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Design, fabrication, and characterization of solar cells for high temperature and high radiation space applications

Zachary Bittner

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Design, Fabrication, and Characterization of Solar Cells for High Temperature and High Radiation Space Applications

by

Zachary S. Bittner

A Thesis Submitted
in Partial Fulfillment
of the Requirements for the Degree of
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in Materials Science & Engineering

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Design, Fabrication, and Characterization of Solar Cells for High Temperature and High Radiation Space Applications

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__________________________________________
Zachary S. Bittner

Date
Dedication

To my parents, for all the loving and unconditional support they have provided.
Acknowledgments

I am grateful for the assistance and support of my advisor Dr. Seth Hubbard for providing me the opportunity, facilities, and the guidance required to conduct this research. His enthusiasm for the work and eagerness to entertain discussion is unmatched. I am proud to be continuing under his advisement as I work towards a PhD.

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Abstract

In this work, novel III-V photovoltaic (PV) materials and device structures are investigated for space applications, specifically for tolerance to thermal effects and ionizing radiation effects. The first focus is on high temperature performance of GaP solar cells and on performance enhancement through the incorporation of InGaP/GaP quantum well structures. Temperature dependent performance of GaP solar cells is modeled and compared to a modeled temperature dependence of GaAs. The temperature model showed that a GaP cell should have a normalized efficiency temperature coefficient of $-1.31 \times 10^{-3} ^\circ C^{-1}$, while a standard GaAs cell should have a normalized temperature coefficient of $-2.23 \times 10^{-3} ^\circ C^{-1}$, representing a 42% improvement in the temperature stability of efficiency. Both GaP and GaAs solar cells were grown using metal organic vapor phase epitaxy and fabricated into solar cell devices. An assortment of optical and electrical characterization was performed on the solar cells. Finally, GaP solar cell performance was measured in an environment simulating the temperatures and light concentrations seen in sub 1 AU solar orbits, simulating the effects on a solar cell as it approaches the sun. A positive normalized temperature coefficient of $2.78 \times 10^{-3} ^\circ C^{-1}$ was measured for a GaP solar cell, indicating an increase in performance with increasing temperature. In addition, comparing results of GaP solar cells with and without quantum wells, the device without MQWs had an integrated short circuit current density of 1.85 $mA/cm^2$ while the device containing quantum wells has a short circuit current density of 2.07 $mA/cm^2$ or a
12.4% short circuit current increase over that of the device without quantum wells, showing that quantum wells can be used effectively in increasing the current generation in GaP solar cells.

The second focus of this thesis is on the ionizing radiation tolerance of epitaxially lifted off (ELO) InP and InGaAs (lattice-matched to InP) for the purpose of assessing device lifetime in high-radiation Earth orbits. Solar cells are characterized through spectral responsivity as well as illuminated and dark current-voltage (I-V) measurements before being subjected to exposure to a 5 mCi $^{210}$Po $\alpha$ source and a 100 mCi $^{90}$Sr $\beta$ source. Device performance is measured with increasing particle fluences. Previously reported results showed epitaxially grown InP solar cells to generate 76.5% of the beginning-of-life (BOL) maximum power under AM0 at a $1MeV \beta$ fluence of $6 \times 10^{15} \text{e}/\text{cm}^2$[1]. In this study, a degradation to 71.1% unirradiated maximum power was seen at a $1MeV \beta$ fluence of $3.19 \times 10^{15} \text{e}/\text{cm}^2$. This demonstrates that ELO InP cells degrade comparably to bulk InP cells under ionizing radiation. An InGaAs cell was measured under 5.4 $MeV \alpha$ radiation and had a 50% BOL performance point at $4.7 \times 10^9$ $5.4MeV \alpha/cm^2$. The 50% BOL performance point for an InP cell in the same conditions was $1.9 \times 10^{10} \alpha/cm^2$, showing similar degradation at 4x the $\alpha$ fluence.
## Contents

Dedication .......................................................................................................................... iii

Acknowledgments ............................................................................................................... iv

Abstract ............................................................................................................................... vi

1 Introduction ....................................................................................................................... 1
  1.1 SOLAR POWER IN SPACE & III-V PHOTOVOLTAICS ........................................ 1
  1.2 SOLAR CELL OPERATION .................................................................................... 4
  1.3 NANOSTRUCTURES IN PHOTOVOLTAIC DEVICES ........................................... 14
  1.4 SOLAR CELL TESTING METHODOLOGIES ....................................................... 14
  1.5 PRIOR WORK ........................................................................................................... 18
     1.5.1 Gallium Phosphide Photovoltaics ................................................................. 18
     1.5.2 Nanostructured Devices in Optoelectronics .................................................. 19
     1.5.3 Radiation Damage in InP Photovoltaics ....................................................... 20
  1.6 ORGANIZATION OF WORK .................................................................................... 21

2 Gallium Phosphide Photovoltaics ..................................................................................... 22
  2.1 MOTIVATION ............................................................................................................ 22
  2.2 THEORY .................................................................................................................... 24
  2.3 MODELING OF TEMPERATURE EFFECTS ............................................................... 26
  2.4 GROWTH AND FABRICATION .............................................................................. 30
  2.5 EXPERIMENTAL SETUP ....................................................................................... 35
  2.6 RESULTS ................................................................................................................. 36
     2.6.1 Current-Voltage Characteristics .................................................................. 36
     2.6.2 Spectral Responsivity and Electroluminescence ............................................ 39
     2.6.3 Temperature Dependent Performance ......................................................... 44
  2.7 CONCLUSIONS ........................................................................................................ 50

3 Radiation Damage in InP & InGaAs Solar Cells ............................................................... 52
  3.1 MOTIVATION ............................................................................................................ 52
3.2 THEORY ................................................................. 53
3.3 EPITAXIAL LIFT-OFF .................................................. 59
3.4 EXPERIMENTAL SET-UP .............................................. 60
3.5 INDIUM PHOSPHIDE .................................................... 63
  3.5.1 Introduction ....................................................... 63
  3.5.2 \( \alpha \) Irradiation .................................................. 66
  3.5.3 \( \beta \) Irradiation .................................................. 72
  3.5.4 Conclusions ....................................................... 76
3.6 INDIUM GALLIUM ARSENIDE ........................................ 77
  3.6.1 Introduction ....................................................... 77
  3.6.2 \( \alpha \) Irradiation .................................................. 79
  3.6.3 Conclusions ....................................................... 85
3.7 COMPARISON OF InP AND InGaAs ................................. 86
3.8 CONCLUSIONS ............................................................ 89

4 Summary, Conclusions, & Future Work ............................. 90
  4.1 GALLIUM PHOSPHIDE PHOTOVOLTAICS ......................... 90
    4.1.1 Summary & Conclusions ....................................... 90
    4.1.2 Future Work .................................................... 91
  4.2 RADIATION DAMAGE IN InP AND InGaAs SOLAR CELLS .......... 92
    4.2.1 Summary & Conclusions ....................................... 92
    4.2.2 Future Work .................................................... 94

References ................................................................. 96
# List of Tables

2.1 *AM0* illuminated *J-V* characteristics for baseline and QW GaP devices .......................... 37  
2.2 Diode parameters for samples A, B, and C .............................................................. 39  
2.3 Diode parameters for samples A, B, and C .............................................................. 42  

3.1 Beginning of life *AM0* illuminated *J-V* characteristics for cells used in radiation study .............................................................. 63  
3.2 Table of InP dark diode parameters ........................................................................ 71  
3.3 Table of Dark *IV* Parameters for InGaAs cell ......................................................... 83
# List of Figures

1.1 Chart of bandgaps and lattice constants of binary and ternary III-V semiconductors [www.lightemittingdiodes.org](http://www.lightemittingdiodes.org)[2]. ................................................. 3  
1.2 Example diode IV curve with and without illumination (left) and example solar cell IV curve with example parameters. ................................................................. 6  
1.3 Solar Cell equivalent circuit ............................................................... 7  
1.4 AM0 and AM1.5G spectra compared to 6000K black body ........................ 8  
1.5 Visualization of thermalization and transmission events in a solar cell. ........ 9  
1.6 Example detailed balance calculation made using a 6000$K$ reference spectrum, separating out transmission and thermalization loss. ................................. 11  
1.7 Block diagram of TS Space Systems solar simulator at RIT. ....................... 15  
1.8 RIT TSS solar simulator spectrum compared to the ASTM AM0 reference spectrum ........................................................................................................ 16  
1.9 Block diagram of a Spectral Responsivity setup. ...................................... 17  
2.1 Example of type-I heterojunction quantum well superlattice. The quantum wells facilitate sub-host-bandgap absorption of ‘red’ photons .................... 25  
2.2 AM0 spectrum with calculated absorption using the Beer-Lambert law in a 2.5 $\mu m$ GaP film. ................................................................. 27  
2.3 Modeled efficiency of GaP and GaAs devices vs temperature. ................. 29  
2.4 GaP MQW solar cell schematic ............................................................... 32  
2.5 a. Nomarski image of surface of Sample B. b. Nomarski image of surface of Sample C. The inclusion of an AlP BSF in sample C leads to poor surface morphology ........................................................................... 33  
2.6 Picture of fabricated GaP wafer ............................................................. 35  
2.7 AM0 illuminated $J$-$V$ curves for baseline and QW GaP devices. Samples B & C contain quantum wells, while Sample A does not. ............................ 36  
2.8 $J_{sc}$-$V_{oc}$ curves and local ideality factors for baseline and QW GaP devices. Sample A exhibits an ideality factor of 6 at low to moderate voltages. ........ 40  
2.9 External quantum efficiency and integrated short circuit current densities of GaP devices. ................................................................. 41
2.10 AM0 Spectrum shown with TS Space Systems AM0 spectrum. As points of reference, the direct and indirect GaP band-edges are shown along with the InGaP (calibration cell) band-edge.

2.11 Electroluminescence of $0.5\, \text{cm}^2\, p-i-n$ GaP solar cell with $5x\, \text{In}_{0.17}\text{Ga}_{0.83}\text{P/GaP}$ QW.

2.12 Diagram showing a, shallow-distant donor-acceptor pair recombination and b, shallow donor-O complex.

2.13 Normalized measured and modeled AM0 illuminated $J-V$ curve parameters for GaAs $p-i-n$ solar cell.

2.14 Normalized measured and modeled AM0 illuminated $J-V$ curve parameters for GaP $p-n$ solar cell.

2.15 Temperature dependent $I-V$ sweeps measured on GaP solar cell.

2.16 Modeled and measured efficiency of a GaP $p-n$ solar cell in simulated sub-1 $\text{A.U.}$ Solar orbits. The red line is modeled black body temperature of an object in orbit around the Sun as a function of distance. The blue dotted line is the modeled efficiency, and the blue data points are measured normalized efficiencies.

3.1 Diagram depicting atom displacement from radiation damage in zincblende lattice (not to scale).

3.2 Diagram depicting change in diode structure with increasing particle fluence. Shown on the top left is the beginning of life structure. The junction is between the $n^+$ emitter and $p$-base. The end of life structure, shown on the bottom right is the junction formed between the type-converted base and BSF.

3.3 $^{90}\text{Sr}/^{90}\text{Y}$ spectrum from $^{90}\text{Sr}$ source.

3.4 Example cell structure and a picture of the cell mounted on the ceramic sub-strate. Since the substrate is removed, the total structure is thin.

3.5 Can containing $^{210}\text{Po}$ source.

3.6 Fixture used to protect user and environment from $^{90}\text{Sr}$.

3.7 TRIM simulation showing $5.4\, \text{MeV}$ $\alpha$ path through InP. The average stopping range is past $20\, \mu\text{m}$.

3.8 TRIM simulation showing displacements in InP from $5.4\, \text{MeV}$ $\alpha$ particles.

3.9 AM0 $J-V$ curves across increasing $\alpha$ particle fluence in InP.

3.10 InP solar cell parameters vs. increasing $\alpha$ particle fluence.

3.11 Series and shunt resistances of InP solar cell as function of $\alpha$ particle fluence extracted from AM0 $J-V$ curves.
3.12 Dark IV curves and calculated local ideality factor of InP solar cell at each $\alpha$ fluence. .............................................................. 70
3.13 External quantum efficiency of InP solar cell across increasing $\alpha$ particle fluence. ............................................................... 72
3.14 SR degradation at $\lambda_{\text{photon}} = 550 \text{ nm}, 700 \text{ nm}, \text{and } 900 \text{ nm}$ vs. increasing $\alpha$ particle fluence in InP solar cell .......................................................... 73
3.15 AM0 JV curves across increasing $\beta$ particle fluence in an InP solar cell. ................................................................. 74
3.16 InP solar cell parameters vs. increasing $\beta$ particle fluence. .................................................................................. 75
3.17 Series and shunt resistances of InP solar cell as function of $\beta$ particle fluence extracted from AM0 JV curves. .............................................................. 76
3.18 External quantum efficiency of InP across increasing $\beta$ particle fluence. ................................................................. 77
3.19 SR degradation at $\lambda_{\text{photon}} = 550 \text{ nm}, 700 \text{ nm}, \text{and } 900 \text{ nm}$ vs. increasing $\beta$ particle fluence in InP ................................................................. 78
3.20 TRIM simulation showing 5.4 MeV $\alpha$ path through InGaAs. The average stopping range is past 20 $\mu$m. .................................................. 79
3.21 TRIM simulation showing displacements in InGaAs from 5.4 MeV $\alpha$ particles. .......................................................... 80
3.22 AM0 JV curves across increasing $\alpha$ particle fluence in InGaAs solar cell. ........................................................ 81
3.23 InGaAs solar cell parameters vs. increasing $\alpha$ particle fluence. ............................................................. 82
3.24 Series and shunt resistances of InGaAs solar cell as function of $\alpha$ particle fluence extracted from AM0 JV curves. .............................................................. 83
3.25 Dark IV curves and calculated local ideality factor at each $\alpha$ fluence in InGaAs. .............................................................. 84
3.26 External quantum efficiency across increasing $\alpha$ particle fluence in InGaAs. .............................................................. 85
3.27 SR degradation at $\lambda_{\text{photon}} = 550 \text{ nm}, 700 \text{ nm}, \text{and } 900 \text{ nm}$ vs. increasing $\alpha$ particle fluence in InGaAs .............................................................. 86
3.28 SR degradation at $\lambda_{\text{photon}} = 550 \text{ nm}, 700 \text{ nm}, \text{and } 900 \text{ nm}$ vs. increasing $\alpha$ particle fluence in InGaAs .............................................................. 87
3.29 Depiction of InP/InGaAs tandem expected EQE at BOL and EOL based on EQE from single-junction InP and InGaAs devices .............................................................. 88
4.1 Example of an InP-based three junction solar cell. ............................................................. 95
Chapter 1

Introduction

1.1 SOLAR POWER IN SPACE & III-V PHOTOVOLTAICS

Early satellites relied upon chemical potential energy to operate, limiting the operational lifetime of the satellite to the energy stored in the included batteries which was restricted to the weight that could be feasibly launched. Vanguard I, the first artificial satellite to include solar cells, was launched on March 17, 1958. Due to lack of faith in the then untested photovoltaic technology, chemical batteries were included as the main power source of the satellite. The on-board batteries lasted nineteen days, but the Vanguard I continued operating on solar power\cite{4} for six years. This paved the way for use of solar power in space. Due to weight and size constraints, PV is the only feasible method of generating power in space in sub-Jupiter solar distances. Modern satellites and space probes have similar weight constraints to those previously launched, but have much higher power demands, thus requiring higher efficiency PV. This need lead to interest in the development of high efficiency III-V solar cells for space power applications.
Silicon, being abundant, inexpensive, and manufacturable was the first material used for space applications. The International Space Station is still today powered primarily with Si PV. The realization of epitaxial deposition of high quality single-crystalline III-V semiconductor materials in the 1980’s sparked the development of high performance III-V photovoltaics. The gap between confirmed device and theoretical efficiencies for III-V and Si devices became comparable[5]. GaAs, one of the earliest III-V materials to be developed was preferable over Si because it has a direct bandgap, which translates to high absorption up to the band edge. High absorption means that the active device can be significantly thinner, alleviating the problems of a short diffusion length. GaAs devices also exhibited increased radiation tolerance over Si devices[6] which improves PV lifetime outside of low-earth orbit (LEO) where satellites are still partially shielded from ionizing radiation by the Earth’s magnetic field.

