increased (i.e. broad peak with the linewidth of 40 arcmin becomes dominant). It indicates strong correlation between insulating properties and structural quality. Employing higher growth temperature or control of initial stage of growth will improve structural quality of AlN layers.

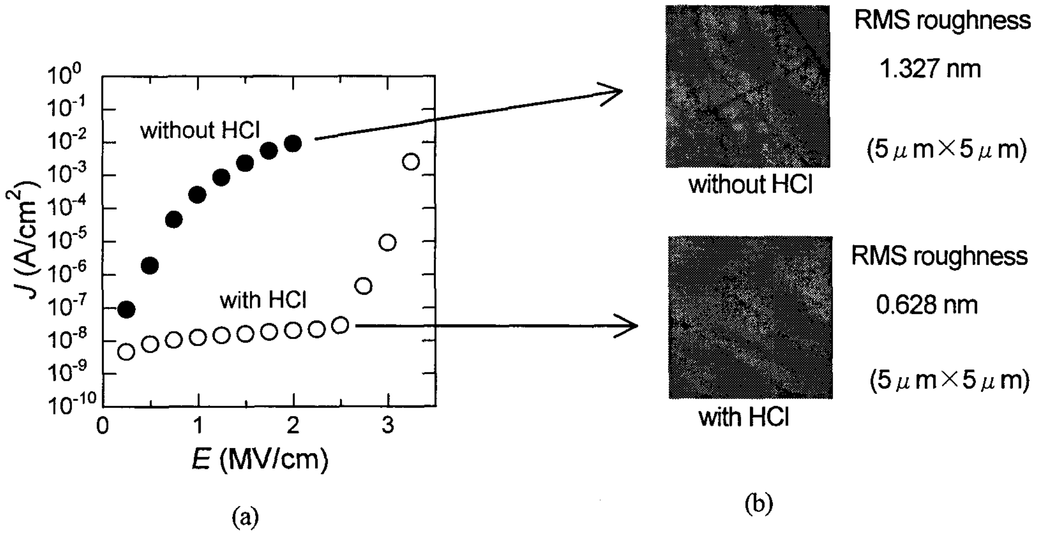

Fig. 1 (a) Insulating properties of AlN layers on 6H-SiC substrates. Thickness of AlN is around 35 nm. Open and close circles are for AlN layers with and without HCl pretreatment. (b) AFM images of those AlN layers.

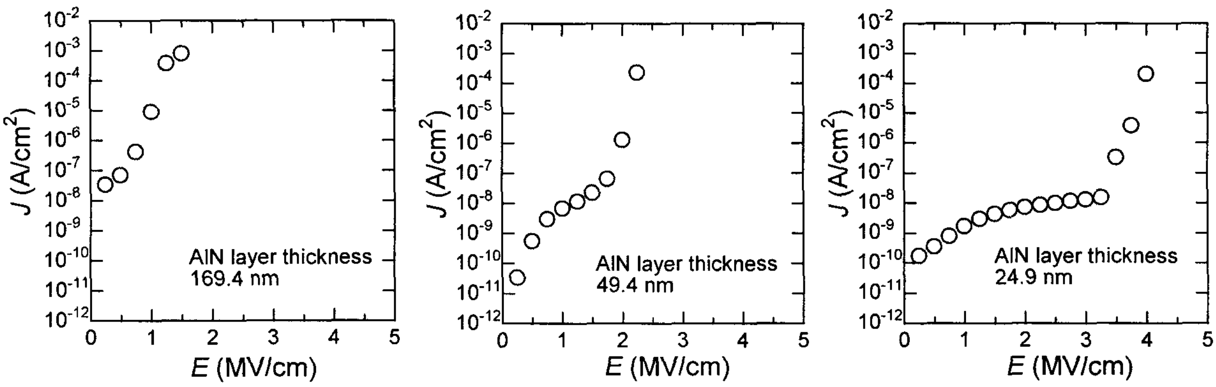

Fig. 2 Layer thickness dependence of insulating properties of AlN layers.

