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[54] COMPOUND SEMI-CONDUCTORS AND CONTROLLED DOPING THEREOF

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Related U.S. Application Data

[63] Continuation-in-part of Ser. No. 276,599, Jul. 18, 1994, Pat. No. 5,463,978, which is a continuation of Ser. No. 8,650, Jan. 25, 1993, abandoned.

[51] Int. Cl.⁶ C30B 23/08
 [52] U.S. Cl. 117/96; 117/97; 437/100
 [58] Field of Search 117/95, 97, 101, 117/106, 913; 437/100

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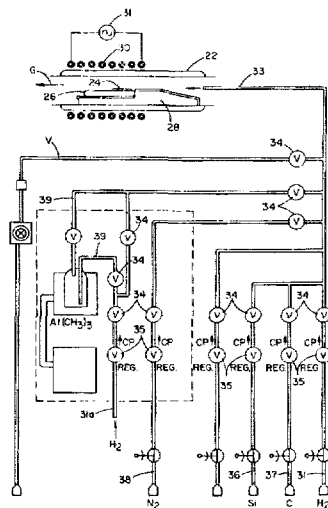
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[57] ABSTRACT

A method of controlling the amount of impurity incorporation in a crystal grown by a chemical vapor deposition process. Conducted in a growth chamber, the method includes the controlling of the concentration of the crystal growing components in the growth chamber to affect the demand of particular growth sites within the growing crystal thereby controlling impurity incorporation into the growth sites.

52 Claims, 6 Drawing Sheets



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FIG. 1

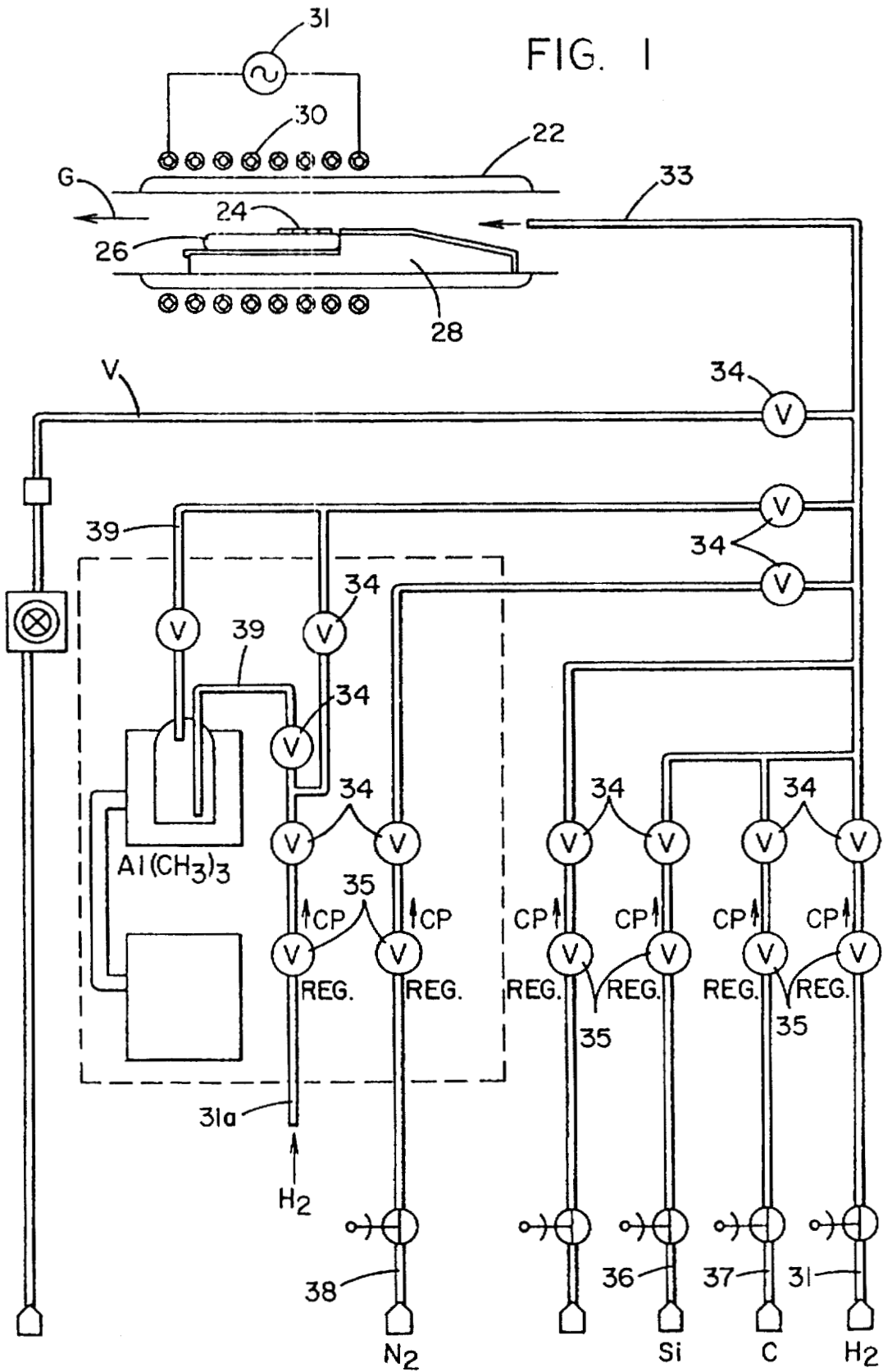


FIG. 2

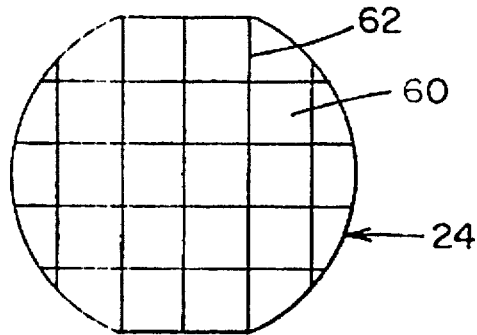


FIG. 3

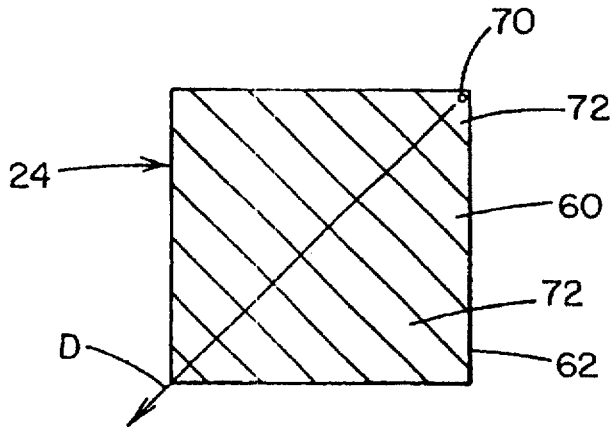


FIG. II

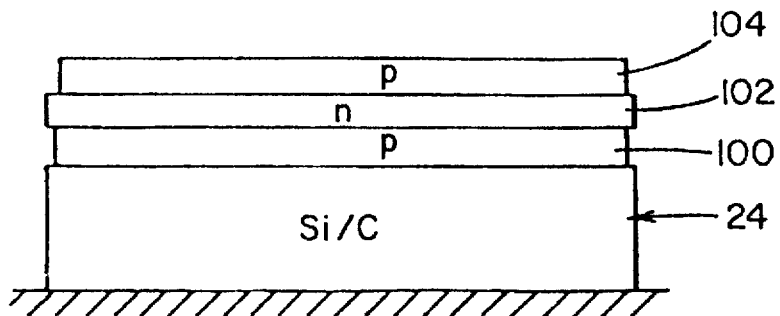


FIG. 4

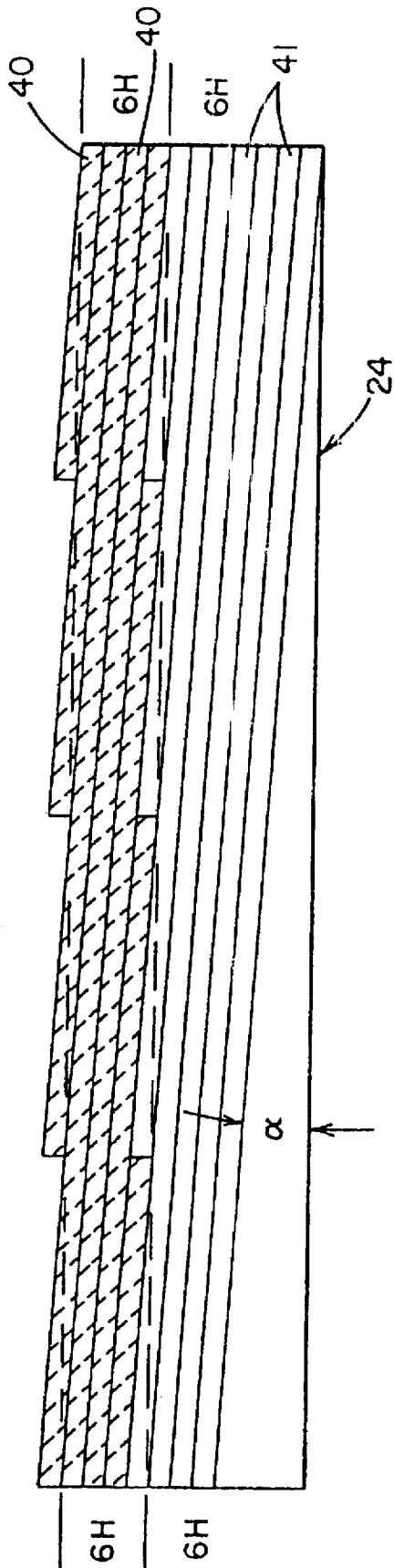


FIG. 5

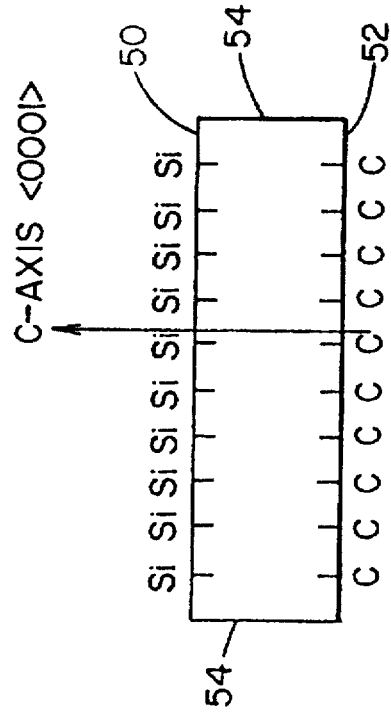
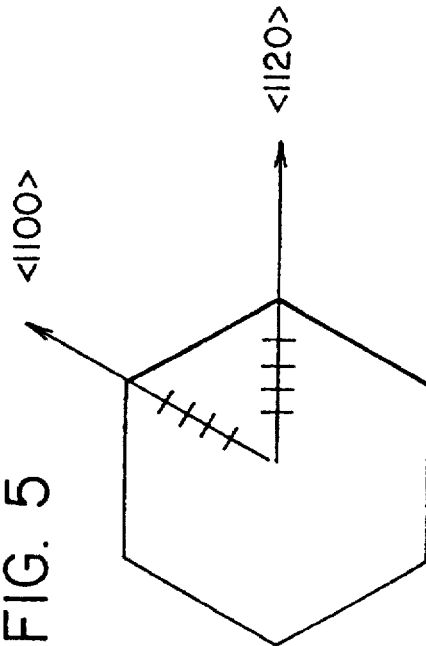