Since these devices relied on epitaxial, or crystalline, deposition onto a substrate, the material system was expanded from a single compound to a wide range of compounds drawing from III-V elements such as Ga, In, Al, N, P, As, Sb. The ability to effectively stack different crystalline materials first resulted in the ability to passivate device surfaces using heterojunction surface fields but grew into the ability to grow entire diode junctions on top of each other, electrically connected through tunnel diodes in order to more efficiently convert photons from the entire spectrum into electricity. The theory of which will be covered in the next section. Shown in Figure 1.1 are III-V materials with bandgaps and lattice constants. III-V material growth can be either lattice-matched or strained, but growth of strained layers can induce defects into the
system, degrading performance.

![Chart of bandgaps and lattice constants of binary and ternary III-V semiconductors](image)

Figure 1.1: Chart of bandgaps and lattice constants of binary and ternary III-V semiconductors [www.lightemittingdiodes.org][2].

The wide range of available bandgaps in the III-V material system, the careful selection of bandgaps in that range, and the ability to grow films of different compositions directly on top of each other with near atomic-layer precision has lead to a current world record efficiency ($\eta$) of 43.5% under the terrestrial $AM1.5D$ spectrum at around 400 suns as held by Solar Junction [7]. Current world record devices under the extraterrestrial $AM0$ spectrum perform at around 35% efficiency under $1 – sun$ $AM0$ with manufacturing lot averages near 30%

In addition to the potential for high efficiency, the variety of material properties gives
the III-V material systems the flexibility for optimization of device performance for ex-
treme environments. The focus of this study is on characterizing device performance in the extreme conditions that can be seen in space, primarily in high temperature or high radiation environments. In this thesis, devices grown on GaAs, GaP, and InP substrates were characterized.

1.2 SOLAR CELL OPERATION

The solar cell is fundamentally a diode, or a junction between a semiconductor with impurities in the atomic lattice that either give up an electron (donors) or grab an electron (acceptors). When light of photon energies greater than the bandgap of the material \( (E_g) \) illuminates the diode, an electron-hole pair is generated. When an injected minority carrier diffuses to the diode junction, charge separation occurs due to the built-in diode electric field. This leads to a light injected current \( (I_L) \). When the diode is held at short-circuit, the current collected is called the short-circuit current \( (I_{sc}) \). The diode \( IV \) curve is effectively shifted downward into the fourth quadrant by \( I_{sc} \) (Figure 1.2 left). For solar cells, this quadrant is called the "power quadrant" because it is the operation range where power is generated. It is traditionally flipped into the first quadrant to show that power is being generated as seen in Figure 1.2 on the right and is expressed using the ideal Shockley diode equation as

\[
I = I_L - I_0 \left( e^{\frac{qV}{nkT}} - 1 \right)
\]  

(1.1)
As a forward bias is applied to the diode, diode forward operation current begins to balance out the photon induced current until a forward bias point is reached where the net current through the diode is 0. This voltage bias point is the open circuit voltage \( V_{oc} \). The point on the \( IV \) curve where the maximum power is generated is called \( P_{max} \). These parameters are shown on the \( IV \) curve in Figure 1.2. Fill factor is calculated from \( P_{max} \) with the equation

\[
FF = \frac{P_{max}}{I_{sc} \cdot V_{oc}} = \frac{I_m \cdot V_m}{I_{sc} \cdot V_{oc}}
\]

The solar cell power conversion efficiency (\( \eta \)) is the ratio of the maximum generated power \( P_{max} \) at a given illumination to the incident illumination power, or

\[
\eta = \frac{P_{max}}{P_{in}}
\]

where \( P_{in} \) is the total power illuminating the cell. In the case of solar cells, this is \( P_{sun} \). The incident power will be discussed later in this chapter with solar spectra.

There are also two lumped parasitic resistance terms that are added to model resistance effects in the solar cell. The first being series resistance \( (R_s) \) which, true to the name, is shown as a resistor in series with the solar cell. Transport through the junction, lateral conduction in the solar cell emitter, conduction in the grid fingers and busbars of the cell, and metal-semiconductor contact resistances are included in this term. An ideal cell would have no series resistance. The second parasitic resistance term is the shunt resistance \( (R_{sh}) \), which characterizes the leakage current through
the diode. The shunt resistance consists mostly of trap assisted tunneling across the diode. An ideal cell has an infinite shunt resistance. The series resistance is most pronounced when there is a voltage drop across the two terminals of the cell and primarily reduces the magnitude of the slope around $V_{oc}$, while the shunt resistance is most pronounced at reverse bias, zero bias, or small forward biases and manifests as an increase in the magnitude of the slope around $I_{sc}$. Adding in the effects of parasitic resistances to the diode equation results in

$$ I = I_L - I_0 \left( e^{\frac{q(V - I_R)}{nk_bT}} - 1 \right) - \frac{V + IR_s}{R_{sh}} $$

(1.4)

Solving at $V = 0$ shows that with parasitic resistances, $I_{sc}$ does not necessarily equal $I_L$ as

$$ I_{sc} = I_L - I_0 \frac{e^{\frac{-IR_s}{nk_bT}} - I_{sc}R_s}{R_{sh}} $$

(1.5)
where the exponential can be neglected due to the small magnitude resulting in

\[ I_{sc} = I_L - I_{sc} \frac{R_s}{R_{sh}} \]  

(1.6)

If the ratio of \( R_s \) to \( R_{sh} \) is not small, the effect on short circuit current can’t be ignored. An equivalent circuit diagram is shown in figure 1.3.

![Solar Cell equivalent circuit](image)

Figure 1.3: Solar Cell equivalent circuit

Photovoltaics are generally tested under a rigidly defined spectrum in order to be able to calculate an \( \eta \) for the target conditions of the device. In the case of solar cells, this spectrum is the solar spectrum. The sun is close in shape to a black body radiator with a temperature of 6000K and is often modeled as such. The precise solar spectrum measured from space is defined as Air Mass Zero (AM0). Since solar power has extensive terrestrial applications as well, a standardized spectrum was chosen at AM1.5G, or the sun through 1.5 times the atmosphere thickness or at a corresponding zenith angle of 48.2°. There is general scattering of light in the visible range, and absorption dips due to specific molecules in the atmosphere such as \( \text{H}_2\text{O} \) and \( \text{CO}_2 \). These three spectra are shown for reference in Figure 1.4. The spectrum that a solar
cell is designed to operate under is critical due to the major internal power loss mechanisms present in solar cells. The maximum thermodynamic, or Carnot, efficiency limit is given by the ratio of the temperatures of the two bodies involved[8].

\[ \eta_{th} \leq 1 - \frac{T_{cell}}{T_{sun}} \]  

(1.7)

where, as mentioned before, the sun is modeled as a 6000K black body radiator. Assuming the cell is operating at 25°C, there is a maximum thermodynamic efficiency of 95%. This value is not useful in determining maximum achievable solar cell efficiency.
because it ignores other power loss mechanisms which will be discussed in detail below.

Further power loss begins with transmission and thermalization. Photons with energies below the bandgap of the material can’t be converted into electrical energy, while a photon with an energy above that of the bandgap can be absorbed. Assuming any photon that can generate a carrier does generate a carrier, there are further power loss mechanisms. Photons with energies above the bandgap of the semiconductor generate 'hot' carriers which relax down to the band-edge and the excess energy is lost to thermalization. Visualizations of these processes are shown in Figure 1.5.

![Figure 1.5: Visualization of thermalization and transmission events in a solar cell.](image)

There is another power loss to entropy from the mismatch in absorption and emission angles. This is known as the Boltzmann loss. Both the Boltzmann and Carnot
losses can be expressed as a reduction in optimal operating voltage [3]

\[ V_{\text{opt}} = E_g \left( 1 - \frac{T_{\text{cell}}}{T_{\text{sun}}} \right) - k_b T_{\text{cell}} \ln \left( \frac{\Omega_{\text{emit}}}{\Omega_{\text{abs}}} \right) \]  

(1.8)

where \( \Omega_{\text{emit}} \) is \( \pi \), and \( \Omega_{\text{abs}} \) is the solid angle of the Sun as seen from Earth, or \( 6.8 \times 10^{-5} \) steradians [3]. Devices under high light concentration can outperform devices at an equivalent one-sun illumination by reducing the Boltzmann loss through the increase in the effective absorption angle from the sun.

Finally, operating current can be calculated by subtracting the absorbed photon flux from the emitted photon flux at the the correct solid angle, defined, where flux, \( n \), is

\[ n(E, T, \mu, \Omega) = \frac{2\Omega}{c^2 h^2} \frac{E^2}{e^{\frac{E - \mu}{k_b T}} - 1} \]  

(1.9)

where \( E \) is energy, \( c \) is the speed of light, \( h \) is Planck’s constant, and \( \mu \) is chemical potential and operating current is

\[ J = e \left( \int_{E_g}^{\infty} n(E, T_{\text{sun}}, 0, \Omega_{\text{abs}}) dE - \int_{E_g}^{\infty} n(E, T_{\text{cell}}, E_g, \Omega_{\text{emit}}) dE \right) \]  

(1.10)

and is based on Kirchoff’s law of thermal radiation. Applying these power loss mechanisms as a function of bandgap results in the plot shown in Figure 1.6 It is clear from this model that at narrow bandgaps, the bulk of the power loss is due to thermalization, while with wide bandgap materials, the bulk of the power loss is due to thermalization. The weaknesses of this model is that it assumes that the diffusion length is infinite,
both the sun and cell are perfect radiators, any photon that can generate an electron-hole pair does so, and all recombination is radiative. The work in this study focuses primarily on the effects of thermalization and transmission for reasons that will be discussed later on [3].

In reality, not every generated electron-hole pair results in collection. Bulk and surface recombination events and reflection loss result in lost potential current. The ability to collect generated carriers, or the quantum efficiency of the device is an important metric in assessing both device design and material quality. This can be directly measured or modeled in terms of spectral responsivity (SR) which is defined as the amount of current (A) collected per unit power (W) illuminating the device at a given wavelength(\(\lambda\)). The Hovel/Woodall model[9] is a series of carrier transport equations.
that can be used along with absorption data to model current collection in a device where a flux at a given wavelength \((F)\), starting with current generated and collected in the front-surface field (FSF).

\[
J_D = \frac{qF\beta L_a}{\beta^2 L_a^2} \left[ \frac{\beta{L_a} + S_a\tau_a}{L_a} \left( 1 - e^{-\beta D \cosh \frac{D}{L_a}} \right) - e^{-\beta D \sinh \frac{D}{L_a}} \right] - \beta{L_a}e^{-\beta D}
\]  

(1.11)

where \(\beta, D, L_a, \tau_a, S_a\) are the absorption coefficient, thickness, diffusion length, lifetime, and surface recombination velocity to air in the FSF material respectively. The next component, the emitter is modeled as

\[
J_{D+d} = \frac{qF e^{-\beta D} \alpha L_g}{\alpha^2 L_g^2} \left[ \frac{\alpha L_g + S_g\tau_g}{L_g} \left( 1 - e^{-\alpha d \cosh \frac{d}{L_g}} \right) - e^{-\alpha d \sinh \frac{d}{L_g}} \right] - \alpha{L_g}e^{-\alpha d}
\]  

\[
+ \frac{J_D}{S_g\tau_g \sinh \frac{d}{L_g} + \cosh \frac{d}{L_g}}
\]

(1.12)

where \(\alpha, d, L_g, \tau_g, S_g\) are the absorption coefficient, thickness, diffusion length, lifetime, and surface recombination velocity to FSF in the emitter material respectively. Next, the space-charge, or depletion region of width \(W\) is modeled. The assumption here is that all generated carriers are collected because the drift field sweeps them out quickly.

\[
J_W = qF e^{-\beta D} e^{-\alpha d} (1 - e^{-\alpha W})
\]

(1.13)

Finally the contributions of the base are considered where

\[
J_{D+d+w} = qF e^{-\beta D} e^{-\alpha d} e^{-\alpha W} L_p \alpha L_p \frac{\alpha L_p}{\alpha L_p + 1}
\]

(1.14)
for a diode with a long base. The SR of the cell is given as

$$SR = \frac{J_{D+d}(\lambda) + J_W(\lambda) + J_{D+d+W}(\lambda)}{qF(\lambda)}$$  \hspace{1cm} (1.15)$$

The value of such a model is that it allows for extraction of material quality parameters such as surface recombination velocities, minority carrier diffusion lengths, and minority carrier lifetimes when fitting to measured data or for the prediction of device performance using estimated or textbook parameters. PC1D, a simulation tool developed by UNSW[10] which uses a similar model to what is shown above was used in this study for modeling purposes.

