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Growth of high quality 6H-SiC epitaxial films on vicinal (0001) 6H-SiC wafers

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Previously reported growth of SiC films on SiC by chemical vapor deposition (CVD) used Acheson and Lely α -SiC crystal substrates. We report the CVD growth and evaluation of high quality 6H-SiC films on 6H-SiC wafers cut from large boules grown by the modified-sublimation process. The single-crystal 6H-SiC films were grown on wafers oriented 3° to 4° off the (0001) plane toward the $\langle 11\bar{2}0 \rangle$ direction. The films, up to $12 \mu\text{m}$ thick, had surfaces that were smooth and featureless. The high quality of the films was demonstrated by optical and electron microscopy, and low-temperature photoluminescence.

Silicon-carbide semiconductor technology has been advancing rapidly during the last several years.¹ Although much work has been focused on the growth of 3C-SiC on Si substrates,²⁻⁵ the most significant recent progress has been in the development of the modified-sublimation growth process.⁶⁻⁸ Large boules of single-crystal 6H-SiC can now be grown in sufficient size to produce 2.5-cm-diam wafers. The results presented herein demonstrate that chemical vapor deposition (CVD) can be used to produce high quality single-crystal 6H-SiC films on wafers cut from these boules. Results presented include optical and electron microscopy, and low-temperature photoluminescence (LTPL).

Until recently, the only SiC substrates were small irregular-shaped α -SiC platelets that were produced in (1) the industrial Acheson process or (2) the laboratory Lely process.⁹ The growth of SiC films on SiC substrates by CVD depends very much on the temperature and substrate orientation. For many years it has been known that a temperature above 1550°C is required to achieve 6H-SiC growth on the basal plane.¹⁰ In 1973, it was found that 6H-SiC could be grown on 6H-SiC in the range 1320 – 1390°C if the growth direction was perpendicular to the $\langle 0001 \rangle$ axis.¹¹ More recently, (1) 3C-SiC films were achieved on the basal planes of 6H-SiC in the range 1350 – 1550°C ,¹²⁻¹⁴ and (2) single-crystal 6H-SiC films were achieved on 6H-SiC surfaces that were oriented several degrees off the basal plane.^{15,16} All of the above growth results were achieved on Acheson or Lely crystals.

The 6H-SiC wafers used in this letter were produced by Cree Research Inc.¹⁷ by the modified-sublimation process. The wafers were 25.4 mm in diameter, $200 \mu\text{m}$ thick, sliced 3° to 4° off from the (0001) plane toward the $\langle 11\bar{2}0 \rangle$ direction, and polished on the carbon side of the wafer. They were transparent and nearly colorless. Hall measurements at NASA Lewis showed they were n type with a resistivity of about $0.05 \Omega \text{ cm}$, net carrier concentration in the low 10^{17} cm^{-3} range, and an electron mobility around $250 \text{ cm}^2/\text{V s}$.

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The wafers were prepared for growth experiments in the following manner. Because of limited availability, the wafers were cut into pieces, yielding nine square substrates, each approximately 6.4 mm on an edge. No attempt was made to cut the edge of the squares along any particular crystal direction. Before loading the substrates into the growth chamber, they were scrubbed with a liquid detergent, rinsed with $18 \Omega \text{ cm}$ water, and etched with HF to remove the native oxide, rinsed again, then dried. The SiC growth was carried out at NASA Lewis in a horizontal CVD system from SiH_4 and C_3H_8 in H_2 carrier gas at 1 atm. This system is described in Ref. 4. The substrates were heated by a rf-heated SiC-coated graphite susceptor. After being loaded into the chamber and prior to growth, the substrates were subjected to a 2 min HCl etch at 1200°C . The temperature was then increased to $1450 \pm 25^\circ\text{C}$ for the film growth. When approximate temperature equilibrium was reached (within 30 s), SiH_4 and C_3H_8 were introduced for the desired growth period. Typically a growth rate of $4 \mu\text{m}/\text{h}$ was achieved with a carrier gas flow of 3 ℓ/min and SiH_4 and C_3H_8 concentrations of 200 and 150 ppm, respectively.