FIG. 6

FIG. 7

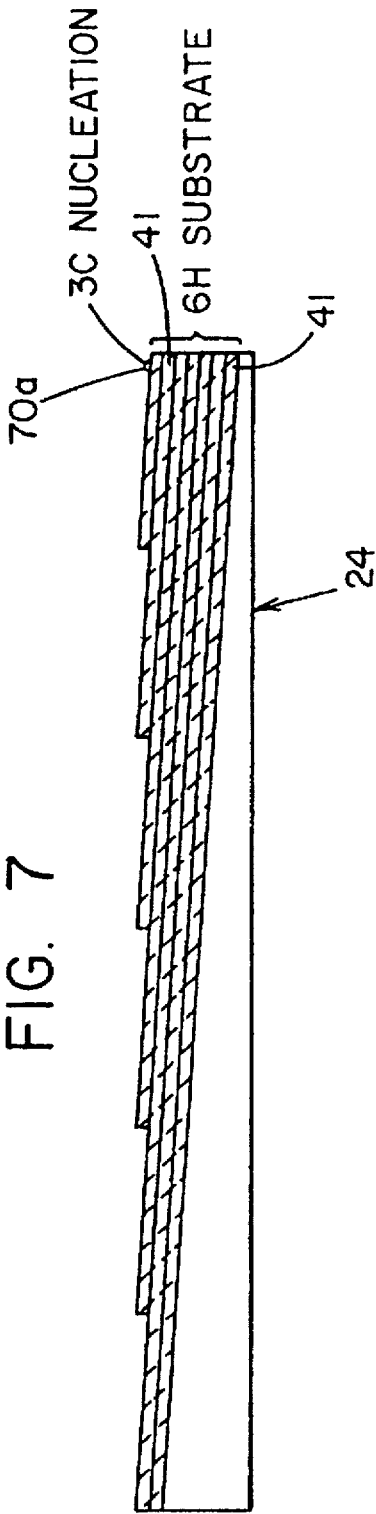


FIG. 8

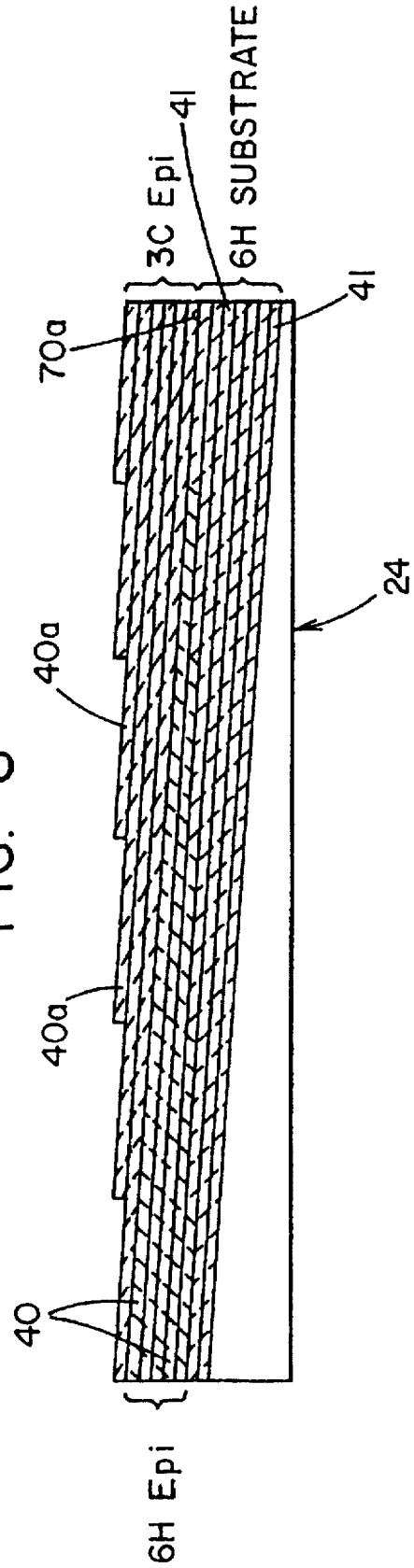


FIG. 9

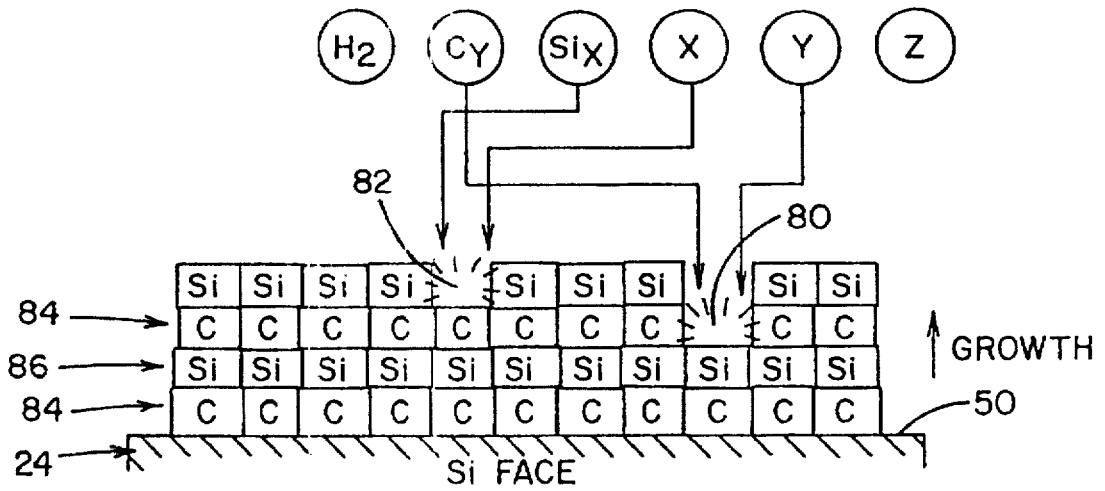


FIG. 10

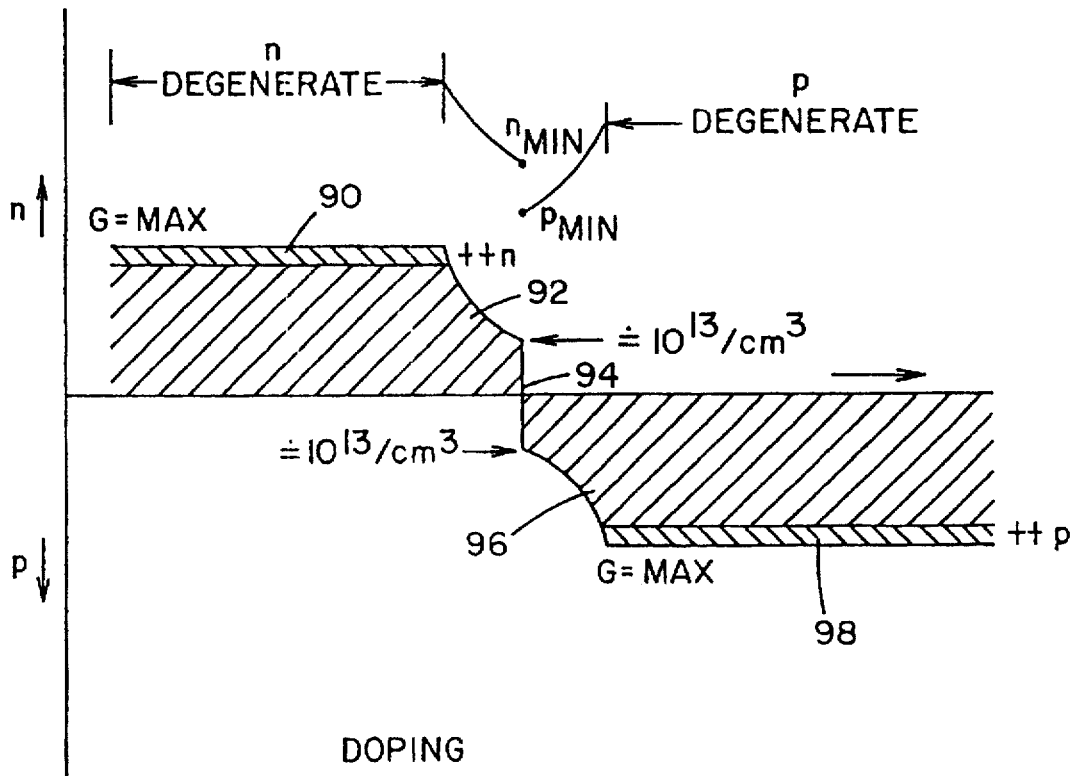
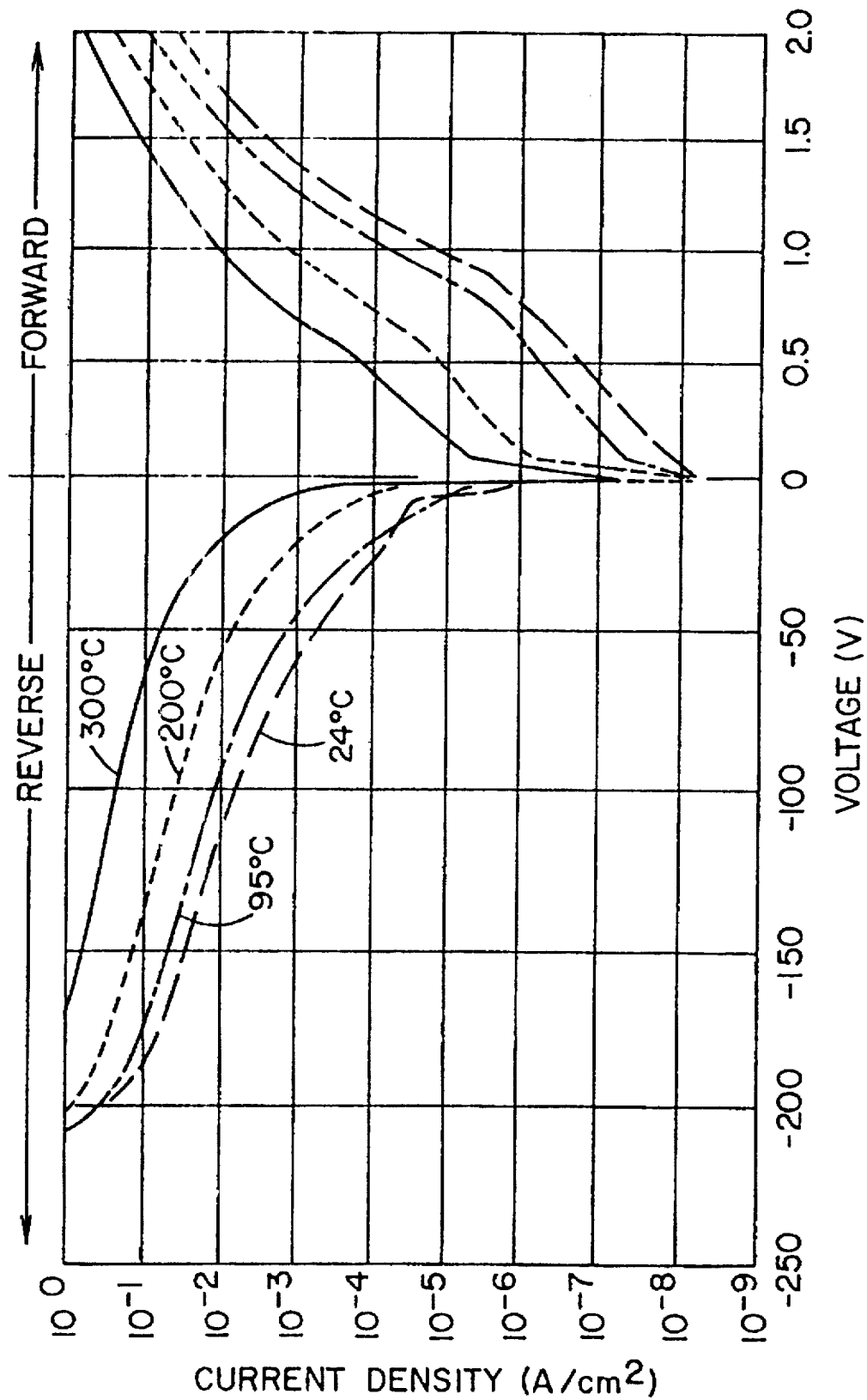


FIG. 12



COMPOUND SEMI-CONDUCTORS AND CONTROLLED DOPING THEREOF

This application is a continuation in part of U.S. patent application Ser. No. 276,599 filed on Jul. 18, 1994 and issued as U.S. Pat. No. 5,463,978 issued on Nov. 7, 1995, which is a continuation of U.S. application Ser. No. 008,650 filed Jan. 25, 1993, now abandoned.

The invention relates to the controlled growth of high-quality semiconductor device crystal films, and more particularly, to a method and system for producing high-quality silicon carbide semiconductors which are highly reproducible. This invention was made with government support under a government contract awarded by NASA. The government has certain rights in the invention.

INCORPORATION BY REFERENCE

U.S. Pat. No. 5,363,800 entitled "Process For The Controlled Growth Of Single-Crystal Films Of Silicon Carbide Polytypes On Silicon Carbide Wafers" issued Nov. 15, 1994 and U.S. Pat. No. 5,248,385 entitled "Process For The Homoepitaxial Growth Of Single-Crystal Silicon Carbide Films On Silicon Carbide Wafers" issued Sep. 28, 1993 which both illustrate the pretreatment of a silicon carbide substrate are incorporated herein by reference. U.S. Pat. No. 5,463,978 entitled "Compound Semiconductor And Controlled Doping Thereof" issued Nov. 7, 1995 is also incorporated herein to illustrate the technique of site competition epitaxy for crystals having two or more components.

BACKGROUND OF THE INVENTION

The invention is particularly applicable to production of silicon carbide crystals (herein used to include crystal films) and it will be discussed with particular reference thereto; however, the invention has much broader applications and can be used for other crystals grown by the chemical vapor deposition process.

Semiconductor devices are used in a wide variety of electronic applications. Semiconductor devices include diodes, transistors, integrated circuits, light-emitting diodes and charge-coupled devices. Various semiconductor devices using silicon or compound semiconductors such as gallium arsenide (GaAs) and gallium phosphide (GaP) are commonly used. In order to fabricate semiconductor devices, it is necessary to be able to grow high-quality, low-defect-density single-crystal films with controlled impurity incorporation (with respect to both net concentration and concentration profiles) while possessing good surface morphology. In recent years, there has been an increasing interest in research of the silicon carbide semiconductors for use in high temperature, high powered and/or high radiation operating conditions under which silicon and conventional III-V semiconductors cannot adequately function.

Silicon carbide has been classified as a compound semiconductor with potentially superior semiconductor properties for use in applications involving high temperature, high power, high radiation and/or high frequency. Silicon carbide has a number of characteristics that make it highly advantageous for various uses. Such advantages include a wide energy gap of approximately 2.2 to 3.3 electron volts, a high thermal conductivity, a low dielectric constant, a high saturated electron drift velocity, a high breakdown electric field, a low minority carrier lifetime, and a high disassociation temperature. Furthermore, silicon carbide is thermally, chemically and mechanically stable and has a great resistance to radiation damage. In addition, a variety of optical

devices, such as light-emitting diodes (LEDs) can be fabricated from silicon carbide and operated at temperatures exceeding 600° C. Despite these many advantages and capabilities of silicon carbide semiconductor devices, large scale commercialization of silicon carbide devices has been slow because of the lack of control over the crystal quality, growth reproducibility, and controlled dopant incorporation into the silicon carbide crystals.

Several properties of SiC contribute to this lack of control. First, it does not melt at reasonable pressures and it sublimes at temperatures above 1800° C. Second, it grows in many different crystal structures, called polytypes. Third, post-growth doping attempts (i.e. diffusion into the crystal from a gas phase species as is used in the silicon industry) are not effective in SiC crystals. Other known post-growth doping techniques (i.e. ion implantation) typically result in considerable crystal damage. Attempts to remove this crystal damage, and therefore improve device performance by post-annealing, commonly result in severe dopant profile redistribution.

Since molten-SiC growth techniques cannot be applied to SiC, two techniques have been developed to grow silicon carbide crystals. The first technique is known as chemical vapor deposition (CVD) in which reacting gases are introduced into a crystal chamber to form silicon carbide crystals upon an appropriate heated substrate. A second technique for growing bulk silicon carbide crystals is generally referred to as the sublimation technique or Lely process. In the sublimation technique, some type of solid silicon carbide material other than the desired single crystal in a particular polytype is used as a starting material and heated until the solid silicon carbide sublimes. The vaporized material is then condensed to produce the desired crystals. Although a large number of crystals can be obtained by either the sublimation method or the epitaxial growth method (CVD), it is difficult to prepare large single crystals of silicon carbide and to control with high accuracy the size, shape, polytype and doping of the silicon carbide crystals.

Silicon carbide crystals exist in hexagonal, rhombohedral and cubic crystal structures. Generally, the cubic structure, with the zincblende structure, is referred to as the β -SiC or 3C-SiC whereas numerous polytypes of the hexagonal and rhombohedral forms are collectively referred to as α -SiC. The most common α -SiC polytype is 6H-SiC. Each of the various silicon carbide polytypes have unique electrical and optical properties which give them advantages over the other polytypes in particular applications. For example, the 6H-SiC polytype has a bandgap of about 2.9 electron volts and a hexagonal structure, wherein the 3C-SiC polytype has a lower bandgap of about 2.2 electron volts and has a higher symmetry structure than the 6H-SiC polytype. These property differences lead to advantages for the 6H-SiC polytype in some applications such as a wider bandgap resulting in blue light-emitting diodes and operation at higher temperatures. On the other hand, the 3C-SiC polytype has a higher electron mobility leading to a higher frequency of operation.

SiC polytypes are formed by the stacking of double layers of Si and C atoms. Each double layer may be situated in one of three positions. The sequence of stacking which determines the particular polytype. The stacking direction is called the crystal c-axis which is perpendicular to the basal plane. For 6H-SiC polytypes, the (0001) plane (the Si-face) or (0001) plane (the C-face) is known as the basal plane and for 3C-SiC, the plane (111) is equivalent to the (0001) basal plane.

Many advances have been made in the growing of higher quality 6H-SiC and 3C-SiC crystals having fewer

dislocations, stacking faults, microtwins, double positioning boundaries (DPBs), threading dislocations and anti-phase boundaries (APBs).

U.S. Pat. No. 5,037,502 (Suzuki) discloses a process for producing a single-crystal silicon carbide by growing in a CVD chamber a single crystal film of α -silicon carbide on a single crystal film of β -silicon carbide. Suzuki discloses that the CVD reaction chamber should be maintained between 1400–1900° C. and preferably 1500°–1700° C. while the β -silicon carbide crystal is growing. Silane (SiH_4) and propane (C_3H_8) are the source gases fed into the growth chamber at Si/C atomic number ratio of 0.01–10 and preferably 0.5–5. The Si/C ratio is set at a constant rate during crystal growth. The setting of the Si/C ratio at a single optimum point maximizes crystal growth while minimizing crystal defects to grow the β -silicon carbide crystal. The α -silicon carbide crystal is grown on top of the β -silicon carbide crystal by maintaining a similar Si/C ratio and reducing the growth chamber temperature to 800–1200° C. and preferably 1000–1100° C. Suzuki discloses that β -silicon carbide crystals must grow at temperatures between 1200–1400° C. and preferably 1300–1350° C. By lowering the chamber temperature below 1200° C., only α -silicon carbide crystals can grow. Suzuki only teaches that temperature affects the type of silicon carbide crystal which can be grown. Suzuki does not disclose (1) any method of doping the β or α -silicon carbide crystal, (2) any type of pretreatment process such as preparing the β -silicon carbide crystal prior to growing the α -silicon carbide crystal, (3) whether the tilting of the β -silicon carbide substrate will affect the growth of the α -silicon carbide crystal, and (4) whether a particular side of the β -silicon carbide substrate grows better α -silicon carbide crystals.

With all the advances in growing 3C—SiC and 6H—SiC crystals, the controlled purity and doping of such crystals are limited. Furthermore, controlled and reproducible degenerate doping or very low doping has of yet been unachievable. Dopant incorporation in a grown crystal is known as intentional impurity incorporation and, contaminant incorporation into the crystal is known as unintentional impurity incorporation (i.e. contamination). During the sublimation or CVD crystal growing process, various compounds and/or elements are intentionally and unintentionally incorporated into the crystal. Various techniques of limited success have been used to exclude contaminants from a crystal to produce highly pure crystals.

One technique for doping silicon carbide crystals is disclosed in an article by H. J. Kim and R. F. Davis entitled Theoretical and Empirical Studies of Impurity Incorporation into β -SiC Thin Films During Epitaxial Growth (1986). In the Kim reference, a study relating to the dopant incorporation of β -SiC crystals are disclosed. The Kim reference teaches that the dopant incorporation in a β -SiC crystal increases as the dopant concentration increases within the SiC growth reactor. Kim also teaches that some dopants are incorporated in concentrations which are higher than predicted from theory. Kim's observation of an unexpectedly high concentration of incorporation, as determined by comparisons of SIMS and electric probe results are hypothesized to be due to three possibilities, namely, (I) line or point defects and/or trapping of impurities at dislocations and stacking faults known to be in the material, (ii) dopant-Si and/or dopant-C interaction, and (iii) location in non-electrically active interstitial sites. The dopant-Si and/or dopant-C interaction that Kim refers to relates to a portion of each of the dopants either 1) not ionized or located on non-electrically active sites or 2) form complexes with Si or

C. Kim's phrase "forms complexes with Si or C" teaches that the dopant atom can combine with available Si and/or C to form distinctly different compounds possessing a non-electrical character within the SiC crystal, thereby creating the discrepancy between the electrically measured dopant concentration versus the atomically probed (i.e. SIMS) elemental dopant concentration. These proposed complexes are inclusions of compounds, such as the compound silicon nitride (Si_3N_4). The high melting point material being referred to in the Kim reference is silicon nitride (Si_3N_4), which has a melting of 1900° C. With regard to the location of non-electrically active interstitial sites, Kim suggests that the dopant atoms could also be located in other non-electrically active sites. Specifically, Kim makes reference to "a non-electrically active interstitial" position in the crystal lattice. Kim teaches that some atoms (e.g., Al and P), follow ideal incorporation behavior while others (i.e. N, B) apparently do not because these other dopant atoms are incorporated into the grown SiC crystal at a much higher concentration than predicted from theory. In summary, Kim teaches that the excess dopant incorporation in the disclosed β -SiC crystal is primarily believed, as theorized by Kim, to result from (1) accumulation of non-electrically active dopant atoms at defect sites such as dislocations and stacking faults, (2) non-electrically active high melting point compounds formed from interactions of dopant atoms with the SiC and/or C or (3) accumulation of dopant atoms at non-electrically active interstitial sites within the SiC lattice. These three possible explanations are well known to the art of SiC growth since the high density of defects in β -SiC (grown on silicon substrates) is one of the main reasons it has been largely abandoned for growth on the superior SiC substrates. Kim does not disclose any techniques for controlling the rate of incorporation of dopant atoms in a SiC crystal by varying the Si/C source gas ratio. The study in Kim was conducted by maintaining the Si and C source gas ratio constant during the entire crystal growth and by varying the dopant concentration. The theoretical considerations and experimental data discussed in Kim do not give any insights into how the varying of the Si and C source gas ratio during crystal growth would affect dopant incorporation into the growing crystal. The study in Kim was limited solely to the effects on dopant concentration in grown crystals in the presence of different partial pressures of dopant gas in the reactor. Kim also does not teach the varying of dopant gas in the reactor once crystal growth has begun. All the experiments and data collected resulted from crystal growth based upon constant Si, C and dopant gas concentrations. The varying of any gas within the reactor during crystal growth is not disclosed in Kim.

Another technique reported for InP and GaAs CVD growth systems are the use of a blocking technique whereby relatively large amounts of a crystal compound are used to block or shield the crystal surface from impurities. *Chemistry of the In—H₂PCL₃ Process* by R. C. Clarke, *Inst. Phys. Conf.*; Ser. No. 45; Chapter 1 (1979) 19–27 and *Doping Behavior of Silicon and Vapor Growth III—V Epitaxial Films* by H. P. Pogue and B. M. Quimlich, *J. Crystal Growth* 31 (1975) 183–89. Although the blocking technique reduces contaminant incorporation into a crystal, the required use of large amounts of a crystal growing compound to affect blocking has adverse effects on crystal quality and surface morphology. In addition to the problems associated with contaminant exclusion, the controlled intentional doping of crystals at relatively low concentrations has of yet been unachievable and/or unreproducible. Furthermore, techniques to control and/or reproduce sharp dopant concentra-

tion profiles within crystals (i.e. from p-type to n-type, degenerate to lightly doped) are also unavailable.

In CVD crystal growth, control of the source gases in the prior art was solely used for setting a single optimum source gas ratio for obtaining smooth single-crystal surface morphology while preventing the formation of unwanted crystal formations (i.e. polycrystalline SiC in SiC crystals), as well as other crystal defects. The teachings and the understandings in the art of CVD crystal growth taught against the changing of the source gas ratio during crystal growth. It was believed that one must first obtain the optimum source gas ratio in order to obtain high quality crystals. Once the source gas ratio set point was obtained, dopant incorporation was controlled solely by controlling the dopant concentration in the reaction chamber. The prior art taught that the varying of the source gas ratio had no effect on the dopant incorporation (i.e. electrical properties) of the grown crystal. Once determined, this optimum source gas ratio was constantly maintained and is specific and distinct for each CVD system during the growth of the entire crystal. The controlled constant amount of Si-source and C-source gases, taught by prior art, resulted in a constant ratio of available Si-sites to C-sites (i.e., constant site composition) for incorporation of either crystal-growth atoms (i.e. Si and C atoms) or dopant-source atoms (e.g., Al, N, Ga, Ti, V, Na, Fe, P, S, O, B, F, etc.). The maximum concentration of a p-type dopant incorporated into a growing SiC crystal is fixed, because the p-type dopant atoms compete with the Si-source atoms for a fixed amount of available Si-sites. When the amount of aluminum dopant-source (Al) is increased in an attempt to obtain a higher carrier concentration than this fixed amount ($p=6 \times 10^{17} \text{ cm}^{-3}$), the excess Al results in Al-containing material deposited on the reactor walls which results in a very rough surface of a heavily p-doped ($p=1.6 \times 10^{19} \text{ cm}^{-3}$) quasi-polycrystalline film. Therefore, it was not possible to grow p-type doped (e.g. Al-doped) single crystal films with carrier concentrations greater than $p=6 \times 10^{17} \text{ cm}^{-3}$.

A variety of prior art techniques have been developed, and tried, in an attempt to control the impurity incorporation in the growing SiC films. For example, molecular beam epitaxy (MBE) utilizes an ultrahigh vacuum system and the growth of the crystal occurs by using a stream of molecules which is formed into a beam and focused onto a heated substrate. This technique does offer some amount of control over dopant-concentration profiles. However, the low rate of growth is a major drawback and poses several problems. First, commercialization is not practical because of the very low growth rate. Second, the impurity incorporation is still limited by the purity of the source gas and cleanliness of the growth reactor. Furthermore, the problem of impurity incorporation is exacerbated by the very slow growth rate.

