Electroluminescence of a-Si/c-Si Heterojunction Solar Cells after High Energy Irradiation

Dissertation

zur Erlangung des akademischen Grades

DOKTOR-INGENIEURIN

der Fakultät für Mathematik und Informatik der FernUniversität in Hagen

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Hagen 2009
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Introduction

Russel Ohl discovered the first silicon solar cell by accident in 1940 (Riordan and Hoddeson, 1997). He was surprised to measure a large electrical voltage from what he thought was a pure rod of silicon when he shone a flashlight on it. The semiconductor revolution of the 1950s followed, which – besides the invention of the transistor - also resulted in the first efficient solar cells in 1954. This caused enormous excitement and attracted front-page headlines at the time. The first commercial use of the new solar cell on a spacecraft started in 1958 with the Vanguard mission. Electrical power supply to satellites was the major commercial application of solar cells until the early 1970s, when the oil-crisis of that period stimulated a re-examination of the cells’ potential closer to home. The industry is now in an explosive period of growth where the subsidised urban-residential use of photovoltaics is providing the main market. The present healthy state of the solar industry is stimulating the fabrication of several large new manufacturing facilities and the commercialisation of new cell technologies. For space applications, solar cells should be optimized for highest power density rather than for highest efficiency. The best power density has so far been obtained with multi-junction III-V based solar cells; a power density value of 3 W/cm2 was recently demonstrated using a InGaP/GaAs cell configuration [1]. Figure 1.1
includes a spectrum of the sun’s light in space (AM0) and in earth (AM1.5). The spectral distribution of light in space is like the theoretical radiation provided by a black body. When the light passes through the atmosphere, some of the light is absorbed or reflected by gasses such as water vapour and the ozone. For this, the typical distribution of light on the surface of the earth is different than the distribution of light in space.

![Figure 1.1. Spectrum of solar radiation in space and on earth](image)

Notice that there are a large number of photons through most of the near infrared region of the spectrum (700 to 1500 nm). At the higher-energy, shorter-wavelength, end of the spectrum, the number of photons drops off dramatically. Current commercial best products based on Ge/III-V multi-junction solar cells can have an AM1.5 efficiency of 42% (for bare unwired cells). Alternatively, thin-film solar cells, sometimes deposited on flexible substrates offer promising possibilities. Even though efficiencies of these cells are much lower (between 7% and 16%), the cost reduction due to the use of less
material can compensate for the loss in efficiency [2]. Typical materials for these kind of solar cells are amorphous hydrogenated silicon (a-Si:H), CdTe and CuInGaSe₂ (CIGS). Promising candidates for space applications are also a-Si:H/c-Si heterojunction solar cells, because it is possible to obtain higher efficiencies than with crystalline silicon homojunction cells. This type of solar cell consists of an amorphous hydrogenated silicon (a-Si:H) film on top of a crystalline silicon (c-Si) absorber and is proposed as technological alternative which avoids the high temperature processing steps implied in conventional c-Si solar cell technology with a diffused p-n junction. The high potential of such a technology was demonstrated 7 years ago [3] by a record cell with the so-called HIT structure (heterojunction with thin intrinsic buffer layer between the a-Si:H emitter and the c-Si base) having an efficiency of 20%. Sanyo, which reached 21%, already demonstrated the high potential of such a devices on n-type CZ silicon substrates [4]. Today most groups are focusing on p-type substrates and recently 17% has been obtained on p-type float-zone (FZ) material, although this process did not completely take place at low temperatures (diffused back surface field) [5]. While the progress with p-type hetero-junction cell is steady, simulation studies indicate that heterojunction solar cells on n-type material are intrinsically better than on p-type material because of the better band alignment [6,7]. It therefore appears that for highly efficient solar cells, n-type material should be used. However, considering the continually increasing efficiencies on p-type, it is not clear whether the superiority of n-type base heterojunction solar cells will experimentally be confirmed. In the Table 2.1 the efficiency of different fabrication are compared.
<table>
<thead>
<tr>
<th>FABRICATION</th>
<th>STRUCTURE</th>
<th>OBTAINED CELL EFFICIENCY</th>
</tr>
</thead>
<tbody>
<tr>
<td>Sanyo (Japan)</td>
<td>HIT 210W module With 17 % Efficiency</td>
<td>19.5 % mass production cell</td>
</tr>
<tr>
<td></td>
<td></td>
<td>21.3 % laboratory cell</td>
</tr>
<tr>
<td>Hahn-Meitner Institut, Berlin (Germany)</td>
<td>p-type FZ silicon with BSF</td>
<td>19.8 %</td>
</tr>
<tr>
<td>Hagen University (Germany)</td>
<td>p-type Flat FZ silicon , with BSF</td>
<td>17.7 %</td>
</tr>
<tr>
<td>IMM-CNR, Bologna (Italy)</td>
<td>n-type CZ silicon, low T (&lt; 250°C)</td>
<td>16.2 %</td>
</tr>
<tr>
<td>C.R. ENEA, Portici (Italy)</td>
<td>p-type CZ silicon BSF (800°C)</td>
<td>17.0 %</td>
</tr>
</tbody>
</table>

Table 2.1 Different a-Si:H/c-Si solar cells types and obtained efficiencies

The heterojunction solar cells might also find application in thin film solar cells, due to the fact that these cells benefit from processing at low temperatures. These cells are often fabricated on or attached to low temperature substrates and therefore processing at low temperatures enables more straightforward process sequences. The most prominent features of the HIT solar cell can be summarized as follow:

- **Simple structure**: a high efficiency can be obtained with no complicated structural techniques, such as partly heavy doping or a partial oxidation method.
- **Simple, low-temperature process**: due to the simple structure. The temperature of process is so low (< 200°C) that degradation of the minority carrier lifetimes for the substrate is negligible even for low-quality silicon materials.
Simultaneous production of surface passivation and p-n junction: with the insertion of a very thin intrinsic a-Si layer at the a-Si/c-Si heterojunction, surface recombination is drastically suppressed.

Stability: the Staebler-Wronski effect, which is seen in a-Si based solar cells, is not seen in the HIT cell. This is probably due to the fact that the a-Si layers are very thin and contribute little to the power generation.

The most usual technique to grow thin silicon films for silicon heterojunction formation is Plasma-Enhanced Chemical Vapour Deposition (PECVD), which has led to commercially available modules.

In addition, the record HIT solar cell, like its predecessors, has a textured surface to minimize reflection losses and to enhance light trapping.