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Structural and Electronic Characterization of Heteroepitaxially grown AlN on Si(111) using surface-reconstruction induced epitaxy

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The multitude of desirable properties of AlN such as its wide bandgap (6.2 eV), high dielectric strength, high temperature stability, high thermal conductivity, chemical inertness, high acoustic velocity, high melting point, and piezoelectric behavior, makes this material suitable for wide range of electronic, optical and mechanical applications. Integration of AlN with Si adds many new dimensions to utilization of AlN. Examples are; integration of optoelectronic components made of direct bandgap group-III nitrides on Si (optical interconnect, LED, Lasers, waveguides), fabrication of solar blind UV and X-ray detectors, as a gate dielectric, and in MEMS. Established Si fabrication technology and the possibility of building the driver circuits on the same wafer provide another impetus for integration.

In this work, AlN was grown on Si(111) using surface reconstruction induced epitaxy. The Si(111)7x7, generated after thermal cleaning of Si, was converted to aluminum induced Si(111)√3x√3 by depositing ~0.3 ML of Al on the Si(111)7x7 surface at temperatures between 650 to 700 °C. In the √3x√3 surface configuration, Al passivates all surface Si atoms, minimizing possible interaction between the Si and the overlayer. In this case, it prevents formation of amorphous Si nitride prior to growth of AlN. In addition, the √3x√3 provides the proper template for hexagonal (001) or cubic AlN growth. The growth was then conducted using thermal Al evaporation from an effusion cell and atomic nitrogen beam from an RF atomic source. Figure 1 shows X-ray

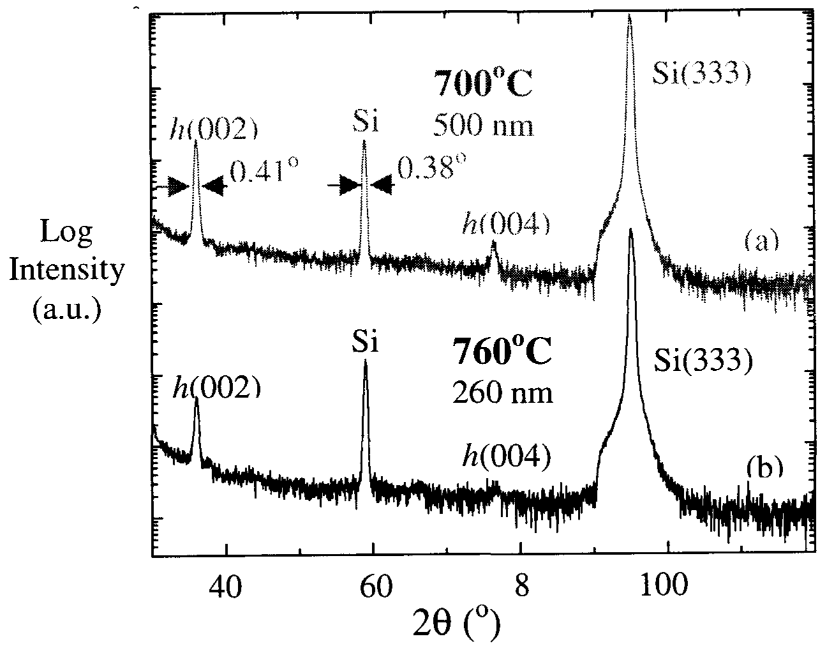

Figure 1. X-ray diffraction from hexagonal AlN(001) grown on Si(111). The FWHM measured from the layer reflection is close to that of the substrate

diffraction from samples gown at 700 and 760 °C. Except for peaks from hexagonal AlN(001) and the Si substrate no measurable intensities can be obtained from other phases or orientations. Moreover, the full width at half maximum (FWHM), measured from the layer peak, is almost equal to that of the substrate (i.e. the resolution of diffractometer) indicating highly oriented AlN layer. Epitaxial growth was achieved over a wide range of Al/N fluxes and growth temperatures extending from ~ 350 to 850 °C. As the growth temperature was lowered and the N/Al flux ratio was increased, a second peak related to cubic AlN(001) became evident indicating growth of a thin interfacial layer of cubic AlN. Figure 2 shows example of an x-ray reciprocal space mapping from an AlN layer grown at 400 °C. The elongation of the cubic peak indicates that the layer is very thin while the FWHM is equal to the resolution of the diffractometer implying a highly ordered epitaxial layer.