The results of the growth experiments are as follows. A typical growth run lasted 3 h and produced a single-crystal 6H-SiC film $\sim 12 \mu\text{m}$ thick. In appearance, the films were transparent, colorless, and mirror-like. Examined by Nomarski optical microscopy, the $12\text{-}\mu\text{m}$ -thick films were mostly smooth and featureless. Generally, along two edges of the 6.4 mm^2 film/substrate, there were small hexagonal platelets and linear (or slightly tapered) features propagating in from the edge. An edge with platelets is shown in Fig. 1. The platelets all have the same relative orientation; they most certainly have the orientation of the 6H-SiC substrate. The linear features were aligned with the $\langle 11\bar{2}0 \rangle$ direction as determined by the orientation of the platelets. Film/substrate 6H-SiC samples were broken and the broken edge was viewed optically and by scanning electron microscopy (SEM). No film/substrate interface could be detected. In several growth runs, a small piece of a wafer was placed polished-face down on the substrate, acting as a mask during

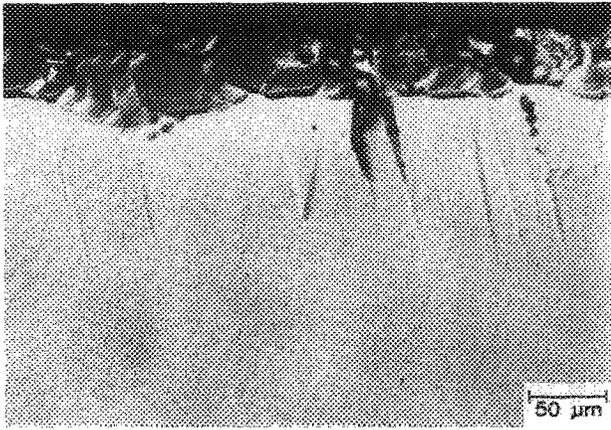


FIG. 1. Nomarski optical micrograph of the edge of a 12- μm -thick 6H-SiC film grown on an "off-axis" 6H-SiC wafer. Note hexagonal platelets along edge.

growth. After the run, a profile trace of the surface across the edge of the mask position confirmed that there was a 12- μm -thick film on the substrate.

The 6H-SiC film/substrates were examined by trans-

mission electron microscopy (TEM) at Case Western Reserve University. Generally, with plan-view TEM, no defects were detected in the film. Since the typical area viewed was about 10^{-5} cm^2 , the actual defect density was $< 10^5 \text{ cm}^{-2}$. In a region near a scratch in one sample, some unexplained paired dislocations were detected. With high-resolution cross-sectional TEM, excellent lattice images were obtained, but no defects and no interface between film and substrate could be detected. In previously published cross-sectional images of a 6H-SiC film/substrate, the interface could be seen because of contrast caused by a large difference in the impurity level between film and substrate.¹⁵ The high quality and lower impurity level of both film and substrate evidently precluded such contrast in our growth.

The 6H-SiC films were studied by LTPL at the University of Pittsburgh. The experimental setup for the LTPL measurements is described in Ref. 18. Samples were immersed in pumped liquid He at a temperature of $< 2 \text{ K}$ and excited by light from a He-Cd laser. Some LTPL results are presented in Fig. 2. The upper spectrum is for one of the 6H-SiC films grown on a Cree substrate; the lower spectrum is for a high quality, bulk Lely-grown 6H-SiC crystal. It is in-

