There is a renewed interest in the integration of high mobility III-V channel materials into future generations of low power CMOS (complimentary metal oxide semiconductor) or TFET (tunnel field effect transistor) devices. A myriad of challenges for incorporation of these materials with existing Si VLSI (very large scale integration) processes exist at present but the contact resistance in these devices now dominate over channel resistances and can significantly affect on-state drive currents and operating frequencies. In order to reduce contact resistance, it is desirable to increase the doping density in these materials but the achievable doping density is limited by the low density of states in III-V materials. Active dopant concentrations above $1 \times 10^{20} \text{cm}^{-3}$ are regularly obtained in Si while III-V materials such as InGaAs and GaAs regularly report active dopant concentrations on the order of $0.5-5 \times 10^{19} \text{cm}^{-3}$. Unlike elemental or group IV compound semiconductors, doping in compound semiconductors is more varied as the choice of amphoteric, group IV, or group VI, species can result in n-type activation. This report provides an overview of considerations related to the selection of source dopants, incorporation method, and processing for maximizing n-type doping levels in InGaAs.

**Dopant Selection Considerations in InGaAs**

Optimal species selection of dopants in InGaAs is dependent on a couple of key factors. Knowledge of dopant incorporation method, dopant diffusivity, and activation limits is required to determine the suitability of a given dopant in a desired application. In source and drain doping in modern devices it is generally desirable to limit dopant diffusivity and maximize the activation of the dopants.

**Incorporation method.**—Historically, ion implantation has been an important means of introducing dopants into source/drain regions of MOSFET structures since these regions could be self-aligned to the gate. Regrowth of the source and drain regions has become more popular since the introduction of strained Si channels despite the necessity of extra etch and regrowth steps. Monolayer doping is a more recent doping technique which has been advanced for future 3-D devices requiring conformal doping of structures that are not easily implanted or where source and drain regrowth could be challenging. There are unique considerations for determining the suitability of a given dopant species and the respective dopant incorporation method.

The introduction of dopants via ion implantation is a considerably non-equilibrium process since dopant incorporations well above the chemical solubility of the dopant in the implanted substrate can be achieved. Ion implantation is also a very energetic process that results in a considerable amount of damage in the implanted material due to atomistic collisions. Damage accumulation from implantation can eventually result in the conversion of crystalline material into amorphous material. Amorphization is generally avoided for devices made from III-V materials since amorphous layer regrowth can be very defective. 3-D devices have even more challenges associated with defective regrowth and as result amorphization must be avoided in 3-D devices made from Si and Ge. The bond strength of these materials plays an important role on the implant dose and implant temperature dependent amorphization threshold. Bond-length disordering in the ternary InGaAs compound even results in lower amorphization threshold of ternary compounds relative to the binary constituents. Elevated implantation temperatures from 80–300 °C are generally sufficient to avoid amorphization with the high dose implants needed for source/drain doping and the use of lighter implant species can further reduce implant damage. Fig. 1 Shows post-anneal damage resulting from non-amorphizing and amorphizing implants in InGaAs from Lind et al. Si and S are both more suitable species for high dose n-type implants into GaAs, InGaAs and InAs since these species will result in less implant damage relative to heavier species such as Se, Sn and Te which cause amorphization at lower doses. Electrical activation of the implanted dopants also requires subsequent equilibrium thermal treatment at high fractions of a given compounds melting temperature to anneal out implanted damage and cause implanted dopants to move onto lattice sites.
The necessity of post-implant thermal annealing is a significant disadvantage for ion implanted dopants since the thermal annealing treatments necessary for dopant activation in III-V materials are often complicated by the preferential loss of the group V species.\(^{22,23}\) For the arsenides, it is necessary to either anneal in group V rich ambient or prevent group V loss via dielectric encapsulation.\(^{24-27}\) A variety of dielectric encapsulant films have been used in previous studies. Nitride and oxide encapsulants are commonly used but any film that can prevent As loss during annealing is likely suitable. Practical considerations often dictate that encapsulant films can be selectively etched, have no adverse interactions with the underlying substrate at high annealing temperatures, and can be deposited at temperatures that avoid surface degradation. Ideally, encapsulant films will also have good uniformity to prevent degradation through pinholes.\(^{22}\) Stress from mismatches in the thermal expansion coefficient can also cause crack formation and pathways for As loss. Cracking can be limited by using thin films no thicker than necessary to prevent As out-diffusion and by reducing the mismatch in thermal expansion coefficient. Sub-15 nm ALD films of Al\(_2\)O\(_3\) have been shown to provide adequate surface protection at high annealing temperatures and have good etchable selectivity.\(^{28-30}\)

The complications associated with implantation and annealing of group III-As are not intractable, but the extra steps and difficulties associated with annealing may make growth doping more attractive in these materials.