Finally, $J_{sc}$ can be calculated from either measured SR data or an SR model by integrating across the spectrum where

$$J_{sc} = \int SR(\lambda) * \phi_{spectrum}(\lambda)d\lambda$$  \hspace{1cm} (1.16)$$

where $R(\lambda)$ is the reflectivity of the cell and $\phi_{spectrum}(\lambda)$ is the desired spectrum that performance is to be measured under. The benefit of this technique is that it allows for the calculation of $J_{sc}$ under any spectrum desired, without the concern of spectral mismatch from the lamps that would be used to simulate a spectrum.
1.3 NANOSTRUCTURES IN PHOTOVOLTAIC DEVICES

As established in the previous section, bulk materials exhibit a sharp absorption cut-off at the band edge of the material. One method around this constraint is to insert nanostructures of a narrow bandgap material into a wide bandgap host material. The original theory, presented by Henry et al. was that a narrow bandgap material could be placed between two wide bandgap materials in order to confine electrons in one dimension. Discrete energy states would result from the one-dimensional finite well, shown in thin samples as gaussian peaks in absorption beyond the band-edge of the wide-bandgap host material[11]. Discussion of nano structures will be continued in Chapter 2.

1.4 SOLAR CELL TESTING METHODOLOGIES

The first standard test done on solar cells is to measure performance under the desired illumination conditions. Since the application in this study was space, and the cost of bringing devices to space make using the actual AM0 spectrum unfeasible, a simulated AM0 spectrum was required. RIT has a TS Space Systems (TSS) dual-source solar simulator that uses a 6 kW hydrargyrum medium-arc iodide (HMI) lamp to provide the visible and UV part of the solar spectrum and a 12 kW quartz-tungsten-halogen (QTH) bulb to fill in the near-IR and IR part of the spectrum. Filters and a dichroic mirror are used to further shape and combine the spectra of the two bulbs. A block diagram of the solar simulator is shown in Figure 1.7.
Figure 1.7: Block diagram of TS Space Systems solar simulator at RIT.

Standard calibration procedure uses an InGaP cell first to calibrate the HMI lamp since the QTH lamp cut-on is past the band edge of InGaP. A GaAs cell is used to calibrate the QTH lamp. The TSS solar simulator is overlaid on the ASTM AM0 spectrum in Figure 1.8.

Since performance at light concentrations greater than one sun are sometimes required, a few methods can be applied to increase concentration. The first is to use a lens to focus the light down. The second is to use a large area pulsed solar simulator (LAPSS) where a simulator that provides a large area of illumination is employed and the cell is moved closer to the source to effectively increase the acceptance angle of light from the source. The final method is to use a high intensity pulsed solar simulator (HIPSS) which uses a high intensity bulb and an array of reflective optics to capture a lot of light.
Figure 1.8: RIT TSS solar simulator spectrum compared to the ASTM AM0 reference spectrum

Spectral responsivity, mentioned in more detail above, is measured using a monochromator, an optical chopper, a lock-in amplifier, and a source meter. A diffraction grating and a tungsten bulb provides the narrow spectral bandwidth required and the grating steps to change the wavelength that illuminates the cell. Since low intensities are used, a lock-in amplifier and optical chopper provide virtually noise-free amplification to the output signal. The solar cell is held at short circuit and the current generated at a given illumination wavelength is measured and normalized to a calibration file to get SR. A block diagram of a spectral responsivity setup is shown in Figure 1.9.

Dark diode $IV$ characteristics can be measured in order to find the reverse saturation current and diode idealities, but since the front of the cell is left unshadowed as
light needs to be able to be absorbed by the device, series resistance in solar cells is significantly higher than in standard diodes. This makes it difficult to measure the diode parameters at the range of voltages where the cell will operate. A workaround for this problem is to illuminate the solar cell, starting with a low intensity, and measure the $J_{sc}$ and $V_{oc}$. The illumination intensity can be slowly increased in order to generate a curve of $J_{sc}$ and $V_{oc}$ values. Since there is no voltage drop across the device at $J_{sc}$ and no current flowing through the device at $V_{oc}$, the result is a diode $JV$ curve with series resistance removed where

$$J_{sc} = J_0 e^{\frac{qV_{oc}}{nkT}}$$  \hspace{1cm} (1.17)

Series resistance is calculated with Ohm’s Law and the difference in voltage between a dark $JV$ and a $J_{sc-Voc}$ curve at the one-sun $J_{sc}$ current value. Shunt resistance can be calculated by measuring the slope between 0 $V$ and a point where the cell is reversed biased.
Absorption is to some extent a reversible process. A material that can convert light into free carriers can also emit light as carriers recombine. A method that takes advantage of this phenomenon can be used to investigate material quality and properties. It involves injecting carriers, either electrically, with a forward bias, or optically, with a laser, and measuring the output spectrum of the solar cell with a spectrometer. Electroluminescence uses an injected current to provide carriers for radiative recombination, which requires a fabricated device. Photoluminescence makes use of carrier excitation from a high-power laser and does not require a fabricated device or a \textit{pn} junction. Direct bandgap materials radiate strongly at the band edge, while indirect band-gap materials do not because the absorption and subsequent emission of photons at the band-gap energy require the assistance of a phonon. Two particle interactions have low probabilities of occurring. Radiative emission can also come from radiative defect states[12] and quantum confined states within the bandgap and is often used to measure radiative transitions in nanostructured photovoltaics[13].

1.5 PRIOR WORK

1.5.1 Gallium Phosphide Photovoltaics

Research into GaP has fallen largely into two categories. The first being high temperature space applications which aim to take advantage of the 2.26 eV bandgap and the resulting operating satiability at high temperatures[14]. Devices grown by Sulima et al had a measured open circuit voltage of 1.62 V and open circuit current density of 1.1
mA/cm² under AM0.

The second motivation involves the utilization of the wide bandgap, but in this case to potentially reduce thermalization loss through stacking a GaP cell on top of a narrower bandgap cell material. This application can also potentially take advantage of the 5.4505 Å lattice constant which is closely matched to Silicon meaning that it could potentially be grown on Silicon substrates for use in a multi-junction device. Allen et al developed a GaP solar cell design for this purpose and reported a measured $V_{oc}$ of 1.53 V and $J_{sc}$ of 0.959 mA/cm² under an AM1.5G spectrum.

The previously reported $J_{sc}$ values for GaP solar cells are significantly lower than the 17.95 mA/cm² maximum $J_{sc}$ calculated from a detailed balance model. The goal of this work is to both improve upon currently attainable current densities both through conventional design and through bandgap engineering through the addition of quantum wells.

1.5.2 Nanostructured Devices in Optoelectronics

Quantum wells were first proposed for use in lasers in 1973 by C.H Henry at Bell Labs who proposed that thin heterostructures could confine carriers and effectively create states within the bandgap of the host material[11]. This was demonstrated through showing absorption spectra demonstrating sub-bandgap absorption peaks. This demonstration started the investigation into quantum confinement for optoelectronic applications.

Quantum wells have been used to tune both absorption and emission properties of
III – V devices. Here at RIT, there has been extensive work towards the inclusion of quantum dots to increase the current generation in the limiting junction for multifunction photovoltaics[13]. Significant work has gone into development of GaAsP/InGaAs quantum well solar cells by Ekins-Daukes et al[15], showing a 2% relative current increase. Walters et al have proposed the application of quantum wells to tune the absorption of the bottom cell for a next-generation multijunction cell design[16]. There has been no published work on the introduction of nanostructures into a GaP PV device.

1.5.3 Radiation Damage in InP Photovoltaics

The interest in radiation tolerant photovoltaics is largely driven by the presence of belts of charged particles trapped by the Earth’s magnetic field which can potentially interfere with operation of electronics such as artificial satellites in orbit around the Earth. Extensive research has been performed at the Naval Research Labs[17][1][18] on radiation effects in both diffused junction and epitaxially grown bulk InP solar cells. A new processing technique called epitaxial lift-off where the substrate has been removed has been incorporated into the manufacturing process for next-generation photovoltaics. This process decreases the weight of the finished devices and can potentially lower cost through recycling of substrates. The radiation tolerance of these devices has not yet been compared to the radiation tolerance of bulk InP devices.
1.6 ORGANIZATION OF WORK

The following chapters introduce application specific design and testing of solar cells for space applications. Chapter 2 focuses on design and testing of gallium phosphide solar cells with indium gallium phosphide quantum wells for high temperature applications. It includes discussion of the theory behind and benefits of bandgap engineering and temperature dependent modeling of solar cell performance. It also includes electrical and optical characterization of gallium phosphide devices and experimental results on temperature and solar orbit dependent performance. Chapter 3 presents data and analysis on the effects of ionizing radiation, specifically $\alpha$ and $\beta$ radiation, on epitaxial lift-off indium phosphide and indium gallium arsenide thin film solar cells obtained from a commercial partner. Chapter 4 presents conclusion for both InP and GaP studies and finally a discussion on future work.
Chapter 2

Gallium Phosphide Photovoltaics

2.1 MOTIVATION

Space missions such as NASA’s Solar Probe+ project involving a close proximity to the sun will require solar cells that can efficiently operate at solar concentrations up to $510$ suns and temperatures well above $250^\circ C$ [19]. The Solar Probe+ project was proposed to study the Solar corona. The first pass near the Sun occurs at a distance of $0.18 \ AU$, and the probe slingshots around Venus seven times to tighten the highly elliptical orbit around the sun and will make 24 near-sun passes, three of which will be at or near a distance of $0.044 \ AU$ or $9.5 \ Solar \ radii$. A black body radiator with $35\%$ reflection would reach a temperature of around $1700 \ K$ at this distance. Previous endeavors, such as the MESSENGER mission to Mercury utilized highly reflective coatings on the PV components and have been successfully operated at Solar distances of $0.4 \ AU$. The current plan for the Solar Probe+ project is to incorporate mirrors on each cell and utilize a $74^\circ$ tilt along with cell cooling using current state of the art (SOA) triple junction concentrator solar cells. This design is required in order to keep cell temperature below
The Solar Probe+ mission is severely weight constrained, so techniques to potentially lower the weight of the power sub-system for similar future missions are of great interest\cite{20}. One such method would be to develop a cell technology capable of operating at significantly higher temperatures with similar, or better levels of reliability, while lowering the weight requirement of the power sub-system. Since solar winds are being investigated, these solar cells also need to be able to withstand an environment rich in ionizing radiation.

Gallium phosphide was investigated because it has a wide bandgap and is available as a substrate material. The wide bandgap coupled with the bandgap independence of the Boltzmann efficiency loss mentioned above means that GaP devices can potentially operate at significantly higher temperatures than devices of narrower bandgap materials. The potentially high radiation tolerance of GaP is due to wide bandgap materials having high binding energies which decreases the number of atomic displacements generated with an equivalent amount of non-ionizing energy loss from incident radiation. Mechanics of radiation damage in solar cells will be discussed more thoroughly in chapter 3. Low temperature annealing of radiation damage in GaP LEDs has been previously investigated\cite{21} and a 6 hour anneal at $350^\circ K$ lead to a reestablishment of the beginning-of-life (BOL) luminous intensity. The increased thermal stability of GaP with respect to GaAs, a current generation commonly used PV material that will be used as a point of reference in this study, causes the theoretical efficiency limit of single junction GaP to surpass that of GaAs at temperatures above $350^\circ C$. High temperature operation of the solar cell may alleviate some of the
requirement for active PV cooling, greatly reducing both weight and complexity of the power and cooling sub-systems.

The major drawback of wide bandgap semiconductors for photovoltaic applications is the power lost due to non-absorption. Bandgap engineering through the addition of nano-structures to solar cells is an effective way to increase current output through increased light absorption. The wide bandgap and temperature stability of GaP combined with bandgap engineering through InGaP multiple quantum wells (MQW) is a novel approach to enhance the overall efficiency of GaP solar cells. Adding quantum wells into the unintentionally doped region of a GaP solar cell allows for the exploitation of the temperature stability of GaP while increasing the current output of the cell by bringing the effective bandgap closer to the ideal bandgap for the solar spectrum.

In this study, the temperature dependent efficiency is modeled for a GaP solar cell and compared to a modeled GaAs solar cell. GaP solar cells with and without InGaP MQWs were grown by organometallic vapor phase epitaxy (OMVPE), fabricated, and characterized electrically and optically. Lastly, temperature dependent AM0 Current-Voltage characteristics of a GaP solar cell was measured and compared to modeled values.

### 2.2 THEORY

GaP has a direct bandgap ($E_g$) of 2.26 eV, and an indirect bandgap ($E_0$) of 2.78 eV at room temperature. The quantum well approach to bandgap engineering relies upon
one-dimensional quantum confinement which occurs from an offset in either the conduction or valence band between two or more materials. The quantum confined states within the bandgap of the host material facilitate the absorption of sub-host bandgap photons which lowers the effective bandgap of the material. In a system with shallow quantum well states, carrier extraction from the quantum wells occurs primarily from interactions with phonons. A schematic of a \textit{pin} solar cell with quantum wells due to a \textit{type-I} heterojunction is shown below in Figure 2.1.

![Figure 2.1: Example of \textit{type-I} heterojunction quantum well superlattice. The quantum wells facilitate sub-host-bandgap absorption of 'red' photons](image)

In the case of this study, the quantum well is from the confinement caused by inserting $\text{In}_{0.17}\text{Ga}_{0.83}\text{P}$ into GaP bulk material which has a conduction band offset of $100 - 120 \text{meV}$ to GaP\cite{23}. The increase in absorption from using MQW InGaP/GaP solar cells can potentially be more pronounced than the short circuit current gain seen from similar approaches in GaAs/In(Ga)As nanostructured solar cells due to the high
photon flux around 550 nm, the indirect absorption edge of GaP.