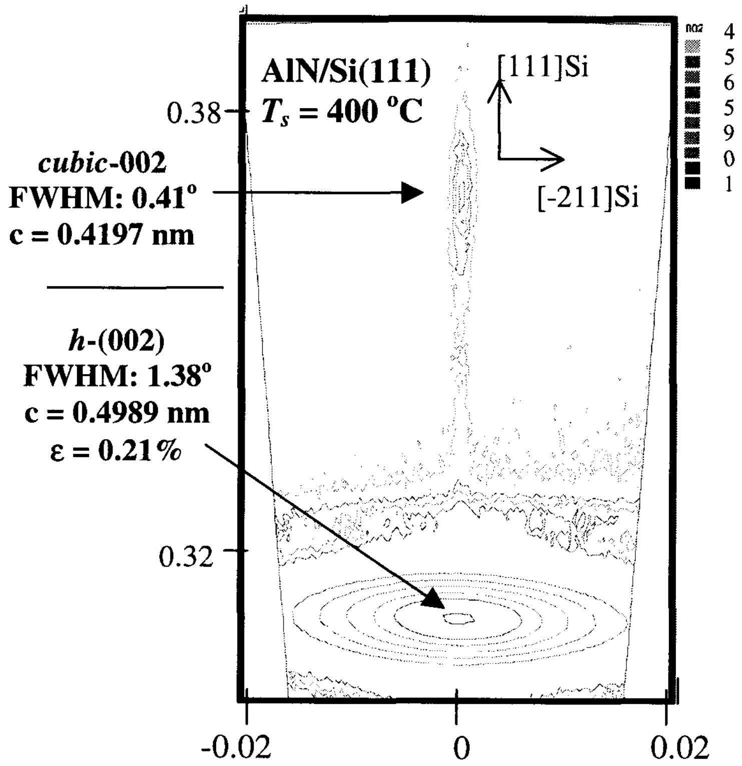

Figure 2. Reciprocal space mapping from AlN layer grown at 400 °C. A strong hexagonal AlN is evident in the map. In addition, a weak, highly oriented cubic-AlN(001) is also present.

Finally, an AlN/Si heterojunction diode was fabricated and tested. A breakdown voltage in excess of 350V was obtained and a leakage current below 100 nA was measured indicating a high quality interface (see Fig. 3).

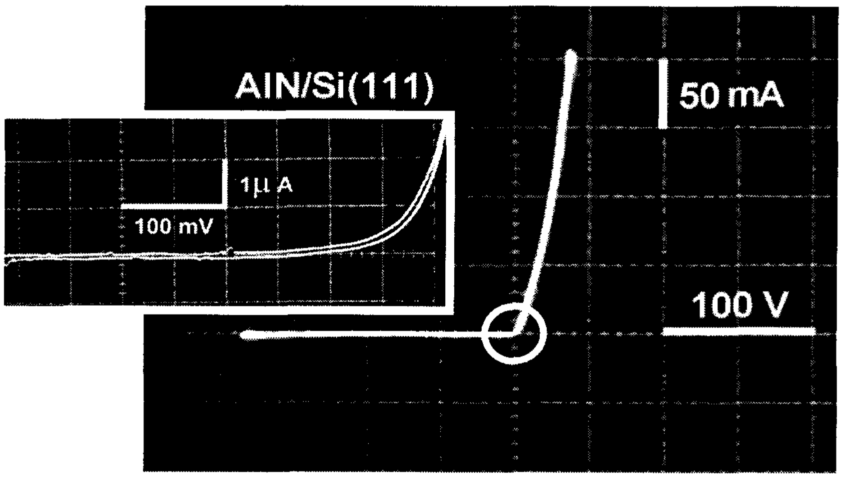

Figure 3. I-V curve from a heterojunction AlN/Si(111) diode showing a breakdown voltage of > 350 V and a leakage current below 100 nA. The breakdown is not shown in the figure.

Acknowledgment: The authors gratefully acknowledge the financial support of the Office of Navel Research (ONR) administered by Dr. Colin Wood and the Defense Advanced Research Project Agency (DARPA) administered by Dr. Daniel Radack. Support from the U.S. Department of Energy (grant DEFG02-96-ER45439) is also appreciated. Special thanks to Mr. John Hudak for an outstanding technical support.