Growth doping requires extra processing steps in device fabrication to etch and regrow source and drain regions. This method is used to great effect regrow SiGe in silicon p-FETs to introduce uniaxial compressive strain for enhanced mobility,\(^{31}\) but in n-FET devices tensile strain is necessary to result in mobility enhancements. Growth doping in the III-V systems are likely to be preferred over ion implantation even without mobility enhancements from strain since growth-based techniques such as MOCVD and MBE generally result in higher achievable n-type activation than ion implantation.

Monolayer doping is a more recently developed technique but in practice it closely emulates the very early methods of source diffusion of dopants in device fabrication which were eventually replaced by ion implantation. In monolayer doping the dopant species is introduced to the surface of the substrate material and encapsulated before a subsequent thermal anneal step used to drive the dopant from the surface into the bulk.\(^{32-34}\) Monolayer doping of III-V materials will present the similar challenges and disadvantages to activation and diffusion associated with ion implantation due to annealing. Monolayer doping may be advantageous over ion implant, however, since it could be applied to 3-D structures which are hard to dope conformally via ion implantation and does not result in radiation damage. Capping considerations for monolayer doping are likely no different than those in implant except that conformal film deposition techniques will be required in 3-D structures. Previous studies have shown that it is possible to create very abrupt dopant profiles with this technique but dopant selection is likely to play a large role in the observed diffusivity and subsequent post anneal concentration profiles. Suitable chemistry for adsorbing dopant monolayers onto the surface of a given substrate material may also limit dopant selection for this technique.

**Dopant diffusivity.**—Very abrupt source and drain junctions are necessary in modern devices to preserve the electrostatics of short channel devices. This requirement further necessitates the limited diffusivity of dopants in heavily doped regions. Vacancy diffusion generally has a higher activation energy than interstitial diffusion so species that diffuse via vacancy mechanisms will generally have lower diffusivities. \(S\) is known to be a fast diffuser relative to Si which could complicate the creation of shallow junctions. A significant body of diffusion results of n-type dopants in GaAs exists but there are relatively few reports for InGaAs.\(^{35}\) However, the observed trends in diffusivity for a given n-type dopant in GaAs are also observed in InGaAs. Si is generally reported to have limited diffusivity in InGaAs but more recent reports suggest that the diffusion of Si in InGaAs is actually heavily concentrated dependant.\(^{36}\) Historically, Si seems to be an attractive dopant for implantation and growth methods of incorporation while Se, Sn, and Te are likely best suited for growth based techniques due to implant damage considerations.

**Dopant activation.**—N-type dopants in InGaAs may come from either group IV or group VI. Group IV dopants are often referred to as amphoteric. It is generally assumed in literature that amphoteric dopants such as C, Si, Ge, and Sn are limited in activation due to the propensity to exist in both a donor and acceptor configuration and thereby cause self-compensation.\(^{37-41}\) MBE experiments are perhaps the most convincing evidence that group IV dopants, and Si in particular, can be used as both a n- and p-type dopants.\(^{42,43}\) It is not clear if p-type doping in MBE growth from species which are nominally n-type via ion implantation or source diffusion remain p-type with subsequent thermal annealing. If a dopant was perfectly amphoteric it would be expected that no activation would result. In practice, implanted and MOCVD grown GaAs and InGaAs doped with Si and Sn result in nominally n-type activation. C is generally p-type for GaAs and n-type for In\(_x\)Ga\(_{1-x}\)As with \(x > 0.8\) suggesting that there are ternary concentrations where C could be completely amphoteric. Carbon contamination could be problematic in the creation of very lightly doped InGaAs films with MOCVD since carbon-containing organic precursors are often used to grow these films. Ge might also exhibit more amphoteric behavior in the GaAs system based on reports of type conversion of implanted Ge with increased annealing temperatures.\(^{49}\)