**Motivation and Outline**

Solar cells operating on board of satellites in space are bombarded by high-energy particles, principally protons and \( \alpha \)-particles. The resulting decrease of the solar cell conversion efficiency presents a challenge to the solar cell manufacturer and to the designers of the satellite electric power system. Research about solar cells for space applications includes therefore studies of radiation damage. Solar cells of compound semiconductors have demonstrated higher conversion efficiencies and radiation-resistance than silicon solar cells but have much higher prices and nowadays the space solar cell market is still dominated by crystalline silicon solar cells. In the last years the HIT (heterojunction with intrinsic layer) solar cell, a heterojunction that consists of a doped hydrogenated amorphous silicon (a-Si:H) layer, deposited at low
temperature on top of the crystalline silicon (c-Si) absorber, with the insertion of a thin intrinsic amorphous silicon layer for c-Si surface passivation, has shown to obtain higher efficiencies than purely crystalline silicon homojunction cells. It should therefore be of great interest to use this type of solar cell also in space applications. In this thesis different types of silicon-based solar cells regarding have been compared regarding their capability to withstand high-energy proton irradiation. Electroluminescence (EL) spectroscopy is one of the techniques used to detect defects and impurities in semiconductors. The technique has the advantages to be non-destructive and not to require special sample preparation. There have already been numerous experiments on electroluminescence as a technique to characterize defects in silicon, however mostly regarding only diffused homojunction silicon solar cells. In this thesis, the effect of high-energy particle irradiation on the electroluminescence efficiency of both heterojunction and homojunction silicon solar cells has been investigated with particular emphasis on the absorber bulk and on the hetero-interface degradation.

In the first chapter, after a general introduction of heterojunction theory, the different a-Si/c-Si heterojunction technologies are presented. Subsequently the fabrication details of the solar cells, characterized in this thesis, have been described. In particular, in this thesis, three different types of a-Si:H(n-type)/c-Si(p-type) heterostructure solar cells have been investigated and compared to a conventional, purely crystalline silicon solar cell. In the second chapter the standard measurement techniques are presented as they were applied in this work to evaluate the properties of the solar cells. In the third chapter the electroluminescence measurement theory and
measurement technique are described. In the fourth chapter, the theory of radiation damage is described, particular focused on the proton radiation damage in silicon. In the fifth chapter the experimental results of this research regarding proton irradiated solar cells are reported and for one type of heterojunction solar cells compared to the degradation due to 1MeV electron. A wide range of proton energies – between 0.8MeV and 65MeV - has been used for the solar cell degradation of standard pn-homojunctions and n-type amorphous on p-type crystalline silicon heterojunctions. While at some energies the irradiation has been done only with a single dose, at two distinct energies, namely at 1.7MeV and at 65MeV, the dose dependence of the irradiation damage has been investigated in detail. These two energies have been chosen, because at 1.7MeV the maximum damage has been observed, however with a non-uniform damage profile, while at 65MeV a uniform damage distribution has been achieved.
Chapter 1

Heterojunction technologies

Heterojunction devices are playing an increasingly important role in optoelectronics. They are produced combining semiconductors that have differing band gap energies but usually closely matching lattice constants. Scientists in the late 1950s envisaged the possibility of using two different semiconductors to form a heterojunction as opposed to a homojunction. The early 1970s brought about increased interest in heterojunction for various areas of devices applications. In fact, the production of a room-temperature heterostructure laser demonstrated the feasibility of a new generation of devices whose performance would have been impossible with the conventional homojunction devices.

1.1 Theory of heterojunction solar cells

The importance of the band discontinuities became apparent in the very early days of heterojunction research. The first fundamental step
was the formulation of Anderson’s electron affinity rule in 1962 [8]. In the following a short introduction into the general heterojunction theory is given.

1.1.1 Band discontinuities

Treatment of the discontinuity problem requires a clear definition of the relevant physical quantities. The used basic parameters are illustrated in Figure 1.1, which shows a simplified picture of the interface of a a-Si(n-type)/c-Si(p-type) heterojunction.

![Figure 1.1 Schematic diagram of a a-Si(n-type)/c-Si(p-type) heterojunction interface](image)

Relevant physical quantities, besides $E_g^A$, $E_g^B$, $\Delta E_V$, and $\Delta E_C$, are:
• The barriers, \( V_1 \) and \( V_2 \), due to the band bending at the two sides of the interfaces
• The corresponding *built-in-potential* \( V_{bi} \)
• The distance between the Fermi level and the band edges, \( \delta_A \) and \( \delta_B \)

Several simple equations link the above relevant quantities. Here is a list of the most frequently used:

\[
\Delta E_g = E_g^A - E_g^B \quad \quad [1]
\]

\[
\Delta E_V + \Delta E_C = \Delta E_g \quad \quad [2]
\]

\[
\Delta E_V + V_{bi} + \delta_A + \delta_B = E_g^A \quad \quad [3]
\]

\[
\Delta E_C - V_{bi} - \delta_A - \delta_B = -E_g^B \quad \quad [4]
\]

Equation 2 shows that, once \( \Delta E_V \) has been determined and we can calculated also \( \Delta E_C \). Equations 3 and 4 emphasize the direct link between band discontinuities and junction barriers heights. Discontinuities in the valence band or in the conduction band at the interface can form barriers for carrier transport. Only carriers with sufficient kinetic energy can pass such a barrier. These barriers have different forms and magnitudes for electrons and holes: under dark conditions and when a forward voltage is applied, electrons have to cross the conduction band discontinuity (\( \Delta E_C \)) plus the band bending when being injected from n-type to p-type, whereas holes flowing in opposite direction experience only the band bending. The nature of the barrier depends on the doping levels (\( N_A \) and \( N_D \)) and the electron affinities (\( \chi_1 \) and \( \chi_2 \)) of the materials.
1.1.2 Defects at a-Si/c-Si heterojunction interfaces

However the principal problem of a-Si:H/c-Si heterojunction is the recombination of carriers at the interface, because many of the transport properties are controlled by the electronic band profile at the heterointerface. If there is a high recombination at the interface, the splitting of the quasi-Fermi levels in the bulk will be strongly decreased. It is possible to reduce this recombination rate with defect passivation techniques.

Different techniques are used in order to reduce the defect density at the surface of the crystalline silicon absorber. It is possible to remove them directly via chemical processes (HF etching of the wafer e. g.) or indirectly by the deposition of passivation layers, that induce a strong band bending (field-effect passivation).

The defects introduced at the a-Si/c-Si heterojunction interface increase the electron-hole recombination rate due to the introduction of non-radiative recombination centers. To get more insight into this mechanism, the a-Si:H/c-Si heterojunction structure has been simulated with the AFORS-HET-Code.

In the Figure 1.2 the resulting dependence of the fundamental solar cell parameter on the a-Si/c-Si interface state density for an a-Si:H/c-Si heterojunction without i-layer is shown.
Figure 1.2 Simulated dependence of the fundamental parameter on the a-S/c-Si interface state density for an a-Si:H/c-Si heterojunction without i-layer.

The strong reduction of the open circuit voltage with increasing interstate density is due to additional interface recombination and can be partially suppressed by ensuring a strong band bending in the crystalline silicon (p-type) absorber, as mentioned before. It is essential to passivate the interface in a way that the resulting defect densities not larger then $N_{\text{trap}} \approx 5 \cdot 10^{10}\text{cm}^{-2}$. 
1.2 Technology of the investigated solar cells

In the beginning of the ‘80s the idea was launched to combine a-Si and c-Si in a solar cells [8]. Such a heterojunction solar cell would benefit from the high efficiency of c-Si and the low production costs of a-Si and therefore has therefore the potential to become a low cost and high efficiency solar cell. Ever since the first heterojunction solar cell (an a-Si:H/µc-Si heterojunction with an efficiency of about 9%) was presented by Okuda et al [9] in 1983, the efficiency of these solar cells has increased steadily.