RHEED studies of In effect on the N-polarity GaN surface kinetics modulation in plasma-assisted molecular-beam epitaxy

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III-nitride materials have attracted a great deal of attention due to its potential applications in optical and electronic devices. Recently, positive In effects on the film quality of GaN have been reported by several groups.[1-3] Under the different growth techniques and growth sources, In plays different roles according to the reports. In cases of MOCVD and NH_3-MBE, In works as a doping-effect to improve the GaN film quality. In case of rf-MBE, surfactant effect of In was found to modify the growth kinetics. In addition, lattice-polarity control of III-nitride epitaxial films recently becomes a hot topic due to its great influence on the optical and electrical properties of the films.[4] We found that the stability of GaN surface with Ga-polarity is much superior to that of GaN surface with N-polarity during the interruption of the growth at high temperature in rf-MBE.[5] This phenomenon is especially important to those who need to interrupt the growth to change the growth conditions, such as for the InGaN growth.

In this paper, we reported the In effect on the N-polarity GaN surface kinetics modulation during the interruption of GaN growth under the nitrogen flux in rf-MBE studied by RHEED. The GaN films with N-polarities were grown on sapphire (0001) substrates rf-MBE. Detail of the N-polarity GaN films preparations and the clarifications of the lattice-polarity has been published elsewhere.[6] During the GaN growth, the N_2-plasma power was 350 W and the N_2 flow rate was 5.0 sccm. The growth temperature was fixed at 700°C and the growth rate was 0.6 μm/hr. In-situ RHEED observations along the [11-20] azimuth during the growth and the growth interruption were carried out and the RHEED images were recorded using a CCD camera. The intensity change of specific RHEED spot/streak was measured from the recorded data using an image processor with a computer.

As a summary, we found that In does have significant effects on the N-polarity GaN surface kinetics modulation in rf-MBE. We proposed a model to explain the phenomena. Based on our model, There are two effects of In on the surface kinetics modulation. First, surfactant effect as reported was observed which made the rough surface (spotty RHEED pattern) change to smooth surface (streaky RHEED pattern). This effect was also be confirmed by the SEM observation of surface morphologies as shown in Fig. 1. Second, In can suppress the evaporation of Ga atoms from the grown surface, and reduce the reaction between Ga and N, which are the main factors resulting in the surface roughening during the growth interruption. Details will be presented at the conference.

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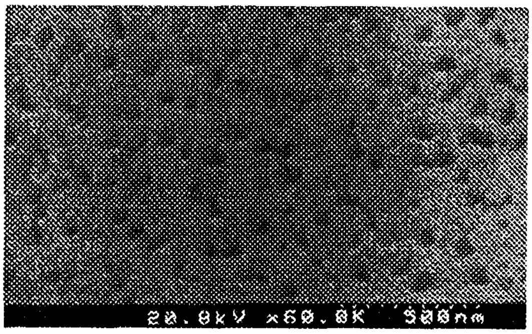
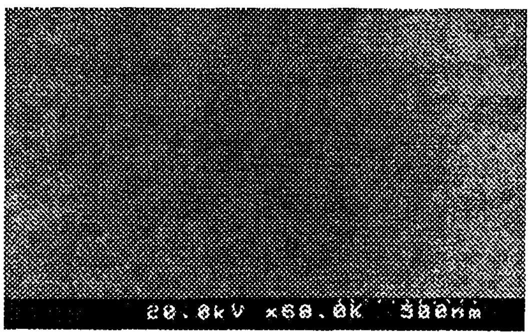


(a)



(b)

Figure 1. SEM observations of the GaN surface morphologies (a) growth interruption without the In exposure, (b) growth interruption with the In exposure

The Growth of $(SiC)_x(AlN)_{1-x}$ Epitaxial Thin Films on 6H-SiC, by Ion-assisted Dual Magnetron Sputter Deposition

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ABSTRACT

The wide-band gap semiconductor alloy of aluminum nitride (AlN) and silicon carbide (SiC), has become a very interesting material for high-power and optical electronic applications. This alloy can eventually be used for band gap engineering by controlling the ratio of AlN:SiC, such that the band gap can be tailored between 3.0 and 6.2 eV, the band gap values of SiC and AlN, respectively. Attempts to grow $(SiC)_x(AlN)_{1-x}$ alloys or solid solutions have been made for several years by different growth techniques.[1,2] However, the quality of the material still need to be enhanced before it can be used in microelectronic applications as a semiconductor.