Amphoteric compensation may be a downside to the use of group IV dopants in III-V but implanted Si often shows the similar or better activation than implanted group VI dopants such as Se when treated to equilibrium thermal processing.\(^{50,51}\) The chemical concentration limit is generally much higher than the electrically active concentration limit of a given dopant for both group IV and group VI species. In ion implanted InGaAs and GaAs the chemical concentrations of the implanted n-type impurity can often exceed \(1 \times 10^{20}\) cm\(^{-3}\) without deleterious dopant/host segregation or the precipitation of second phases. In these cases, the electrically active impurity concentration is shown to not exceed \(0.5–1.5 \times 10^{19}\) cm\(^{-3}\). This result indicates that both group IV and group VI dopants are electrically limited at high doping levels due to some electrical compensation mechanism. Both group IV and group VI dopants show remarkably similar upper activation limits in a given III-V arsenide. This is surprising since species independent doping limits seem counterintuitive given that group IV dopants are presumably unable to self-compensate as proposed in amphoteric, group IV dopants. Multiple compensation mechanisms have been proposed to explain the limited n-type activation of dopants in InGaAs and GaAs and evidence for and against these mechanisms is discussed in the following section.

The high active carrier concentrations from growth-doped methods have been shown to be unstable with subsequent equilibrium annealing treatments above the growth temperature.\(^{51-55}\) Fig. 2 shows the post anneal electrical activation of ion implanted Si dopants and the deactivation of grown in Si dopants in InGaAs with data taken from.\(^{52,53}\) This result indicates a practical limit on back-end thermal processing. Processing steps which exceed the growth temperature of heavily n-doped regions will likely result in deactivation. The results also indicate that ion implanted Si dopants have the same thermodynamic limit as growth doped material once similar thermal anneals have been performed.

**Discussion of Activation Limiting Mechanisms of Group IV and Group VI Dopants**

There has been much speculation on the part of previous experiments as the nature of electrical self-compensation of n-doped InGaAs and related III-V materials such as GaAs. Three historically hypothesized reasons for upper activation limits of n-type dopants in InGaAs are 1) chemical solubility limited activation for group IV or group VI dopants 2) self-compensation in the case of amphoteric, group IV dopants or 3) defect limited activation. It has been commonly assumed that the maximum achievable n-type doping concentration in InGaAs
and GaAs with group IV dopants is due self compensation and in the case of group VI dopants the activation limits are due to chemical solubility issues or defect complexing.

**Chemical solubility limited activation.**—Limited activation of S, Se, and Te has been attributed to clustering or interstitial, rather than substitutional, occupation in the case of ion implanted GaAs and InGaAs. Early experimenters sought to improve activation of group VI dopants by increasing the occupation of the group V sublattice by group VI dopants by modulating the site stoichiometry with co-implantation. With co-implantation, excess group III species were implanted to generate excess group V vacancies that implanted group VI dopants would preferentially occupy. Most early experimenters report modest improvements with the addition of co-implants but these reports are generally based on sheet number results or from differential Hall effect techniques. It is also not well understood if co-implants could influence dopant diffusion and subsequent changes with sheet numbers are not often converted to carrier concentrations based on SIMS results. Furthermore, the gains in sheet number are generally very modest and much less than the dose of the co-implant species suggesting that co-implantation has limited effectiveness in the III-V arsenides. One deficiency with many previous experiments investing maximum doping limits is that electrical evidence is the only evidence presented but direct evidence of lattice site or interstitial site occupations would provide the best evidence for improved chemical solubility coming from co-implantation. RBS/C studies of high dose Si and S implants into GaAs have indicated that even for doses which show high amounts of electrical compensation there is limited evidence of implanted dopants sitting interstitially and that most of the dopants are substitutional. This does not rule out the possibility of vacancy complexing causing the observed compensation, however, since RBS would not be sensitive the vacancy complexes with substitutional species.