2.3 MODELING OF TEMPERATURE EFFECTS

Solar cell efficiency was modeled with a maximum operating power calculated with the temperature dependencies of $J_{sc}$, $V_{oc}$, $FF$, and $\eta$ and the AM0 solar irradiance [24], starting with current generation. The quantum efficiency was assumed to be one ($QE = 1$) up to the temperature dependent direct band-edge [25] [26] of the solar cell material. This allows calculation of an approximate temperature dependent $J_{sc}$. A more precise model of $J_{sc}$ would require temperature dependent absorption of GaP. The direct band-edge was used for calculating the short circuit current density because the deposition rate of epitaxially grown materials as well as the relatively short minority carrier lifetime limit device thickness which translates to low absorption of photons with energies between the indirect and direct band edges of GaP, shown in Figure 2.2. Absorption in thin films, calculated using the Beer-Lambert law drops off drastically past the direct band edge. The designed device thickness is in turn limited by the low minority carrier diffusion lengths seen in the grown GaP. Temperature dependent intrinsic carrier concentration is calculated as a function of effective densities of states in the conduction and valence bands [27].

$$n_i(T) = \sqrt{N_c(T)N_v(T)e^{-\frac{E_g(T)}{k_BT}}}$$  \hspace{1cm} (2.1)
where $N_c(T)$ and $N_v(T)$ are the temperature dependent densities of states in the conduction and valence bands respectively.

The temperature dependent intrinsic carrier concentration along with diffusivity and minority carrier diffusion length values extracted from a PC1D device model fit to the spectral responsivity of a cell were used to calculate a reverse saturation current density of a $p^+-n$ diode [28] with the equation

$$J_0(T) = \frac{qD_p}{L_pN_d}n_i(T)^2$$  \hspace{1cm} (2.2)$$

where $q$ is the electron charge, $N_d$ is the donor ion density, $D_p$ is hole diffusivity and $L_p$
is the hole diffusion length in the \( n \)-base. While lifetime and diffusion length do exhibit a temperature dependence, it is not expected to be as dramatic as other effects and since data on the temperature dependencies of these parameters was unavailable, they were neglected.

Next, open-circuit voltage was approximated using the ideal diode equation \([24]\) and the calculated \( J_{sc}(T) \).

\[
V_{oc}(T) = \frac{n k_b T}{q} \ln \left( \frac{J_{sc}(E_0(T))}{J_0(T)} \right) + 1 \tag{2.3}
\]

where \( k_b \), the Boltzmann constant is approximately \( 8.617 \times 10^{-5} \) and the diode ideality factor, \( n \), is an empirical fitting parameter for diode operation. Common \( n \) values are 1 and 2, where \( n = 1 \) is associated with radiative recombination and \( n = 2 \) is associated with non-radiative recombination through a defect at midband, or Shockley, Reed, Hall recombination. The reverse saturation current density equation from above assumes radiative recombination and all recombination in the model was assumed to be radiative so \( n = 1 \) was used.

The temperature dependence of diode fill-factor was calculated using an empirical model from Green \([29]\). This does not take into account fill factor degradation from changes in series or shunt resistance.

\[
FF(T) = \frac{V_{oc}}{k_b T} - \ln \left( \frac{V_{oc}}{k_b T} + 0.72 \right) \tag{2.4}
\]
The temperature dependencies of these parameters were used to calculate a temperature dependent efficiency for GaP and GaAs, shown in Figure 2.3. Normalized temperature coefficients have traditionally been reported as linear parameters, and while some curvature exists, remain relatively linear until the open circuit voltage of the cell begins to approach $3k_bT$. The normalized temperature coefficient of efficiency, or the slope extracted from figure 2.3, was modeled to be $-2.23 \times 10^{-3} \, ^oC^{-1}$ for a GaAs solar cell, and $-1.31 \times 10^{-3} \, ^oC^{-1}$ for a GaP solar cell. As expected, GaP showed a lower temperature sensitivity than GaAs. The parameters used in this model will be compared to experimental data for both GaAs and GaP devices.

Figure 2.3: Modeled efficiency of GaP and GaAs devices vs temperature.
2.4 GROWTH AND FABRICATION

In this study, GaP devices were grown by organo-metallic vapor phase epitaxy (OMVPE). In this process, group-\textit{III} (tri-methyl indium, tri-methyl gallium, and tri-methyl aluminum) and group-\textit{V} (arsine and phosphine) containing precursors are flown into a reactor chamber. The chamber temperature is increased to a temperature where the precursors are ‘cracked’, or pyrolyzed. The byproducts, mainly methane and hydrogen, are pumped out of the system while a fraction of the \textit{III}/\textit{V} adatoms stick to the surface and diffuse around until reaching a proper vacancy and binding with the crystal structure. As an example,

\begin{equation}
Ga(CH_3)_3 + PH_3 \rightarrow GaP + 3CH_4
\end{equation}

This process yields high quality crystalline material. Unlike molecular beam epitaxy (MBE), OMVPE does not require the system to be pumped to ultra-high vacuum (UHV) while maintaining high quality of grown materials so samples can be loaded and unloaded much more quickly which improves the feasibility of large-scale manufacturing. All cells were grown via OMVPE at NASA Glenn Research center.

A high group \textit{V} overpressure was required, shown as \textit{V}/\textit{III} ratio, in order to prevent phosphorous out-diffusion from the surface. The required over-pressure is dependent on growth temperature. In this study, a \textit{V}/\textit{III} ratio in excess of 160 was used. The first challenge encountered in the growth process was a high density of hexagonal pyramid structures visible in Nomarski microscope images of the surface. The
origination of the pyramidal structures was determined to be stacking faults caused by adsorption at the surface coupled with low surface mobility of the adatom species. This means that adatoms are inhibited from diffusing to the generated defect sites \[30\]. In order to correct this, GaP was grown at \(700^\circ C\), where the formation of large defects that propagate to the surface was suppressed.

Previously published results by Sulima\[14\] and Allen\[31\] were fitted in PC1D to extract minority carrier diffusion lengths. If the emitter or base of the cell is too wide, generated carriers will recombine before diffusing to the junction which inhibits charge separation and carrier collection. A PC1D fit to Sulima’s results\[14\] yielded diffusion lengths of \(750 \text{ nm}\) in the emitter and \(1 \mu\text{m}\) in the base, while a fit to the cells from this study yielded diffusion lengths of \(300 \text{ nm}\) in both base and emitter. A \(150 \text{ nm} \ 2 \times 10^{18} \text{ cm}^{-3}\) Zn doped emitter was used in order to compensate for the low minority carrier lifetime. Both GaP and AlP back surface fields (BSF) were investigated. The base was \(2 \mu\text{m}\) thick and was Si doped to \(1 \times 10^{17} \text{ cm}^{-3}\). A heterojunction BSF more effectively passivates the active layers and assists in reflection of diffusing carriers back towards the junction. The effects of the BSF material will be more thoroughly discussed after the introduction of the grown samples.

In this study, a GaP control \(pn\) solar cell (sample A) and a GaP \(pin\) solar cell with a 5-\textit{period} \(\text{In}_{0.17}\text{Ga}_{0.83}\text{P}\)/GaP MQW grown pseudomorphically in an \textit{i}-region (sample B) were grown with AlP heterojunction front surface field, a GaP back surface field, and a heavily doped GaAs contact layer in order to investigate the effects of the addition of InGaP MQWs and of the temperature dependence of \(AM0\) efficiency. An additional
device wafer (sample C) was grown under the same conditions as sample B, but the GaP back surface field was replaced with an AlP heterojunction back surface field. Substrates were 300 µm thick 2" diameter $2 \times 10^{18} \text{ cm}^{-3}$ sulfur doped (100) n-GaP liquid encapsulated Czochralski (LEC) substrates obtained from SurfaceNet GmbH. Zn and Si doping was used for p and n-type films respectively. The device structure is shown in Figure 2.4. A heavily Zn doped GaAs contact layer was grown on all samples to facilitate ohmic contact on the front side of the wafer.

![Figure 2.4: GaP MQW solar cell schematic](image)

Samples A & B, the samples with a GaP BSF, exhibited smooth surface morphology. There was a low density of hexagonal pyramids with a diameter of $10 - 20 \mu m$ across the wafer surface, shown in Figure 2.5a. Sample C (Figure 2.5b) however exhibited a nearly complete surface coverage of hexagonal pyramids with diameters ranging from $50 - 100 \mu m$. This is similar to the surface morphology seen in GaP films grown below $700^\circ C$ and mentioned previously. The presence of the pyramidal
structures in only the sample with an AlP BSF suggests that the pyramids nucleate in AlP layers. The formation of pyramids early in the epitaxy process means they will continue to expand during growth.

Figure 2.5: a. Nomarski image of surface of Sample B. b. Nomarski image of surface of Sample C. The inclusion of an AlP BSF in sample C leads to poor surface morphology.

The general process flow was back surface metallization, annealing of contacts, frontside metallization, Mesa isolation etching, contact layer removal, and finally frontside metal anneal. Samples were fabricated using Ge/Au/Ni/Au thermally evaporated backside n-type evaporated contacts. The backside metallization was annealed at 520°C for 6 minutes[32]. It is generally preferable in processing for metallization to occur at the end of the process so it isn’t subjected to chemical processing, but the high anneal temperature required for backside metal meant that it must be performed prior to frontside metallization.

A lift-off process with a bi-layer resist scheme was used to facilitate frontside patterning of the grid array. LOR 10A lift-off resist was spun on the wafers and cured.
S1830 photoresist was deposited on top of the LOR film. The photoresist was exposed using a contact aligner and placed in developer. The high etch rate of LOR in the developer results in a sharp undercut profile. This is critical to the lift-off process because it insures that the deposited metal film will be discontinuous from the resist surface to the trenches that extend down to the wafer surface. The discontinuity in the metal films allows the solvent used to strip the photoresist to undercut the metal coating on the resist. \( Au/Zn/Au \) front \( p \)-type metal contacts were thermally evaporated onto the wafers.

A wet chemical mesa etch was performed in Hydrochloric Acid: Nitric Acid: Acetic Acid (1:1:1) in order to isolate the \( p - n/p - i - n \) structures. Etching GaP presented a challenge because the chemistry used exhibits a high etch rate of \( Au \) used for the backside metallization. Photoresist was painted on the backside of the wafer using a pipette in order to prevent removal of the rear contact. \( AlP \), used for the front surface field, is water soluble to the extent that water vapor in the air will readily etch it away, so it requires a GaP cap. This means etch selectivity is required between the contact layer and GaP. The GaAs contact layer was used to improve the ohmicity of the front contact and provide etch selectivity between contact layer and cell. Removal of contact layer is critical for getting light into the optically active layers of the solar cell.

Finally the frontside metal was annealed at \( 407^\circ C \) for 7 minutes. An assortment of \( 1 \times 1, 1 \times \frac{1}{2}, \) and \( 1 \times \frac{1}{4} \) \( cm^2 \) cells were fabricated. Grid shadowing was approximately 4%. Anti-reflection coatings were not applied. A fabricated wafer is shown in Figure 2.6.
2.5 EXPERIMENTAL SETUP

Illuminated one sun $AM0$ current density-voltage ($J-V$) curves were taken with a Keithley 2400 Source Meter and a TS Space Systems close match dual source solar simulator. The lamps were calibrated using $InGaP$ and GaAs solar cells, both calibrated at NASA Glenn Research Center (GRC). Dark $J - V$ and $J_{sc} - V_{oc}$ was measured at NASA GRC using a custom built probe station. Spectral Responsivity was measured using an Optronics Laboratories OL750 monochromator. Electroluminescence (EL) was captured using an Ocean Optics HR2000 Spectrometer and a Janis cryostat with liquid nitrogen. Finally, solar orbital conditions were simulated using a hotplate and the Large Area Pulsed Solar Simulator (LAPSS) at NASA GRC.
2.6 RESULTS

2.6.1 Current-Voltage Characteristics

One Sun $AM0$ current density-voltage ($J - V$), $J_{sc} - V_{oc}$ and dark diode curves were measured to inspect the diode quality and one-sun performance of the GaP devices. Figure 2.7 illustrates the one sun $AM0$ illuminated $J - V$ curves for the three GaP devices and Tables 2.1 & 2.2 show diode parameters.

![Graph](image)

Figure 2.7: $AM0$ illuminated $J$-$V$ curves for baseline and QW GaP devices. Samples B & C contain quantum wells, while Sample A does not.

The samples with quantum wells both exhibited a $J_{sc}$ enhancement, shown in Table ???. This an exciting initial result as it is correlated with the addition of MQWs.
Table 2.1: AM0 illuminated J-V characteristics for baseline and QW GaP devices

<table>
<thead>
<tr>
<th>Sample</th>
<th>$J_{sc}$ (mA/cm²)</th>
<th>$V_{oc}$ (V)</th>
<th>FF (%)</th>
<th>η (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>A : &quot;No QW&quot;</td>
<td>2.27</td>
<td>1.43</td>
<td>56.1</td>
<td>1.34</td>
</tr>
<tr>
<td>B : &quot;5x QW&quot;</td>
<td>2.47</td>
<td>1.21</td>
<td>57.3</td>
<td>1.26</td>
</tr>
<tr>
<td>C : &quot;5x QW, AlP BSF&quot;</td>
<td>2.56</td>
<td>1.29</td>
<td>75.0</td>
<td>1.83</td>
</tr>
</tbody>
</table>

Sample A exhibited an open circuit voltage of 1.43 V which was lower than previously reported values of 1.62 V under AM0 conditions [14]. It is important to note that both the measured $V_{oc}$ for this study and the previously reported $V_{oc}$ are significantly lower than the detailed balance $V_{oc}$ limit of 1.88 V calculated from thermodynamic and Boltzmann voltage losses from $E_g$. Sample B shows a Voc degradation of 220 mV when compared to sample A. The inclusion of quantum wells could be degrading the open circuit voltage of the cell through the introduction of strain-induced defects in the emitter and i-region of the cell or through lowering in the effective quasi-Fermi level splitting of the junction from thermalization into the quantum well states. Sample C exhibited a $V_{oc}$ of 1.29 V, or an increase of 80 mV compared to sample B. This may partially be the result of the inclusion of the heterojunction back surface field. Both samples with a GaP BSF showed the effects of high series resistance and low shunt resistance and had low fill factors. Further investigation of diode behavior was performed through $J_{sc}$-$V_{oc}$ measurements to investigate the effects of both open circuit voltage loss and to quantify the parasitic resistances.