An important advancement was the use of high quality wide band gap window layers for junction formation such as amorphous silicon carbide (a-SiC:H) [10], low temperature deposited amorphous silicon (a-Si:H) [11], electron cyclotron resonance (ECR) plasma deposited microcrystalline silicon carbide (µc-SiC:H) [12,13,14], or very high frequency (VHF)-plasma deposited microcrystalline silicon (µc-Si:H) [15,16].

The wide band gap of these window layers ensures a low amount of light absorption in the highly doped region of the solar cell. Tanaka et al [17] showed the importance of incorporating a thin (about 5nm) intrinsic a-Si:H buffer layer at the heterojunction interface. These so called HIT structure (heterojunction with Intrinsic Thin layer) showed a reduced dark current, higher open circuit voltage, and higher fill factors than heterojunctions without such buffer layer. Heterojunction solar cells incorporating these window and buffer layers exhibited efficiencies in the range of 13-15%.

Additional improvement of efficiency was possible by depositing a so-called back surface field (BSF) layer on the backside of the wafer. This
is a highly doped layer that reduces the recombination losses at the back of the cell by shielding minority carriers from the highly defective c-Si back surface. Finally, texturization of the front and back surface was used to increase light scattering and increase the red response of the cell. Sawada et al [18] demonstrated an efficiency of 20.0% for a 1.0cm² heterojunction solar cell using a texturized n-type c-Si wafer and a double side a-Si passivation.

Four different types of solar cells from different manufacturers have been investigated: two of them are n-type a-Si:H on p-type c-Si heterojunction structures, one with (cell 1469) and one without (cell 1458) a thin intrinsic a-Si:H buffer layer deposited before the n-type a-Si:H deposition, one of them is n-type a-Si:H on p-type c-Si heterojunction but without anti-reflective coating and the last type is a crystalline silicon homojunction solar cell with the top n-layer obtained by diffusion and intended for concentrator applications [19].

Some fabrication details of the four different cells are summarized again in Table 1.1. It should be noted that the investigated cells have all modest initial conversion efficiencies, because for our study not completely optimized solar cells have been used.

All four types of solar cells have been shown to reach efficiencies around 17% for heterojunctions and 20% for homojunction, when optimized. In the comments of Table 1.1 the not optimized features are mentioned. In particular: the homojunction cell SCH1 has been manufactured without anti-reflective coating (ARC), the heterojunction cell Hjp14a has been fabricated without top transparent conductive oxide (TCO) layer and the heterojunction 1461 has been done without back-surface-field (BSF).
Table 1.1 Comparison of the different structures and initial efficiency of two types of a-Si:H/c-Si heterojunction solar cells and one homojunction crystalline silicon solar cell

<table>
<thead>
<tr>
<th>Sample name</th>
<th>Cell type</th>
<th>Base material</th>
<th>Base thickness</th>
<th>Cell area (cm²)</th>
<th>Comments</th>
<th>Initial efficiency</th>
</tr>
</thead>
<tbody>
<tr>
<td>SCH1 (ENEA)</td>
<td>c-Si homojunction</td>
<td>FZ (100) 0.3 Ωcm</td>
<td>p-type 230 µm</td>
<td>1.1</td>
<td>no ARC</td>
<td>12.2 %</td>
</tr>
<tr>
<td>Hip14a (ENEA)</td>
<td>a-Si/c-Si heterojunction</td>
<td>CZ (100) 1 Ωcm</td>
<td>p-type 350 µm</td>
<td>1.98</td>
<td>no TCO</td>
<td>11.2 %</td>
</tr>
<tr>
<td>1461 (Hagen)</td>
<td>a-Si/c-Si heterojunction</td>
<td>CZ (100) 1 Ωcm</td>
<td>p-type 230 µm</td>
<td>1.00</td>
<td>no BSF</td>
<td>12.7 %</td>
</tr>
<tr>
<td>1458 (Hagen)</td>
<td>a-Si/c-Si heterojunction</td>
<td>FZ (100) 0.8 – 1.2 Ωcm</td>
<td>p-type 230 µm</td>
<td>1.00</td>
<td>no i-layer and no BSF</td>
<td>12.8 %</td>
</tr>
</tbody>
</table>

1.2.1 Silicon homojunction cells

The starting material is a strongly p-type doped (0.3 Ωcm) monocrystalline <100> oriented silicon Float Zone (FZ) 250 Ωm thick wafer. After conventional RCA cleaning the n-type emitter is fabricated on the top of the wafer by a diffusion process starting from POCl₃ gas mixture and performing a subsequent drive-in process at 940 °C of temperature. During this step the backside of the wafer is protected by a spin on glass oxide (SOG). The emitter formation ends with a wet chemical etching of SOG and POCl₃ residual in a 10 % HF bath. A junction depth of 1 µm has been chosen in order to minimize recombination and avoid current crowding inside the emitter layer for high charge carrier concentrations. To reduce the surface recombination velocity the wafers have been thermally oxidized at a temperature of 850 °C. For the thermal oxide a thickness of 16 nm has been chosen in order to achieve a good compromise between passivation effect and antireflection coating refractive index matching.
The oxide present on the back side of the wafer has been removed before the deposition of the metallic back contact by dipping the wafer in a 10% HF bath while the front side of the wafer has been covered with a photoresist. The back contact of the cell has been manufactured by a blanket 2 µm Al e-beam evaporation. To ensure a good ohmic contact between wafer and Al, an annealing step at 550°C in N₂ followed. Finally a metallic grid has been obtained by a lift-off procedure on a trilayer (50 nm Ti, 50 nm Pd and 10 µm Ag) electron beam evaporated. Further details can be found in Ref. [20]. This structure is shown Figure 1.3.

![Figure 1.3 Cross section of the investigated homojunction solar cells](image)

### 1.2.2 Silicon heterojunction cells

Three different types of silicon heterojunction solar cells have been characterized in this thesis. All of them have a p-type silicon absorber and a n-type silicon emitter. The cells (named 1458 and 1461), that have been fabricated at the Fernuniversität in Hagen, did not
incorporate a back surface field (BSF). One of the solar cells is realized without (cell 1458) and one of them with (cell 1461) a thin intrinsic a-Si:H buffer layer deposited before the n-type a-Si:H deposition. The device, deposited at the ENEA Research Center in Portici (type Hjp14a), incorporated as well a BSF layer as an intrinsic a-Si:H buffer layer, but was realized without anti-reflective coating.