Other classical signs of chemical solubility limited behavior in these dopants are noticeably absent from previous literature. Fig. 3 shows the post anneal SIMS profiles of a 10 keV, \(5 \times 10^{14} \, \text{cm}^{-2}\) Si\(^{+}\) implant into In\(_{0.53}\)Ga\(_{0.47}\)As as a function of annealing time and temperature determined by SIMS. Heavily concentration dependent diffusion is observed at all annealing temperatures. Si concentrations above \(3 \times 10^{19} \, \text{cm}^{-3}\) are shown to be mobile. Dopant clustering is often identified in diffusion studies by the presence of immobile dopant peaks but for high dose Si implants in InGaAs the electrically inactive concentrations are shown to be mobile. Chemical solubility limited activation would also be obvious with improvements in electrical activation occurring with increasing annealing temperature. Generally, increasing the annealing temperature improves the chemical solubility and electrical activation of an implanted dopant. Fig. 4 shows the calculated active carrier concentration of a 10 keV, \(5 \times 10^{14} \, \text{cm}^{-2}\) Si\(^{+}\) implant into In\(_{0.53}\)Ga\(_{0.47}\)As as a function of annealing time for various annealing temperatures. Carrier concentration was determined by post-anneal SIMS and Hall effect as done in previous reports. Fig. 3 indicates that there is no significant dependence of maximum activation level as a function of annealing temperature or temperature. The diffusion and activation observations show no evidence for chemically limited solubility of Si in InGaAs. Very high dose Sn implants in GaAs have been shown to cause problems with precipitation of impurities and other classic signs of chemical solubility limits but the onset of compensated electrical activation is well before the formation of precipitates. Arguments for chemical solubility limited activation of n-type dopants have perhaps the least convincing evidence given the lack of increasing dopant solubility with increasing anneal temperature or the lack of evidence of large amounts of immobile Si clusters in post anneal SIMS of Si in InGaAs or GaAs or the results of RBS/PXIE experiments indicating that for the most part Si and S is substitutional in GaAs even for heavily compensated doses.

**Figure 2.** Comparison of Si activation as function of annealing temperature for ion implanted and growth doped In\(_{0.53}\)Ga\(_{0.47}\)As. Data is for 10m anneals taken from the reports by Lind and Watanabe. Thermal annealing treatments above the growth temperature deactivate MBE doped substrates. Ion implanted substrates activate to the same stable deactivation limit (\(\approx 1.5 \times 10^{19} \, \text{cm}^{-3}\)) for grown in Si dopants.

**Figure 3.** As-implanted and post anneals SIMS of of a 10 keV, \(5 \times 10^{14} \, \text{cm}^{-2}\) Si\(^{+}\) implant into In\(_{0.53}\)Ga\(_{0.47}\)As for various annealing temperatures and times. The shouldering in the post-annea SIMS profiles are consistent with highly concentration dependent diffusion. Si concentrations above \(3 \times 10^{19} \, \text{cm}^{-3}\) are observed to be highly mobile.

**Figure 4.** Active carrier concentration of a 10 keV, \(5 \times 10^{14} \, \text{cm}^{-2}\) Si\(^{+}\) implant into In\(_{0.53}\)Ga\(_{0.47}\)As as a function of annealing time. Annealing temperatures range from 550–750°C. Active carrier concentration was determined from post anneal SIMS, some of which are shown in Fig. 3, and the measured sheet number from van der Pauw Hall effect.
Amphoteric limited compensation.—Attempts to modulate group IV dopant sublattice occupation with co-implants of either group III or group V species to improve n or p-type activation have shown similar results to those of co-implant experiments seeking to improve group VI dopant solubility. Generally, the reported improvements in electrical activation from co-implantation of a group III or group V dopant with a group IV dopant is much less than the co-implant dose and some studies indicate that there is no improvement in activation from co-implantation. Without direct measurement of Si sublattice occupation it is impossible to say that co-implantation is completely ineffective at modulating the degree of amphoteric behavior of a group IV dopant but there appears to be limited practical improvement in electrical activation. Dual implant studies of group IV and group VI species or multiple group IV dopants suggest that the total activation of dopants in InGaAs and GaAs is not the sum of the individual electrical solubility of the group IV dopants and the group VI dopants. This result seems surprising given that group VI dopants are electrically active and potentially influences site selection of group IV dopants. For intermediate Si concentrations in GaAs the background doping concentration has been shown to effect the site occupation of Si with p-type substrates preferring the SiGa donor configuration and n-type substrates preferring the SiAs acceptor configuration suggesting that Si lattice site behavior is modulated predominately by the background electrical characteristics rather than dopant species.