Another technique for doping SiC is the use of ion implantation, which is a post-growth doping technique used to introduce the desired dopants. See *Large-Band-Cap, III-V Nitride, and II-VI Zn-Se-Based Semiconductor Device Technologies*, by H. Morkoc et al; J. App. Phys. 76(3), p. 1368 (1994). This method produces a large amount of damage to the crystal structure and typically requires a post-anneal step to reduce the high density of defects (which greatly affects device quality) generated by this technique. Furthermore, as a result of the unusually high temperatures ($>1800^\circ \text{C}$.) needed for the only partially effective annealing of SiC crystals, the dopant concentration changes by a factor of four (4x) for p-type and is lost via out-diffusion for the n-type dopants.

When degenerately doped layers are desired, such as the case when metal contact layers are needed, obtaining degenerate

p-type via CVD is limited by problems such as gas phase nucleation (from an excessively high concentration of p-type source gas needed) which results in very poor film morphologies.

As stated above, silicon carbide (SiC) is emerging as a material of choice for fabrication of high power and/or microwave-frequency semiconductor devices suitable for operation in high temperature, high radiation, and corrosive environments. The intrinsic material advantages of silicon carbide are currently being exploited in the development of high power and high frequency semiconductor devices for service in high temperature, corrosive, and high radiation environments. These microelectronic devices include high voltage Schottky rectifiers and power metal-oxide semiconductor field-effect transistors (MOSFETs), microwave and millimeter-wave devices, and high temperature, radiation resistant junction field-effect transistors (JFETs). However, in order for the theoretically calculated advantages of using SiC to be fully realized, advancements are needed both in the bulk growth and in the epilayer growth of SiC. For example, improvements in the bulk growth of SiC are needed for elimination of device limiting micropores and micropipes. Other advancements are needed in the epilayer growth process. In particular, dopant incorporation during the growth of SiC epilayers must be understood and reliably controlled. In order for the inherently superior high-temperature semiconductor properties of SiC to be realized in advanced electronic device applications, control over the electronic properties of chemical vapor deposited (CVD) epitaxial layers must be improved. Prior to this work, control over dopant incorporation for CVD SiC epilayers had been limited, with reproducible doping typically confined to doping concentrations ranging from $N_D=2 \times 10^{16} \text{ cm}^{-3}$ to $5 \times 10^{18} \text{ cm}^{-3}$ for n type and from $N_A=2 \times 10^{16} \text{ cm}^{-3}$ to $1 \times 10^{18} \text{ cm}^{-3}$ for p-type 6H-SiC epilayers. Expanding the reproducible doping range to include lower concentrations would enable the fabrication of multikilovolt SiC power devices, whereas the availability of higher doping concentrations would result in devices with increased performances because of lower parasitic resistance.

As a result of the inadequate techniques available to control dopant and/or contaminant incorporation into crystals grown in a CVD process, there is a demand for a method to selectively exclude impurities from a CVD grown crystal.

THE INVENTION

The present invention relates to a method of growing high-quality crystals and controlling the impurity or dopant incorporation during crystal growth.

In accordance with the principal feature of the invention, there are provided a system and a method of controlling the amount of a selected element deposited in a given growth area of a crystal grown by the chemical vapor deposition (CVD) process. One type of crystal grown is a SiC crystal wherein silicon (Si) is deposited in Si growth sites and carbon is deposited in C growth sites at the growth area. The selected element competes for either the Si growth site or C growth site during SiC crystal growth. The improvement comprises flowing of a first amount of a gaseous Si compound through the growth chamber and flowing a second amount of a gaseous C compound through the growth chamber and controlling the ratio of the first amount of Si compound relative to the second amount of C compound to control the amount of the selected element deposited in the SiC crystal at the crystal growth area. The selected element is an impurity and is either a dopant or a contaminant.

A crystal growth chamber is used to grow the crystals by chemical vapor deposition. The crystal growing chamber includes a crystal growing area upon which crystals are grown. The crystal growing chamber may also include a heating element to control the temperature within the crystal growing chamber. The crystals are formed within the crystal growing chamber by introducing vaporized crystal growing compounds into the chamber. A carrier gas may be used to introduce the crystal growing compounds into the crystal growing chamber. Many different types of crystals may be grown in the crystal growing chamber such as silicon carbide, gallium arsenide, gallium phosphide, etc. The grown crystals are formed by crystal components being deposited in crystal growing sites. For example, SiC crystals are grown by introducing a silicon compound and a carbon compound into the crystal growing chamber. In the crystal growing chamber, carbon atoms dissociate from the precursor of the carbon compound and silicon atoms dissociate from the precursor of the silicon compound. The SiC crystal is then grown by carbon atoms being deposited in carbon sites and silicon atoms being deposited in silicon sites. The SiC crystal is formed from stacked double layers or film layers of silicon and carbon atoms. Each film layer is formed from a layer of carbon atoms bonded to a layer of silicon atoms. Due to the double layer stacking of SiC film layers, the SiC crystal has a silicon face and a carbon face. The crystal growing sites, such as carbon sites and silicon sites for a SiC crystal, are highly specific as to what types of atoms or molecules may enter the site during crystal growth. The high specificity of the crystal growing site is due in part to the physical configuration of the site. Each atom and molecule has a distinct size and atomic configuration. Unless an atom or molecule have the same or a similar size and configuration as the crystal growing site, the atom or molecule will have a low probability of bonding at the growing site. The crystal growing site specificity is also affected by bonding forces in and about the growing site. Each atom and molecule has an electron shell, having some charge, which interacts with electron shells of an atom or molecules present about the growing site, thereby being either attracted to or repelled from the growth site. The purity of any crystal grown by the CVD process can be controlled by controlling the impurity deposition in the growth sites during crystal growth. During crystal growth, a specific growth site has a particular demand for a compatible atom or molecule. The growth site demand can be manipulated by controlling the amount of compatible atoms or molecules near the growth site. The competition for a particular growth site between a crystal atom component and an impurity atom or molecule is controlled by controlling the concentration of the respective atoms or molecules at the growth site. This concept is herein referred to as site-competition epitaxy which serves to increase the reproducibility of CVD crystal epilayers while greatly expanding the doping range for both n-type and p-type CVD crystal epilayers. Surprisingly, the growth site-competition can be accurately and reproducibly controlled by controlling the ratio of the crystal growing compounds introduced into the crystal growing chamber. Many crystals other than SiC crystals which have two or more growth sites potentially can be grown by a CVD process and use the site competition technique to control impurity incorporation. The recognition and implementation of growth site-competition technique allows for substantially greater latitude and control as compared to prior growth techniques over crystal film growth conditions, so as to optimize crystal growth rate, surface morphology, impurity profiles and other film characteristics during crystal growth.

In accordance with a broader aspect of the present invention, there are provided a system and a method of controlling the amount of a non-crystal element deposited in a given growth area of a crystal formed from at least two crystal elements as the crystal is grown at the crystal growing area by a CVD process conducted in a growth chamber. The crystal comprises a first crystal element and a second crystal element and the grown crystal has at least two crystal growing sites. The first crystal element is deposited in the first crystal growing site and the second crystal element is deposited in the second crystal growing site. The non-crystal element is competitive for at least one of the growth sites. The improved method comprises flowing a controlled amount of gaseous crystal element compounds through the growth chamber, wherein each of the crystal element compounds includes at least one of the crystal elements, and controlling the ratio of the controlled amount of gaseous crystal element compounds to control the amount of the non-crystal element deposited in a particular growth site of the crystal at the crystal growth area. The non-crystal element is an impurity such as a dopant or a contaminant. The crystal can be formed by two, three, four or more crystal elements to form a binary, tertiary, etc., type of crystal. Each crystal element is deposited in a particular crystal growth site. The non-crystal element competes with at least one particular crystal growth site. The non-crystal element normally only competes with one crystal growth site due to the geometry and charge of the non-crystal element and the geometry and charges of the crystal growing sites. If a non-crystal element can deposit itself in more than one crystal growing site, the non-crystal element usually has a much larger affinity for one over the others. The types of crystals that possibly can be grown include ZnSe, GaAs, InP, $\text{In}_x\text{Ga}_{1-x}\text{As}$, GaAsP, GaP, InAs, $\text{In}_x\text{Ga}_{1-x}\text{As}_y\text{P}_{1-y}$, etc.

In accordance with another aspect of the present invention, a dopant material is introduced into the crystal growing chamber and the rate of dopant incorporation into the crystal during crystal growth is controlled. Various types of dopant materials may be used to form n-type or p-type layers in the crystal. The dopant profile of a crystal is very important for the type of electronic device the doped crystal is incorporated therein. The dopant material is introduced into the growing chamber either in its pure form (i.e. nitrogen) or as a compound which includes the dopant material bound to a dopant precursor (i.e. trimethyl aluminum). The dopant material which is introduced into the growth chamber is selected such that the dopant material has an affinity for at least one of the growth sites of the crystal. Almost all dopants have an affinity for only one growth site due to the physical characteristics of the dopant and of the growth site. The select few dopants which may bind in multiple crystal growth sites have a higher affinity for one growth site as compared to the others. Once the dopant material is introduced into the crystal growing chamber, the dopant material primarily competes with a crystal element component for a particular growth site. The competition (i.e. demand) of the dopant material with a crystal atom component for a particular growth site is controlled by the manipulation of the availability of the growth site. The manipulation of the growth site demand is accomplished by adjusting the ratios of the crystal atom components within the crystal growing chamber. By proper manipulation of the ratios of the various crystal atom components, a particular growth site will become more or less available to the dopant material during the crystal growing period, thus affecting and controlling the growth site demand and the dopant concentration profile within a particular crystal.

In accordance with still another aspect of the present invention, the crystal is grown on a pretreated substrate to expand the growing parameters for producing high quality crystals. The substrate may or may not be made of the same material as the grown crystal film. A crystal film grown on a substrate may be grown homoepitaxially (substrate and crystal film have the same crystal elements and structure) or grown heteroepitaxially (substrate and crystal film have different crystal elements and/or structure). The substrate is pretreated to remove impurities from the substrate surface which may cause defective crystal films during the crystal growth. Abnormal crystal growth can take place at sites where there is contamination or defects or some other surface disturbance on the substrate surface. These contaminants or defects result in unwanted nucleation resulting in crystals having unwanted polytype structure, inferior surface morphology, stacking faults, APBs, low quality dopant profiles, high contaminant concentrations, etc. Contaminants and surface defects are removed from the substrate by proper cutting and polishing of the substrate surface and subsequent etching of the substrate surface. The etching is carried out so as not to alter the substrate surface in a manner that would impair crystal growth upon the substrate surface. The clean, low-defect substrate surface surprisingly expands the operation parameters for growing quality crystals. By reducing the stress to the crystal growing process, previous Si/C concentration ratios within a growth chamber were limited in range and did not stray far from a narrow ratio-range which is dependent on the choice of a hydrocarbon employed (Si:C:1:1: for $\text{SiH}_4/\text{C}_2\text{H}_4$, Si:C::2:3 for $\text{SiH}_4/\text{C}_3\text{H}_8$). With any given crystal growing process, there are stress components to the process such as temperature, pressure, crystal component concentrations, dopant concentrations, impurity concentrations, growth surfaces, etc., which affect the crystal film layer during growth. The elimination or reduction of stresses during crystal growth caused by a defective and/or contaminated substrate surface allows for a greater variance in stress levels from other stress components. The reduced stresses caused by a pretreated substrate enable the use of greater crystal compound ratios in the growth chamber for enhanced control of impurity incorporation into the crystal.

In accordance with still another aspect of the present invention, there is provided a method for growing crystals whereby the crystals are grown on a particular surface face of a substrate to reduce the stresses to the crystal growing system. A substrate formed from two or more different atoms will have a crystal structure having at least three distinctive faces. Substrates formed of Si—C crystals have a silicon face (Si-face), a carbon face (C-face) and a carbon-silicon face (A-face). Each face of the crystal structure has different physical characteristics (e.g. polarity) which can increase or reduce the stresses to the crystal growing system, which will affect the quality of the crystals grown on the substrate face.

The primary object of the present invention is the provision of a method to grow high-quality, reproducible crystal films by chemical vapor deposition.

Another object of the present invention is the provision of a method for controlling impurity incorporation in a crystal film layer during the growth of a crystal.

Still another object of the present invention is the provision of a method to grow high-quality, low-defect reproducible crystals having a desired impurity profile.

Yet still another object of the present invention is to provide a method for growing crystals on a pretreated substrate whereby contaminants and surface defects have

been removed from the surface of the substrate upon which crystals are to be grown.

Another object of the present invention is to provide a method for growing crystals on a particular face of the substrate to produce high quality crystals having low defects and a smooth surface morphology.

Yet another object of the present invention is to provide a method for degenerate doping of crystals for both p-type and n-type epilayers having a high-quality surface morphology and low defects.

Yet still another object of the present invention is to provide a method for growing crystals having very abrupt changes in dopant n-p type profiles.

Another object of the present invention is to provide a method of growing extremely abrupt dopant profile interfaces.

Yet another object of the present invention is to provide a method for growing low-defect, low-concentration doped single-crystals on various substrates.

Another object of the present invention is to provide a method for growing crystals having abrupt changes in dopant concentration profiles.

Still yet another object of the present invention is to provide a method for growing high-quality, low-defect crystals having gradient dopant profiles.

Yet another object of the present invention is to provide a method for increasing the range of concentration ratios usable to grow high-quality, low-defect crystals.

Yet still another object of the present invention is the provision of a method to grow high-quality, low-defect crystals having very high dopant concentrations.

These and other objects and advantages will become apparent to those skilled in the art upon reading the following description taken together with the accompanying drawings.

BRIEF DESCRIPTION OF THE DRAWINGS

FIG. 1 is a schematic drawing of a CVD system employed for the growing and doping of crystals in accordance with the present invention;

FIG. 2 is a plan view of a substrate that has been divided into selected growth regions;

FIG. 3 is a plan view schematic drawing of a selected growth region having parallel lines which illustrate some of the atomic-scale growth steps;

FIG. 4 is a schematic diagram illustrating the homoepitaxial growth of 6H—SiC films upon a 6H—SiC substrate;

FIG. 5 illustrates a plan view of a 6H—SiC crystal showing the dimensional axis of the crystal;

FIG. 6 is a cross-sectional view of a SiC substrate showing the Si-face and the C-face and the A-face of the substrate;

FIG. 7 is a cross-sectional view of a 6H—SiC substrate nucleated with 3C—SiC;

FIG. 8 is a substrate as shown in FIG. 7 illustrating the expansion of 3C—SiC growth from a nucleation point on the 6H—SiC substrate;

FIG. 9 illustrates the site competition at the C site and Si sites during the growing of a SiC crystal;

FIG. 10 is a graphic illustration of a dopant concentration profile of a crystal;

FIG. 11 is a cross-sectional view of p and n epilayers of a SiC crystal grown upon a SiC substrate; and

FIG. 12 illustrates the forward and reverse current-voltage characteristics on a logarithmic scale of a 3C—SiC crystal at several temperatures.

PREFERRED EMBODIMENT

Referring now to the drawings, wherein the showings are for the purpose of illustrating preferred embodiments of the invention only and not for the purpose of limiting the same, the invention describes an improved chemical vapor deposition (CVD) method for obtaining improved control of impurity incorporation and dopant profiles of CVD films and also improving the quality of the grown crystal films. While the method may be applied to many different crystals, the method will be specifically described with respect to the growing of silicon-carbide (SiC) crystals. The improved CVD method includes pretreating a substrate, heating the substrate in a reaction chamber, introducing a carrier gas, vaporizing the crystal growing compounds, introducing the vaporized compounds in the reaction chamber via the carrier gas, maintaining proper energy levels and material flow rates in the reaction chamber for a sufficient time to grow a crystal film having a desired smooth surface morphology, a uniform thickness, a low-defect density and a controlled impurity profile. The crystals may be intentionally doped to form n-type and/or p-type crystals. The improved CVD crystal growing method is based on the discovery that impurity incorporation into a growing SiC film layers is very sensitive to the ratio of silicon compound (Si containing precursors) and carbon compound (C containing precursors) in the reaction chamber during crystal growth. By varying the Si/C compound ratio at or near the crystal growing surface, the impurity incorporation into the growing crystal film is controlled.

The method of the invention can be carried out with a conventional chemical vapor deposition (CVD) system similar to that used in Si, SiC and GaAs semiconductor technology. The gases used in a SiC CVD system are hydrogen (used as a carrier gas), Silane (used as a source of Si), HCL (used for cleaning and etching the substrate surface), propane (used as a source of C), nitrogen (N₂) (used as a n-type dopant), and trimethyl aluminum (TMA) (used as a p-type dopant). Other gases may be used as the Si or C source or used to dope the crystal. If organic compounds are used as the Si and C source, the process is commonly referred to as metal-organic vapor phase epitaxy (MOVPE). Any CVD system that can deliver these gases to a suitable reaction chamber at the proper flow rates under high purity conditions can be used for the inventive method.

Referring now to FIG. 1, there is shown a schematic, partial view of a suitable CVD reaction system for carrying out the process of the invention. The CVD reaction system includes a reaction chamber 22 comprising a double-walled quartz tube such that the inner quartz tube can be water cooled. The inside diameter of reaction chamber 22 is preferably 50 mm. A SiC substrate 24 is supported by a SiC coated graphite susceptor 26, which in turn is supported by quartz support 28. To produce the desired temperature of the surface of substrate 24, a radio-frequency (RF) induction coil 30 is disposed around reaction chamber 22. Induction coil 30 is powered by frequency generator 31. The RF field produced by induction coil 30 heats substrate 24 via susceptor 26 to the desired temperature. When the SiC film layers are grown, substrate 24 is preferably a SiC substrate. The gaseous crystal compounds are introduced into reaction chamber 22 by primary line 33. Primary line 33 is located at one end of reaction chamber 22 and directs the gases to flow in direction G across substrate 24 and out the opposite end

of chamber 22. The various gaseous crystal compounds are connected to primary line 33 and the gas flow is regulated by valves 34 and regulators 35 connected to each gas line. Line 36 is the silicon gas line that controls the silane flow into primary line 33, and line 37 is the carbon gas line that controls the propane flow into primary line 33. The dopants are introduced into primary line 33 by line 38 and line 39. Line 38 is the n-type dopant line and preferably controls the nitrogen gas (N₂) flow rate. Line 39 is the p-type dopant line and preferably controls the trimethyl aluminum (TMA) flow rate. Carrier gas line 31 carries all the gaseous crystal compounds and dopants through primary line 33 and into reaction chamber 22. The carrier gas is preferably a gas such as hydrogen gas (H₂). Carrier gas line 31 is partially diverted into line 31a to supply line 39 so that the carrier gas can be bubbled through the liquid TMA. A vacuum line V connected to a vacuum can be connected to primary line 33 to evacuate reaction chamber 22 of gases.