1.2.2.1 Heterojunction cell without i-layer and BSF

The samples are prepared on polished 4 in. float zone (FZ) wafers (0.8–1.2 Ω cm) with (100) orientation. In order to remove the contaminations on the front side a standard RCA process has been used. The oxide is removed by a 2% hydrofluoric acid (HF) dip. After the dip the samples are immediately inserted into the standard parallel plate MRG three-chamber plasma enhanced chemical vapor deposition (PECVD) setup. The a-Si:H layers are deposited at 13.56 MHz and an RF-power of 4 W. Process gases are semiconductor-grade silane (SiH₄) and phosphine (3% PH₃ diluted in SiH₄) with gas fluxes of 24 and 12 sccm, respectively. The sample temperature and the process pressure during deposition are 230°C and 300 mTorr, respectively. In order to reduce reflection losses and to enhance the transversal conductivity Indium Tin Oxide (ITO, 3×10⁴ Ω cm) is deposited by dc sputtering from a ceramic target. Finally, the front and back contacts are prepared by E-gun evaporation. 30 nm Cr and 4 μm Ag layers structured by photolithography are used on the front side and 2 μm Al (full area) is deposited on the backside [21]. This structure is shown in Figure 1.4.
1.2.2.2 Heterojunction cell without BSF

These samples have been prepared on 400 µm thick, 1 Ωcm CZ-crystalline silicon wafers with <100> orientation. The a-Si:H layers are deposited by PECVD growth at 230°C. Before the deposition of the 15 nm thick amorphous n-type emitter layer, an about 5 nm thick intrinsic a-Si:H layer has been deposited. In order to reduce reflection losses and to enhance the transversal conductivity Indium Tin Oxide (ITO, 3·10⁻⁴ Ωcm) has been deposited by sputtering on top of the amorphous silicon n-layer. Finally, the front and back contacts have been prepared by e-gun evaporation. 30 nm Cr and 4 µm Ag layers structured by photolithography are used on the front side and 2 µm Aluminum (full area) has been deposited on the backside. Wet chemical etching of the ITO has been used to separate the 1 cm² large solar cell test structures. Further details can be found in Ref. [22]. This structure is shown Figure 1.5.
1.2.2.3 Heterojunction cell without TCO

The following growth technology and sequence has been used: a 400 µm thick, <100> oriented, 1 Ωcm p-type crystalline CZ-silicon wafer is exposed to KOH etchant for surface texturing, resulting in an irregular inverted pyramid structure with about 20 µm surface roughness. After etching the average wafer thickness is about 350 µm. As next step aluminium screen-printing and subsequent annealing at 900 °C do the back contact metallization. This annealing step is needed to produce a back surface field (BSF). After a cleaning procedure, including a dip in 1 % HF for native oxide removal, the 5 nm thick intrinsic and the 15 nm thick n-type a-Si:H layers have been deposited at about 200 °C in a continuous deposition step without interruption and changing to n-type deposition only by changing the gas phase composition. This continuous deposition has been chosen in order to
avoid unstable plasma conditions during the beginning of the n-layer deposition. Finally the front contact has been fabricated in a grid shape by photolithographic step on 500 nm thick e-beam evaporated Ag, resulting an active area of 1.98 cm². Further details can be found in Ref. [23]. This structure is shown Figure 1.6.

Figure 1.6 Schematical drawing of the heterojunction type Hjp14a
In the present chapter the "standard" measurement techniques, which allow to characterize the performance of solar cells are reviewed. The current collection mechanism of a p-n heterojunction solar cell is similar to that of a p-n homojunction solar cell. The main contribution to the photocurrent comes from the base region, with a smaller contribution coming from the top emitter layer and the depletion region. The photocurrent of a p-n heterojunction solar cell can be derived in a similar way as that of a p-n homojunction solar cell discussed earlier. The p-n heterojunction solar cell usually has a better short-wavelength response, lower series resistance, and better radiation tolerance than a conventional p-n homojunction solar cell. In order to obtain a maximum short-circuit current, open circuit voltage, and conversion efficiency, it is essential that materials selected...
for fabricating the heterojunction cells must have compatible thermal expansion coefficients.

Energy band discontinuities at the heterointerface of a heterojunction cell must be minimized to avoid barrier formation (spike) at the heterointerface where photocurrent collection can be severely degraded.

### 2.1 Electrical characterization

Current-Voltage (I-V) curves are the most commonly used characterization tools for solar cell devices [24]. In this technique, the current is measured as a function of voltage for a cell, in both illuminated and dark conditions. The solar cell acts as a diode, and the I-V curves usually reflect these properties.

#### 2.1.1 Dark current-voltage characteristics

The dark J-V characteristics when evaluated as a function of temperature and plotted on a semi-logarithmic scale (Figure 2.1) can be extremely useful in identifying the recombination mechanisms and hence the potential improvement in $V_{OC}$.
Figure 2.1 An illustrative dark (log J)-V plot, indicating the effect of series resistance $R_s$, shunt resistance $R_{sh}$ and two distinct dark current mechanisms [25]

The ideality factor $n$ and the saturation current $I_0$ can be evaluated from the linear regions of the (log I)-V plot by using the slope and by extrapolating the straight line to zero voltage respectively. All the common current mechanisms have the exponential dependence of equation 1, and their voltage and temperature dependence are indicated in Table 2.1 [27].

$$I = I_{ph} - I_0 \left[ \exp \left( \frac{qV}{nkT} \right) - 1 \right]$$

[1]

All the various recombination mechanisms listed in Table 2.1 occur in parallel and one or more of them may be dominant in a given solar cell.
Table 2.1 Dark current mechanisms and their functional dependence on voltage and temperature [25]

For example, in the characteristics of Figure 2.1, space charge region recombination is the likely mechanism at low voltages (since $n = 2$), while minority carrier bulk diffusion dominates at higher forward voltages (until it is limited by $R_S$).

The minority carrier bulk diffusion sets the ultimate limit on the maximum $V_{OC}$ obtainable from a solar cell, and hence attempts are made to suppress or eliminate all the other bucking current components in designing the optimal cell with a given materials system.
2.1.2 Photo-current-voltage characteristics

The illuminated current-voltage curve is similar as the dark one, only shifted downwards of a negative current due to the photo-generated carriers under the influence of the internal field. This negative current occurs when holes move towards the back contact and electrons move towards the front contact, while when a forward voltage is applied, carriers are moved in the opposite direction producing a positive current. When the total current is zero the measured voltage is the Open Circuit Voltage \( V_{OC} \), while when the applied voltage is zero the measured current is the Short Circuit Current \( I_{SC} \). The series resistance, \( R_S \), is the inverse of the slope measured for \( V \to \infty \), and the shunt resistance \( R_{sh} \) is the inverse of the slope for \( V \to -\infty \). A bad back contact will yield a device with a higher series resistance. An ideal I-V curve is shown in Figure 2.2 with the parameters of interest inserted for both the illuminated and dark conditions.

![Figure 2.2 Ideal I-V characteristics curve for a solar cells](image-url)
Other parameters of interest can be derived from those obtained directly from the I-V curve. One of these parameters is the fill factor (FF) that is a measure of the efficiency of a particular cell. The definition of the fill factor is given in equation 2:

\[ FF = \frac{P_{\text{max}}}{V_{\text{oct}} \times I_{\text{sc}}} \]  

[2]

where \( P_{\text{max}} \) is the total power of the incident illumination (typically 1 sun (AM 1.5) or 100mW/cm\(^2\)). Another parameter of interest is the efficiency (\( \eta \)). For a solar cell the efficiency is defined by the equation 3:

\[ \eta = \frac{P_{\text{out}}}{P_{\text{in}}} = \frac{V_{\text{M}} \times I_{\text{M}}}{E \times A_{C}} \]  

[3]

where \( P_{\text{out}} \) is the maximum electrical power output of the cell, \( P_{\text{in}} \) is the input light irradiance (\( E \), in W/m\(^2\)) under standard test conditions (STC) and the surface area of the solar cell (\( A_{C} \), in m\(^2\)). These basic parameters depend strongly on the intensity and on the spectrum of the light, used for characterization. Therefore it is important to have a calibrated light source that matches as close as possible the intensity as well as the spectrum of the sunlight. For all presented measurements the AM1.5 spectrum (100 mW/cm\(^2\)) has been used.