Amphoteric site occupation and electrical compensation at high Si doping concentrations also does a poor job of explaining Si diffusion behavior at high concentrations in InGaAs and GaAs. DFT calculations and quantum well intermixing studies suggest that Si diffuses primarily on the group III sublattice. Si shows heavily concentration dependent diffusion but there is limited theoretical prediction that SiAs donors or inactive SiGa-SiAs next-nearest neighbor neutral pair configurations are mobile in GaAs despite experimental evidence indicating that heavily compensated Si is highly mobile and inactive in InGaAs. The plots of Si diffusion in Fig. 3 are indicative of heavily concentration dependent diffusion where Si concentrations above $3 \times 10^{19} \text{cm}^{-3}$ is highly mobile. In this case the active Si concentration is also limited to $1.5 \times 10^{19} \text{cm}^{-3}$. It is impossible to claim that amphoteric dopant behavior limits group IV dopant activation from electrical activation results alone. Direct, quantitative evidence of lattice site occupation of group IV dopants at high doping concentrations would be necessary to prove that amphoteric behavior sets an upper limit to dopant. Currently, RBS and electrical results alone cannot rule out other compensation mechanism such as dopant-defect complexing.

Point defect limited activation.—Perhaps the best evidence against the chemical solubility limited and amphoteric limited compensation of dopants at high doping concentrations is the preponderance of evidence for point-defect limited activation in the III-V arsenides. The columbic attraction of positively charged dopant species with negatively charged vacancies species may result in the creation of inactive vacancy-dopant complexes. The amphoteric native defect model proposed by Wałukiewicz and experimental DFT calculations suggest that with heavy n-type doping the Fermi level shifts toward the conduction band and there is a resultant decrease in the enthalpy of formation for negatively charged $V_{\text{III}}$. The formation energy of these defects is dependent on the background electrical characteristics such that any dopant species which can cause the requisite Fermi level shift would result in lowering of the formation energy of vacancies. Previous reports indicate that Si and Se have nearly the same maximum electrically active carrier concentration in InGaAs and this result seems remarkably coincidental if the two species are compensated via entirely different mechanisms (amphoteric and chemical solubility limited respectively). Vacancy-dopant complexing better explains the species independent and non-additive doping of multiple species in InGaAs and GaAs since vacancy creation and subsequent compensation of donors depends primarily on the Fermi level of the material.

Concentration dependent Si diffusion in InGaAs and GaAs also is shown to increase rapidly at high n-type doping concentrations which is consistent with a large increase in $V_{\text{III}}$ population at high Fermi level shifts. DFT experiments predicts that Si diffusion in GaAs is mediated by the $(S_{\text{Ga}}-V_{\text{III}})$ mechanism. This mechanism would be consistent with the observation of electrically inactive Si that is highly mobile at high concentrations. Large increases in $V_{\text{III}}$ point defects can also explain why some other dopants likely exhibit limited diffusion. Sn and Te are n-type but may prefer diffusion on the group V sublattice. As a result, Fermi level shifts which reduce the formation energy of $V_{\text{III}}$ may not significantly enhance the diffusion of species which do not diffuse on the group III sublattice. Quantum well intermixing studies of AlGaAs/GaAs heterostructure also indicate that diffusion of species which occupy the group III sublattice is enhanced by the introduction of n-type group IV and group VI dopants. Direct evidence of vacancies using STM have been confirmed in n-type GaAs and positron annihilation results also indicate that n-type doping results in higher concentration of vacancies than is observed in un-doped or p-type GaAs. There are also reports of superlattice of GaAs and InAs with n-type doping and dopant-vacancy complexes have been suggested as the cause for the unexpected increases in lattice parameter at high doping concentrations. Microscopic evidence of interstitial loops formed by ion implantation of iso-electronic species have also been shown to be less stable in heavily n-type substrates presumably due to a large number of excess vacancies in the n-type material. Dopant-defect complexing seems to provide the best explanation of the observed equilibrium electrical activation limits and diffusion phenomenon in n-InGaAs substrates and there is a large body of experimental evidence suggesting that high vacancy concentrations exist in heavily n-type GaAs substrates. In summary, there is little evidence for chemical solubility limited electrical activation in the III-V materials based on dopant behavior beyond indirect electrical activation measurements. Non-equilibrium growth methods such as MBE have been able to preferentially cause Si to occupy group III or group V sites which have been used as evidence for amphoteric behavior of these dopants but there is limited evidence otherwise that amphoteric occupation of group IV dopants on the group III and group V sublattice results in the upper equilibrium activation limits of Si doped GaAs and InGaAs. Instead, the equilibrium activation and diffusion behavior of Si dopants as well as other group VI dopants can be explained consistently by the presence of negatively charged cation vacancy defects which form compensating vacancy-dopant complexes in InGaAs and GaAs.