$J_{sc}$-$V_{oc}$ characteristics of the three samples can be seen in Figure 2.8. A more standard practice in semiconductor device characterization is the measurement of dark IV characteristics. In photovoltaics, the primary interest is electrical characteristics
around $V_{\text{max}}$ and $V_{\text{oc}}$. Dark $IV$ characteristics at these voltages are dominated by series resistance effects, so the $J_{\text{sc}}-V_{\text{oc}}$ method is applied instead. The inset in Figure 2.8 shows the local ideality factor, calculated as a numerical derivative of the diode $IV$ curve. Sample A exhibits an ideality factor significantly greater than $n = 1\text{-}2$ which is generally expected of diodes. The high ideality factor region is not present on the two devices with an $i$-region containing the MQWs. Samples B and C have diode ideality factors close to $n = 2$ over the entire range of open circuit voltages measured indicating recombination in the space-charge region dominates diode characteristics. The poor surface morphology of Sample C did not appear to degrade the properties of the cell, and a diffuse reflection measurement established that it did not have any major anti-reflective effects. Diode parameters near $V_{\text{oc}}$ were extracted and shown in Table 2.2.

Sample B exhibited a shunt resistance ($R_{\text{sh}}$) of $1.31 \, k\Omega$ which is nearly an order of magnitude lower than the shunt resistance of Sample A ($11.8 \, k\Omega$). This could be due to the addition of the strain-uncompensated nanostructures introducing material defects which create shunt pathways through the cell[13]. Sample C showed a greatly improved shunt resistance which is interesting because it shows that the stacking faults from the AlP BSF are not creating an increase in shunt pathways through the junction. Expected shunt resistances with good material quality are on the order of hundreds of $M\Omega$ to $G\Omega$.

Samples A and B both exhibited high series resistance ($R_s$) of $285 \, \Omega$ and $150 \, \Omega$ respectively, shown in Table 2.2. This is significantly higher than the measured $R_s$ of
Table 2.2: Diode parameters for samples A, B, and C

<table>
<thead>
<tr>
<th>Sample</th>
<th>$J_0$ (A/cm²)</th>
<th>Diode Ideality</th>
<th>$R_{series}$ (Ω)</th>
<th>$R_{shunt}$ (kΩ)</th>
</tr>
</thead>
<tbody>
<tr>
<td>A: &quot;No QW&quot;</td>
<td>4.14 x 10⁻¹⁵</td>
<td>4.09</td>
<td>285</td>
<td>11.8</td>
</tr>
<tr>
<td>B: &quot;5x QW&quot;</td>
<td>1.43 x 10⁻¹²</td>
<td>2.17</td>
<td>150</td>
<td>1.31</td>
</tr>
<tr>
<td>C: &quot;5x QW, AlP BSF&quot;</td>
<td>8.91 x 10⁻¹⁵</td>
<td>1.87</td>
<td>36.2</td>
<td>151</td>
</tr>
</tbody>
</table>

Sample C of 36.2 Ω. Samples A and B exhibited slightly rectifying behavior on the front contacts and a specific contact resistance on the order of 1 Ω-cm². The high doping in the contact layer means that the non-ohmic behavior is likely not caused by the metal-GaAs contact. One potential cause would be from difficulty of dopant incorporation into the wide-bandgap AlP FSF. It is interesting to note that Sample C did not exhibit problems with contact quality, so a problem from fabrication can not be entirely ruled out.

2.6.2 Spectral Responsivity and Electroluminescence

EQE measurements of samples A-C are shown in Figure 2.9. The EQE peak at 446 nm is just above the 2.78 eV direct band edge of GaP. As seen in the inset, there is an absorption shoulder around 475 nm in samples B & C and an extension of photon conversion past 575 nm while there is an abrupt decline in EQE past 550 nm on sample A. This is believed to be caused by the addition of MQWs. The increase in the sharpness of the peak at 440 nm in sample C when compared to samples A & B is the likely result of the inclusion of a heterojunction back surface field which results in lower surface recombination velocities.
Figure 2.8: $J_{sc}$-$V_{oc}$ curves and local ideality factors for baseline and QW GaP devices. Sample A exhibits an ideality factor of 6 at low to moderate voltages.

An integration of EQE with the $AM0$ spectrum provides an evaluation of the short-circuit current density under low bias which minimizes the effects of $R_{series}$. Sample A has an integrated $AM0$ $J_{sc}$ of 1.85 m$A$/cm$^2$. This is substantially lower than the simulated $AM0$ measured $J_{sc}$. Broad peaks at 422 nm and 436 nm in the hydrargyrum medium-arc iodide (HMI) lamp spectrum which are not present in the ASTM E490 $AM0$ spectrum were measured and shown in Figure 2.10. This explains the discrepancy seen between integrated SR and measured $AM0$ $J_{sc}$. A more accurate calibration would require a GaP calibration cell. An integrated SR $J_{sc}/J_{VJsc}$ ratio was calculated in order to characterize the calibration mismatch due to the use of
Figure 2.9: External quantum efficiency and integrated short circuit current densities of GaP devices.

an InGaP (lattice matched to GaAs) calibration cell. Samples A, B, and C exhibited a SR/J – V \( J_{sc} \) mismatch factor ranging from 0.81-0.84. This is equivalent to a solar concentration of 1.19-1.23 \( S_{uns} \). What this means is the calibration using an InGaP cell for GaP results in too great a flux in the wavelength range where GaP absorbs light. Calibration of the system using a GaP cell would provide a more accurate measurement.

The integrated \( J_{sc} \) from this study compares favorably with previously reported \( AM0 \ J_{sc} \) values of 1.1 \( mA/cm^2 \) under \( AM0 \) without an AR coating [14]. This improvement is partially due to the thinning of the emitter mentioned above which improves
Table 2.3: Diode parameters for samples A, B, and C

<table>
<thead>
<tr>
<th>Sample</th>
<th>Full Spectrum</th>
<th>Beyond 446 nm</th>
<th>SR/J-V Jsc Mismatch</th>
</tr>
</thead>
<tbody>
<tr>
<td>A : “No QW”</td>
<td>1.85</td>
<td>0.43</td>
<td>0.81</td>
</tr>
<tr>
<td>B : “5x QW”</td>
<td>2.08</td>
<td>0.58</td>
<td>0.84</td>
</tr>
<tr>
<td>C : “5x QW, AlP BSF”</td>
<td>2.07</td>
<td>0.66</td>
<td>0.81</td>
</tr>
</tbody>
</table>

collection when the diffusion length in the film is short. The two MQW samples, B & C, exhibited an integrated $J_{sc}$ of 2.08 mA/cm$^2$ and 2.07 mA/cm$^2$ respectively. Sample B exhibited an increase of 0.15 mA/cm$^2$ in integrated $J_{sc}$ past the direct band edge of GaP, when compared to sample A represents an increase of 8% in short circuit current density.

Electroluminescence (EL) was performed in order to further investigate the reasons behind the differences in SR between the ‘baseline’ and QW containing samples. Low temperature EL was performed in an attempt to prevent thermal escape from quantum confined states and to increase the emission intensity which is low because of the indirect bandgap of GaP. Figure 2.11 shows the EL response of samples A and B measured at 80 K. An injection current density of 2 $A/cm^2$ was required to measure a signal distinguishable from the noise floor of the detector with sample A, and an injection current of 4 $A/cm^2$ was required to produce a measurable luminescence signal in sample B. The need for such a large injection current is due to the low probability of radiative recombination at an indirect transition. A broad peak is seen at 1.75 eV in both samples A and B. This is attributable to shallow defect-O complexes [12] present in the material.
Sample A shows a single sharp peak at 2.16 eV while Sample B exhibits a doublet with peaks at 2.12 eV and 2.22 eV. Electroluminescence signal near the indirect band-edge has been seen from GaP, and is thought to be a result of shallow distant donor-acceptor pair recombination or from bound exciton states, not band-to-band recombination [12]. Both of these recombination methods are shown in Figure 2.12. The addition of an intrinsic region in sample B corresponds with the requirement of a higher injection current in order to generate a measurable EL signal. The 2.12 eV peak in Sample B is 100 meV below the higher-energy peak, which is near the value of the expected conduction band offset. The peak may be attributed to the additions
of the MQWs, but further investigation of the properties of the InGaP such as a more exact bandgap value and more precise composition information would be instrumental in further modeling of quantum confined states.

### 2.6.3 Temperature Dependent Performance

AM0 1-sun illuminated IV sweeps were taken at temperatures between 25°C and 85°C for both GaAs and GaP solar cells. A previously fabricated GaAs pin solar cell was measured as a 'baseline' in conjunction with a GaP solar cell (Sample A) in order to compare the merits of GaP to currently utilized space PV technology. Modeled
parameters and data were normalized to performance at $26^\circ C$. Normalized data allows for the extraction and direct comparisons of trends, isolating the temperature effects, and removing the effects assumption of idealized behavior of the model, mainly the two assumptions that all recombination was radiative and that all photons that could be converted up to the direct band-edges would result in collected carriers.

The normalized temperature coefficient of efficiency for the GaAs solar cell was measured to be $-2.24 \times 10^{-3} \, ^\circ C^{-1}$, shown in Figure 2.13, which is very close to $-2.23 \times 10^{-3} \, ^\circ C^{-1}$ seen from Figure 2.3. This provides confirmation of the validity of the model for GaAs. As seen in Figure 2.14, the efficiency of the GaP solar cell appears to track closely with the current increase. A normalized temperature coefficient of efficiency of $2.78 \times 10^{-3} \, ^\circ C^{-1}$ was exhibited. This represents an increase in efficiency with increasing temperature as opposed to the local modeled value of $-1.31 \times 10^{-3} \, ^\circ C^{-1}$ from Figure 2.3.
The experimental $J_{sc}$ increased more sharply than modeled. On possible explanation for the greater than expected increase in short circuit current density is that some degradation in $J_{sc}$ when compared to $J_L$ is seen due to the high series resistance and low shunt resistance. If $R_s$ recovers substantially enough with temperature, $J_{sc}$ can increase. The fill factor recovery that was seen with temperature points towards a reduction in $R_s$. The temperature dependent $J-V$ sweeps shown in Figure 2.13 indicates that the fill factor recovery is due to a reduction in series resistance. Another possible explanation would be change in indirect absorption with temperature. Since absorption with an indirect bandgap requires a two-particle interaction with photons
Figure 2.14: Normalized measured and modeled $AM0$ illuminated $J-V$ curve parameters for GaP $p-n$ solar cell.

and phonons, increasing the availability of phonons could increase the probability of the event occurring. The change in temperature wasn’t too large however and assessment of this possibility would require temperature dependent absorption in GaP. SR could possibly be used to assess this effect.

As mentioned previously, the modeled GaP short circuit current density neglects absorption from the indirect band-edge because of the low spectral responsivity at these wavelengths due to the thin device structure at room temperature. A relative increase in efficiency is still seen with the replacement of the measured fill factor with a modeled fill factor. Temperature dependence of a GaP MQW solar cell warrants
Figure 2.15: Temperature dependent $I$-$V$ sweeps measured on GaP solar cell.

further investigation.

The high temperatures the GaP cells are designed for are caused by relative proximity to the Sun, so heating the cell doesn't necessarily provide the proper operating conditions. Increased solar concentration must also be considered. In order to extend the temperature model for GaP to simulated performance in a high temperature space environment, cell temperature was modeled as a function of distance from the sun using the Stefan-Boltzmann equation which depends on intensity, $I$ which was scaled by a concentration factor calculated from the distance from the sun.

\[
\alpha I = (\epsilon_f + \epsilon_r)\sigma T^4
\]  

(2.6)
where $\alpha$ is the absorptivity, $\epsilon_f$ and $\epsilon_r$ are front and rear emissivities, and $\sigma$ is the Stefan-Boltzmann constant. The ratio of absorptivity to emissivity is assumed to be 1, or the cell is assumed to be a perfect black body radiator.

Assuming no power is converted from heat to electricity provides a worst-case temperature assuming the cell is at open-circuit and no incident energy is converted to electricity. Light concentration effects due to solar proximity were also taken into account. The modeled temperature dependent short circuit current density of GaP was replaced with the experimental temperature coefficient in order to correct for the extreme deviation seen. Shown in Figure 2.16 is modeled cell temperature as well as modeled and measured normalized cell efficiency as a function of simulated solar proximity. The left $y$-axis is relative efficiency, the right $y$-axis is cell temperature. The $x$-axis is the distance from the sun in astronomical units, where the surface of the sun is 0 AU and the distance between the sun and the earth is 1 AU. Data was collected up to 362°C (equivalent Mercury orbit) because the metallization was not chosen for temperature stability. Increasing the temperature higher would require high temperature compatible contacts.

The measurements were taken using a hotplate and the LAPSS at NASA GRC. No decrease in relative efficiency increase was measured up to a simulated 0.6 AU solar orbit. However, cell efficiency started decreasing sooner than projected. This is attributed to the effects of high series resistance coupled with the high light concentration. Since there is an $P^2R$ relationship between current, resistance, and power dissipated, a large series resistance term will result in a sharp reduction in fill factor as current increases
with concentration. A GaP device to be used for concentration would need a lower series resistance to operate efficiently. Despite the early drop-off in $\eta$, this still demonstrated the ability of a GaP solar cell to operate at temperatures well above the standard operating temperature limits.