In this work, we demonstrate that ion-assisted dual magnetron sputter deposition in ultra-high-vacuum environment is an alternative route to grow a high quality of this material. $(SiC)_x(AlN)_{1-x}$ thin film have been grown on vicinal (3.5°) 6H-SiC substrates at the growth temperature of 1000°C. The sputtering was carried out at the total pressure of 10 mTorr of a pure (99.999999%) gas mixture between nitrogen and argon. An elemental Al-disc and a polycrystalline stochiometric SiC-disc were used as targets for the magnetrons. The alloy films were grown by co-sputtering from both targets. The composition ratio of the alloy was controlled by varying the power of each magnetron.

In-situ reflection high-energy electron diffraction (RHEED) was used to monitor the films during the growth. The cross-sectional microstructure of the alloy films and interfaces were investigated by high-resolution transmission electron microscopy (HREM). As seen in figure 1, the high resolution cross-sectional electron micrograph of the $(SiC)_x(AlN)_{1-x}$ thin film/ SiC substrate interface shows an epitaxial film, which has a hexagonal phase since the beginning of the nucleation. The film has a dense structure with a large domain size, with a domain width of about 30 nm at the base. The quality of the alloy thin films is also reflected in the opto-electronic properties. The preliminary cathodoluminescence (CL) spectrum from one of these semiconductor alloys shows promising results for the band gap tailoring of this alloy with an emission at the energy of 3.40 eV, as illustrated in figure 2. Auger electron spectroscopy (AES) has been used to determine the composition of the alloy thin film for each different growth condition, as well as, the atomic force microscopy (AFM) for characterizing the surface morphology of the films. A detailed discussion about the growth of this alloy will be presented.

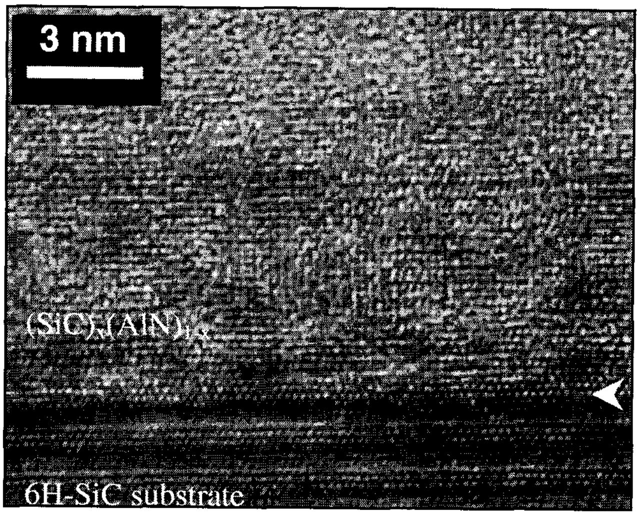

Fig. 1 High resolution electron micrograph of $(SiC)_x(AlN)_{1-x}$ alloy thin film on 6H-SiC, grown by ion-assisted dual magnetron sputter deposition, the arrow indicates the interface between film and substrate.