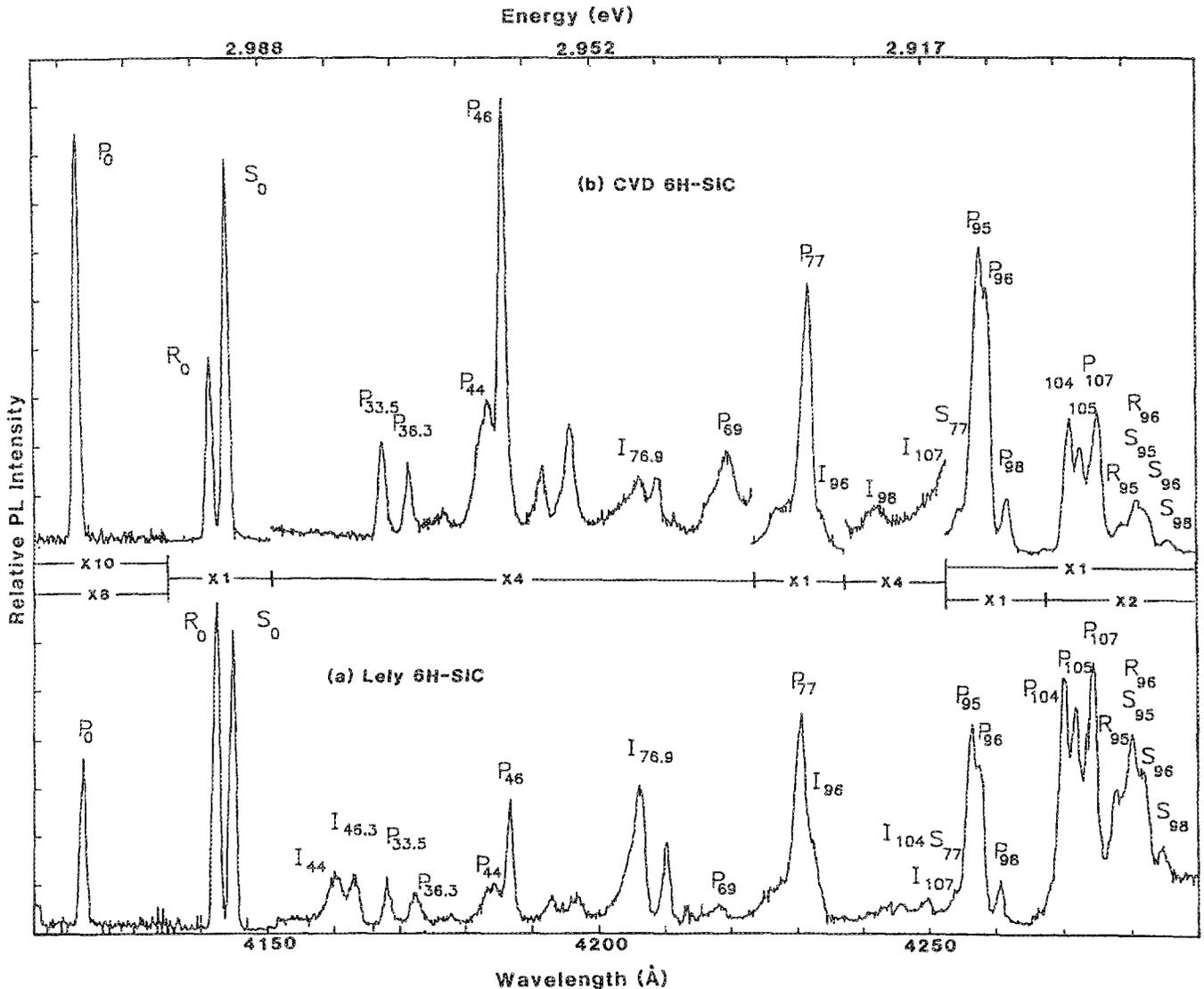


FIG. 2. 2 K photoluminescence spectra for (a) a high quality Lely-grown 6H-SiC crystal and (b) a CVD-grown 12- μm -thick 6H-SiC film.

cluded for comparison because the two spectra are nearly identical. The spectrum for Lely 6H-SiC crystals was originally measured and published in 1962.¹⁹ The spectral lines are due to exciton recombination. Most are no-phonon lines and momentum-conserving phonon replicas due to excitons bound to neutral nitrogen donors. Since nitrogen substitutes for carbon and there are three inequivalent carbon sites, there are three different no-phonon lines and associated replicas. These are denoted by *P*, *R*, and *S* in Fig. 2. The subscripts in this figure denote energy shifts (in meV) from the no-phonon lines (P_0, R_0 , or S_0). In the spectrum of very pure Lely crystals (e.g., the lower spectrum of Fig. 2), there are additional spectral lines due to intrinsic unbound excitons; these are denoted by *I*. The spectrum of the CVD-grown 6H-SiC film exhibits not only the bound exciton lines, but also some lines due to intrinsic unbound excitons. Hence, based on LTPL we can conclude that the film is of very high quality.

Previously reported CVD 6H-SiC films on "off-axis" Acheson 6H-SiC crystals were much thinner (1–3 μm) and were grown at lower growth rates (1–2 $\mu\text{m}/\text{h}$).^{15,16} The higher growth rate (4 $\mu\text{m}/\text{h}$) achieved for our films is certainly desirable for commercial device fabrication. The dominant factors limiting maximum CVD growth rate are unknown at this time.

The platelet and linear features mentioned above are evidence supporting the step-controlled growth model proposed by Matsunami *et al.*¹⁶ In their model for a vicinal (0001) 6H-SiC substrate, growth occurs at steps on the surface in a direction parallel to the basal plane which is at some small angle to the growth surface. In our case, film growth parallel to the basal plane creates the hexagonal platelets that are observed along two edges of the Cree substrate. The linear defects probably start at some point on the original substrate surface; during film growth, the defect propagates approximately along a basal plane in the direction of the step growth. The defect ends at the intersection of the basal plane and the final film surface.

In conclusion, we have demonstrated that high quality 6H-SiC films can be grown on vicinal (0001) 6H-SiC wafers produced by the modified-sublimation process. In addition, we achieved higher growth rates and thicker films than previously reported CVD SiC growth on vicinal (0001) SiC substrates.

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