![Figure 2.16: Modeled and measured efficiency of a GaP $p-n$ solar cell in simulated sub-1 A.U. Solar orbits. The red line is modeled black body temperature of an object in orbit around the Sun as a function of distance. The blue dotted line is the modeled efficiency, and the blue data points are measured normalized efficiencies.](image)

2.7 CONCLUSIONS

Multiple quantum well enhanced gallium phosphide solar cells have been grown and fabricated for the first time. Integrated spectral responsivity demonstrated a 0.15 mA/cm² (8%) short circuit current density enhancement from a five period superlattice in a GaP
cell with a measured $J_{sc}$ of 2.08mA/cm² (no ARC) A normalized increase of efficiency of $2.78 \times 10^{-3} \, ^{\circ}C^{-1}$ with increasing temperature was measured for GaP. The temperature stability was further tested along with increased light concentration, and no decrease in relative $\eta$ was seen at a thermally and optically equivalent orbit conditions to that of Venus without any active or passive cooling required. This demonstrates that GaP is a potentially effective material for near-sun photovoltaic applications and can potentially enable a great weight savings for probes used for exploring objects in sub-1 AU orbits.
Chapter 3

Radiation Damage in InP & InGaAs Solar Cells

3.1 MOTIVATION

A major concern for the longevity of satellites in Medium Earth orbit (MEO), or orbital radii from 1.8 to 2.5 times the Earth’s radius ($R_e$) is the high energy radiation effects, the main cause of solar cell degradation in space, from passing through the Van Allen belts where charge particles are trapped by the Earth’s magnetic field. Electron fluxes are as high as $9.4 \times 10^9$ $e^-/cm^2$ and proton fluxes as high as $2 \times 10^8$ $protons/cm^2$ are seen[33]. Incident particles collide with atoms in the crystalline lattice and lose energy through ionizing and non-ionizing interactions. Additionally, at lower fluxes is a presence of an assortment of trapped ions, mainly $\alpha$ particles and $O^+$. An $\alpha$ particle is a high energy helium nucleus (two protons and two neutrons). While radiation tolerance in both bulk epitaxial devices and diffused junction devices has been characterized, little characterization has been performed on newer device designs such as the epitaxial lift-off devices investigated in this study.
3.2 THEORY

Ionizing energy loss occurs when a particle collides with a bound electron. The collision knocks the electron out of position, ionizing the atom it surrounded. This does not permanently damage the device because another electron can be captured, annihilating the trapped charge. Cell degradation is caused by non-ionizing energy loss (NIEL) events where the high energy particle collides with an atomic nucleus. The primary knock-on (PKO) atom is displaced if the energy absorbed from the collision is enough to break the chemical bonds and force the atom out of position, generating a vacancy and, when it stops moving, resting in an interstitial site in the lattice. The generated vacancy/interstitial defect pair is known as a Frenkel defect. If the PKO atom absorbs enough energy, it can generate secondary vacancies. Most of the damage to displacements is concentrated near the stopping range of the particle. An example of an atomic displacement is shown in Figure 3.1 which is a depiction of a zincblende lattice.

In general, around 99% of the energy lost by a particle is to ionization\[34\]. The atomic diameter is on the order of 1 Å while the nuclear diameters are on the order of \(10^{-5} \text{ Å}\). Since the electron cloud occupies \(10^{15}\) times more volume, an electron collision is much more probable. The generation of defects in the crystalline lattice has the effect of shortening the minority carrier diffusion length in the material by increasing the probability of carrier trapping and scattering. Radiation damage can be quantified through the use of damage constants. Minority carrier diffusion length
Figure 3.1: Diagram depicting atom displacement from radiation damage in zincblende lattice (not to scale).

The damage constant is expressed as $K_L$ expressed as

$$ K_L = \frac{1}{L_0} \left( \frac{1}{L_0^2} - \frac{1}{L_0^2} \right) $$

(3.1)

where $L_\phi$ is the minority carrier diffusion length at a given fluence ($\phi$) and $L_0$ is the beginning-of-life (BOL) diffusion length. Carrier removal rate is expressed as a linear constant that is independent of BOL doping expressed as

$$ N(\phi) = N_0 - R\phi $$

(3.2)
where $N$ is doping and $R$ is the carrier removal rate. With high fluences, carrier removal effects can decrease the effective doping of a material which often manifests as a widening in depletion width ($w_d$). If the rate of NIEL is known for a given particle and energy, it can be used to calculate an effective displacement damage dose ($D_d$). Since $D_d$ is a direct measurement of displacement damage, it is consistent across all particle types and energies.

In some cases, such as in InP, carrier removal effects can even proceed as far as to cause type-conversion in the material, changing the polarity of the layer. This effect was demonstrated by Messenger et al.\cite{18} in InP solar cells with proton fluences of $4 \times 10^{13} \text{ p}^+/\text{cm}^2$ with the use of electrochemical capacitance-voltage (ECV) measurements. ECV is a process where an etchant that provides an electrochemical Schottky contact to the sample material so capacitance-voltage measurements can be taken. The material is slowly etched by the electrolyte and a $C-V$ depth profile is measured. Majority carrier concentration and polarity is extracted, giving the doping value for the layer. Starting with a $4.5 \times 10^{16} \text{ cm}^{-3}$ $p$-base, type conversion was seen at a $3 \text{ MeV}$ proton fluence of $4 \times 10^{13} \text{ cm}^{-2}$ where an $n$-type doping of around $10^{15} \text{ cm}^{-3}$ was measured. At this point, Messenger concluded that use of a linear carrier removal rate in the base with increasing fluence must be discontinued.

One result of this type conversion is enhanced long-wavelength spectral responsivity at high particle fluences. It is caused by the shifting of the $pn$ junction from the original emitter-base interface to the base and heavily doped back surface field interface which is caused by type-conversion of the base, shown in the diagram in
Figure 3.2. At low fluences, an InP cell will show consistent degradation, but at the onset of type-conversion $I_{sc}$ shows a recovery from the widening of the depleted region. The recovery in $I_{sc}$ causes a plateau in $\eta$ drop-off. At full type-conversion however, collection from the emitter degrades rapidly and both cell $I_{sc}$ and $\eta$ plummet[18].

![Diagram](image)

Figure 3.2: Diagram depicting change in diode structure with increasing particle fluence. Shown on the top left is the beginning of life structure. The junction is between the $n^+$ emitter and $p$-base. The end of life structure, shown on the bottom right is the junction formed between the type-converted base and BSF.

Indium phosphide, one of the materials of focus in this study, is unique in having shown the ability to show nearly full recovery of radiation damage from low temperature annealing of radiation induced defects at temperatures around $100^\circ C[17]$ in diffused junction cells. This phenomenon has only been seen in diffused junction cells, not in epitaxially grown InP solar cells. A discussion with Dr. Walters yielded the theory that the epitaxially grown InP cells were recovering as they were irradiated, the result of which being a higher radiation tolerance without further recovery effects seen at end-of-life (EOL).

The best way to perform radiation studies is with a collimated monochromatic beam
of protons or electrons so the exact energies, fluences, and path lengths are known. Devices can either be measured in-situ, requiring one sample for the entire study, or an assortment of cells can be exposed to specific fluence levels and tested at the end of the exposure regimen. Degradation is characterized in terms of remaining factor, or a given cell parameter normalized to the beginning-of-life (BOL) value.

In this study, the devices were exposed to radioisotopes because a beamline source was not available. There are a few disadvantages of using a radioisotope. One is the spacial nonuniformity of source activity which can be characterized using a phosphor screen or, in the case of an $\alpha$ emitter, by moving a pinhole around the surface. Another disadvantage is the spectral width, which in the case of $\beta$ emitters is extremely broad. The spectrum can be further widened if the radioisotope decays into something similarly unstable which emits at a different range. A third disadvantage is nonuniformity of path length through semiconductor as emission angles can range from 0 to $2\pi$ where the acceptance angle of the device is less than $\pi$ and changes with the distance from the source.

A radioisotope is an unstable elemental isotope which has a probability of decay. The nuclear decays that occur can generate either an $\alpha$, $\beta$, or $\gamma$ particle. The decay rate is proportional to the probability of decay. The activity of a source is measured in curies ($Ci$) where $1\ Ci$ is $37$ billion decays per second. The radioactive lifetime of an isotope is given in the half-life, or time in which half of the sample will have decayed. In this study, $\alpha$ particles emitted from $^{210}Po$ and $\beta$ particles emitted from $^{90}Sr$ were used. $\alpha$ particles are not present in particularly high fluxes in space but damage coefficients
and NIEL can be used to compare damage across types of radiation.

\(^{210}Po\) has a half-life of 138.376 days and emits primarily \(\alpha\) particles, with around one decay in every 100,000 additionally producing a \(\gamma\) ray. \(^{210}Po\) is nearly monochromatic and emits 5.4 MeV \(\alpha\) particles as it decays into \(^{206}Pb\). \(^{90}Sr\) has a half-life of 28.8 years and decays into \(^{90}Y\), emitting a \(\beta\) particle which then decays rapidly into \(^{90}Zr\) emitting another \(\beta\) particle. Shown in Figure 3.3 is the \(^{90}Sr\) spectrum. An average \(\beta\) energy used for this study is 950 keV.

Figure 3.3: \(^{90}Sr/^{90}Y\) spectrum from \(^{90}Sr\) source.
3.3 EPITAXIAL LIFT-OFF

Epitaxial lift-off (ELO) is a process where the removal of epitaxially grown films is facilitated through the growth of a sacrificial layer which can be selectively etched away. A strain is applied to the wafer surface and it is submerged in the etchant. As the sacrificial film etches, the device peels upward, improving the inflow of etchant which facilitates the removal of the substrate[35]. Other substrate removal techniques involve conventional etching of the substrate while devices are protected from the front of the wafer and an etch-stop film protects the bottom of the device layers.

The are three primary benefits of the ELO approach. It facilitates the reuse of substrates for subsequent growths. It also has the potential to increase the mass-specific power of a PV array from around 210 W/kg to greater than 700 W/kg[36]. The third benefit is what has been driving the next-generation state-of-the-art (SOA) PV technology, which is the growth of inverted metamorphic multijunction (IMM) structures.

Multijunction solar cells have historically been limited to lattice matched materials such as Ge/GaAs/InGaP, but the ideal bottom junction with a GaAs/InGaP middle and top cell would have a 1 eV bandgap. Lattice matched 1 eV dilute nitride GaInNAs alloys were investigated by Geisz et al, but intrinsic defects in the material limited the device performance [37]. Another option is to use graded buffer layers to change the lattice constant while minimizing the presence of threading dislocations through the device. Using ELO technology, the top and middle InGaP and GaAs cells can be grown on a GaAs template, and graded buffer layers allow for the growth of high quality 1eV InGaAs. The entire device structure can then be removed from the substrate and
"flipped," resulting in an IMM solar cell. Looking forward, this technique opens the possibility of growth of 4-6 junction devices.

### 3.4 EXPERIMENTAL SET-UP

The cells measured were ELO InP and InGaAs cells lattice matched to InP and grown on InP substrates. Thin-film cells with a total thickness around 3-4 \( \mu m \) were bonded to thermally conductive, electrically insulating ceramic substrates. The cells were provided by a corporate partner for an assessment of radiation tolerance. A picture of a cell and an example of the device structure is shown in Figure 3.4.

![Figure 3.4: Example cell structure and a picture of the cell mounted on the ceramic substrate. Since the substrate is removed, the total structure is thin.](image)

The short half-life of \( ^{210}Po \) makes the measurement of activity near the time of the study critical. The 5 \( mCi \) \( ^{210}Po \) source flux was measured using a pinhole in parafilm and a geiger counter resting on the can. The geiger counter head was much larger than the pinhole so an acceptance angle of nearly \( \pi \) was assumed. A background
\( \gamma \) ray measurement was required using an unbroken piece of parafilm in order to get a precise reading of the number of counts solely coming through the pinhole. The pinhole was moved across the surface of the source in order to ensure that it was uniform. Because of the short penetration depth of \( \alpha \) particles, no additional shielding of the experiment was required. The flux of the \( \alpha \) source ranged from \( 1-2 \times 10^6 \ \alpha/cm^2/s \) over the course of the study. No additional shielding of the \( \alpha \) source was required because of the relatively low flux and the short stopping distance of \( \alpha \) particles. As an example, alpha particles can be completely stopped by a normal thickness sheet of paper, 37.6 \( \mu m \) of skin cells and even 40.7 \( mm \) in air. This means that the only exposure concern with \( ^{210}Po \) is internal exposure, such as ingestion. The can containing the \( ^{210}Po \) source is shown in Figure 3.5. Cells are placed directly on the can.

![Figure 3.5: Can containing \( ^{210}Po \) source.](image)

The \( \beta \) source, exhibiting a half-life of 28.8 years, and showing too much activity to be measured with the equipment at RIT, had an activity of \( 9.11 \times 10^8 \ \beta/cm^2 \) calculated as in integration across the volume of the source and the surface of a reference cell. It was assumed to be consistent over the course of the study. Due to the activity of the beta source, it was contained in a Delrin polymer fixture (Figure 3.6) shown below.
Plastic is used because the tangled hydrocarbon chains effectively stop $\beta$ particles. The fixture is shielded by an enclosure of lead bricks when not in use, bringing the flux one foot away from the plastic and lead shielding to under $0.1 \text{ mRem/hr}$. A microscope slide was exposed to the $\beta$ source to produce a dark spot over the hotspot of the source. The solar cell was mounted onto the glass slide centered on the hotspot and slid over the source.

![Figure 3.6: Fixture used to protect user and environment from $^{90}\text{Sr}$.

Cells were irradiated to a target fluence, then spectral responsivity was measured on an Optronics Laboratory OL750 monochromator and an Agilent B1500A Semiconductor Device Analyzer was used to perform both $AM0$ illuminated and dark $J - V$ measurements. One concern with performing a radiation study is the mitigation of annealing effects of carrier injection through both optical and electrical injection in an attempt to separate radiation damage effects from any repairing that might occur.
Table 3.1: Beginning of life AM0 illuminated J-V characteristics for cells used in radiation study

<table>
<thead>
<tr>
<th>Sample</th>
<th>Radiation</th>
<th>$J_{sc}$ (mA/cm$^2$)</th>
<th>$V_{oc}$ (V)</th>
<th>FF (%)</th>
<th>$\eta$ (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>InP 5.4 MeV $\alpha$</td>
<td>33.0</td>
<td>0.802</td>
<td>77.4</td>
<td>15.1</td>
<td></td>
</tr>
<tr>
<td>InP 0.95 MeV $\beta$</td>
<td>32.2</td>
<td>0.776</td>
<td>74.8</td>
<td>13.7</td>
<td></td>
</tr>
<tr>
<td>InGaAs 5.4 MeV $\alpha$</td>
<td>56.0</td>
<td>0.369</td>
<td>68.1</td>
<td>10.4</td>
<td></td>
</tr>
</tbody>
</table>

simultaneously. In order to mitigate these effects, the cells were kept dark when measurements were not being taken, and cells were shrouded right before and right after measurements in the solar simulator. Solar cell parameters were extracted from the AM0 illuminated I-V sweeps and effective series and shunt resistances were calculated by measuring the slope around $V_{oc}$ and $I_{sc}$ respectively. Dark I-V curves were analyzed for reverse saturation currents and diode idealities, and external quantum efficiency was calculated from SR and an analysis of damage in cell base and emitter was analyzed from fluence dependent SR at specific wavelengths. Shown in table 3.1 are the beginning of life solar cell parameters for the devices used in this study.