Effects of Equilibrium Thermal Processing on Dopant Activation and Deactivation in InGaAs

It is well established that growth-based incorporation techniques are capable of producing higher doping concentrations than ion implantation but the historically proposed reasons for this effect are largely speculative. Ion implantation requires subsequent thermal treatment to recover implant damage and incorporate the implanted dopants onto lattice sites. Monolayer doping similarly requires thermal treatment initiate diffusion of dopants from the surface into the bulk. Some have hypothesized that ion implant damage is compensating and can reduce the activation of implanted dopants but more recent studies of Si in InGaAs have indicated that damage does not significantly limit dopant activation. It is suspected that vacancies created from implantation also improve activation at short annealing times by providing more sites for dopants to move onto. Damage arguments for limited dopant activation from implantation lack convincing evidence and more recent result suggest that the thermal treatments used to activate dopants in ion implanted substrates result in the formation of compensating defects. Previous studies have indicated that high doping levels of Si and Se in MBE grown InGaAs substrates are metastable. In particular, Watanabe et al. observed that at annealing temperatures below 700 °C Si deactivation in InGaAs is accompanied by a corresponding increase in lattice parameter which was attributed to the movement of...
Si onto Ga sites or movement from substitutional sites to interstitial sites.\textsuperscript{52} The observed lattice dilation and deactivation might also be explained by the creation of vacancies and subsequent complexing of Si dopants. A similar paper exploring Se deactivation in InGaAs indicated that annealing temperatures greater than 500°C were sufficient to cause deactivation of the incorporated Se dopants and Hall effect studies suggested that the improvements in mobility upon deactivation was due to the formation of electrically neutral impurities.\textsuperscript{51} This result is also consistent with the formation of neutral vacancy dopant-complexes in InGaAs. A study comparing the activation and deactivation of Si in growth doped and ion implanted InGaAs was also performed.\textsuperscript{53} The results of this study are included in Fig. 3 along with the results by Watanabe. This study shows that the upper activation limit of ion implanted Si in InGaAs is the same as the deactivation level of Si doped MBE InGaAs substrates if both substrates receive identical thermal treatments. Ion implantation and monolayer doping require temperatures that are often much higher than the temperatures used to create grown in source and drain dopants. Typical processing temperatures for MBE or MOCVD growth of InGaAs range from 400–600°C while activating anneals for implants regularly require annealing temperatures from 700–800°C to maximize activation and anneal out implant damage. The results of deactivation studies suggest that the higher temperature thermal treatments required to move dopants onto lattice sites and recover damage in ion implanted material likely allows formation of compensating vacancy defects upon post growth thermal annealing of heavily n-doped InGaAs. This explanation may better explain the discrepancy in maximum activation levels in growth doped InGaAs with ion implant techniques which use higher temperature activating anneals under equilibrium conditions.

**Strategies for Maximizing n-type Dopant Activation in InGaAs**

Avoiding amorphization and the regrowth-related defect from ion implantation with high doses is possible with the use of hot implants from 80–300°C and light ions.\textsuperscript{26} The non-conservative nature of implantation may still result in the formation of sub-threshold dislocation loops at high enough doses. Ion implantation is likely to be limited in electrical activation to \( \approx 1.5 \times 10^{19} \) cm\(^{-3} \) for Si in InGaAs and non-equilibrium growth-based doping techniques are likely necessary for achieving doping levels above this limit. There is no obvious benefit to performing ion implants with peak chemical concentrations above \( 2 \times 10^{19} \) cm\(^{-3} \) since there is no evidence for electrically active Si concentrations above \( 2 \times 10^{19} \) cm\(^{-3} \) in InGaAs and the heavily concentration dependent diffusion of Si concentrations above \( 2 \times 10^{19} \) cm\(^{-3} \) may create problems in maintaining shallow junctions at high annealing temperatures. Increasing implant doses will also introduce excess damage.