The current-voltage measurements have been performed using a four-terminal probe setup. The use of separate channels for current injection
and for the voltage measurements allows minimizing measurement errors due to losses that are related to cables, connections and wiring.

The current-voltage measurement setup is shown in Figure 2.3. A programmable Voltage/Current Source (Keithley 228A) supplied the voltage in a chosen range, the current and the effective voltage were measured using two multimeters (Hp3478 A). For the current measurement a calibrated 1Ω shunt resistance (Yokogawa 2792) has been used. In this case a current measurement input resistance is known and can be corrected for.

![Figure 2.3 Setup for current-voltage measurements under illumination](image)

An X-25 solar simulator from Spectrolab equipped with a 3 kW xenon short-arc source lamp with a maximum irradiance value of 120mW/cm² has been used as solar simulator. The lamp current is regulated by a variable shunt in constant current mode. In Figure 2.4 the schematic of the optics of this simulator is shown.
The lamp is attached to a special strain-free adjustable mount for positive alignment, moreover there are interchangeable optical packages and special filtering and a collimating lens accessory (M1, M2 and M3) [26].

### 2.2 Spectral Response

The optical performance of a solar cell can be acquired by a measure of its spectral response (SR) that describes the sensitivity of the solar cell to optical radiation of different wavelengths. This quantity is the ratio of the current generated by the solar cell to the power incident on the solar cell:

$$SR(\lambda) = \frac{I(\lambda)}{P_\lambda}$$  \[4\]
Figure 2.5 shows a typical spectral response measurement of a heterojunction solar cell.

![Spectral response of a heterojunction solar cell](image)

**Figure 2.5 Typical spectral response of a heterojunction solar cell**

The spectral response is conceptually similar to the quantum efficiency (QE). The quantum efficiency is defined as the number of measured electrons divided by the number of photons, incident on the solar cell. The relation between the two entities is given by the subsequent equation:

\[ SR(\lambda) = QE(\lambda) \frac{q\lambda}{hc} \]  \hspace{1cm} [5]

The photocurrent of a cell can be calculated by the following integration:

\[ J_{AM1.5} = q \int_{\lambda_{min}}^{\lambda_{max}} SR(\lambda) \Phi_{AM1.5}(\lambda) d\lambda \]  \hspace{1cm} [6]
It is possible to define two types of different quantum efficiency, one is the internal quantum efficiency (IQE) and one is the external quantum efficiency (EQE). They are defined as follows:

\[
IQE = \frac{\text{number of measured electrons}}{\text{absorbed photon number}}
\]

\[
EQE = \frac{\text{number of measured electrons}}{\text{incident photon number}}
\]

The setup used for the measurement of the spectral response or the derived external quantum efficiency is shown in the Figure 2.6.

The monochromatic probe light is generated using a halogen lamp and a monochromator and gets modulated when passing the mechanical
chopper. After passing a beam splitter, a fraction of the light is measured by a radiometer in order to determine the incident photon number on the solar cell and a PC controlled motor changes the monochromator wavelength. The photogenerated current of the solar cell is first amplified by a transimpedance preamplifier and then further amplified using a lock-in amplifier. The use of the lock-in technique renders the system insensitive to external electrical noise and to variations of the external light not passing through the chopper and hence through the monochromator and enables also to add a bias light to the chopped light.

The presented results are, however, all obtained without bias light.

### 2.2.1 Diffusion length determination from spectral response measurements

Diffusion length determination using spectral response data is often carried out with an Schottky barrier device, using a semitransparent (approximately 100 Å thick) metal film. For illumination from the top, the short-circuit photocurrent is approximately given by

\[
I_{sc} \approx q\phi(\lambda)T(\lambda) \left[1 - \exp(-\alpha W) + \frac{\alpha L}{1 + \alpha L} \exp(-\alpha W)\right] \tag{8}
\]

where \(\phi(\lambda)\) is the incident photon flux per second, \(T(\lambda)\) is the transmission through the upper metal layer, \(\alpha W\) is the absorption
coefficient in the semiconductor, L is the minority carrier diffusion length and W is the width of the depletion region.

In the classical method of obtaining L from spectral response the short circuit photocurrent is measured as a function of λ over a range wherein \( \alpha(\lambda)W < 1 \). Then equation 8 may be simplified to

\[
\left( \frac{I_{sc}}{q\phi(\lambda)T(\lambda)} \right)^{-1} = \frac{1}{L} \left[ L + \frac{1}{\alpha(\lambda)} \right]
\]

[9]

Thus a plot of the inverse of the short-circuit photocurrent per incident photon (in the semiconductor), i.e. the inverse quantum efficiency, against \( 1/\alpha(\lambda) \) will yield a straight line with a horizontal axis intercept of magnitude equal to the diffusion length. While conceptually simple, this method suffers from the need to know the precise values of the photon flux incident on the cell and the transmission coefficient T of the Schottky metal.

Of course, if the assumptions leading to equation 8 are not valid, L can be deduced from the measured spectral response by curve fitting to exact theoretical expressions (including the effect of back-surface recombination velocity [28,29]).

### 2.3 Light Beam Induced Current

Light beam induced current (LBIC) is a non-destructive semiconductor characterization technique that has been used to examine p-n junction.
The LBIC configuration used to measure the spatial distribution of the solar-cell photocurrent is shown in Figure 2.7.

![Figure 2.7 LBIC setup [30]](image)

The light source is selected from laser diodes of different wavelengths. Current to the laser diode is electronically modulated to produce an AC laser beam, and the modulation also provides the reference signal for a lock-in amplifier. Attenuation over a wide dynamic range is set with digital control. The beam is expanded, the polarization is selected and the beam is sampled for both intensity and back-reflection from the cell. The tested device is moved under the fixed laser beam on computer-controlled $X$, $Y$ translation stages. The photogenerated current is measured by an $I-V$ converter and sent to a PC. In Figure 2.8 a typical LBIC image of polycrystalline silicon solar cell, illuminated with a focused laser diode at a wavelength of 650nm, is shown. One can clearly identify two fingers of the top-metallization grid, where the light gets completely reflected and hence no photocurrent is collected,
and the boundaries between different grains of the polycrystalline silicon.

Figure 2.8 LBIC images of polycrystalline silicon solar cells [31]

Figure 2.9 illustrates this situation in a schematic cross-section through an illuminated cell implying a shunt.

Figure 2.9 The equivalent circuit of local photocurrent generation near a strong ohmic shunt in a solar cell [30]

Another possible application of LBIC imaging is the detection of local shunts in solar cells. In Fig. 2.7 the electrical of an equivalent circuit of local photocurrent generation near a strong ohmic shunt in a solar cell is shown. One observes, that the generated photocurrent is divided into
two parts, the $I_{LBIC}$ flowing to the current amplifier and $I_{shunt}$ flowing across the shunt. Assuming the input resistance to be zero, the LBIC signal is given by:

$$I_{LBIC} = I_{photo} \frac{R_{shunt}}{R_{path} + R_{shunt}} \quad [13]$$

If $R_{shunt} \gg R_{path}$, the beam is sufficiently distant from the shunt, so that the photocurrent flows to the current amplifier. Otherwise at least some part of the photocurrent is drained by the shunt, leading to a dark LBIC contrast around sufficiently strong shunts.