3.5 INDIUM PHOSPHIDE

3.5.1 Introduction

As mentioned previously, the majority of the displacement events happen near the stopping point of the energetic particle bombarding the semiconductor. The simulation software the Stopping and Range of Ions in Matter (SRIM) and the Transport of Ions in Matter (TRIM)[38] were used to calculate stopping ranges and to assess whether the displacement damage was uniform through the cell thickness. TRIM performs a
Monte Carlo simulation to model an ion’s path through a material. In the model, ion energy is lost to ionization events, vacancy generation, and phonon generation. The software also models the effect of atomic recoil events in the lattice. Figure 3.7 shows that the stopping range of $5.4 \, MeV$ $\alpha$ particles is $22 \, \mu m$ into InP.

![Depth vs. Y-Axis](image)

Figure 3.7: TRIM simulation showing $5.4 \, MeV$ $\alpha$ path through InP. The average stopping range is past $20 \, \mu m$.

The device structure was on the order of $3 \, \mu m$, which is significantly shorter than the ion range. Displacement damage was also investigated by plotting a calculated displacement per $\AA$ per ion, shown in Figure 3.8. The number of displacements per $\AA$ is flat until around $10 / \mu m$ into InP, which validates the assumption that the damage profile through the entire device is consistent.
TRIM does not simulate the path of $\beta$ particles, but due to the relatively small size of an electron, stopping distances are very long, requiring thick plastic and lead shielding of the source. The damage caused by $\beta$ particles with an average energy of 950 keV can therefore assumed to be consistent through the entire device structure.

In this study, the primary interest is the lifetime of the InP solar cells under $\alpha$ and $\beta$ irradiation. Assessment was performed using the experimental set-up mentioned above. Of particular interest is SR near the band-edge of InP as this provides information on type-conversion effects in the base of the devices.
3.5.2 α Irradiation

One of the InP cells was subjected to the $^{210}\text{Po}$ α source. Shown in Figure 3.9 are the AM0 IV curves at increasing particle fluences. One of the first things to note is the rapid degradation in fill factor with increasing fluences. This degradation appears to be dominated by the change in slope around $I_{sc}$, or effective shunt resistance of the cell. Also there appears to be a very minor change in $I_{sc}$ going from $2 \times 10^9 \text{α/cm}^2$ to $9 \times 10^9 \text{α/cm}^2$ as well as an introduction of a weak second diode around $0.35 \text{V}$ which is investigated through SR and dark I-Vs. In order to further investigate the relative rates of degradation, solar cell parameters were plotted vs α fluence in Figure 3.10.

![AM0 JV curves across increasing α particle fluence in InP.](image)

Figure 3.9: AM0 JV curves across increasing α particle fluence in InP.
Figure 3.10 shows that cell fill factor degrades more quickly than either $I_{sc}$ or $V_{oc}$. At an interpolated value of $1.9 \times 10^{10} \ \alpha/cm^2$ half the power of an unirradiated device is produced. This plot also shows that there is clearly little change in $I_{sc}$ in the exposure range of $2 \times 9 \times 10^9 \ \alpha/cm^2$. This is similar to previously demonstrated effects and is likely due to a change in device structure through type conversion in the base[18] creating an enhancement in $I_{sc}$ at long wavelengths which will be further investigated through spectral responsivity measurements. The degradation in fill factor can be explained by a change in series resistance, a change in shunt resistance, or a change in diode ideality near the open circuit voltage of the cell. In order to investigate the role parasitic resistances play in fill factor degradation, series resistance and shunt resistance were extracted from the $AM_0 \ IV$ curves.

Figure 3.11 shows the 1-sun effective series and shunt resistance as a function of particle fluence. Beginning of life effective shunt resistance was measured to be $0.926 \ k\Omega$ and series resistance was measured to be $1.08 \ \Omega$. The lowest effective shunt resistance measured was at $9 \times 10^9 \ \alpha/cm^2$. It had a value of $187 \ \Omega$, or $20\%$ of the starting $R_{sh}$. $R_s$ remains relatively flat until high particle fluences where it increased from $1.13 \ \Omega$ at beginning of life to $4.78 \ \Omega$ at the measurement at $2 \times 10^{10} \ \alpha/cm^2$, for a relative increase of $445\%$. These results show that both parasitic resistance terms are degraded under $\alpha$ fluence, though degradation is primarily seen in $R_{shunt}$. The increase in series resistance is likely caused by degradation in current transport characteristics of the semiconductor material from the injected defects (decreased mobility), while the decrease in shunt resistance could be caused by trap assisted tunneling through
injected defect states. Shunt resistance degrades immediately because the introduction of shunt pathways occurs rapidly and requires a lower defect density, while the change in series resistance is caused by transport through bulk layers and requires a higher threshold defect density. Dark diode $IV$ curves were taken to further assess the change in electrical properties of the irradiated device.

Local ideality factor was calculated from the dark diode curves by taking a numerical derivative of $n$ with respect to voltage in the diode equation. Changes in the ideality factor at low voltages can give information about shunting, while changes in the reverse saturation current correlate with changes in the open circuit voltage. An
interesting phenomenon was seen in the trend in changing voltage dependent ideality factor of the diode with increasing exposure. The unirradiated device had an ideality factor near $n = 2$, indicating diode current is mainly from recombination in the space-charge region\[27\]. This is as expected, at least at low to moderate forward bias. At a low fluence, however, the diode ideality factor shifts upwards to around $n = 7$ at low forward bias indicating the possibility of non-radiative recombination through a trap that is not at mid-band, which is assumed for Shockley-Reed-Hall recombination\[27\]. More interestingly, the diode ideality factor at low voltages appears to trend downwards with increasing exposure. This could be caused by the widening of $w_d$ creating
a larger space-charge region.

Diode reverse saturation current densities and ideality factors fit near \( V_{\text{max}} \) are shown in Table 3.2 The leftward shift in the \( IV \) curves with increasing radiation exposure is indicative of a decreasing open circuit voltage with increasing \( \alpha \) fluence and is expected as material quality degrades through the injection of vacancies and interstitials in the crystal lattice. For further analysis of radiation effects, degradation in spectral responsivity is considered.

Figure 3.12: Dark \( IV \) curves and calculated local ideality factor of InP solar cell at each \( \alpha \) fluence.

Looking at figure 3.13, the degradation in external quantum efficiency with increasing fluence appears to be primarily at long wavelengths. The expectation is that the lightly doped base should exhibit the greatest degradation because carriers have the
Fluence ($\alpha/cm^2$) | 0 | $1.94 \times 10^7$ | $1.94 \times 10^8$ | $9.69 \times 10^8$
---|---|---|---|---
$J_0(A/cm^2)$ | $9.69 \times 10^{-11}$ | $1.25 \times 10^{-10}$ | $9.0 \times 10^{-11}$ | $5.05 \times 10^{-11}$
$n$ | 1.60 | 1.63 | 1.58 | 1.49
Fluence ($\alpha/cm^2$) | $2.02 \times 10^9$ | $9 \times 10^9$ | $19.5 \times 10^9$
---|---|---|---
$J_0(A/cm^2)$ | $5.29 \times 10^{-11}$ | $3.99 \times 10^{-10}$ | $3.46 \times 10^{-8}$
$n$ | 1.47 | 1.58 | 2.12

Table 3.2: Table of InP dark diode parameters

longest distance to diffuse and the density of injected vacancies is closer to the doping level than in the heavily doped emitter. A reduction in the diffusion length in the base would explain the trend seen. At $9 \times 10^9 \alpha/cm^2$ however, EQE at long wavelengths begins to increase, and shows a further increase at $2.0 \times 10^{10} \alpha/cm^2$. This is indicative of type-conversion of the base as mentioned above. In order to more clearly display the wavelength specific degradation, the normalized spectral responsivity vs. particle fluence was plotted at specific wavelengths.

Figure 3.14 shows normalized spectral responsivity at 550 nm, 700 nm, and 900 nm vs increasing particle fluence. At relatively low fluences, the rate of SR degradation increases with increasing wavelength. At $2 \times 10^9 \alpha/cm^2$, the spectral responsivity at 900 nm begins to increase. This is a further indication of type conversion in the base and a shifting of the junction depth to the BSF. Looking back to Figure 3.13, it appears that at the last fluence step, collection at short wavelengths drops drastically. This is also indicative of a shift in the junction away from the heavily doped, short diffusion length emitter.
Figure 3.13: External quantum efficiency of InP solar cell across increasing α particle fluence.

3.5.3 β Irradiation

A similar experiment to the one shown above was performed on an InP cell under β irradiation. The average particle energy was 950 keV. The first plot, Figure 3.15 shows diode IV characteristics vs particle fluence. As before, the cells appear to become increasingly shunted with increased radiation exposure. A comparison of solar cell parameters is shown in Figure 3.16. Of interest, however, is an apparent net increase in $I_{sc}$ at a fluence of $9.89 \times 10^{15}$ $e/cm^2$. This will be further investigated through spectral responsivity later on in this study.

As with the α irradiation study shown before, fill-factor degrades at a quicker rate...
than either $I_{sc}$ or $V_{oc}$. This cell was not degraded past 67.6% starting efficiency due to the high fluences required and the limited rate in which the cell could be damaged on the $\beta$ source, but a logarithmic fit yielded an extrapolated 50% $\eta$ point with a fluence of $9.5 \times 10^{16} \beta/cm^2$. $I_{sc}$ and $V_{oc}$ appeared to degrade at close to the same rate. The final $\eta$ of the cell at $9.98 \times 10^{16} e/cm^2$ suggests that an equivalent 5.4 $MeV$ $\alpha$ particle exposure is in the range of mid $10^9 \alpha/cm^2$.

Similar to the $\alpha$ experiment, this cell exhibited primarily shunt resistance degradation, shown in Figure 3.17. The $R_s$ exhibited a 49% maximum increase from 1.47 $\Omega$ while the shunt resistance was reduced to only 24% of the initial value of 733 $\Omega$. 

Figure 3.14: SR degradation at $\lambda_{photon} = 550 \ nm$, 700 nm, and 900 nm vs. increasing $\alpha$ particle fluence in InP solar cell
The relatively minor change in $R_s$ compared to the previous cell that underwent $\alpha$ irradiation is likely due to the lower level equivalent damage done to the cell at the final fluence target. Previously reported results for epitaxially grown InP solar cells showed an irradiated max power point of 76.5% beginning-of-life power at $6 \times 10^{15}$ $MeV e/cm^2$[1]. The cell used in this study measured after a $3.19 \times 10^{15}$ equivalent $1 MeV e/cm^2$ fluence generated 71.1% of the unirradiated max power.

Spectral responsivity of the InP cell was measured at each fluence target. A subset of the calculated external quantum efficiency from SR is shown in Figure 3.18. A similar effect is seen as before where the bulk of the loss in EQE occurs at long wavelengths near the InP band edge. Long wavelength current enhancement begins
Figure 3.16: InP solar cell parameters vs. increasing $\beta$ particle fluence.

to be seen at a fluence of $4.41 \times 10^{15} \text{ e/cm}^2$. This, again, suggests an equivalent 5.4 $MeV$ $\alpha$ exposure of mid $10^9/cm^2$. No sharp reduction of collection around 500-550 nm is seen at the final fluence target. A break-away of normalized spectral responsivity is shown in Figure 3.19.

The wavelengths 550 nm, 700 nm, and 900 nm were again used to analyze the wavelength dependent spectral responsivity degradation with increasing particle fluences. As seen before, a greater degradation is seen at increasing wavelengths at low fluences, and the near-band-edge collection hits a point of inflection and begins increasing, this time at a fluence around $2 \times 10^{15} \text{ e/cm}^2$, indicating the shifting of the
Figure 3.17: Series and shunt resistances of InP solar cell as function of $\beta$ particle fluence extracted from AM0 $J$-$V$ curves.

junction as seen before for $\alpha$ particles.

3.5.4 Conclusions

The ELO InP cell used in this study measured after a $3.19 \times 10^{15}$ equivalent 1 $MeV$ $e/cm^2$ fluence generated 71.1% of the unirradiated max power. This fluence is the equivalent exposure of up to 33.7 years in the Van Allen electron belt[33] with a $\beta$ flux of $3 \times 10^6$ $e/cm^2/s$. Previously reported results for epitaxially grown InP solar cells showed an irradiated max power point of 76.5% beginning-of-life power at $6 \times 10^{15}$ 1 $MeV$ $e/cm^2$[1]. A monochromatic source would provide a better comparison, but it
can be concluded that ELO InP cells perform similarly to standard epitaxial InP cells. Further analysis is required to calculate $D_d$ in order to more fully compare results from $\alpha$ and $\beta$ exposure, but the study suggests that a 950 keV $\beta$ fluence of $1 \times 10^{16}$ e/cm$^2$ is equivalent to a mid $10^9$ 5.4 MeV $\alpha$/cm$^2$ exposure.

### 3.6 INDIUM GALLIUM ARSENIDE

#### 3.6.1 Introduction

As with InP, a TRIM simulation was performed on InGaAs to model the effects of $\alpha$s in InGaAs A $\beta$ study was not performed because the InP was occupied the source for
Figure 3.19: SR degradation at $\lambda_{\text{photon}} = 550 \text{ nm}$, $700 \text{ nm}$, and $900 \text{ nm}$ vs. increasing $\beta$ particle fluence in InP

the entire duration of the study. Shown in Figure 3.20 is a plot of ion paths through InGaAs. The results are similar to those of InP and result in very similar stopping ranges.

The device structure was around $4 \mu m$ thick, which is again significantly shorter than the ion range. Figure 3.21 shows that the number of displacements is consistent until the ions pass through around $10 \mu m$ of material. The assumption, as before, is that displacement damage throughout the entire solar cell is uniform.

Since InGaAs has too narrow of a bandgap to be a suitable material for a high efficiency single-junction solar cell, of particular interest in this study is how well the cell
Figure 3.20: TRIM simulation showing 5.4 $MeV$ $\alpha$ path through InGaAs. The average stopping range is past 20 $\mu m$.

holds up to radiation in comparison with InP. If an InP/InGaAs tandem was fabricated, it would be important to identify what cell would be current-limiting at both BOL and after radiation exposure.

3.6.2 $\alpha$ Irradiation

An $\alpha$ exposure experiment similar to the one performed on the InP solar cell above was performed on an InGaAs solar cell lattice matched to InP. Shown in Figure 3.22 is the results of $AM0 IV$ sweeps at each particle fluence step. The degradation in fill factor of this cell does not appear to be as sharp as the degradation seen in the InP
Figure 3.21: TRIM simulation showing displacements in InGaAs from 5.4 MeV α particles.

cell, but the $V_{oc}$ appears to degrade much more quickly. No $J_{sc}$ recovery is seen at high fluences. Solar cell parameters were plotted vs increasing $\alpha$ fluence in order to more clearly show trends between the degradation of each parameter.