Vacancy creation at high doping levels may be due to the formation of Frenkel pairs. Injection of interstitials from Frenkel pair formation could have other effects such as enhancing the diffusion of Be or Zn which migrate via interstitial mechanisms in InGaAs.\textsuperscript{116} Other n-type dopants which do not diffuse on the group III sublattice may be less susceptible to highly concentration dependent diffusion since concentration dependent Si diffusion is likely mediated by the presence of \( V_{bg} \), but the creation of Frenkel pairs may still complicate diffusion of other species in InGaAs regions adjacent to the n-type material.

Monolayer doping with S and Si in InGaAs seems to provide similar activation levels to those reported for ion implant but less than what is commonly achieved with growth-based methods.\textsuperscript{107–119} Monolayer doping is potentially advantageous over ion implantation for the conformal doping of 3-D structures and the lack of implant damage. Incorporation of high doses of dopants may prove problematic since the dose being driven into the substrate may be limited by the amount of dopant that can be adsorbed on the surface. S and Si species have been used for monolayer doping in InGaAs and InAs already but the adoption of other dopant species may require further development of gas or liquid phase adsorbates.

It seems likely that the activation limits will be the same for both ion implantation and monolayer doping techniques since both methods require thermal treatments which result in atomistic diffusion of dopant atoms and seemingly inevitable dopant-vacancy complexing. It is unclear if it is possible to perform thermal annealing treatments that can effectively prevent dopant-vacancy complexing while still promoting movement of the dopant species onto host lattice sites required for activation. Techniques such as flash-assisted RTP and laser annealing may have limited effectiveness in increasing activation given the observation of temperature independent activation but these methods may be successful in reducing dopant diffusion and surface degradation.

Both MOCVD and MBE doping methods have shown doping levels above \( 5 \times 10^{19} \) cm\(^{-3} \) in InGaAs with group IV and group VI dopants.\textsuperscript{20–122} MOCVD growth is likely to be preferred over MBE for VLSI applications since MBE methods have lower throughput and require much higher vacuum levels for deposition. There may be some benefit to judicious choice of particular dopants for growth-based techniques as well. Te is known to have surfactant effects and improve crystalline growth quality in InGaAs\textsuperscript{86} but Te also has a strong memory effect which may complicate the formation of abrupt junctions in epitaxially grown layers.\textsuperscript{120} Processing conditions such as growth temperature and the ratio of group III to group V precursor overpressure also significantly affect post growth electrical activation. A survey of the growth doping literature suggests that lower growth temperatures improve the dopant activation in growth doped substrates.\textsuperscript{51–53,121,123–126} Fig. 5 plots the reported maximum Si activation in InGaAs as a function of growth temperature from a number of studies. It is shown that lower temperature growths generally result in higher Si dopant activation. The highest reported active concentrations for Si dopants in MBE were grown at 370°C but very low growth temperatures can reduce the quality of epitaxy in growth doped films.\textsuperscript{123} Growth conditions that can limit the formation of dopant-vacancy complexes will enhance the electrical activity of incorporated dopants. These results suggest that lower growth temperatures may more effectively prevent dopant-vacancy complexing.

**Conclusions**

Processing techniques such as ion implantation and monolayer doping may be limited in activation due to the requirement of thermal annealing to cause dopant incorporation. Device specific...
fabrication and process flow considerations may make a particular dopant incorporation technique or species more advantageous than another if maximum doping concentrations below 1.5 × 10¹⁹ cm⁻³ can be tolerated. N-type doping incorporations above 1.5 × 10¹⁹ cm⁻³ are achieved only with non-equilibrium growth doping. Growth dopant exhibits metastability upon thermal annealing presumably due to dopant-vacancy complexing.

All forms of doping at high concentrations seem to be limited by dopant vacancy compensation mechanism. Improving the activation of ion implanted or monolayer doped substrates will likely need to be achieved by the prevention of dopant-vacancy complexing upon activating anneals. Lower growth temperatures and proper adjustment of dopant-vacancy complexing upon activating exhibits metastability upon thermal annealing presumably due to growth temperature. Thermal annealing on ion implanted or monolayer doped substrates is unlikely to result in deactivation. Understanding of how point defects and in particular, the cation vacancy, can influence dopant diffusion and activation should inform engineering decisions regarding dopant selection and incorporation in device creation.

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