### 2.4 Electron Beam Induced Current

Connection of an electronic device to a current amplifier while irradiated by the electron beam of a scanning electron microscope (SEM) permits to generate highly resolved charge-collection images. This is also known as EBIC technique (Electron Beam Induced Current). The term EBIC is used generally to represent a collection of techniques based on measurement of: (a) minority charge carriers generated by energy deposition of a primary beam; and/or (b) the flow of the primary beam current itself, the latter being less common. Since electron beams may be focused and EBIC utilizes minority carriers, the potential of the technique for studying thin-film solar cells with sub-grain scale resolution was recognized early on. Best known among the EBIC techniques is the cross-section or lifetime curve method; the
device is cleaved and the beam scanned perpendicularly to the layer stack. In Figure 2.10 an EBIC setup is reported.

\[ I_{EBIC} = I_0 \exp \left( -\frac{|x|}{L} \right) \]  

[14]

where \( x \) is the distance from the junction, \( I_0 \) a constant and \( L \) the minority-carrier diffusion length of electrons or holes. In the Figure 2.11 an SEM image of an intrinsic amorphous silicon/p-type crystalline silicon heterojunction is reported. It is possible to see that EBIC line-scan, the white line, linearly increase in the p-type crystalline silicon and the maximum point is in the intrinsic amorphous silicon.
The cross-sectional EBIC therefore allows to monitor the junction position and homogeneity and can be used to measure the minority carrier diffusion length. In Figure 2.12, the EBIC image of an intrinsic amorphous silicon/p-type crystalline silicon heterojunction is shown.

Figure 2.11 EBIC line scan (white line) and SEM image of an intrinsic amorphous silicon / p-type crystalline silicon heterojunction

Figure 2.12 EBIC image of an intrinsic amorphous silicon/ p-type crystalline silicon heterojunction
2.5 Time Resolved Microwave Conductivity

Time-Resolved Microwave Conductivity (TRMC) is another powerful technique used to evaluate the degradation of semiconductor material by irradiation. This technique is based on the measurement of the relative change of the microwave power reflected from a semiconductor \([\Delta P(t)/P]\) upon a small increase in the conductivity. It was shown that the TRMC signal, \(\Delta P(t)/P\), is proportional to the induced conductivity, \(\Delta \sigma(t)\), in a small perturbation approximation. A particular example of such a setup is shown in Fig. 2.13 and consists of three parts connected to the respective ports of a circulator.

![Figure 2.13 Schematical representation of the TRMC measurements setup [32]](image)

The microwave source is connected to the entrance port (1) of the circulator. Connected to the second port there is the sample, which is in this way irradiated by the incident microwave. The microwave signal, reflected by the sample, is then directed by the circulator to the detector, which is connected to the third port. Microwave source, circulator, sample and detector are connected with waveguides. The
power reflection coefficient $R$ (defined as the ratio between the reflected power from the sample $P(t)$ and the incident power $P$, in the reference plane) describes macroscopically the interaction between the sample and the microwaves. Microwave power loss between sample and circulator is neglected. Under these conditions, the reflection coefficient at the detector is the same as the reflection coefficient $R$ in the reference plane.

For a small uniform $\Delta \sigma(t)$, an approximation to the TRMC signal $\Delta P(t)/P$ is given by [33]:

$$\frac{\Delta P(t)}{P} = A \Delta \sigma(t)$$  \hspace{1cm} [15]

$A$ is called the sensitivity factor and is given by:

$$A = \left( \frac{1}{R(\sigma)} \right) \left( \frac{\partial R(\sigma)}{\partial \sigma} \right)$$  \hspace{1cm} [16]

Hence, $A$ can be determined by measuring $R$ for different values of $\sigma$ in a fixed cell geometry. However, the photoconductivity is rarely uniform.

The photoconductivity can be non-uniform because of the non-uniformity of both the initial induced carrier distribution and the charge-carrier decay processes (e.g., surface recombination). The generalization of Equation 17 to a non-uniform $\Delta \sigma(z,t)$ is
\[
\frac{\Delta P(t)}{P} = \int_{0}^{d} A(z) \Delta \sigma(z, t) dz
\]

where \( A(z) \) is the generalized sensitivity coefficient for a non-uniform response and \( d \) is the thickness of the sample. In the general case, the change in the TRMC signal \( \Delta P(t)/P \) with time is due to the decay of the photoconductivity and the change in the distribution of the excess conductivity with time.
Chapter 3

Electroluminescence

3.1 Optical processes in semiconductors

Light absorption and emission in a semiconductor depend strongly on the band structure of the semiconductor. Direct band gap semiconductors have a higher probability for photon absorption and also for radiative recombination than indirect band gap semiconductors. This difference is illustrated with Figure 3.1 and can be explained based on the energy and momentum conservation required for electronic transitions.
Figure 3.1 (a) Photon absorption in a direct band gap semiconductor, (b) Photon absorption in an indirect band gap semiconductor

In direct band-gap materials, the minimum of the conduction band and maximum of the valence band are coincident in the k-space, the probability of radiative recombination is high and the radiative lifetime is typically of the order of a few nanoseconds. The radiative recombination requires an additional phonon, which significantly reduces the probability and rate of that process, resulting in a radiative lifetime of around 100 µs above 20K [38,39] (depending on doping), given by:

\[ \tau_{rad} = \frac{1}{B(p_0 + n_0 + \Delta n)} \]  

[1]

where B is the coefficient for radiative recombination, \( p_0 \) and \( n_0 \) the dopant densities for p-type and n-type material, respectively, and \( \Delta n \) the injected carrier density. At the same time, there are competing non-radiative processes, such as Shockley-Read-Hall (SRH) recombination.
through deep traps \([40,41]\) with minority carrier lifetime in the low-injection regime given by

\[
\tau_{SHR} = \frac{1}{N_T \nu_{th} \sigma_T} \tag{2}
\]

where \(N_T\) is the density of traps, \(\nu_{th}\) is the thermal velocity of the minority carriers and \(\sigma_T\) is the capture cross-section for the minority carriers.

The optical transitions are substantially two: absorption processes, that happens exciting an electron to a higher energy level by absorbing a photon, and emission processes, electron relaxing to a lower energy state by emitting a photon.