Preferably, substrate 24 is pretreated to remove any contaminants or impurities on the surface of the substrate so as to facilitate the growing of high-quality, low-defect epitaxial films. SiC substrate 24 is prepared by slicing a section from a SiC boule. Substrate 24 may be cut such that the surface is slightly misoriented relative to the basal plane by some tilt angle. If 3C—SiC film layers are to be heteroepitaxially grown on an α -SiC substrate 24, the tilt angle preferably is less than 1° and additional surface preparation is necessary. If SiC film layers are to be grown homoepitaxially on SiC substrate 24, the tilt angle is preferably greater than 0.1°. The tilt direction is preferably toward the <1100> or <1120> direction, as illustrated in FIG. 5, to produce the optimum growth rates and quality of the SiC epitaxial films grown on substrate 24. The surface of substrate 24 is polished preferably on one surface with a diamond paste. SiC substrate 24 has three faces, a Si-face 50, a C-face 52 and the A-face 54, as illustrated in FIG. 6. Any of the faces may be polished and prepared for growth of the SiC epitaxial layers. Preferably, Si-face 50 is polished and used for epitaxial growth. It has been found that Si-face 50 produces the highest-quality epitaxial layer films which have the best surface morphology and lowest defects.

If heteroepitaxy is preferred, then substrate 24 is further preparedly creating boundaries or grooves 62 on the face of substrate 24 which form growth regions 60, as illustrated in FIG. 2. Grooves 62 forming growth region boundaries 60 are preferably cut with a precision dicing saw with a 25 micrometer thin blade to minimize crystal damage; however, boundaries 60 may be formed by other processes such as photolithography, ion etching and/or photochemical or electrochemical etching processes. The width of groove 60 need only be less than 1 micrometer but larger widths can also be used. The depth of groove 60 is preferably about 50 micrometers but may be larger or smaller.

Once the substrate surface has been polished and growth regions 62 have been formed, substrate 24 is placed in reaction chamber 22. Prior to growing the crystal film layers on substrate 24, the substrate is pretreated with a pregrowth etch to remove contaminants and defects on the surface of the substrate that could act as unwanted sites for heterogeneous nucleation of the SiC film layers. These defects on the surface of the substrate can be generated during the cutting and polishing of the substrate. Preferably, the pregrowth etch involves subjecting substrate 24 to a high temperature gaseous etch in a mixture of hydrogen chloride gas and hydrogen within the reaction chamber. The etch is monitored such that the substrate is not altered in a way that unwanted sites for heterogeneous nucleation are introduced to the

surface of the substrate. Preferably, the etch uniformly removes at least one atomic layer from the surface of substrate **24** to ensure a low-defect, highly-pure surface. A typical etch is carried out for about 25 minutes at a temperature of 1350° C. using about 3–4% hydrogen chloride gas in an H₂ carrier gas with a flow of about 3 liters per minute. Preferably, the concentration of the hydrogen chloride gas ranges between 1–5% during the pregrowth etch. Lower hydrogen chloride gas concentrations may not properly remove all the contaminants and surface defects from the substrate. Higher hydrogen chloride gas concentrations may produce a rough surface morphology or pits on the substrate, which may cause lateral growth of the epitaxial layers and create random nucleation sites throughout the surface of the substrate. The temperature during the etch ranges between 1200°–1500° C. Lower temperatures will not properly eliminate unwanted heterogeneous nucleation sites. Temperatures greater than 1500° C. will too rapidly etch the substrate surface around the peripheral edge of the substrate and introduce unwanted heterogeneous nucleation sites upon the surface of the substrate. Other pregrowth treatments, such as oxidation or reactive ion etching, may also be used to further remove potential unwanted nucleation sites prior to growing the crystal epilayers.

Referring now to FIG. 3, nucleation sites **70** on substrate **24** may be formed for intentional heterogeneous nucleation. Nucleation sites **70** can be formed by an intentional localized alteration of the surface of substrate **24**. These sites **70** can be formed by indenting substrate **24** with a diamond scribe at a predetermined location, damaging substrate **24** by using an electrical probe striking an arc between the surface, and/or implanting a single-crystal whisker of some desired polytype on substrate **24**. If 3C—SiC crystals are to be grown on a 6H—SiC substrate **24**, as shown in FIG. 8, a whisker of 3C—SiC polytype crystal is implanted at nucleation site **70** in a desired growth region **62** of substrate **24**, as illustrated in FIGS. 3 and 7. For growing SiC epilayers, the optimum location for nucleation site **70** is at the corner of growth region **62** and on the topmost terrace or step **72**, as illustrated in FIG. 3. As illustrated in FIGS. 3 and 7, the surface of substrate **24** comprises several steps **72** of crystal layers. The multiple steps **72** are formed when the substrate is cut at any non-zero angle relative to the basal plane. For instance, if the substrate surface is tilted at an angle α of 3° (relative to the basal plane), as illustrated in FIG. 4, a substrate surface will cut across several crystal layers to form multiple crystal layer steps **72**. Nucleation site **70** should be located on the uppermost step **72**, as illustrated in FIG. 3. The optimum direction of the substrate tilt is along the diagonal **D** of growth region **60**, as illustrated in FIG. 3.

Once substrate **24** has been pretreated, reaction chamber **22** is prepared for crystal growth. Reaction chamber **22** is preferably evacuated by vacuum via vacuum line **V** and subsequently purged with an inert gas to remove impurities. Hydrogen gas may be used to purge the reaction chamber. Once the reaction chamber is purged, the carrier gas flow rates and the temperature within the reaction chamber are brought to equilibrium. Hydrogen gas is preferably used as the carrier gas, but other gases (e.g. inert gases) can be used. Once the temperature and flow within the reaction chamber have reached equilibrium, generally within less than one minute, silane and propane are added to the carrier gas to initiate SiC growth. Preferably, the silane concentration within the carrier gas is approximately 200 ppm resulting in a 200 ppm atomic concentration of Si. The amount of propane introduced into the carrier gas is approximately 130 ppm to 600 ppm resulting in a atomic concentration of C

between 390 ppm to 1800 ppm. The prescribed pretreatment of substrate **24** allows for significantly greater deviations from the optimum Si/C ratio than was previously thought possible for growing high-quality, low-defect SiC crystals. The ratio of the atomic concentrations of Si to C may be varied to create different growth rates and different types (i.e. n or p-type) of SiC epilayers. The ratio may range between 0.01–1.0 and preferably is between 0.1–0.5.

Referring now to FIG. 4, there is shown an atomic-scale cross-sectional drawing of 6H—SiC substrate **24** comprising several layers of 6H—SiC film **41** and several epitaxial 6H—SiC film layers **40** deposited on the surface of the substrate. SiC epilayer growth rates from a carrier gas containing 200 ppm silane and 600 ppm propane resulted in a vertical epilayer film growth rate parallel to the c-axis of about 5.5 micrometers per hour. Once the crystal begins to grow, multiple layers **40** will form on top of each other thus producing a multiple-layer SiC epitaxial film, as shown in FIG. 4.

FIGS. 7 and 8 illustrate the growth of 3C—SiC epilayers **40a** on substrate **24** made of 6H—SiC epilayers **41**. As the crystal growth continues over time, the 3C—SiC epilayers **40a** grow laterally from 3C—SiC nucleation site **70a** until the 3C—SiC epilayers (film layers) **40a** completely cover the growth region. The growing of 3C—SiC boundary film layer **40a** on 6H—SiC film layer **40** is called a heteroepitaxial layer. A heteroepitaxial layer is an epitaxial film layer that is of a different material, different polytype, or is lattice-mismatched from the film layer upon which it grows. The heteroepitaxial layer is under stress due to compression or tension along the plane between the two different polytype epilayers. Although there is stress between the two polytype epilayers, the 3C—SiC film layers **40a** have few, if any, DPBs and stacking faults because nucleation of the 3C—SiC epilayer takes place at one location.

Referring now to FIG. 9, during the growth of the SiC epilayers Si atoms are deposited in Si sites **80** and C atoms are deposited in C sites **82**. As a result, the SiC epilayers are formed by layers of Si atoms **86** and C atoms **84** forming on substrate **24**. One epilayer or film layer is represented as one double-stack layer of a Si atom layer **86** and C atom layer **84**. The SiC epilayers grow in stacking alternating layers of Si layers **86** and C layers **84**, which increase the thickness of the SiC crystal. The type of SiC epilayer grown on the substrate can be controlled by nucleation sites and/or the polytype of the substrate. As illustrated in FIG. 4, 6H—SiC epilayers **40** are grown on a 6H—SiC polytype substrate **24**, which is referred to as homoepitaxial growth. As illustrated in FIG. 8, 3C—SiC epilayers **40a** are grown on a 6H—SiC polytype substrate using 3C—SiC nucleation site **70a**, which is known as heteroepitaxial growth.

FIG. 9 is an atomic scale schematic illustration of Si atom layers **86** and C atom layers **84** forming on substrate **24**. Si-face **50** is prepared for SiC epilayer growth resulting in C atom layer **84** being the first atom layer to form on the surface of the substrate. The growth of C atom layer **84** on Si-face **50** results in the growth of the highest-quality, lowest-defect SiC epilayers. During the growth of the SiC film layers, there are a number of different types of atoms or molecules at or near the film surface. H₂ represents the carrier gas in the reaction chamber. C_γ and Si_χ represent a C atom and a Si atom respectively which have dissociated from their respective precursors and are in the process of depositing themselves in their respective C sites or Si sites. X, Y and Z represent impurities which intentionally or unintentionally exist in the reaction chamber. Specifically, X represents a dopant that competes with an Si_χ (Si atom) for

Si site 82 and Y represents a dopant that competes with a C_γ (C atom) for C site 80. Z is a contaminant that was not removed during the purging of the reaction chamber or was unintentionally introduced into the reaction chamber. Z may compete for C site 80, Si site 82, both sites or no site.

During the growth of the SiC film layer, the amount and type of impurity incorporated into the film layer is controlled to produce a SiC crystal with specifically designed properties. Dopants can be intentionally added to alter the electrical and/or optical characteristics of the SiC crystals. Phosphorous and nitrogen dopants can be added to the SiC film layers to form n-type layers and aluminum and boron can be added to SiC film layers to form p-type layers. When an n-type SiC epilayer is formed, phosphorous or nitrogen, as represented as Y in FIG. 9, competes with C_γ for C sites 80 during the formation of the C atom layer 84. Similarly, p-type dopants such as aluminum and boron, as represented by X in FIG. 9, compete with Si_α for Si sites 82 during the formation of the Si atom layer 86. The impurity incorporation into each atom layer is significantly affected by the ratio of Si/C in the reaction chamber. During the growing of a p-type film layer, wherein aluminum, sodium, iron, gallium, titanium, vanadium, or boron is incorporated into Si atom layer 86, the amount of p-type dopant incorporated is increased by decreasing the Si/C ratio in the reaction chamber. Conversely, the amount of p-type dopant incorporated into Si atom layer 86 is decreased by increasing the Si/C ratio in the reaction chamber. Similarly, when a n-type film layer is to be grown, wherein oxygen or nitrogen is incorporated into C atom layer 84, the amount of n-type dopant incorporation increases as the Si/C ratio increases and n-type dopant incorporation decreases as the Si/C ratio decreases. The inventors have determined that for n-type doping, fluorine, chlorine, sulfur and nitrogen incorporate into Si-sites; and oxygen and nitrogen incorporate in C-sites. For p-type doping, sodium, iron, aluminum, gallium, titanium, vanadium, and boron incorporate into Si-sites, and boron incorporates into C-sites.

Site-competition epitaxy is an advancement for the control of dopant incorporation such as for both p-type and n-type doped epilayers, which has resulted in increased doping range and improved doping reproducibility for the growth of chemical vapor deposition (CVD) SiC epilayers. Use of site-competition epitaxy will lead to improved device performance which includes high voltage diodes, ohmic as-deposited contacts, and high temperatures JFETs. The usefulness of site-competition epitaxy relies upon each particular dopant atom substituting primarily for either a Si atom in the Si-site or for a C atom in the C-site of the growing SiC epilayer. Site-competition epitaxy has been successfully used for control of both Al and N doping, partly because Al substitutes for Si in the Si-sites whereas N mainly substitutes for C in the C-sites of the SiC lattice when grown on the Si-face. The silicon-to-carbon (Si/C) ratio within the growth reactor strongly affects dopant incorporation for epilayers grown on the (0001)6H—SiC silicon face (Si-face), SiC epitaxy on SiC(1210) A-face substrates and on 3C—SiC(111) and 4H—SiC(0001) Si-face basal plane substrates and on SiC(0001) C-face substrates.

The site-competition epitaxy process is based on the discovery that the concentration of dopant atoms incorporated into a growing SiC crystal is very sensitive to the Si-source/C-source gas ratio. That is, the concentration of dopant incorporated into a growing SiC crystal can be controllably varied solely by purposely changing the Si/C ratio for each selected single amount of a dopant source in the reaction chamber. The use of site-competition epitaxy

affords production of very high quality p-type SiC single crystal films including carrier concentrations with $p > 5 \times 10^{18} \text{cm}^{-3}$. In fact, concentrations at least as high as $p > 5 \times 10^{20} \text{cm}^{-3}$ have been obtained. The site-competition epitaxy process provides for greater incorporation of p-type such as Al by decreasing the Si-source/C-source ratio. This results in 1) a decreased amount of Si-source to compete with the p-type dopant source for available Si-sites, and 2) an increase in the amount of available Si-site (i.e., from increasing the ratio of available Si-site to C-sites) for enhanced p-type dopant incorporation. This is not possible by using any prior art teachings wherein a fixed Si/C ratio was used because the Si-source/C-source is maintained constant and therefore the relative amount of Si-site is also to be constant. Analogously, films of much lower doping concentrations can be produced by using site-competition epitaxy. For example, Al doping levels can be lowered by increasing the Si-source/C-source ratio, thereby decreasing the relative amount of available Si-sites while simultaneously increasing the amount of Si-source to out compete the Al-dopant source for available Si-sites, both of which favor exclusion of Al from the growing SiC film. The superior doping ranges for n-type doping are also obtained by using site-competition epitaxy to grow single crystal films including carrier concentrations with $n > 2 \times 10^{19} \text{cm}^{-3}$. N-type concentrations as high as at least about $n > 5 \times 10^{20} \text{cm}^{-3}$ have been obtained. In addition, the Si/C ratio can be changed and/or selected at a single ratio to form low dopant concentrations for both p-type and n-type dopants in single crystal films which are less than $1 \times 10^{16} \text{cm}^{-3}$ and have been reproducibly obtained at dopant levels as low as less than about $8 \times 10^{13} \text{cm}^{-3}$. These high dopant concentrations (i.e. $> 5 \times 10^{18} \text{cm}^{-3}$) and low dopant concentrations (i.e. $< 2 \times 10^{16} \text{cm}^{-3}$) were previously unachievable in grown single crystal films. High dopant concentrations could be achieved only by further crystal processing (i.e. ion implantation). However, such post growth crystal processing destroyed or damaged the crystal and further did not create a uniform dopant distribution in the crystal.

Although the inventors do not want to be held to one theory for this physical phenomena, it is believed that the controlled incorporation of dopants into the SiC film layers is accomplished by the manipulation of site competition at C sites 80 and Si sites 82. Although dopant concentrations in the film layers can be controlled somewhat by regulating the amount of dopant introduced into the reaction chamber, very low dopant concentrations, very high degenerate dopant concentrations, sharp p-n junctions or n-p junctions, and/or reproducible dopant concentrations are unattainable by solely regulating the dopant concentration. The additional degree of control over dopant incorporation is carried out by manipulating the demand of a particular dopant at Si site 82 or C site 80 during the film layer growth. During the growth of each SiC film layer, the rate at which a layer of Si atoms 86 and C atoms 84 are formed depends on the availability of atoms and/or molecules that can fill a particle site in the Si atom 86 or C atom 84 layer. When the concentration of C atoms in the reaction chamber is increased relative to the Si atom concentration, the demand for atoms to fill C sites 80 decreases since the available amount of C atoms has increased. As a result, the C atom layer 84 is formed at a faster rate and C sites 80 are disproportionately filled with C atoms. However, the relative increase in C atoms effects a relative decrease in Si atoms available to fill Si sites 82. The relative decrease in available Si atoms results in a slower filling of Si sites 82 and a greater demand for any type of atom or molecule to fill Si site 82.

The greater Si site demand results in an increase in non-Si atoms (i.e. dopants, contaminants) filling Si sites 82 to form Si atom layer 86. Conversely, a relative increase in Si atoms to C atoms in the reaction chamber decreases the demand of atoms or molecules at Si sites 82 and increases the demand of atoms or molecules at C sites 80. The decrease in Si site 82 demand, due to a relative increase in Si atom availability, results in a disproportionately high Si atom occupation in Si atom layer 86. Further, the increase in C site 80 demand reduces the rate at which the C atom layer 84 is formed and increases the amount of impurities relative to C atoms incorporated into C atom layer 84.

The simultaneous control of the multiple growth sites of the crystal can be used to control the impurity profile of crystals irrespective of the impurities contained within the reaction chamber. For instance, during the growth of SiC crystals, a n-type crystal can be grown even if equal amounts of n-type and p-type dopant were present in the reaction chamber. By properly controlling the Si/C ratio (i.e. increasing the Si/C ratio) in the reaction chamber, the simultaneous demand manipulation of Si site 82 and C site 80 is affected whereby Si site 82 demand decreases and C site 80 demand increases. The increase in C site 80 demand results in larger n-type dopant incorporation into C atom layer 84 than p-type dopant incorporation into Si atom layer 86, thereby forming a n-type SiC crystal. The simultaneous growth site manipulation may be possibly used to grow other types of crystals having at least two growth sites.

The phenomena of controlling impurity incorporation are also believed to be affected by growth site physical properties, site geometry and the growth rate of Si atom layer 86 and C atom layer 84. It is postulated that the competition for C site 80 between a C atom and an impurity is in favor of the C atom, and the competition for Si site 82 between a Si atom and an impurity is in favor of an Si atom. In a crystal lattice structure, a particular growth site has a specific geometry into which an atom or molecule can bond. For instance, in Si site 82, the geometry is of the shape of a Si atom. The electron cloud about the Si atom specifically bonds with the atoms surrounding Si site 82 to form a highly stable bond. An impurity which does not have the proper geometry to fit into Si site 82 has a very low probability of bonding. Si site 82 also has a particular charge, due to atoms located about Si site 82, which is receptive to a Si atom or another atom or molecule having a charge similar to a Si atom. Furthermore, there exists an equilibrium reaction between the bonding and unbonding of a particular atom to Si site 82. It is believed that during relatively rapid Si atom layer growth, wherein the Si/C ratio is large, the equilibrium of reaction is in favor of the Si atom. In addition, it is believed that the equilibrium of reaction in favor of the Si atom reduces as the growth rate of the Si atom layer decreases. Therefore, in a situation wherein a Si atom and an impurity simultaneously reach Si site 82, the Si atom has a higher probability to bond at Si site 82 due to its geometry, charge and equilibrium of reaction with Si site 82. The postulated competition for Si atom in Si sites 82 equally applies to C atom competition in C atom sites 80.

The control of the Si/C ratio in the reaction chamber enables the production of quality, reproducible SiC crystals which can be degenerately doped with dopant concentrations exceeding $1.0 \times 10^{19} \text{cm}^{-3}$ or very light dopant concentrations at least as low as $8.0 \times 10^{13} \text{cm}^{-3}$.