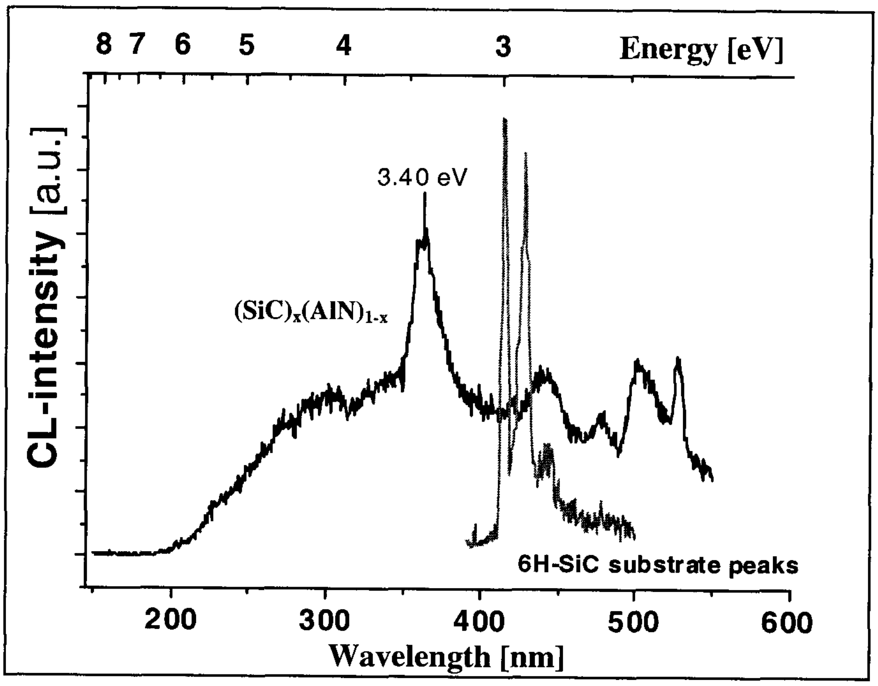

Fig.2 CL spectrum from $(SiC)_x(AlN)_{1-x}$ on 6H-SiC at 5 K, the gray line represented the CL spectrum of 6H-SiC substrate

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Growth and Characterization of GaGdN and AlGdN on SiC by RF-MBE

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In the development of superior electronic devices, the existence of lattice-matched materials and lattice-matched substrates is the most important problem. For example, in the case of GaAs substrate, the lattice-matched AlGaAs or InGaP gives high quality epitaxial films with dislocation densities less than $10^4 cm^{-2}$. For the growth of II-VI compounds on GaAs substrate, the discovery of a lattice-matched ZnMgSSe system improves film quality and the lifetime of the laser diodes [1]. On the other hand, there is no lattice-matched AlGaInN to SiC substrates. The III-nitrides lattice-matched to SiC substrates can be formed by using boron, but it is difficult to use boron in molecular beam epitaxy (MBE) growth. In this report, we will discuss another possible candidate to form a new nitride system. By the addition of Gd to GaN and AlN, GaGdN and AlGdN alloy semiconductors have been obtained, respectively, for the first time.

The films are grown on (0001) SiC substrates by MBE using RF plasma-excited nitrogen. The sample structures are as follows: GaGdN(250nm)/GaN (250nm)/AlN(20nm)/SiC and AlGdN (120nm)/AlN(180nm)/SiC.

After the growth of GaGdN, the RHEED pattern shows a 4x4 reconstruction pattern with some unusual diffraction spots, which may relate to the segregation of Gd atoms but the segregation is not observed in SEM surface observations.

Fig.1 shows the X-ray diffraction pattern for GaGdN. In addition to the diffraction from the SiC substrate and the GaN epitaxial layer, another diffraction peak caused by the GaGdN mixture is observed. The strained lattice constants for the a-axis and the c-axis of GaGdN film estimated from the (11-24) asymmetric XRD pattern are 3.20A and 5.21A, respectively. From the XPS measurement (Fig.2), the composition of Gd and Ga are estimated to be 0.06 and 0.94, which results in the film composition of $Ga_{0.94}Gd_{0.06}N$. From these results, the strained a-axis and c-axis lattice constants for wurtzite structure of GdN are evaluated to be 3.48A and 5.49A, respectively. The crystal structure of GdN is usually a rock-salt structure with a lattice constant of 4.999A [2]. Lattice constants of wurtzite GdN are estimated for the first time.

In the cathodoluminescence (CL) spectrum measured at room temperature (R.T.), band-edge, deep level and Gd^{3+} related emissions are observed, which is shown in Fig.3. The wavelength of the band emission is 370nm and slightly longer than that of GaN (363nm), which means that bandgap shrinkage by incorporating Gd atoms occurs. The emission peak around 640nm is considered to relate to Gd^{3+} according to the analogy of GaN:Eu [3].

Fig.4 shows the R.T. CL spectrum for AlGdN film. The composition of Gd is estimated to be around 2% by XPS. There are three peaks located at 312, 317, and 322nm and the center peak is dominant. These peaks are probably due to Gd^{3+} but further investigations are needed. A similar spectrum is obtained for AlGdN films with 13% Gd composition but with extra emissions from 350 to 600nm.