### 3.1.1 Absorption processes

The optical absorption is a process that happens when an electron is transferred from the valence band to the conduction band. The absorption is described by the absorption coefficient \(\alpha\) and depends on the wavelength \(\lambda\), the Figure 3.2 shows a schematic of the optical absorption phenomena.
Figure 3.2 Different types of optical absorption phenomena

- **Bandedge and excitonic absorption**: The energy to create an exciton (electron-hole pair) is smaller than the band gap and can be estimated by use of the Bohr atom model:

\[
E_{ex}^l = \frac{-13.6 \, m^*_r}{l^2} \, \left( \frac{l}{\epsilon_r} \right)^2 (eV) \quad l = 1, 2, 3 \ldots \quad [3]
\]

where \( m^*_r \) is the effective mass of the exciton, \( \epsilon_r \) is the dielectric susceptibility of material. Exciton related absorption is more important in semiconductors at low temperatures but can also be observed above room temperature. There are two types of excitons: Frenkel exciton and Wannier exciton. The length scale of the interaction between the electron and hole distinguishes them. The tightly bound Frenkel exciton is highly localized and the distance between electron and hole is just about a few angstroms (typical unit cell dimensions). The Frenkel exciton has a large binding energy and is typically found in insulators or organic materials. It moves in the material by hopping from one site to another. An electron and a hole that are separated by about 10 to 100 Å form the Wannier exciton. It has a small binding energy and moves freely in the material. The Wannier exciton is often found in inorganic
semiconductors. Absorption of a photon is due to excitation of an electron from the valence band to the conduction band. It is possible to distinguish direct band gap semiconductors from indirect band gap semiconductors by measuring their absorption spectra. From the theory we know that the absorption coefficients varies, for direct band gap, as the square-root of energy above the band edge:

$$\alpha_{\text{direct gap}}(h\nu) = A_{\text{direct gap}} \left[ (h\nu - E_g) \right]^{1/2} u_1 (h\nu - E_g)$$ [4]

and, for the indirect band gap, as the square of energy above onset:

$$\alpha_{\text{indirect gap}}(h\nu) = A_{\text{phonon absorption}} \left[ (h\nu - (E_g - E_{ph})) \right]^2 u_1 (h\nu - E_g + E_{ph}) + A_{\text{phonon absorption}} \left[ (h\nu - (E_g + E_{ph})) \right]^2 u_1 (h\nu - E_g - E_{ph})$$ [5]

- **Free carrier absorption**: This process does not generate electron-hole pairs, because the photon energy is absorbed by free-carriers in either the conduction or valence band, moving the carrier to a higher energy state within that band. Free-carrier absorption is heavily dependent on wavelength, it may still significantly affect the spectral response. This has implications for the device characterization.

- **Interband transition**: an electron in the valence band absorbs a photon and ends up in a conduction band.

- **Impurity absorption**: When both donor and acceptor impurities exist in a semiconductor, an electron bound to a donor can recombine with a hole bound to an acceptor, thus producing donor-acceptor pair
(DAP) emission. After photoexcitation, both donor and acceptor are neutral and the corresponding energy is given by Equation 4 if the weak Van der Waals interaction is ignored:

\[ E_i = E_g - E_D - E_A \]  \[6\]

\( E_D \) and \( E_A \) are the ionization energies for the donor and the acceptor, respectively. When the electron initially bound to the donor recombines with the hole bound to the acceptor, both donor and acceptor become ionized and only the Coulomb interaction between the ionized donor and acceptor exists. Thus, the energy for the final state can be expressed as:

\[ E_f = -\frac{q^2}{\varepsilon \cdot r} \]  \[7\]

\( \varepsilon \) is the dielectric constant of the semiconductor and \( r \) is the distance between donor and acceptor. Hence, the emitted photon energy is the energy difference between the initial and the final states:

\[ E_{DAP} = E_i - E_f = E_g - E_D - E_A + \frac{q^2}{\varepsilon \cdot r} \]  \[8\]

According to Equation 10, the emitted photon energy is dependent on the donor and acceptor separation distance \( r \), which indicates a large set of transitions corresponding to different pair separations. The closer pairs give rise to higher transition energies due to the larger Coulomb interaction energy. The closer pairs will also have shorter radiative
lifetimes because the transition probability decreases with increasing pair separation. The more distant pairs have lower transition probabilities and thus have longer lifetimes, this is very useful to determine whether a transition is DAP or not. Phonons can also participate in DAP recombination, thus the phonon replicas of a DAP transition will be observed on the lower energy side with separation of one or more phonon energies.

3.1.2 Emission processes

Emission is the inverse process of the absorption process, where the energy is removed from an incident electromagnetic wave while an electron-hole pair is created. An electron-hole pair within the medium is destroyed with the emission of electromagnetic radiation. This process involves the recombination of free electrons and free holes [42]. Let us define $\tau_r$ as the radiative recombination time of one electron and one hole. If the free electron and hole concentrations are, respectively, $n_e$ and $n_h$ then the rate of emission of photons by their recombination is given by

\[
\text{rate of emission} = \frac{n_e n_h}{\tau_r} \quad [9]
\]

assuming that $\tau_r$ is the same for all possible choices of recombining pairs. In electroluminescence, where it is possible to inject one extra minority carrier into a semiconductor containing an equilibrium distribution of electrons and holes, it is usual to define minority carrier
radiative lifetime $\tau_{\text{rad}}$ as the time for this extra carrier to be annihilated radiatively by the majority carriers. As a concurrent reaction to radiative recombination also non-radiatively recombination has to be taken into account. The total decay rate of the electron-hole pair population is given by:

$$\frac{1}{\tau_{\text{tot}}} = \frac{1}{\tau_{\text{rad}}} + \frac{1}{\tau_{\text{nonrad}}}$$

[10]

where $\tau_{\text{nonrad}}$ is the non-radiative recombination rate. In non-radiative processes the energy of the electron-hole pair is dissipated as heat via excitation of phonons. The internal quantum efficiency or radiative recombination efficiency is defined as

$$\eta_{\text{rad}} = \frac{R}{R_{\text{rad}} + R_{\text{nonrad}}}$$

[11]

The recombination processes in semiconductors are illustrated in Figure 3.3.

**Figure 3.3** Radiative and non-radiative recombination processes in semiconductors.
3.2 Electroluminescence of silicon solar cells

In the 2004, Bresler at al. made the first studies on the light emission properties of the a-Si/c-Si heterojunction [45]. These devices have efficient emission of silicon band gap light under forward bias and have the potential for high conversion efficiencies. Green et al. showed, that an optimised solar cell structure can work as an LED with an external power conversion efficiency close to $\eta_{\text{ext}} = 1\%$. [45] A solar-cell design optimized for high efficiency (24.7%) was applied to produce a high-efficiency silicon LED (0.85% power-conversion efficiency). [46]

3.2.1 Light emitting of silicon

Light emission in silicon comes from a low probability phonon-mediated transition that unfavourably competes with fast non-radiative recombination paths. The absence of efficient light emission in silicon has impeded, up to now, the integration of electronic and optical devices on mass-produced silicon chips. In this paragraph we will show a review of some significant light emitting devices based on silicon that have been proposed in the literature. Several strategies have been considered and explored. A first approach is based on the introduction of erbium ions in silicon, i.e. light emitting impurities. Another approach is based on low dimensional systems, for example porous silicon, silicon nanocrystals and Si/SiO$_2$ quantum wells. The quantum confinement has several effects, for example increases the probability of radiative recombination and decreases the non-radiative
recombination routes. In details, we discuss the porous silicon and silicon nanocrystals.

3.2.1.1 Silicon Photonics

Silicon Photonics promises to revolutionise the next generation of integrated circuits ICs by providing solutions for optical interconnections between chips and circuit boards, optical signal processing, optical sensing, and the “lab-on-a-chip” biological applications. The current generation of optical devices are designed for high-performance and relative low-volume applications. With silicon photonics higher density integrated optical circuits may be possible in the future.