FIG. 10 illustrates a dopant profile of an n-p type SiC crystal. The SiC crystal was initially doped with a n-type dopant to produce several degenerate n-type film layers 90 containing n-type dopant concentrations exceeding $1.0 \times$

10^{19}cm^{-3} . The degenerate layers generally are used as connect points between the crystal and metal connects (Ohmic contacts). The dopant profile illustrates SiC crystal film layers 92 containing decreasing amounts of n-type dopant resulting from a decrease in dopant atoms and/or a decrease in the Si/C concentration in the reaction chamber. A sharp n-p 94 junction is illustrated. Junction 94 can be produced by decreasing or terminating the n-type dopant flow into the reaction chamber and introducing or increasing p-type dopant into the reaction chamber. Further, the regulating of the Si/C concentration ratio in the reaction chamber can produce a highly defined, controllable and reproducible n-p junction. The p-type dopant incorporation is increased in film layers 96 by increasing the dopant concentration and/or decreasing the Si/C concentration ratio. The dopant incorporation is increased until p-type degenerate film layers 98 are formed. Alternatively, a p-n junction can be formed by first forming p-type film layers and then forming n-type film layers.

FIG. 11 illustrates a SiC crystal having p-type SiC layer 100 growing on the surface of substrate 24, a n-type layer 102 grown on layer 100 and a p-type layer 104 grown on layer 102 to form a pnp SiC crystal which is the basic building block of all electronic devices. Many different SiC crystals with varying dopant profiles can be made by the inventive method. Controlled and reproducible dopant concentration below $1.0 \times 10^{16} \text{cm}^{-3}$ and degenerate film layers exceeding $1.0 \times 10^{19} \text{cm}^{-3}$ (resulting in "Ohmic as deposited" metal contacts), which were previously unattainable, can be formed by growth site manipulation through the control of Si/C concentration in the reaction chamber. The SiC film layers can be grown on a substrate surface with little or no tilt angle or with large tilt angles. The control of growth site competition by varying the Si/C concentration in the reaction chamber appears not to be affected (within reasonable limits) by the tilt angle of the surface of the substrate. However, the tilt angle has been found to affect the polytype and crystal quality of the grown SiC crystals.

The purity of a crystal grown by a CVD process can be substantially controlled by the manipulation of site competition at particular growth sites of the SiC film layers. During crystal growth, contaminants, illustrated as Z in FIG. 9, can incorporate themselves into the SiC crystal during crystal growth. Purging the reaction chamber with an ultra-purified gas, such as hydrogen, reduces some but not all the contaminants within the reaction chamber. The source of the contaminants can be the precursors for the crystal atoms, impurities which inadvertently leak and/or are introduced into the reaction chamber and/or atoms or molecules within the reaction chamber prior to crystal growth. Preferably, the crystal component precursors are selected so as not to behave as a contaminant source. A common contaminant in SiC crystal growth is aluminum. Aluminum is commonly used as a p-type dopant. Upon entering the reaction chamber, aluminum, due to its metallic properties, has a tendency to remain in the reaction chamber after the doping process has ended. During the growth of a pure SiC crystal, aluminum remaining within the reaction chamber can bond in Si sites 82 and potentially degrade the electrical characteristics of the SiC crystal. The amount of aluminum contaminant incorporation into the SiC crystal can be significantly reduced by increasing the Si/C concentration ratio in the reaction chamber and reducing the Si growth site demand. By identifying the primary contaminant which is incorporated within a SiC crystal and the growth site at which the contaminant bonds, the reduction of contaminant incorporation in the SiC crystal can be controlled by prop-

erly adjusting the crystal growing compound concentration ratio within the reaction chamber to reduce the demand of the contaminant at the growth site.

The inventive method can be applied to the fabrication of semiconductor device structures of many kinds. The technique for control over the doping is far superior to any known conventional doping technique for growing SiC crystals. The technique also provides for abrupt in situ doping during crystal growth and is capable of fully utilizing the doping concentration ranging from degenerately p-type doped (for "Ohmic as deposited" contacts) through near intrinsic (extremely low doped n-type or p-type) up to and including degenerate n-type SiC. The ability to control the dopant in such a predictable and reproducible manner is crucial in the fabrication of most SiC devices. Many modifications of the inventive method are possible; for example, the inventive method could be carried out in an ultrahigh vacuum system (e.g. a molecular beam epitaxial (MBE) system).

EXAMPLE 1

A n-type SiC crystal containing about $1.0 \times 10^{15} \text{cm}^{-3}$ was formed using the improved CVD method. For the method described in this example, a commercial 6H—SiC substrate cut from a 6H—SiC boule was used. The Si-face (0001) of the substrate was polished with a diamond paste and cut with a 25 mm dicing saw to form boundaries on the substrate surface. The substrate was subsequently placed within the reaction chamber and subjected to an etch for about 20 minutes at a temperature of 1375° C. using about 3%–4% HCL gas in a H₂ carrier gas with a flow of about 3 L/min. After 25 minutes, the flow of HCL gas was terminated and the temperature of the reaction chamber was increased to 1450° C. and allowed to come to equilibrium in about 30 seconds. Silane and propane were added to the reaction chamber to begin SiC growth. Silane was added at 200 ppm (Si=200 ppm) and propane was added at 600 ppm (C=1800 ppm) to the H₂ carrier gas to create a Si/C ratio within the reaction chamber of 1:9. A n-type dopant, N₂, was added to the H₂ carrier gas at 40 ppm (N=80 ppm) to begin dopant incorporation into the SiC crystal. A growth rate of 5.5 micrometers/hour was achieved using the improved CVD method. After about one hour, the flow of silane, nitrogen, and propane in the H₂ carrier gas was terminated and the carrier gas was allowed to continue to flow in the reaction chamber for about 10 minutes during the cool down of the reaction chamber.

EXAMPLE 2

A method essentially the same as Example 1, wherein a n-type SiC crystal containing about $6.0 \times 10^{15} \text{cm}^{-3}$ was grown. Silane was added at 200 ppm (Si=200 ppm) and propane was added at 350 ppm (C=1,050 ppm) to the H₂ carrier gas to create an Si/C ratio of 1:5.25. A n-type dopant, N₂, was added to the H₂ carrier gas at 66 ppm (N=132 ppm) to dope the SiC crystal.

EXAMPLE 3

A method essentially the same as Example 1, wherein a p-type SiC crystal containing about $3.0 \times 10^{16} \text{cm}^{-3}$ was grown. Silane was added at 200ppm (Si=200ppm) and propane was added at 350 ppm (C=1,050 ppm) to the H₂ carrier gas to create an Si/C ratio of 1:5.25. A p-type dopant, trimethyl aluminum (TMA), was added to the H₂ carrier gas by bubbling H₂ into liquid TMA (held at 21° C.) such that 10 sccm (standard cubic centimeters) was introduced into the reaction chamber to dope the SiC crystal.

EXAMPLE 4

A method essentially the same as Example 1, wherein a degenerate p-type SiC crystal for Ohmic contacts containing about $2.0 \times 10^{19} \text{cm}^{-3}$ was grown. Silane was added at 200 ppm (Si=200 ppm) and propane was added at 600 ppm (C=1800 ppm) to the H₂ carrier gas to create an Si/C ratio of 1:9. A p-type dopant, TMA, was added to the H₂ carrier gas at 21 sccm to dope the SiC crystal. About five minutes before cooling the reaction chamber, the silane flow was terminated and H₂, propane, and TMA flows were continued.

EXAMPLE 5

A method essentially the same as Example 1, wherein a degenerate n-type SiC crystal for Ohmic contacts containing about $2.0 \times 10^{19} \text{cm}^{-3}$ was grown. Silane was added at 200 ppm (Si=200 ppm) and propane was added at 130 ppm (C=390 ppm) to the H₂ carrier gas to create an Si/C ratio of 1:1.95. A n-type dopant, N₂, was added to the H₂ carrier gas at 1153 ppm (N=2306 ppm) to dope the SiC crystal. About five minutes before cooling the reaction chamber, the propane flow was terminated and H₂, silane, and N₂ flows were continued.

EXAMPLE 6

A method essentially the same as Example 1, wherein a multi-polytype SiC crystal containing a p/n junction is grown. The 6H—SiC substrate is nucleated with a 3C—SiC crystal. The tilt angle of the substrate was 0.2°–0.3°. Crystal growth is begun until the 3C—SiC crystal fans out over the complete growth region. During the initial 3C—SiC growth, degenerate n-type doping, as discussed in Example 5, is carried out. The degenerate n-type SiC epilayers are followed by the growth of low doped n-type epilayer, which are grown substantially the same as epilayers grown in Example 2. The pn junction is formed by terminating n-type epilayers and to begin growing degenerate p-type epilayers, as described in Example 4. The positional distribution of 3C—SiC epilayer mesas versus 6H—SiC epilayer mesas on the substrate was random, and the percentage of 3C—SiC mesas was roughly 50%. A 2000 Å thick aluminum etch mask defining circular and square diode mesas ranging in area from $7 \times 10^{-6} \text{cm}^2$ to $4 \times 10^{-4} \text{cm}^2$ was applied and patterned by liftoff. The diode mesas were etched to a depth of approximately 10 μm using reactive ion etching (RIE) in 80% SF₆:20% O₂ under 300 W rf at a chamber pressure of 250 mTorr. The aluminum etch mask was stripped by wet etch, and then a cleanup dip in boiling sulfuric acid was performed. The samples were wet oxidized for 6 hours at 1150° C. to form SiO₂ at least 500 Å thick. After the wafers had been patterned for contacts, vias were etched in the oxide using 6:1 buffered HF solution. Aluminum was then E-beam deposited and lifted off to complete device fabrication. The 3C diode exhibits rectification to 200 V reverse bias at ambient temperature (25° C.), which represents a 4-fold improvement in 3C—SiC diode voltage handling capability. The breakdown is repeatable (i.e., the curve can be taken numerous times with no change in device characteristics) when the current flowing during reverse breakdown is restricted to less than one milliamp; unlimited current flow results in permanent damage to the diode. On the device tested on the substrate, coronas of microplasmas were observed exclusively around device boundaries during breakdown, suggesting that reverse failures are occurring at the mesa perimeter and are not due to a bulk mechanism. Once the diode was catastrophically damaged by excessive

current flow during breakdown, the multi-microplasma corona was replaced by a single microplasma which presumably was the point of catastrophic device failure along the mesa edge. FIG. 12 illustrates the forward and reverse current voltage characteristics on a logarithmic scale at several temperatures. Although the improvement in reverse leakage naturally depends upon the voltage and temperature selected as a basis of comparison, these 3C—SiC diodes clearly represent at least an order of magnitude improvement in reverse leakage current density over any previously published 3C—SiC pn diode. Because the reverse current is not proportional to the square root of the applied voltage, it is surmised that mechanisms other than thermal generation are responsible for the reverse leakages. The exponential regions of the forward characteristics exhibit record-low saturation current densities for CVD-grown 3C pn diodes. However, the change in ideality factors with temperature is not well understood at this time.

The method described in the above examples illustrates the controlled and reproducible doping of SiC crystals having a smooth surface morphology and a low density of defects. The improved CVD method can possibly be used to grow and control purity and doping profiles of crystals other than SiC having at least two growth sites, even though such crystals involve a different chemistry, growth parameters, dopants, contaminants, etc. from that of SiC crystals. One type of crystal the improved method can possibly be used with is GaAs crystals. The following prophetic examples illustrate the controlled doping of gallium arsenic (GaAs) crystals.

EXAMPLE 7

For the method described in this example, a degenerate p-type GaAs crystal containing at least $1.0 \times 10^{19} \text{cm}^{-3}$ for Ohmic contacts is to be formed. A GaAs substrate is polished, growth boundaries cut and the surface etched prior to crystal growth on the substrate. The reactor chamber is brought to the proper crystal growth temperature and the carrier gas flow is to be brought to equilibrium in the chamber. Gallium and arsenic containing compounds are to be combined with the carrier gas to begin GaAs crystal growth. For the case of a p-type dopant which preferentially competes for the Ga site, the gallium concentration is to be decreased relative to the arsenic concentration, which increases the demand of Ga growth sites and, therefore, increases the incorporation of the p-type into the GaAs crystal. The amount of p-type dopant to be introduced into the reaction chamber should be the maximum amount allowed which does not adversely affect the GaAs epilayer surface morphology. Five minutes before reactor cooling, the gallium compound flowing into the reactor should be terminated and the flow of the p-type dopant, arsenic compound, and carrier gas continued.

EXAMPLE 8

A method essentially the same as prophetic example 7, wherein a degenerate n-type GaAs crystal containing at least $1.0 \times 10^{19} \text{cm}^{-3}$ is to be grown. For the case of a n-type dopant which preferentially competes for the As site, the gallium concentration is increased relative to the arsenic concentration to increase the As site demand to increase the incorporation of the n-type dopant into the GaAs crystal. The amount of a n-type dopant to be introduced into the reaction chamber should be the maximum amount allowed which does not adversely affect the surface morphology of the GaAs crystal. Five minutes before reactor cooling, the

arsenic compound flowing into the reactor should be terminated and the flow of the n-type dopant, gallium, and carrier gas continued.

EXAMPLE 9

A method essentially the same as prophetic example 7, wherein a p-type GaAs crystal containing much less than $1.0 \times 10^{19} \text{cm}^{-3}$ is to be grown. The gallium concentration relative to the arsenic concentration is increased to create a relatively large Ga/As ratio (relative to the Ga/As ratio of example 7) in the reaction chamber. An appropriate amount of p-type dopant is then added to the reactor.

EXAMPLE 10

A method essentially the same as prophetic example 8, wherein a n-type GaAs crystal containing much less than $1.0 \times 10^{19} \text{cm}^{-3}$ is to be grown. The gallium concentration is to be decreased in the reactor chamber such that a relatively small Ga/As ratio (relative to the Ga/As ratio of example 8) exists. An appropriate amount of a n-type dopant is then added to the reactor.

EXAMPLE 11

The SiC epilayers were grown at 1450°C . on commercially available n-type Si-face SiC substrates polished 3° off-axis from the (0001) plane toward the (1210) plane. The SiC substrates were placed onto a Si—C-coated graphite susceptor and loaded into a water-cooled quartz reactor of an atmospheric pressure CVD system. The SiC epilayers were grown using silane (3% in H_2) and propane (3% in H_2) with a hydrogen carrier gas, resulting in $\sim 3 \mu\text{m/h}$ growth rates, and doped using either nitrogen (for n-type) or trimethylaluminum (TMA) (for p-type). The epilayers were characterized using secondary ion mass spectrometry (SIMS) analysis, mercury-probe or p-n diode capacitance voltage (C-V), low-temperature photoluminescence spectroscopy (LTPL), and Hall mobility measurements. Site-competition epitaxy was used by (dopant control technique) adjusting the Si/C ratio within the growth reactor to effectively control the amount of dopant incorporated into substitutional SiC crystal lattice sites. These sites, i.e. carbon lattice sites (C sites) or silicon lattice sites (Si sites), located on the active growth surface of the silicon carbide crystal were manipulated by site-competition epitaxy. The model for site-competition epitaxy was based on the principle of competition between nitrogen and carbon for the C sites and between aluminum and silicon for the Si sites of the growing silicon carbide epilayer. The concentration of n-type (nitrogen) dopant atoms incorporated into a growing silicon carbide epilayer were decreased by increasing the carbon-source concentration so that C out competed N for the C sites. Analogously, the amount of p-type dopant (aluminum) incorporated was decreased by increasing the silicon-source concentration within the growth reactor so that Si out competed Al for the Si sites. A series of p-type and n-type doped single crystals were grown by varying the propane concentration to effectively change the Si/C ratio, while maintaining a constant silane and dopant-source concentration. For the p-type doping of crystals, the propane concentration was the only parameter varied (400–000 at. ppm) to effectively change the Si/C ratio within the growth reactor while the silane (200 at. ppm) and TMA (2 sccm H_2 flow through a constant room-temperature TMA bubbler) flows were held constant. The p-type doping revealed the relative increase in silicon concentration (from an increase in the Si/C ratio) resulted in the exclusion of aluminum from a growing SiC epilayer. It

was found that using a Si/C ratio of 0.1 yielded degenerately doped p-type epilayers with net carrier concentrations of $p > 1 \times 10^{19} \text{ cm}^{-3}$. Additional SiC crystals were grown using a Si/C ratio of 0.5, which resulted in p-type epilayers with net carrier concentrations of $p = 5 \times 10^{16} \text{ cm}^{-3}$. It was found that as the Si/C ratio was increased from 0.1 to 0.5, the relative amount of Si increased and out competed the Al for available Si sites of the growing SiC lattice, which ultimately resulted in decreased Al dopant incorporation.

During the growth of n-type doped epilayers, the silane (200 ppm) and molecular nitrogen (90 ppm) concentrations were held constant while the propane concentration was varied between 133 and 600 ppm. The nitrogen dopant profiles were characterized using SIMS and mercury-probe C-V. The SIMS profile exhibited variations of atomic nitrogen concentration within the grown epilayer from varying only the Si/C ratio (between 0.5 and 0.1) during epilayer growth. The net carrier concentration profile obtained from mercury-probe C-V for the same sample correlated well with the SIMS profile. The results revealed that site-competition epitaxy controlled the dopant incorporation into electrically active crystal sites of the growing SiC epilayer.

To verify the reproducibility of site-competition epitaxy, a number of crystals were grown using the Si/C ratio extremes (0.1 and 0.5) while maintaining constant Si (200 at. ppm) and N (100 at. ppm) concentrations. By using a Si/C=0.1 ratio, grown crystals consistently resulted in intentionally doped n-type epilayers with net carrier concentrations of $n \approx 3 \times 10^{15} \text{ cm}^{-3}$. Further, by using a Si/C=0.5 ratio, grown crystals consistently resulted in n-type epilayers with $n \approx 3 \times 10^{17} \text{ cm}^{-3}$. It is believed that decreased concentration of carbon relative to silicon (from increasing the Si/C ratio from 0.1 to 0.5) allowed the nitrogen to out compete the carbon for C sites of the growing SiC lattice, resulting in an increased atomic nitrogen incorporation.

Very low-doped epilayers were produced by using site-competition epitaxy to exclude the unintentional dopant atoms contained in the growth reactor from incorporating into the grown SiC epilayers. These unintentionally doped epilayers, grown using a Si/C=0.1, were examined using LTPL spectroscopy to determine the crystalline perfection and relative concentration of dopant incorporated during growth. The LTPL results from our lowest unintentionally doped p-type epilayer resulted in an $I_{77}/P_0 = 150$ and $I_{77}/S_0 = 4.7$, indicating the most intrinsic SiC reported to date. The dopant concentration in SiC was correlated with the I_{77}/P_0 and I_{77}/S_0 peak-height ratios for low-doped SiC, where P_0 and S_0 are the nitrogen donor no phonon line intensities at 3.00 eV and at 2.99 eV, respectively, and I_{77} is the peak intensity of the SiC 77-meV intrinsic phonon replica at 2.947 eV. Using these ratios, a $1 \times 10^{14} \text{ cm}^{-3}$ estimate of unintentional dopant concentration was made by LTPL, which was consistent with the mercury-probe C-V room-temperature measurements of $p < 5 \times 10^{14} \text{ cm}^{-3}$. Site-competition epitaxy was used as a dopant-exclusion growth technique for the production of low-doped SiC epilayers. Analogously, degenerately doped epilayers were produced by applying site-competition epitaxy to promote enhanced inclusion of dopant atoms during SiC epilayer growth. Very thin degenerately doped p-type and n-type contact layers were formed by ceasing the source flow of Si or C, respectively, during the final 30–40 s of epilayer growth and during substrate cooling. This allowed the dopant atoms to incorporate into the topmost growing p-type or n-type epilayer without competition from the Si or C atoms, respectively. Degenerate n-type epilayers were produced by ceasing only the propane flow during the remaining 30–40 s of

epilayer growth, while both the silane (200 at. ppm) and nitrogen (300 at. ppm) flows were maintained.