References

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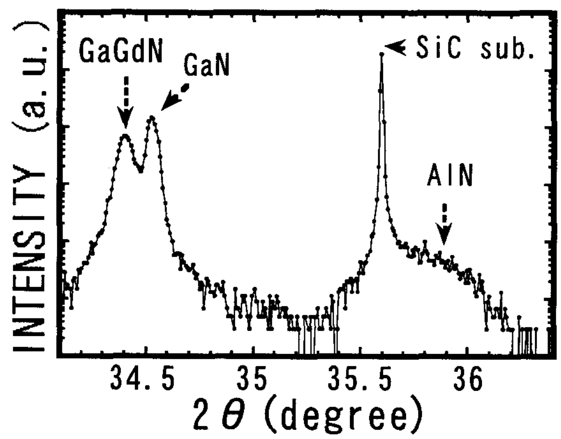

Fig.1. XRD pattern for GaGdN film.

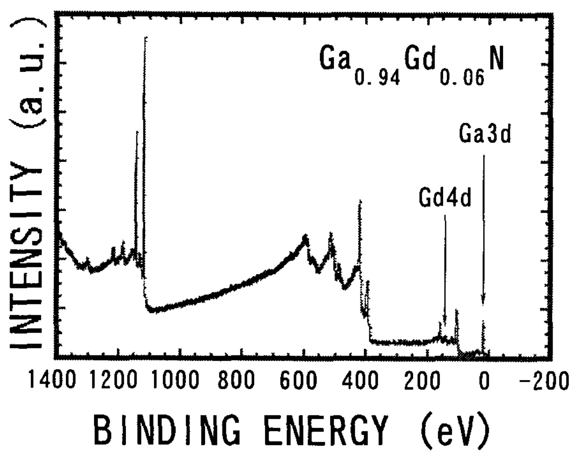

Fig.2. XPS spectrum for GaGdN film.

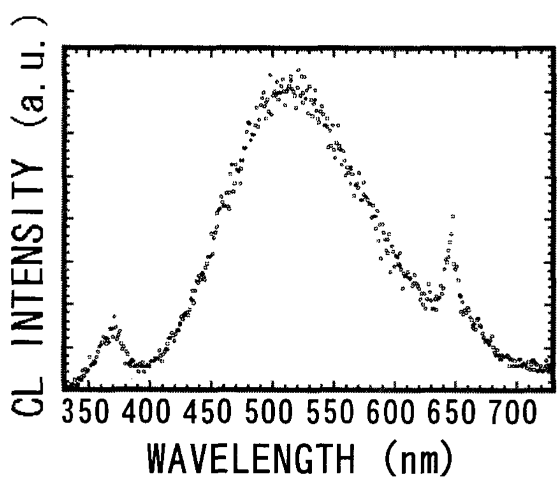

Fig.3. Room temperature CL spectrum for $Ga_{0.94}Gd_{0.06}N$ film.

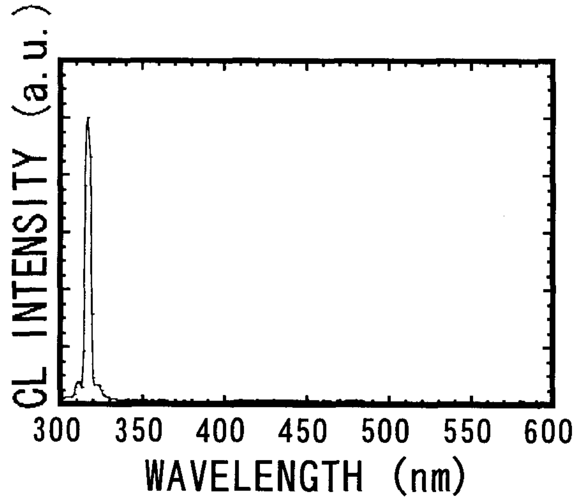

Fig.4. Room temperature CL spectrum for AlGdN film.