It is clear in Figure 3.23 that unlike the case with InP, the fill factor is not the primary factor showing degradation. In this case $V_{oc}$ degrades the quickest. This effect is expected of narrow bandgap semiconductors. There is an interpolated 50% starting $\eta$ point at $4.7 \times 10^9 \, \alpha/cm^2$. This plot also clearly shows that no $J_{sc}$ recovery or reduction in rate of decline is seen at high particle fluences. This partially verifies that InP, shown in the previous section, is in fact a special case. No type conversion effects
Figure 3.22: AM0 JV curves across increasing $\alpha$ particle fluence in InGaAs solar cell.

were expected. Analysis of parasitic resistances was performed despite the relatively small change in fill-factor and shown in Figure 3.24.

Virtually no change in series resistance with increasing particle fluence was seen in the InGaAs cell and the shunt resistance decreases to 32% of the unirradiated value which, while drastic, is not as drastic as that seen in the InP cells. However, little change in $R_s$ was observed, in contrast to the InP cell was seen at the same $\alpha$ fluence. It is possible that the accelerated rate of $V_{oc}$ degradation seen in InGaAs was responsible for $\eta$ degradation beyond 50% of the starting $\eta$ before enough defects were injected to create significant mobility degradation, which leads to increased series resistance.
Though not as drastic as with InP, the InGaAs diode showed a sharp increase in low-bias diode ideality at low fluences and the diode ideality trended downward with increasing particle fluences (shown in Figure 3.25). Ideality for all diodes remained within the range of $n = 1.25 - 2$ suggesting that diode operation is between totally non-radiative and totally radiative recombination limited operation. The InGaAs cells may have a lower ideality factor than the InP cells because of the $n_i$ dependence of recombination in the quasi-neutral regions[27]. A leftward shift in diode $I/V$ characteristics is again indicative of an increasing reverse saturation current with increasing particle fluence. Table 3.3 shows diode parameters with increasing fluence. The reverse
saturation current shows a slow increase with increasing fluence, while diode ideality factor at low voltage bias decreases and at high voltage bias, increases. Both of these factors coupled together correlate with the decreasing open circuit voltage.

<table>
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<th>Fluence (α/cm²)</th>
<th>J₀ (A/cm²)</th>
<th>n</th>
<th>Fluence (α/cm²)</th>
<th>J₀ (A/cm²)</th>
<th>n</th>
</tr>
</thead>
<tbody>
<tr>
<td>0</td>
<td>2.60 × 10⁻⁶</td>
<td>1.47</td>
<td>1.14 × 10⁹</td>
<td>9.33 × 10⁻⁶</td>
<td>1.34</td>
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<td>2.72 × 10⁻⁶</td>
<td>1.43</td>
<td>2.26 × 10⁹</td>
<td>1.94 × 10⁻⁵</td>
<td>1.39</td>
</tr>
<tr>
<td>1.1 × 10⁸</td>
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<td>1.33</td>
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<td>3.76 × 10⁻⁵</td>
<td>1.31</td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

Table 3.3: Table of Dark IV Parameters for InGaAs cell.

The EQE, shown in Figure 3.26, shows a similar trend to that in InP, with the
rate of collection loss increasing with wavelength, again showing that diffusion length degrades quickest in the lightly doped base. No recovery in long wavelength collection or sharp drop-off in short wavelength collection is seen. This further suggests that no type-conversion and subsequent shifting of junction depth is occurring in the InGaAs cell. This is more clearly demonstrated in Figure 3.27.

Spectral responsivity, normalized to the pre-irradiation value, was taken at 700 nm, 1200 nm, and 1600 nm from the data used for Figure 3.26. This clearly shows the spread in degradation in EQE with increasing wavelength. Also of note is the complete lack of recovery in collection at 1600 nm, near the band-edge of the material that was seen in the InP devices.
Figure 3.26: External quantum efficiency across increasing $\alpha$ particle fluence in InGaAs.

### 3.6.3 Conclusions

In conclusion, an interpolated 50\% starting $\eta$ point was seen at $4.7 \times 10^9 \alpha/\text{cm}^2$. No plateau and plummet effect in $\eta$ or $I_{\text{sc}}$ was observed. There was no evidence of a type-conversion effect occurring in the base of the cell. There is little published data on radiation tolerance of InGaAs solar cells, as the bandgap is too narrow to produce a high efficiency single-junction cell. Growth of a monolithic single-junction InGaAs cell for a future study would be required to assess the radiation tolerance of the InGaAs subcell. A more valuable point of comparison would be between an InP/InGaAs ELO tandem cell and a monolithic GaAs/Ge tandem cell as these are two potentially
Figure 3.27: SR degradation at $\lambda_{\text{photon}} = 550\,\text{nm}, 700\,\text{nm}, \text{and} 900\,\text{nm}$ vs. increasing $\alpha$ particle fluence in InGaAs competing technologies.

### 3.7 COMPARISON OF InP AND InGaAs

The InGaAs cell had a 50% remaining efficiency at $\frac{1}{4}$ the fluence of the InP cells, or $4.7 \times 10^9\,\alpha/cm^2$ and $1.9 \times 10^{10}\,\alpha/cm^2$ for InGaAs and InP respectively. Degradation in terms of remaining $\eta$ factor is shown in Figure 3.28. It showed a similar trend between InP and InGaAs at high fluences, however around 95% remaining $\eta$, the difference in fluences between InP and InGaAs is greater than an order of magnitude, showing that
the InP cell degrades much more slowly at low fluences, so while the difference near EOL may be within one order of magnitude of exposure, the average lifetime efficiency of $InP$ would be much higher.

Since the primary interest in InGaAs was as a candidate for the bottom cell in an InP/InGaAs tandem, BOL and post-$\alpha$-exposure SR was plotted. Since no tandem device was available, InGaAs EQE past the InP band edge was considered. Figure 3.29 shows from expected tandem cell EQE that at BOL, the InP and InGaAs cells would be closely current matched since the absorption overlap decreases the amount of light available for the InGaAs cell. Since the InGaAs cell is primarily absorbing...
at longer wavelengths where degradation occurs at low fluences, the integrated SR drops rapidly, and it becomes a heavily current limiting junction. This demonstrates that a dual-junction InP/InGaAs cell would degrade quickly. One method of improving the lifetime would be to stack a wide bandgap top-cell over the InP cell which would increase the output voltage, but through absorption overlap with InP would decrease the output current meaning the InGaAs cell would no longer become the limiting junction.

Figure 3.29: Depiction of InP/InGaAs tandem expected EQE at BOL and EOL based on EQE from single-junction InP and InGaAs devices
3.8 CONCLUSIONS

The ELO InP cell used in this study measured after a $3.19 \times 10^{15}$ equivalent $1 \text{ MeV} \ e/\text{cm}^2$ fluence generated $71.1\%$ of the unirradiated max power. Previously reported results for epitaxially grown InP solar cells showed an irradiated max power point of $76.5\%$ beginning-of-life power at $6 \times 10^{15} \text{ MeV} \ e/\text{cm}^2$[1]. Further analysis is required to calculate $D_d$ in order to more fully compare results from $\alpha$ and $\beta$ exposure, but the study suggests that a $950 \text{ keV} \beta$ fluence of $1 \times 10^{16} \ e/\text{cm}^2$ is equivalent to a mid $10^{9} 5.4 \text{ MeV} \ \alpha/\text{cm}^2$ exposure.

In the ELO InGaAs cell, an interpolated $50\%$ starting $\eta$ point was seen at $4.7 \times 10^{9} \alpha/\text{cm}^2$ which is around $\frac{1}{4}$ the fluence required in InP to show a similar degradation, but little data is available on radiation damage in InGaAs in literature to compare this to. There was no evidence of a type-conversion effect occurring in the base of the cell. Because of this, no plateau and plummet effect in $\eta$ was observed.

An InP/InGaAs tandem without a top cell would not be suitable for high radiation environments because of the rate of current loss in the InGaAs cell. Adding a top cell would alleviate this concern because it would reduce the amount of current generated in the InP cell through absorption overlap, partially stabilizing the current output of the cell.
Chapter 4

Summary, Conclusions, & Future Work

4.1 GALLIUM PHOSPHIDE PHOTOVOLTAICS

4.1.1 Summary & Conclusions

Multiple quantum well enhanced gallium phosphide solar cells have been grown and fabricated for the first time. Integrated spectral responsivity demonstrated a $0.15 \text{mA/cm}^2$ (8%) short circuit current density enhancement from a five period superlattice in a GaP cell with an integrated $J_{sc}$ of $2.08 \text{mA/cm}^2$.

A normalized increase in efficiency of $2.78 \times 10^{-3} \text{C}^{-1}$ with increasing temperature was measured for GaP. The increase in efficiency was not predicted with the model and is partially caused by improvement in fill factor with increasing temperature. A further study into contacting GaP devices with and without conduction through an AlP layer would be required to locate the cause of the high contact resistances measured.

Temperature stability was further tested along with increased light concentration in order to measure performance in simulated solar orbits within 1 AU. No decrease in relative $\eta$ was seen at a thermally and optically equivalent orbit to that of Venus.
without any active or passive cell cooling required. A cell with lower series resistance could potentially operate efficiently at high solar concentrations, and the modeling of moderate cooling such as the effects of adding a passive heat sink or an IR reflector could extend the effective operating range of the cell to conditions closer to the sun than measured in this study.

This work demonstrates that GaP is a potentially effective material for near-sun photovoltaic applications and can potentially enable a great weight savings for probes used for exploring objects in sub-1 AU orbits. It also demonstrates that the addition of quantum wells could be an effective way to improve performance beyond that of bulk GaP, but further analysis of temperature effects in quantum wells is required.

4.1.2 Future Work

With a re-optimization of device structure and improvement of contact quality, GaP appears to be a promising material for high temperature photovoltaics for applications but metallization with a higher melting point than the alloyed contacts used for this study is required for temperatures above 360°C. One possible candidate is Pd/In[39] which requires an anneal temperature of 600°C but is stable at temperatures in excess of 350°C. An optimized concentrator GaP solar cell with grid shadowing optimized for collection of the generated current at a specific concentration should be designed in order to characterize performance at high solar concentration and the temperature dependent performance of an GaP/InGaP MQW solar cell merits further investigation.
Further optimization of growth and doping of AlP could potentially improve contact quality as well as surface morphology when it is used as a BSF material. The first step requires a study of growth temperature effects on AlP surface morphology and would be focused on reducing the occurrence of hexagonal pyramidal defects on the surface. The second step requires a study of dopant incorporation. Secondary ion mass spectrometry can provide details on degree of molecular incorporation of dopant, but electrical testing such as Hall Effect measurement is required to extract the concentration of electrically active dopant in the AlP film.

More quantum mechanical modeling and analysis of the quantum well structure in the solar cell beyond a simple application of the Schrödinger equation is required to determine whether the InGaP concentration used has a direct or indirect bandgap and to fully understand the effects of quantum confinement between two indirect bandgap materials would be. Strain effects and band bending at the MQW heterojunction interfaces were not taken into account. Finally, further investigation of the radiation tolerance of GaP solar cells is required to fully characterize the performance of these cells in the harsh conditions required for missions such as the Solar Probe+.

4.2 RADIATION DAMAGE IN InP AND InGaAs SOLAR CELLS

4.2.1 Summary & Conclusions

The ELO InP cell used in this study measured after a $3.19 \times 10^{15}$ equivalent $1\ MeV\ e/cm^2$ fluence generated $71.1\%$ of the unirradiated max power. Previously reported
results for epitaxially grown InP solar cells showed an irradiated max power point of 76.5% beginning-of-life power at $6 \times 10^{15}$ $1$ MeV $e/cm^2[1]$. The similar relative radiation tolerance demonstrated through this comparison suggests that the ELO devices behave similar to bulk devices under radiation exposure. The extrapolated half-$\eta$ $950$ keV $\beta$ exposure point at a calculated fluence was $9.5 \times 10^{16}$ $\beta/cm^2$. This is the equivalent exposure of up to $33.7$ years in the Van Allen electron belt$[33]$ with a $\beta$ flux of $3 \times 10^6$ $e/cm^2/s$. Further analysis is required to calculate $D_d$ in order to more fully compare results from $\alpha$ and $\beta$ exposure, but the study suggests that a $950$ keV $\beta$ fluence of $1 \times 10^{16}$ $e/cm^2$ is equivalent to a mid $10^9 5.4$ MeV $\alpha/cm^2$ exposure.

In the ELO InGaAs cell, an interpolated $50\%$ starting $\eta$ point was seen at $4.7 \times 10^9$ $\alpha/cm^2$ which is around $\frac{1}{4}$ the fluence required in InP ($1.9 \times 10^{10}$ $\alpha/cm^2$) to show a similar degradation. Little data is available on radiation damage in InGaAs in literature to compare this to. There was no evidence of a type-conversion effect occurring in the base of the cell, as would be indicated by recovery in near band-edge collection at high fluence levels. Because of this, no plateau and plummet effect in $\eta$ was observed. One method that could be applied to improve the radiation tolerance in the InGaAs cell would be to apply a backside reflector, taking full advantage of the ELO process. With a backside reflector, the base could be thinned, alleviating minority carrier diffusion length issues at high fluence.

An InP/InGaAs tandem without a top cell would not be suitable for high radiation environments because of the rate of current loss in the InGaAs cell. Adding a top cell would alleviate this concern because it would reduce the amount of current generated
in the InP cell through absorption overlap, partially stabilizing the current output of the cell.

4.2.2 Future Work

The next step would be to use displacement damage dose calculations to more fully compare damage across ionizing radiation types and energies. Further assessment of radiation tolerance in InP ELO solar cells would be to characterize temperature and carrier injection (both through optically and electrically injected current) effects on the annealing of radiation damage in InP based photovoltaics and confirm previous reports that low temperature annealing only occurred in diffused junction cells[17]. A radiation study on an actual InP/InGaAs tandem would be useful in supporting the conclusions of this study. One step further would be to perform detailed balance calculations for InAlAsP alloys on InP/InGaAs middle and bottom cells, shown in Figure 4.1 could provide a suitable top-cell material. Finally, a beamline system could be used to expose the cells to a more precise spectrum without variation in incident angle and with a more precise flux.
Figure 4.1: Example of an InP-based three junction solar cell.
References