The metal contacts deposited on both the p-type and n-type degenerate epilayers were "ohmic-as-deposited" (i.e. unannealed) for a number of metals, which include aluminum, titanium, nickel, and molybdenum. Contact resistivities determined for as-deposited molybdenum and titanium were calculated [linear transmission line method (TLM)] to be $p_b < 5 \times 10^{-5} \Omega \text{ cm}^2$ and $p_c < 2 \times 10^{-5} \Omega \text{ cm}^2$, respectively, on both n-type and p-type degenerately doped epilayers. At room temperature, the degenerately doped n-type contact epilayers had measured low-electric-field free-carrier concentrations of $n = 4 \times 10^{18} \text{ cm}^{-3}$ and yielded mobilities near $60 \text{ cm}^2 \text{ V}^{-1} \text{ s}^{-1}$. When incomplete ionization was taken into account, the $n = 4 \times 10^{18} \text{ cm}^{-3}$ low-field free-carrier concentration theoretically translates into atomic nitrogen concentrations in excess of $N_D = 2 \times 10^{19} \text{ cm}^{-3}$. The higher electric fields exerted by the C-V technique (relative to the Hall technique) led to higher measured free-carrier concentrations because of field-enhanced ionization that takes place within the high-field depletion region of a Schottky diode. Relatively lighter-doped n-type epilayers with $n = 8 \times 10^{16} \text{ cm}^{-3}$ had room-temperature Hall mobilities ranging from 247 to $253 \text{ cm}^2 \text{ V}^{-1} \text{ s}^{-1}$ and were found to be less than 10% compensated. Superior SiC electronic devices were produced by utilizing site-competition epitaxy. These devices exhibited superior electrical performances because the increased doping range resulting from utilizing site-competition epitaxy allowed fabrication of SiC devices with much higher blocking voltages and lower parasitic resistances. Included among the device accomplishments were high voltage 6H—SiC diodes (2000 V) and 3C—SiC diodes (300 V) which exhibited the largest 6H and 3C—SiC blocking voltages reported to date.

EXAMPLE 12

The 6H—SiC epilayers were grown on commercially available n-type (0001) SiC Si-face boules-derived wafers in an atmospheric pressure CVD system. The SiC substrates were pre-cleaned using a standard degreasing solution, followed by immersion in boiling sulfuric acid for 10 min. with a final deionized-water rinse and then dried with filtered nitrogen. The cleaned substrates were placed onto a SiC-coated graphite susceptor and then loaded into a water-cooled fused-silica reactor using a fused-silica-carrier. The samples were heated via the radio frequency (rf) coupled susceptor which was temperature controlled at 1450° C . using an optical pyrometer. Silane (3% in H_2) and propane (3% in H_2) were used as the sources for SiC epilayer growth, whereas 3% hydrogen chloride gas in hydrogen was used during a 1350° C . in situ 4 min etch just prior to epilayer growth. All gases were mass flow controlled, including the ultra-pure hydrogen carrier gas which was purified by using a heated-palladium diffusion cell. The epilayers were doped p-type by the addition of diborane (100 ppm or 500 ppm B_2H_6 in H_2) into the reactor during epilayer growth.

Post growth annealing of the crystals was carried out in a CVD-polycrystalline SiC cavity at 1700° C . for 0.5 h in an atmospheric pressure, ultra-pure argon environment. Secondary ion mass spectrometry (SIMS) was performed using a CAMECA IMS-4f double-focusing, magnetic sector ion microanalyzer. Cesium bombardment was used for determination of hydrogen, boron, and nitrogen atomic concentration profiles by using the detector in a negative secondary ion detection mode to monitor (H) and the diatomic species B(+C), N(+C), respectively. The aluminum and higher accuracy boron elemental concentrations were determined by

using oxygen bombardment and a positive secondary ion detection mode.

The amount of boron incorporated into the 6H—SiC epilayers was controlled by using site-competition epitaxy. It was first determined whether the specific dopant atom occupied either the C-site or the Si-site of the SiC lattice. Boron had been reported to occupy both the Si-site and the C-site, which would preclude effective use of site-competition epitaxy for control of boron doping in SiC. However, CVD experimental results indicated that boron preferentially occupied the Si-site of the SiC lattice. It was found that the boron incorporation into the SiC epilayer decreases as the silane concentration increases (i.e. increasing the Si/C ratio), which was consistent with the increased amount of silicon out competing the boron for available Si-sites during growth of the SiC epilayer. Conversely, boron incorporation was increased by decreasing the Si/C ratio by decreasing the silane concentration or, alternatively, increasing only the propane concentration. The propane concentration was varied for effective control of the boron doping in which the Si/C ratio was decreased stepwise during epilayer growth by successive increases in propane concentration while maintaining a constant silane and diborane concentration. The boron incorporation in the epilayer increased as the propane concentration was increased due to the relative decrease in the Si/C ratio within the growth reactor. This relative decrease in silicon concentration enabled the boron atoms to out compete the silicon atoms for more of the available Si-sites on the surface of the growing SiC epilayer, resulting in increased boron incorporation into the SiC epilayer.

To determine the reproducibility of the doping control, numerous boron-doped SiC epilayers were grown during separate 2 h growth experiments, each using a constant Si/C ratio ranging from Si/C=0.1–0.5. For selected epilayers, SIMS determined elemental boron concentration was compared to the net carrier concentration measured using mercury-probe C-V. A typical epilayer grown using a Si/C=0.51 had a C-V measured net carrier concentration of $p=5 \times 10^{15} \text{ cm}^{-3}$ as compared to a net carrier concentration of $p=3.5 \times 10^{17} \text{ cm}^{-3}$ for an epilayer grown using a Si/C=0.11 wherein both crystals grown used an identical diborane concentration (1.6 ppm). Secondary ion mass spectroscopy analysis revealed an elemental boron concentration of $[B]=6.5 \times 10^{16} \text{ cm}^{-3}$ for the lower doped ($p=5 \times 10^{15} \text{ cm}^{-3}$) epilayer and a $[B]=1 \times 10^{18} \text{ cm}^{-3}$ for the more highly boron-doped ($p=3.5 \times 10^{17} \text{ cm}^{-3}$) SiC epilayer. The large increase in boron incorporation resulting solely from a change in the Si/C ratio illustrated the strong dependency of the boron incorporation on the Si/C ratio used during epilayer growth, and the preferential B occupancy of the Si-site vs the C-site. Three boron-doped SiC epilayers, grown using identical Si/C ratios but with different diborane concentrations, were subsequently examined using low temperature photoluminescence spectroscopy (LTPL). Identical Si/C ratios were used in order to eliminate potential effects of different propane concentrations on hydrogen incorporation, thereby isolating the hydrogen incorporation effect to only the change in diborane concentration. The resulting spectra reveal that a significant amount of hydrogen was contained in each of the boron-doped epilayers. The analysis of hydrogen in the crystal indicated that the hydrogen incorporation was directly proportional to the amount of boron incorporated into the SiC epilayer. Epilayers containing stepped increases in boron concentration $[B]$, from only varying the Si/C ratio, were prepared and subsequently analyzed for hydrogen using SIMS. This was done to confirm that hydrogen and

boron incorporation were related, and determine whether the hydrogen was removed or if its optical activity was simply altered as a result of annealing. The increase in hydrogen concentration was observed to correspond with the increase in boron concentration within the epilayer. This indicated that the hydrogen incorporation is directly related to the boron incorporation in the 6H—SiC epilayers. After the 1700° C. anneal for 0.5 h in argon, the sample was again SIMS depth profiled for determination of boron and hydrogen concentration. The results indicated that the hydrogen had diffused out of the SiC epilayer as a result of the 1700° C. anneal. The amount of hydrogen remaining within the boron-doped epilayer was below the hydrogen background concentration ($<2 \times 10^{17} \text{ cm}^{-3}$) in the SIMS instrument. Also, it was noted that boron did not undergo appreciable solid state diffusion as a result of the 1700° C. anneal, which was evidenced by the continued sharpness of the $[B]$ profile. It was found that prior to the anneal, hydrogen was passivating the acceptor atoms of the B-doped SiC epilayers. The increase in carrier concentration, due to the reduction in hydrogen-passivation with post-anneal, was approximately 4.5x, 3x and 3.25x for the three samples with post-anneal net carrier concentrations of $p=9 \times 10^{15} \text{ cm}^{-3}$, $p=1.9 \times 10^{16} \text{ cm}^{-3}$, and $p=6.5 \times 10^{16} \text{ cm}^{-3}$, respectively.

It was found that site-competition epitaxy was effective for the control of B-doping in CVD 6H—SiC(0001) epilayers, with B incorporation inversely proportional to the Si/C ratio used within the reactor during SiC epilayer growth. Both LPTL and SIMS analysis of the boron-doped epilayers indicated that the hydrogen concentration contained in the epilayers increased with increasing boron incorporation.

EXAMPLE 13

The 6H—SiC epilayers were grown on commercially available n-type 6H and 4H (0001) SiC Si-face and C-face 6H (0001) SiC boule-derived wafer in an atmospheric pressure CVD system. The SiC substrates were precleaned using a standard degreasing solution, followed by immersion in boiling sulfuric acid for 10 minutes, with a final deionized-water rinse and then dried with filtered nitrogen. The cleaned substrates were placed onto a SiC-coated graphite susceptor and then loaded into a water-cooled fused-silica reactor using a fused-silica-carrier. The samples were heated via the RF-coupled susceptor which was temperature controlled at 1450° C. using an optical pyrometer. Silane (gas cylinder containing 3% in H_2) and propane (3% in H_2) were used as the sources for SiC epilayer growth, whereas 90 sccm flow of ultrapure hydrogen chloride gas in a 3 sLpm flow of hydrogen was used during a 1350° C. in situ etch just prior to epilayer growth. All gases were mass flow controlled, including the ultra-pure hydrogen carrier-gas which was purified by using a heated-palladium diffusion cell. The epilayers were doped n-type by the addition of phosphine (200 ppm PH_3 in H_2) or nitrogen (0.1% N_2 in H_2) and p-type by the addition of diborane (100 ppm B_2H_6 in H_2) or trimethylaluminum (bubbler configuration) into the reactor during epilayer growth. Secondary ion mass spectrometry (SIMS) was performed using a CAMECA IMS-4f double-focussing, magnetic sector ion microanalyzer. Cesium bombardment was used for determination of hydrogen, boron, phosphorous, and nitrogen atomic concentration profiles by using the detector in a negative secondary ion detection mode to monitor H—, P— and the diatomic species B(+C)—, N(+C)—, respectively. For the 6H— and 4H—Si(0001) Si-face samples, nitrogen dopant incorporation into Si-face epilayers was found to be directly related to

the Si/C ratio within the reactor. The Si/C ratio was purposely varied by changing only the propane flow during the epilayer growth experiment while both the silane and nitrogen flows were maintained constant. The SIMS depth profiles indicated that the atomic nitrogen concentration decreased as the propane flow was increased during epilayer growth. This was consistent with an increased amount of carbon out competing the nitrogen for vacant carbon-sites (C-sites) of the SiC crystal lattice on the growing SiC epilayer surface.

Phosphorous (P) dopant incorporation was determined to be inversely related to the Si/C ratio. As the propane flow was decreased, while maintaining a constant silane and phosphine flow, the SIMS determined P incorporation decreased within the Si-face 6H—SiC epilayer. These changes in P incorporation with variation in the Si/C ratio were consistent with the P atom occupying the silicon-sites (Si-sites) of the SiC crystal lattice. As the propane flow was increased, the amount of vacant Si-sites increases which could allow enhanced P incorporation into the SiC epilayer. The p-type doping on the Si-face was accomplished using trimethylaluminum (TMA) for aluminum (Al) and diborane for boron (B) doping. SIMS analysis revealed that the Al dopant incorporation was inversely related to the Si/C ratio used during epilayer growth on both 6H and 4H Si-face substrates. The SIMS determined Al incorporation increased as the propane flow was increased and also when the silane flow was decreased within the CVD reactor, which demonstrated competition between Si and Al for the Si-sites on the growing Si-face 6H—SiC epilayer.

Similarly, boron doping was found to be related to the Si/C ratio. In addition to boron incorporation, a significant concentration of hole-passivating hydrogen was also incorporated during the growth of the B-doped epilayer. SIMS analysis revealed that the hydrogen could be removed from the epilayer by annealing at 1700° C. in argon for 30 minutes which results in a 3× increase in the mercury-probed net hole concentration for these B-doped epilayers. P-type doping of epilayers grown on C-face 6H—SiC(0001) samples were found to be similar to that determined for the Si-face substrates. Both Al and B dopant incorporation on the C-face exhibited similar dependence on the Si/C ratio. However, both Al and B incorporation was not as great on the C-face substrate when compared to Si-face substrate that was doped during the same growth run. For Al doped C-face epilayers, the SIMS determined Al dopant incorporation was approximately 50 times less than that of the corresponding Si-face epilayer. Similarly, B-doped C-face epilayers also exhibited a 50 times lower atomic B concentration compared to simultaneously B-doped Si-face epilayers. For B-doped Si-face epilayers, hydrogen was detected for the B-doped C-face epilayers, where the variation in the SIMS determined H concentration profile paralleled that of the SIMS determined, stepped B-concentration profile.

It was found that Si-face site-competition for 6H—SiC(0001) was similar to that of the 4H—SiC(0001) substrates. The site-competition effect for Al doping was consistent with Al substantially occupying the Si-sites of the SiC lattice for epilayers grown on both Si-face and C-face 6H—SiC(0001) substrates. A similar site-competition effect was obtained in which B was believed to primarily occupy the Si-sites for B doping on both faces. For n-type doping with phosphorus, the site-competition effect was most consistent with the P atom mainly occupying the Si-sites. This effect is believed to be best explained by considering that the lattice site occupied by each dopant atom as compared to the size of the Si or C atom. Specifically, the non-polar covalent radii

for the elements of interest are: Si(1.17Å); C(0.77Å); Al(1.26Å); B(0.82Å); P(1.10Å) and N(0.74Å). Using atomic size as a first approximation (i.e. by neglecting chemical bonding arguments), both Al and P should substitute for Si and not for C in the SiC lattice, which is consistent with the experimental results. The more closely matched atomic size of B to C suggests that B should substitute mainly for C in the C-sites, which is in direct conflict with experimental evidence. However, a considerable amount of H is simultaneously incorporated into the growing B doped epilayer and therefore a "B—H" complex explains these results. A "B—H" species has a size (1.10Å) more closely matched with that of Si(1.17Å) and therefore should occupy a Si-site, which is in agreement with the experimental results. This B—H complex reveals that the Si-site substitution is favored because of the larger effective size of the proposed "B—H" species.

The N doping on the Si-face was consistent with the site-competition mechanism, in which the relatively small N atoms compete with C atoms for available C-sites. It was also found that the small size of N also enabled N to compete to a less extent for the Si-sites. This effect is most pronounced when the SiC crystal is grown on the C-face of a substrate. The results of using site-competition on the C-face were consistent with the N-dopant atoms preferentially substituting into the Si-sites, with a significant contribution from N incorporation into the C-sites.

Site-competition epitaxy was successfully used for p-type and n-type dopant control on both the Si-face and C-face of 6H—SiC(0001) off-axis substrates. The doping results for the C-face indicated that the site-competition effect was highly dependent on the SiC polarity but not significantly affected by the polytype. The results for both the 6H and 4H SiC Si-face samples were similar whereas the 6H SiC C-face samples yielded significantly different results, specifically for N-doped epilayers. In comparing the results from simultaneous growth of N-doped epilayers on the C-face and Si-face, that the site-competition mechanism was believed to be dependent on the surface structure of the SiC substrate and relatively less dependent on gas phase interactions. For p-type doping, the Al and B dopant incorporation increased as the Si/C ratio was decreased on both polar faces. Similarly, the P dopant incorporation also increased on both the Si-face and C-face as the Si/C ratio was intentionally decreased. This was consistent with the P, Al, and B competing with Si for the available Si-sites. The site preference for each dopant was found to be dependent on its atomic size relative to the size of Si(Si-site) and C(C-site). Here, the P and Al were clearly too large for incorporation into the C-site without changing the lattice spacing. Because a significant amount of H was incorporated with B, the relative size of B to be of a "B—H" complex was considered. For this situation, the effective size of the "B—H" complex was also too large and would not easily fit into the C-sites without significantly altering the SiC lattice dimensions. In contrast, the relatively small size of the N dopant atom allows incorporation into either of the lattice sites. Therefore, N doping on the Si-face results in N incorporation into both the Si-site and the C-site, with N mainly incorporating into the C-site. In contrast, for N doping on the C-face, N mainly incorporates into the Si-sites with relatively less incorporation into the C-sites. Therefore, the site-competition mechanism for N dopant incorporation into both Si-sites and C-sites, is apparently dependent upon the growth surface polarity.

EXAMPLE 14

The 3C—SiC and 6H—SiC epilayers were grown at 1450° C. on commercially available 6H(0001) silicon-face

SiC substrates with tilt angles ranging from 0.2° to 0.6° and 3° to 4°, respectively. The SiC substrates were placed onto a SiC-coated graphite susceptor and loaded into a water-cooled quartz reactor CVD system. The grown SiC epilayers were examples using mercury-probe or pn diode CV to determine active doping concentrations. The Si—C ratio within the growth reactor had a strong influence on intentional and unintentional dopant incorporation of the grown 6H, 3D, and 4H SiC(0001) epilayers. Specifically, the active n-type (nitrogen) carrier concentration was found to be directly proportional to the Si/C ratio, whereas, the active p-type (aluminum) concentration was found to be inversely proportional to the Si—C ratio for epilayer growth on the SiC(0001) basal plane. As the Si/C ratio within the growth reactor was decreased, the active nitrogen concentration in the grown SiC epilayer decreased. When the Si/C ratio was decreased further to Si/C≈0.1, the unintentionally doped epilayers changed from n-type to p-type. As a result, we have obtained both p-type and n-type epilayers with room temperature carrier concentrations of $1 \times 10^{14} \text{ cm}^{-3}$, as measured by both mercury-probed CV and low temperature photoluminescence (LTPL). Previously, the unintentionally doped epilayers produced in our lab were exclusively n-type with the lowest net carrier concentrations typically limited to about $n=2-3 \times 10^{16} \text{ cm}^{-3}$.