Heteroepitaxial Growth of Defect-Free 3C-SiC on Step-Free Hexagonal (0001) SiC Mesas

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Abstract: Previous efforts to grow 3C-SiC heteroepitaxial films on Si and α-SiC substrates all yielded films containing extended defects such as double-positioning boundaries (DPB's) and/or stacking faults (SF's), leading to poor electrical performance of devices fabricated in these films. The formation of SiC mesa surfaces as large 0.2 x 0.2 mm completely free of even a single atomic step was recently reported [1]. As described in [1], these surfaces are produced on 4H- or 6H-SiC wafers (on-axis) by first dry etching trench patterns into the wafer surface to form an array of isolated growth mesas. Pure stepflow epitaxial growth, carried out under conditions that suppress 2D terrace nucleation, is then used to grow all initial surface steps on top of each mesa over to the edge of the mesa, leaving behind a top mesa surface completely free of atomic steps. However as reported in [1], mesas that initially contain screw dislocation defects cannot be flattened due to the continual spiral of new growth steps that emanate from screw dislocations during epitaxial growth.

The heteroepitaxial growth of 3C-SiC films completely free of DPB's and SF's has now been achieved at NASA Glenn on step-free 4H/6H-SiC mesas. In the absence of steps that provide a template for maintaining hexagonal substrate polytype during homoepitaxial growth, a single variant of the 3C-SiC polytype can be controllably nucleated and grown without any extended crystal defects on the step-free (0001) basal plane surface. Our experiments confirm that such defect-free growth is not possible without a step-free surface, as the presence of any steps on the nucleation surface produces disorder (i.e., extended defects) in 3C-SiC heteroepitaxial films grown thereon [2]. In particular, SF and DPB defects are observed on 3C-SiC films grown on 4H-SiC substrate mesas that could <u>not</u> be rendered step-free prior to 3C nucleation because they contained substrate screw dislocations. In contrast, under conditions of initial low nucleation rate on the step-free mesas, perfect 3C-SiC films (i.e., no observed defects) were reproducibly grown on mesas up to 0.4 mm by 0.4 mm in size.

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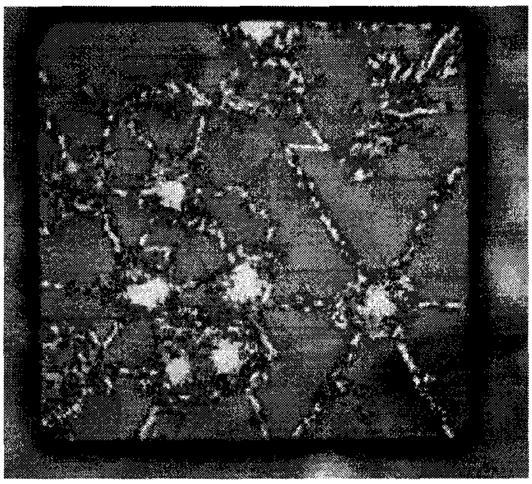

Fig. 1: Defective 3C-SiC heteroepitaxial layer with DPB's and SF's grown on a 0.2 mm x 0.2 mm 4H-SiC mesa that was not step-free because it contained screw dislocations.

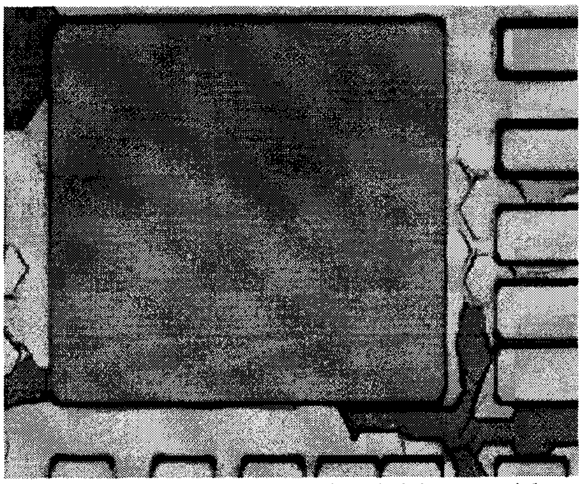

Fig. 2: 3C-SiC heteroepitaxial layer with no DPB's, and no SF's on a 0.3 mm x 0.3 mm 4H-SiC mesa. Both Fig. 1 & 2 oxidized to map polytype (dark = 3C) and defects.