Dopants in SiC are believed to occupy specific sites, specifically nitrogen primarily occupies the carbon site on the Si-face while aluminum occupies the silicon site of the SiC lattice on both sides of the crystal face. The relative increase in carbon concentration "out competes" the nitrogen for the C-sites of the growing SiC lattice. The analogous situation exists for an increased Si/C ratio, in which the relative increase in silicon concentration "out competes" the Al for the Si-sites of the growing SiC lattice. 100 ppm of N (50 ppm N_2) was introduced into the growth reactor and then attempted to exclude it from the growing epilayer by only decreasing the Si/C ratio from about Si/C=0.44 to Si/C=0.1. The Si/C of 0.1 results in consistently producing an intentionally doped n-type SiC epilayer with a net carrier concentration of $n=1 \times 10^{15} \text{ cm}^{-3}$. In contrast, growth using the more typical Si/C=0.44 with 100 ppm of N results in n-type epilayers of $n=1-2 \times 10^{17} \text{ cm}^{-3}$. For epilayer growth using the Si/C ratio of 0.1, the increased amount of carbon is believed to have out competed the nitrogen for the C-sites of the growing SiC lattice. P-type epilayers were also produced using this method. For these series of experiments a flow of TMA was introduced into the reactor during epilayer growth. The resulting p-type epilayer grown with a Si/C ratio of 0.44 was measured to be $5 \times 10^{16} \text{ cm}^{-3}$, while the epilayer grown using a Si/C ratio of 0.1 yielded a degenerately doped p-type epilayer with an estimated net carrier concentration of $1 \times 10^{19} \text{ cm}^{-3}$. As the Si/C ratio was decreased from 0.44 to 0.1 the relative amounts of Si competing with the Al for the Si-sites of the SiC lattice also decreased, which resulted in an increased Al incorporation.

Site-competition epitaxy was also successfully used to obtain very abrupt changes in dopant concentrations in SiC epilayers. In conventional CVD systems, the abruptness of the dopant profile is limited by the purging of the dopant-source from the growth reactor. One advantage of epilayer growth using site-competition epitaxy is that more abrupt dopant profiles can be obtained by excluding the remaining unwanted dopant by changing the Si/C ratio along with the dopant-source gas. Conversely, very abrupt, enhanced dopant incorporation can be accomplished for production of highly degenerately doped epilayers. One highly useful example of this was the formation of very thin degenerately

doped p-type and n-type contact layers by ceasing the source-flow of Si or C, respectively, during the last minutes of epilayer growth. Subsequently deposited metal contacts were "ohmic as deposited" for a wide variety of metals on both p-type and n-type epilayers. The contact resistivities were determined for as deposited molybdenum on both n-type and p-type degenerately doped epilayers were calculated to be $p_c < 5E-5 \Omega \text{ cm}^2$ using the linear TLM method. Preliminary Hall measurements on intentionally doped 6H—SiC epilayer samples were carried out. N-type epilayers with $n=8 \times 10^{16} \text{ cm}^{-3}$ were found to be less than 10% compensated with room temperature Hall mobilities ranging from 247 to 253 $\text{cm}^2/\text{V-s}$. At room temperature, the heavily doped n-type contact epilayers yielded mobilities near 60 $\text{cm}^2/\text{V-s}$ and measured low-electric-field free carrier concentrations of $n=4 \times 10^{18} \text{ cm}^{-3}$. When incomplete ionization was taken into account, the $4 \times 10^{18} \text{ cm}^{-3}$ low-field free carrier concentration theoretically translated into atomic nitrogen concentrations in excess of $2 \times 10^{19} \text{ cm}^{-3}$. The "ohmic-as-deposited" state of the contacts precluded conventional C-V profiling of the heavily doped epilayers. For the lowest doped n-type and p-type epilayers, low temperature photoluminescence (LTPL) was employed to determine crystalline perfection and relative dopant incorporation. The LTPL results from our lowest unintentionally doped p-type epilayer resulted in a $I_{77}/P_0=150$ and an $I_{77}/S_0=4.7$. The estimate of unintentional dopant concentration by LTPL ($1 \times 10^{14} \text{ cm}^{-3}$) was consistent with the mercury-probe CV room temperature measurements of $p \approx 5 \times 10^{14} \text{ cm}^{-3}$.

The novel growth method, based on the use of appropriate Si/C ratios during epilayer growth to affect control over dopant concentration, was demonstrated for SiC(0001) basal plane 6H and 3C—SiC. The carrier concentration change as the Si/C ratio was varied from about 0.1 to 0.8. the donor carrier concentration in the grown epilayer was proportional to the Si/C ratio whereas the acceptor carrier concentration was inversely proportional to the Si/C ratio. The resulting surface morphologies and crystal quality were excellent for all Si/C ratios between 0.1 to 0.8.

OVERVIEW OF SOME APPLICATIONS

The electrical device applications of the invention center around the advantageous use of the wider doping ranges that this new growth technique makes possible. Since the invention allows for the incorporation of higher concentrations of dopant into compound semiconductor crystals than was previously possible, any electrical device structure whose performance is affected by the degree of degenerate doping achieved in any region of the device will be influenced by the present invention.

The improved CVD method significantly reduces contact resistances at the Ohmic contacts. Performances of many semiconductor device structures, such as transistors, are dependant on contact resistance. Contact resistance arises due to the physical interface where the metal (which carries electrical signals and power to and from the semiconductor device) contacts (or connects to) a semiconductor surface. The performance of most semiconductor electrical devices are maximized by minimizing the contact resistances. It is a limiting factor in the electrical capability of many devices, such as transistors, which in turn limit the capabilities of electrical circuits and systems made using those devices. It is widely recognized that the resistance of contacts to 6H—SiC or 3C—SiC is the major limiting factor in the electrical capability of many 6H—SiC or 3C—SiC devices respectively. The contact resistances for semiconductor devices are minimized by maximizing the doping in the

semiconductor where it makes physical contact with metal Ohmic contact. Because the improved CVD method enables device quality (i.e. high quality) epilayers of compound semiconductors (like SiC for example) to be grown with higher doping concentrations, contact resistances to these epilayers (and therefore properly designed electrical devices fabricated in these epilayers) can be reduced thereby increasing the performance capabilities of many compound semiconductor based electrical devices and circuits.

The improved CVD method also reduces bulk semiconductor resistances in the grown crystals. Another source of resistance that can affect semiconductor electrical device performance is the resistance of the bulk semiconductor itself. In almost all semiconductor devices, current flows through undepleted regions of the bulk semiconductor. The resistance associated with charge flow through these regions is often significant and can sometimes be an undesired factor which limits device performance. The resistances associated with these regions can be minimized by maximizing the doping of the semiconductor in these regions. The maximization of doping in these regions through concentration levels previously unattainable is again facilitated by this new inventive method.

The present invention can also be used to form Delta-doped semiconductor devices. These structures are based upon very thin, very heavily doped semiconductor layers. The electrical performance of Delta-doped devices is contingent upon the ability to incorporate a maximum amount of dopant into the thinnest possible layer. The ability to better control the level of degenerate doping of compound semiconductors, which is made possible by the improved CVD method, should significantly enhance the performance characteristics obtainable in Delta-doped device structures.

The improved CVD method further can be used to develop semiconductor devices with smaller depletion widths. The depletion widths found within semiconductor devices at various junctions (e.g. pn junctions, metal semiconductor junctions, hetero-junctions) are largely determined by the semiconductor doping at or around the junction. It is well known that the depletion width decreases as the doping concentration increases. There are many semiconductor devices whose performance and/or function relies on heavy doping to obtain narrow depletion widths and the associated physical effects. Zener diodes and Esaki diodes rely on carrier tunneling through a very narrow depletion region made possible by heavy doping on both sides of a pn junction. Low leakage diodes and diode junction charge storage capacitors are based on narrow depletion widths, so as to minimize generation current and maximize charge storage density. Because the improved CVD method enables heavier doping and results in smaller depletion widths, the improved CVD method has the potential to improve the performance of these devices by narrowing the device depletion widths.

The improved CVD method can also be used to develop semiconductor devices having slightly narrower bandgaps. By increasing the degenerate doping in a semiconductor, a physical phenomenon known as band tailing or bandgap narrowing takes place. The improved CVD process can be used to control the degree of degenerate doping in semiconductors used in devices in which bandgap narrowing enhances device performance, such as light-emitting diodes (LEDs) and bipolar transistors.

In addition to the controlled growth of high quality compound semiconductor epilayers with higher doping concentration, the improved CVD process also enables the

controlled growth of high quality compound semiconductor epilayers with lower dopant concentrations than was previously possible. Any compound semiconductor electrical device whose performance stands to gain from the lowered doping concentrations that this improved CVD method enables will benefit. Such devices which will benefit include fundamental semiconductor junctions. Almost all semiconductor devices and circuits contain fundamental semiconductor junctions. Most transistors composed of these fundamental junctions are arranged in various configurations appropriate to accomplish the desired electrical functions (e.g. amplification, switching). Such devices will be greatly improved by the improved fundamental junction doping of the present invention. One fundamental building block of semiconductor electrical device technology is the intimate junction of p-type material to n-type material, more commonly referred to as p-n junction. It is well known to those skilled in the art that the electrical characteristics (such as junction capacitance, junction breakdown voltage, junction leakage current) of the p-n junction are governed by the physical characteristics (i.e. dopant density, defect density) of the lighter-doped side of the junction. The ability to produce a more lightly doped compound semiconductor, while obtaining the desired polarity (n-type or p-type) via this improved CVD method, dramatically broadens the range of electrical operating characteristics possible in devices that incorporate pn junctions. It is well known that the breakdown voltage of a p-n junction diode increases with reductions in the doping of the lighter doped side of the junction. Therefore, a factor of 10 or greater improvement (i.e. reduction) in doping concentrations has been demonstrated by the improved CVD method, when applied to 6H—SiC crystals, could improve 6H—SiC p-n step junction diode blocking voltages from their current demonstrated maximum of slightly more than 1,100 volts to somewhere around 10,000 volts. In a similar manner, the doping of the semiconductor also plays a major role in the electrical characteristics of another fundamental junction, the metal semiconductor junction. Similar improvements with similar ramifications can be expected for properly designed rectifying metal-SiC Schottky diode junctions. The improved CVD method can also be used in the design of transistors and circuits to permit higher circuit operating voltages.

The improved CVD method can also be used to make semiconductors with reduced junction capacitance. It is well known that diode junction capacitance decreases as doping decreases. The lighter doping, made possible by the improved CVD method, can also lead to the reduction in the depletion capacitance of compound semiconductor junctions, which will reduce the parasitic capacitances that can often limit device and circuit performance. One of these factors that limits the switching speed and high frequency performance of planar Field Effect Transistors (FET's) is the parasitic capacitance formed by the drain-to-substrate p-n diode junction. The decreased doping obtainable by this improved CVD method can decrease this junction capacitance and possibly enable compound semiconductor FET's to operate with higher frequencies and switching speeds than previously possible.

This improved CVD method can further be used to form semiconductor devices with high internal resistivity. It is known that resistivity increases as shallow ionization energy dopant concentration decreases. In many semiconductor device and integrated circuit applications, it is advantageous to build devices in high resistivity (low dopant concentration) material. Such material often provides isolation between adjacent devices on the same chip, and the

degree of isolation determines the spacing between adjacent devices within a chip, which is a big concern on integrated circuits with hundreds of thousands of transistors. Since the degree of isolation is in part a function of the semiconductor purity (i.e. doping density and defect density) between the individual devices, the improved CVD method may prove useful in advantageously shrinking device spacing to allow for more devices on a single chip. Clearly, the method is useful for any device or circuit situation where higher-resistivity compound semiconductor epilayers are desired.

The improved CVD method can also be used to grow crystals used in semiconductor devices in which the crystals have increased carrier mobilities. It is known that carrier mobilities increase as doping and defect concentrations decrease. This is due to the fact that there is less potential perturbations (caused by impurities and defects) for a carrier (electron or hole) to "plow into" as it is moving through the crystal lattice under the influence of an electric field. The fact that the improved growth technique enables the growth of purer material with lower unintentional dopant incorporation should increase the carrier mobility in the compound semiconductors. It is well known to those skilled in the art that carrier mobilities directly affect the performance of many devices and circuits, and that increased carrier mobilities are attractive because they enhance the performance of most semiconductor devices, especially transistors. In most transistors, crucial electrical performance factors (switching speed, maximum operating frequency, current carrying capability, gain) are generally enhanced by increased carrier mobilities.

The improved CVD method can be used to grow crystals having increased carrier lifetimes. Carrier lifetime is another physical property of compound semiconductors that can be enhanced by growing purer crystals with lighter background defect concentrations. It is known to those skilled in the art that the bulk carrier lifetime increases as the crystal defect concentration decreases. An improvement in this property could lead to smaller recombination/generation rates within a device resulting in improved performance for certain types of devices, such as minority carrier based devices. Minority carrier based devices, such as bipolar transistors and solar cells for example, could benefit from longer life-times.

Possible industrial uses of SiC crystals grown using the improved CVD method includes semiconductor devices and sensors for use in high temperature environments, such as advanced turbine engines, space power systems, deep-well drilling, advanced automobile engines, etc.; semiconductor devices and sensors for use in high radiation environments, such as found near nuclear reactors; semiconductor devices for power electronics applications, such as will be required for power conditioning electronics for electric vehicles, space power systems, and for electrical actuators on advanced aircraft; semiconductor devices for high frequency applications, such as found in communication satellites, high speed computers, and microwave power transistors used in radar systems; pressure transducer diaphragm material for high temperature and/or corrosive environments; and, light-emitting diodes (LEDs) for use in high temperature environments.

As stated above, a significant improvement in producing doped crystals is the controlling of the contaminant content and/or dopant concentration of a crystal during crystal growth by varying the concentration of crystal source components during crystal growth. Several techniques which have been previously used to dope crystals formed by chemical vapor deposition (CVD) have significant drawbacks and do not form a crystal which can be repeatedly

5 duplicated and has a controlled amount of dopant as the crystals formed by the process disclosed and claimed in the present invention. One of the major limitations to progress in silicon carbide (SiC) crystal growth is the limited, reproducible doping range available. The prior art doping of SiC epilayers with net carrier concentrations of less than $3 \times 10^{16} \text{cm}^{-3}$ or greater than $5 \times 10^{18} \text{cm}^{-3}$ was not possible, until now, in a reproducible, controllable manner. The lower bound on the doping range ($3 \times 10^{16} \text{cm}^{-3}$) limited high voltage electronic applications. The upper bound on doping concentration ($5 \times 10^{18} \text{cm}^{-3}$) necessitated the need for metal contacts to be annealed using specific high temperature processes. In contrast to prior art teachings, the present invention provides for a greatly expanded, reproducible doping range from $1 \times 10^{14} \text{cm}^{-3}$ or less to greater than $1 \times 10^{19} \text{cm}^{-3}$. Very low doped epilayers are needed for production of electronics which can withstand much higher voltages than were previously available. Very high doped epilayers allow for metal contacts which do not require the prior art process of high temperature annealing.

20 The significant improvement in controlling the amount of contaminant and/or dopant in a crystal grown in a vapor deposition process was achieved by manipulating the site competition for a particular contaminant and/or dopant (site-competition epitaxy). Many types of crystals can be grown by the CVD method such as a silicon carbide crystal. The growth of silicon carbide crystals is accomplished by using a source of silicon (Si) and a source of carbon (C), where the Si atoms occupy only Si-sites and C atoms occupy only the C-sites of the silicon carbide (SiC) crystal lattice. The Si sites are also called Si-growth-sites, just as the C-sites are also referred to as C-growth-sites. When a silicon carbide crystal is doped, a dopant atom such as phosphorous, nitrogen, aluminum, boron, etc. substitutionally occupies a Si-site or a C-site in the SiC crystal lattice. It has been found that the amount of dopant which occupies a particular Si-site or a C-site in the SiC crystal lattice can be controlled by manipulating the Si/C concentration ratio within the CVD reaction chamber. In particular, it was found that specific p-type dopant incorporation can be increased or decreased by varying the Si/C concentration ratio while maintaining a constant dopant concentration within the CVD reaction chamber. An increase in the Si/C concentration ratio allows the Si to compete with a dopant atom for the Si-site while a decrease in the Si/C concentration ratio allows the C to compete with a dopant atom for the C-site. By manipulating the Si/C concentration ratio, the amount of dopant incorporated into a growing SiC crystal can be accurately controlled by adjusting the Si/C concentration ratio. The manipulation of the Si/C concentration ratio directly affects the site competition at both the Si site and the C site. It was also found that site competition epitaxy can be used for both sides of a substrate (i.e. Si-face and C-face for Si (substrate)). For SiC crystals, site competition epitaxy has been successfully used for various types of crystal lattices (i.e. GH, 4H and 3C). By using the doping method disclosed and claimed in the present invention, crystals grown by CVD process having a particular dopant concentration can be accurately duplicated in subsequently grown crystals. Prior methods of crystals grown by CVD process failed to reliably produce duplicative crystals having a uniform dopant concentration. By use of site-competition epitaxy, multiple numbers of crystals having an almost identical dopant concentration can be produced having a concentration which is significantly lower (when lower is desired) and significantly higher (when higher is desired) than dopant concentrations previously obtained by prior methods. By properly

treating the substrate onto which the crystal is to be grown, the grown crystal will be of a higher quality with much fewer defects. Such pre-treatment includes etching, polishing and forming grooves on the substrate to form a desired surface for crystal growth. The substrate may be intentionally nucleated so as to manipulate the growth direction of the crystal and to facilitate in the start of crystal growth during the CVD process. The substrate may also be angled to facilitate a certain type of crystal growth.

Further modification of the invention is evident. For instance, the crystal element ratio can be selected to effectively exclude a particular dopant atom from a growing crystal. This can be accomplished within a broad doping range. For example, the exclusion of unintentionally present Al (i.e. in the reactor) from incorporation into n-type epilayers to improve MOS device characteristics were achieved using site-competition epitaxy techniques. The prior art is absent any teachings for doping crystals (i.e. SiC) to exclude compensating dopant-atoms (e.g. Al in N-doped layers) from incorporation into a growing crystal. A $5E16\text{ cm}^{-3}$ nitrogen doped (n-type) epilayer can grow using a specific flow of nitrogen (n-type dopant) for each reactor during epilayer growth. The specific nitrogen flow varies only slightly to allow for variations between growth reactors and other equipment differences. Once determined, this "recipe" then becomes the only specific method for a particular reactor to produce a specific dopant concentration exhibiting a specific electrical character. The incorporation of unintentionally present Al into the crystal adversely compromises the electrical characteristics of the crystal. Until the present invention, these limitations could not be overcome. It has been discovered that doping such as with nitrogen to obtain a $n=5E16$ doping level can be accomplished using a spectrum of nitrogen flows, each coupled with the appropriate Si/C ratio. Therefore, the same n-type doped epilayer (i.e. $5E16$) can be accomplished by using a wide range of specific Si/C ratios with a range of nitrogen flows. Thus, a method for the exclusion of counter-dopant atoms which are typically unintentionally present (residual) in the growth reactor can be achieved. Therefore, a specific Si/C ratio can be chosen which excludes a specific unwanted counter-dopant (aluminum as a p-type counter-dopant) by using a large Si/C ratio while achieving the needed $n=5E16$ n-type doping (i.e. nitrogen doped) level.

This is important if a p-type counter-dopant is unintentionally present (as an impurity) and needs to be excluded from the intentionally n-type doped epilayer. Specifically, aluminum has been reported to adversely affect MOS (metal-oxide-semiconductor) device performance in Si—C-based devices by causing formation of defects suspected in the oxide within the SiO₂/SiC MOS-structure. $N=5E16$ n-type epilayers produced using a relatively large Si/C ratio (to exclude residual Al) improved MOS device performance whereas otherwise identical $n=5E16$ n-type epilayers produced using a relatively smaller Si/C ratio resulted in decreased MOS-device performance. Also, samples which were grown using a large Si/C ratio contained less residual (unintentional) Al whereas large amounts of (unintentional) Al were incorporated into samples grown with relatively small Si/C ratios.

It will be appreciated that multiple "recipes" for crystals such as SiC crystals can be prepared for various crystal doping ranges. For example, it has been established that single CVD grown SiC crystals can be grown and doped using a wide range of Si/C ratios and corresponding dopant flow rates. The concept of using a multiple of Si/C ratios to produce a crystal with a specific amount of dopant incor-

poration is contrary to the teachings in the prior art of selecting a single optimum Si/C ratio and exclusively controlling the dopant concentration by varying the flow rate of dopant into the growth chamber. These "recipes" can be stored in a computer database or the like for dopant control of the grown crystal. As described above, if a SiC crystal is to be grown with n-type dopant incorporation of a specific concentration, a recipe can be selected to grow such a crystal. Further, if a very pure n-type SiC crystal is needed, a specific recipe is selected which adjusts the selected Si/C ratio and n-type dopant flow rate such that the proper amount of n-type dopant is incorporated into the Si/C crystal but contaminants such as Al are excluded from being incorporated in the SiC crystal. Alternatively, if a very pure p-type SiC crystal is required, a recipe is selected for a particular Si/C ratio and p-type dopant flow rate such that the proper amount of p-type dopant is incorporated into the Si/C crystal but contaminants such as N are excluded from being incorporated in the SiC crystal. As can be appreciated, if a certain amount of p-type and n-type dopant is needed in the grown crystal, the crystal element ratios can be set accordingly. The significant discovery that the amount of dopant incorporation and/or contaminant exclusion is dependent on the crystal element ratio is a significant improvement of regulating crystal doping. Not until the invention has there been a method to exclude contaminants from growing in CVD grown crystals. The control method is especially important for low dopant concentration wherein contaminants incorporated into the crystal have a much greater adverse affect to the electrical properties of the crystal.

Another embodiment to the invention is the use of dopant to exclude contaminant incorporation into the CVD grown crystal. For example, a recipe for a crystal was selected to intentionally out-compete dopant atom-A (unintentionally present) by intentionally introducing dopant atom-B found to decrease incorporation of dopant-A into the growing single crystal epilayer. This example has advantages in applications in which certain dopant atoms (present even in background concentrations) adversely affect the performance of a final device. For instance, aluminum and boron are typically trace impurity dopants found in intentionally-nitrogen doped n-type epilayers. In addition to the unwanted electrical-compensation effect of aluminum and boron on nitrogen, aluminum has been related to poor MOS device performance. A solution to this problem of unintentional aluminum incorporation (which leads to poor MOS devices) was the following:

During nitrogen-doped n-type epilayer growth, the use of an additional n-type dopant atom (such as phosphorus) which competes for the same lattice sites as the unwanted p-type dopant atom (i.e. Al) was simultaneously introduced into the reactor with the nitrogen dopant (which incorporates into a site different from the P and Al). Specifically, phosphorus was simultaneously introduced into the reactor during crystal growth to exclude any residual aluminum from incorporation into the n-type (mainly accomplished using nitrogen as the n-type dopant) epilayer. The exclusion of Al was successful because both phosphorus and aluminum compete for the Si-lattice sites in SiC. The incorporation of phosphorus did not adversely affect the crystal because phosphorus is also a n-type dopant atom. The mixture of phosphorus and nitrogen as intentional n-type dopants provided a method to exclude the unwanted Al as a residual, compensating p-type dopant. In addition to phosphorus, fluoride, chlorine and sulfur can also be used to exclude aluminum from Si-sites.

As will be appreciated, other p-type dopants such as boron, sodium, iron, gallium, titanium and vanadium which occupy Si-sites will be excluded during n-type doping by use of n-type dopants which occupy Si-sites such as nitrogen, fluorine, chlorine and/or sulfur. Alternatively, n-type dopants such as nitrogen, fluorine, phosphorus and sulfur will be excluded from Si-sites during p-type doping by use of sodium, iron, aluminum, gallium, titanium and vanadium. By use of this method, a very pure p-type doped crystal will be grown by using a p-type dopant which occupies C-sites such as boron, selecting a Si/C ratio which excludes unintentional n-type dopant incorporation into Si sites, and introducing a p-type dopant which competes for Si-sites (i.e. Na, Fe, Al, Ga, Ti, V) to further exclude n-type dopant incorporation into the crystal.

A brief list of applications which utilize this method are as follows:

1. Use of P (phosphorus) to dope n-type to exclude:
 - a. Al (p-type) from Si-sites (i.e. 1) reduce the electrical compensation of nitrogen doped n-type epilayers; 2) to further hasten the change in polarity during growth of crystal epilayers from p-type to n-type).
 - b. N from Si-sites; especially on C-face samples (also on Si-face and A face samples) because N incorporation from the C-sites could be preferred over N incorporation into the Si-sites.
 - c. N from Si-sites and simultaneously using relatively high Si-source/C-source ratios for better exclusion of N from incorporation into Si-sites.
 - d. B (Boron) from Si-sites to force B mainly into C-sites; 1) to exclude H from incorporation into crystal (H is incorporated only when B is in the Si-site but not when B is in the C-site).
 - e. Ga (gallium), Ti (titanium); V (vanadium), Na (Sodium) from Si-sites.
2. Use of B (boron) to dope p-type to exclude:
 - a. Al from Si-sites, thereby excluding Al (intentional and/or unintentional) from incorporation into crystal.
 - b. N from Si-sites, especially on the C-face but also some on the Si-face, thereby forcing N incorporation into C-sites while excluding N from Si-sites.
3. Use of O (oxygen) to dope n-type to exclude:
 - a. B from C-sites.
 - b. N from C-sites because N incorporation into Si-sites could be preferred over N incorporation into C-sites.
4. Use of P (phosphorous) to dope n-type to exclude:
 - a. Na from Si-sites.
 - b. S from Si-sites.

Another modification of the invention is the selection of a particular crystal face in combination with site-competition epitaxy. In particular, dopant incorporation rates have been found to be dependent on the particular face of the crystal. For instance, nitrogen primarily incorporates into Si-sites on the C-face of the SiC substrate. However, nitrogen primarily incorporates into C-sites on the Si-face and A-face of a SiC substrate. Consequently, nitrogen can be used to exclude p-type dopants such as aluminum from Si-sites when the crystal is grown on the C-face of a substrate. Conversely, nitrogen can be excluded from a p-type crystal by growing the crystal on the C-face and adding Al, Na, Fe, Ga, Ti and/or V in the growth chamber to exclude N from the Si-site.

The invention has been described with reference to a preferred embodiment and alternates thereof. It is believed that many modifications and alterations to the embodiment as discussed herein will readily suggest themselves to those skilled in the art upon reading and understanding the

detailed description of the invention. It is intended to include all such modifications and alterations insofar as they come within the scope of the present invention.

I claim:

1. A method of regulating an amount of at least one non-crystal element deposited in a given growth area of a single crystal formed from at least two crystal elements as said crystal is grown in a growth chamber, said crystal elements comprising a first crystal element and a second crystal element, said crystal having at least two crystal growing sites wherein said first crystal element is deposited in a first growth site and said second crystal element is deposited in a second element growth site, said at least one non-crystal element being competitive for said first site, said method comprising the steps of:
 - a. selecting an amount of said non-crystal element to be deposited in said given growth area;
 - b. selecting a ratio of said first crystal element to said second crystal element, said ratio being dependent on said selected amount of said non-crystal element to be deposited in said given growth area;
 - c. flowing a controlled amount of at least one gaseous crystal element compound through said chamber to grow said crystal, said controlled amount corresponding to said selected ratio of said first crystal element to said second crystal element in said growth chamber, and each of said element compound including at least one of said crystal element;
 - d. selecting a new amount of said non-crystal element to be deposited in said given growth area;
 - e. selecting a new ratio of said first crystal element to said second crystal element to obtain said new amount of said non-crystal element to be deposited in said given growth area; and,
 - f. changing said ratio of said crystal elements in said growth chamber during the growing of said crystal to regulate said amount of said at least one non-crystal element deposited in said competitive crystal growth site at said growth area to obtain said new amount of said non-crystal element being deposited in said given growth area.
2. A method as defined in claim 1, including a step of changing the amount of said at least one non-crystal element being deposited in said first growth site by changing said ratio of said second crystal element to said first crystal element.
3. A method as defined in claim 2, wherein said at least one non-crystal element is a dopant.
4. A method as defined in claim 3, including a step of pre-treating said growth area to remove sites for heterogeneous nucleation during said crystal growth.
5. A method as defined in claim 4, wherein said pre-treating includes pregrowth etching of said growth area.
6. A method as defined in claim 4, wherein said pre-treating includes polishing said growth area.
7. A method as defined in claim 4, wherein said SiC crystal is grown on the growth area of a substrate.
8. A method as defined in claim 7, including a step of nucleating on said substrate to grow a particular crystal polytype.
9. A method as defined in claim 8, wherein said nucleating of said substrate is the introduction of an impurity at said site to stimulate the growth of a desired crystal structure.
10. A method as defined in claim 9, wherein said nucleating of said substrate is positioned in a corner of said growth area.

11. A method as defined in claim 7, wherein said growth area of said substrate has a non-zero inclination relative to a basal plane.

12. A method as defined in claim 11, wherein said non-zero inclination is greater than about seven degrees.

13. A method as defined in claim 11, including a step of dividing said growth area into growth regions defined by boundaries by introducing grooves along the boundaries of said growth regions.

14. A method as defined in claim 7, wherein said growth area of said substrate has a zero inclination relative to a basal plane.

15. A method as defined in claim 1, wherein said growth chamber is heated to a temperature of 800–2200° C.

16. A method as defined in claim 1, including a step of using a non-reactive carrier gas to introduce at least one of said crystal elements into said growth chamber.

17. A method as defined in claim 1, wherein said crystal is an SIC crystal.

18. A method of growing crystals containing a non-crystal element, of a concentration of less than about $1.0 \times 10^{16} \text{cm}^{-3}$, said crystal formed from at least two crystal elements being grown by a process conducted in a growth chamber, said crystal having at least two growth sites and said non-crystal element competitive for a growth site, said method comprising:

- a) identifying said growth site said non-crystal element is competitive for;
- b) identifying the crystal element which deposits in said growth site said non-crystal element is competitive for;
- c) selecting a crystal element ratio of said two crystal elements which contains an amount of the competing crystal element which will compete with said non-crystal element during said crystal growth to form said crystal with said selected amount of non-crystal element; and
- d) flowing said selected amount of crystal elements into said growth chamber to grow said crystal.

19. A method as defined in claim 18, wherein said crystal element ratio is selected to form a crystal having a non-crystal element concentration of less than about $5.0 \times 10^{15} \text{cm}^{-3}$.

20. A method as defined in claim 19, wherein said non-crystal element concentration of less than about $1.0 \times 10^{15} \text{cm}^{-3}$.

21. A method as defined in claim 20, wherein said crystal element ratio is selected to form a crystal having a non-crystal element concentration of less than about $8.0 \times 10^{13} \text{cm}^{-3}$.

22. A method as defined in claim 21, including a step of selecting a second concentration of said non-crystal element; selecting a new crystal element ratio to form said crystal with said second concentration of said non-crystal element; and changing said crystal element ratio to said new crystal element ratio during the growing of said crystal to form a crystal having multiple selected concentrations of said non-crystal element.

23. A method as defined in claim 18, wherein said crystal is an SIC crystal.

24. A method containing a non-crystal element concentration of greater than about $1 \times 10^{19} \text{cm}^{-3}$, said crystal formed from at least two crystal elements being grown by a process conducted in a growth chamber, said crystal having at least two growth sites and said non-crystal element competitive for a growth site, said method comprising:

- a) identifying said growth site said non-crystal element is competitive for;

b) identifying the crystal element which deposits in said growth site said non-crystal element is competitive for;

c) selecting a crystal element ratio of said two crystal elements which contains an amount of the competing crystal element which will compete with said non-crystal element during said crystal growth to form said crystal with said selected amount of non-crystal element; and

d) flowing said selected amount of crystal elements into said growth chamber to grow said crystal.

25. A method as defined in claim 24, wherein said crystal element ratio is selected to form a crystal having a non-crystal element concentration is greater than about $3.0 \times 10^{19} \text{cm}^{-3}$.

26. A method as defined in claim 25, wherein said crystal element ratio is selected to form a crystal having a non-crystal element concentration is greater than about $5.0 \times 10^{19} \text{cm}^{-3}$.

27. A method as defined in claim 26, wherein said crystal element ratio is selected to form a crystal having a non-crystal element concentration is greater than about $1.0 \times 10^{20} \text{cm}^{-3}$.

28. A method as defined in claim 27, wherein said crystal element ratio is selected to form a crystal having a non-crystal element concentration is greater than about $5.0 \times 10^{20} \text{cm}^{-3}$.

29. A method as defined in claim 27, including a step of selecting a second concentration of said non-crystal element; selecting a new crystal element ratio to form said crystal with said second concentration of said non-crystal element; and changing said crystal element ratio to said new crystal element ratio during the growing of said crystal to form a crystal having multiple selected concentrations of said non-crystal element.

30. A method as defined in claim 18, including a step of selecting a second concentration of said non-crystal element; selecting a new crystal element ratio to form said crystal with said second concentration of said non-crystal element; and changing said crystal element ratio to said new crystal element ratio during the growing of said crystal to form a crystal having multiple selected concentrations of said non-crystal element.

31. A method as defined in claim 24, including a step of selecting a second concentration of said non-crystal element; selecting a new crystal element ratio to form said crystal with said second concentration of said non-crystal element; and changing said crystal element ratio to said new crystal element ratio during the growing of said crystal to form a crystal having multiple selected concentrations of said non-crystal element.

32. A method as defined in claim 24, wherein said crystal is an SIC crystal.

33. A method of growing a crystal containing a selected ratio of at least one n-type dopant and at least one p-type dopant, said crystal formed from at least two crystal elements being grown by a process conducted in a growth chamber, said crystal having at least two growth sites and said at least one n-type dopant primarily competitive for a n-type growth site and said at least one p-type dopant primarily competitive for a p-type growth site, said method comprising of:

- a) identifying said crystal element which primarily deposits in said p-type growth site;
- b) identifying said crystal element which primarily deposits in said n-type growth site;
- c) selecting said ratio of n-type dopant to said p-type dopant incorporation;

41

- d) selecting an element ratio which contains a certain amount of said n-type growth site crystal element to said p-type growth site crystal element to obtain said amount of n-type dopant and p-type dopant to be deposited in said crystal during crystal growth, said ratio being dependent on said selected ratio of said n-type to p-type dopant ratio; and
- e) flowing said at least two crystal elements into said growth chamber in said selected ratio amounts.

34. A method as defined in claim 33, wherein said crystal is an SIC crystal.

35. A method of growing single crystals containing a selected concentration of a non-crystal element, said single crystal formed from at least two crystal elements being grown by a process conducted in a growth chamber, said single crystal having at least two growth sites and said non-crystal element competitive for a growth site, said method comprising:

- a) identifying said growth site said non-crystal element is competitive for;
- b) identifying the crystal element which deposits in said growth site said non-crystal element is competitive for;
- c) controlling a crystal element ratio which contains a certain amount of the competing crystal element which will compete with said non-crystal element during said crystal growth so that $\geq 1.0 \times 10^{19} \text{cm}^{-3}$ of non-crystal element is incorporated in said crystal.

36. A method as defined in claim 35, wherein said crystal is an SIC crystal.

37. A method of growing single crystals containing a selected concentration of non-crystal element, said single crystal formed from at least two crystal elements being grown by a process conducted in a growth chamber, said single crystal having at least two growth sites and said non-crystal element competitive for a growth site, said method comprising:

- a) identifying said growth site said non-crystal element is competitive for;
- b) identifying the crystal growth which deposits in said growth site said non-crystal element is competitive for;
- c) controlling a crystal element ratio which contains a certain amount of the competing crystal element which will compete with said non-crystal element during said crystal growth so that $\leq 1.0 \times 10^{15} \text{cm}^{-3}$ of non-crystal element is incorporated in said single crystal.

38. A method as defined in claim 37, wherein said crystal is an SIC crystal.

39. A method of selecting the amount of a dopant element deposited in a given growth area of a single crystal and reducing the amount of a contaminant element deposited in a given growth area of the single crystal, said single crystal formed from at least two crystal elements as said crystal is grown at an area by a CVD process conducted in a growth chamber, said crystal elements comprising a first crystal element and a second crystal element, said single crystal having at least two crystal growing sites wherein said first crystal element is deposited in a first growth site and said second crystal element is deposited in a second element growth site, said dopant element being competitive for at least one of said first and second growth sites, said contaminant element being competitive for a growth site other than said growth site of said dopant element, said method comprising the steps of:

- a) identifying said growth site said dopant element is competitive for;
- b) identifying said growth site said contaminant element is competitive for;

42

- c) identifying a first crystal element which deposits in said growth site said dopant element is competitive for;
- d) identifying a second crystal element which deposits in said growth site said contaminant element is competitive for;
- e) selecting a ratio of said two crystal elements to obtain said selected amount of said dopant element deposited in said competitive crystal growth site and simultaneously reducing said amount of said contaminant element deposited in said competitive crystal growth site; and
- f) flowing said selected ratio of crystal elements into said growth.

40. A method as in claim 39, including a step of changing the rate of said dopant element deposited in said competitive growth site by changing said ratio of said crystal elements being flowed into said growth chamber.

41. A method as defined in claim 40, wherein said crystal element ratio is selected to produce a crystal containing a dopant element concentration of less than $1.0 \times 10^{16} \text{cm}^{-3}$.

42. A method as defined in claim 41, wherein said dopant element concentration is less than $5.0 \times 10^{15} \text{cm}^{-3}$.

43. A method as defined in claim 40, wherein said crystal element ratio is selected to produce a crystal containing a dopant element concentration of greater than $1.0 \times 10^{19} \text{cm}^{-3}$.

44. A method as defined in claim 43, wherein said dopant element concentration is greater than $5.0 \times 10^{19} \text{cm}^{-3}$.

45. A method as defined in claim 40, including a step of varying at least once the ratio of said crystal elements being flowed into said growth chamber during the growing of said crystal to form a crystal having at least two different dopant element concentrations.

46. A method as defined in claim 45, including the step of adding a second dopant element to reduce incorporation of said contaminant element in said crystal, said second dopant element having the same electrical polarity as said dopant element and said second dopant competitive for said growth site said contaminant element is competitive for.

47. A method as defined in claim 46, including the step of adding a second dopant element to reduce incorporation of said contaminant element in said crystal, said second dopant element having the same electrical polarity as said dopant element and said second dopant competitive for said growth site said contaminant element is competitive for.

48. A method as defined in claim 39, including a step of varying at least once the ratio of said crystal element being flowed into said growth chamber during the growing of said crystal to form a crystal having at least two different dopant element concentrations.

49. A method as defined in claim 48, including the step of adding a second dopant element to reduce incorporation of said contaminant element in said crystal, said second dopant element having the same electrical polarity as said dopant element and said second dopant competitive for said growth site said contaminant element is competitive for.

50. A method as defined in claim 39, including a step of adding a second dopant element to reduce incorporation of said contaminant element in said crystal, said second dopant element having the same electrical polarity as said dopant element, and said second dopant competitive for said growth site said contaminant element is competitive for.

51. A method as defined in claim 50, including a step of increasing the flow rate of said second dopant to further reduce said amount of said contaminant incorporated in said crystal.

52. A method as defined in claim 39, wherein said crystal is an SIC crystal.

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