- **Silicon based waveguides**

Recently, silicon has been identified as a suitable material for the production of high confinement waveguide structures, because it is transparent at the silica optical fiber communication wavelengths of 1.3 μm and 1.55 μm, has a large refractive index, and optical circuitry can be monolithically integrated with electronic circuitry, making it suitable for optoelectronic applications at these wavelengths. There are two means for obtain the optical isolation from the substrate: containment in lowly doped epitaxy on a higher doped substrate or the use of silicon-on-insulator (SOI) technology. The SOI w
There exist different waveguide structures. The simplest is the slab waveguide (Figure 3.4a), where the light may be confined in the middle layer that have a larger reflection index. This structure offers only lateral confinement and cannot confine light horizontally. Another type of structure, offering full confinement, is the strip waveguide (Figure 3.4b). It can be realized by applying photolithography and etching techniques on the slab structure. After etching a top cladding layer is typically deposited to provide a uniform low-index surrounding for the core. This structure can be varied in order to obtain a rib waveguide (Figure 3.4c), in which the high-index core layer is not fully etched. The difference between the effective index of the partially etched slab and
the rib is sufficient to confine the light within the rib section. To obtain a modest but sometimes sufficient optical confinement one can use a strip-loaded waveguide (Figure 3.4d), which is realised with a strip of cladding material on top of a slab waveguide. The refractive index difference is achieved by local doping with an element that changes the refractive index. This results in a diffused waveguide (Figure 3.4e).

- **Silicon based Detectors**

In fiber optical communications, the typical wavelengths are in the near infrared region, in particular at around 850nm, 1300nm and 1550nm. The latter two wavelengths are not covered by silicon photodetectors. Researchers of Intel started to use Germanium absorber regions in silicon integrated circuits. In this way it becomes possible to extend the maximum detectable wavelength of optical detectors on a silicon platform to a 2nd (1300nm) and 3rd (1550nm) window in the optical absorption spectrum of the silica fiber. In the Figure 3.5 a schematic photodetector produced by Intel is shown.

![Figure 3.5 Integrated Ge based photodetector [Intel]](image)
3.2.1.2 Porous Silicon

Uhlir discovered porous silicon in 1956. In 1990, Canham et al. showed that porous silicon materials have a high photoluminescence (PL) efficiency at room temperature in the visible range. An example for the emission spectrum of porous silicon is shown in Figure 3.6.

![Figure 3.6 Photoluminescence emission from porous silicon layer grown from gas phase etchants (100 nm/hour). The PL was excited with the 514 nm line of an Ar+- ion laser [47]](image)

The reason of this was the partial dissolution of silicon, which causes the formation of small silicon nanocrystals in the porous silicon material, the reduction of the effective refractive index of porous silicon with respect to silicon, the spatial confinement of the excited carriers in small silicon regions. These tiny pores allow the penetration of chemical and biological substances, gases, liquids, cells and molecules that changes the optical behaviour of the original system and
open the way to numerous sensor applications. In general, porous silicon is an interconnected network of air holes (pores) in silicon. Porous silicon is classified according to the pore diameter, which can vary from a few nanometers to a few micrometers depending on the formation parameters (see Figure 3.7).

![Figure 3.7 Examples of porous silicon structures: microporous (left), mesoporous (center) and macroporous (right) [48]](image_url)

### 3.2.1.3 Silicon nanocrystals

Increasing the quantum efficiency of crystalline silicon is possible by limiting the movement of an exciton by confining it to a limited space, so reducing the probability to encounter a defect, and consequently, the incidence of the non-radiative phenomena. Confining the excitons in a small region, about few nanometers wide, we may increase the radiative recombination rate and the quantum efficiency. The confinement effect in a region of nanometric size is referred to as quantum confinement effect, and the physical structure in which the quantum confinement of excitons occurs is termed as a nanostructured material. The emission quantum efficiency in silicon nanocrystals remains low compared to the efficiency in direct gap III-V or II-VI semiconductor materials. The size distribution of the nanocrystals can play an essential role in the
excitation exchange between the clusters. There are different methods for preparation of silicon nanocrystals, that has predominantly sizes ranging from 2 – 6 nm: silicon ion implantation, chemical vapour deposition, magnetron sputtering, electron beam evaporation and some others. It is possible to control the mean nanocrystal size via the excess silicon concentration. This involve that the energy of the emitted photons increases as the nanocrystal size decreases. In Figure 3.8 the resulting emission spectra are shown for different materials.

![Graph showing PL spectra of silicon nanocrystals in SiO₂.](image)

**Figure 3.8 PL spectra of silicon nanocrystals in SiO₂.** [48]

The crucial step to fabricate a light source based on silicon nanocrystals embedded in SiO₂ is the difficulty of carrier injection in a semi-insulating material, since the carrier injection determines the excitation of the silicon nanocrystals and therefore the light emission.
3.2.2 Solar cells based LED’s

Silicon has a band gap of 1.12eV at room temperature, which corresponds to a cut-off band-edge response wavelength of 1.11µm, but the phonons extend this weak absorption edge up to about 1.2µm. The mechanisms of light emission from silicon include band gap modifications, quantum confinement, carrier confinement, defect and interface states. When carriers are excited in a semiconductor, they recombine either by radiative or non-radiative processes. Non-radiative processes can occur at defects, in which essentially a ladder of energy levels is provided, enabling a non-detectable series of long wavelength radiative events to occur as the electron loses energy or by Auger processes. Here the electron loses its energy to another conduction band electron, causing it to be excited and then emit this energy as a series of phonons while returning to the bottom of the conduction band. Green et al. [50] have noticed that top-quality solar cells can work as an LED, because they are characterized by extremely long carrier recombination lifetimes of the order of some milliseconds and hence by an external power conversion efficiency close to $\eta_{\text{ext}} = 1\%$. This value was larger by almost two orders of magnitude compared to what had been reached before. That was possible due to the optimized light trapping scheme and the reduced non-radiative recombination in the diode. So it could be demonstrated, that optimized solar cell structures might be useful for the inverse application as LED. Figure 3.9 show a schematical drawing of the device that Green has studied.
To increase the light extraction efficiency, the LED surface was texturized so that most of the internally generated light was impinging on the external surface of the cell with an incident angle lower than the critical angle for total internal refraction. To reduce the free-carrier absorption to a minimum, the electrodes, and hence the heavily doped regions, were confined to very thin and small lines. In Figure 3.10 the Luminescence spectrum and absorption spectrum of the solar cells used as an LED are shown.
We have analyzed the possibilities of using our solar cells as light-emitting diodes compatible and in the Chapter 7 the results are shown.

### 3.2.3 Imaging of defects in solar cells

It is well known that defects in materials create recombination centers, that reduce the local generation/recombination lifetimes of charge carriers, which affects their diffusion lengths. There are many different methods to measure carrier diffusion lengths. A new technique for analyzing the defects in solar cells consists in the analysis of electroluminescence images of the silicon solar cells. Fuyuki et al. [52] showed, that the electroluminescence intensity is proportional to the product of the injected minority carrier density and the effective diffusion length. The electroluminescence imaging apparatus is schematically shown in Figure 3.11.

![Experimental apparatus for EL image](image1)

**Figure 3.11 Experimental apparatus for EL image [52]**

A cooled silicon CCD camera captures the emitted light from a forward biased solar cell. The sensitive wavelength range of detection was 300 - 1100 nm. In the Figure 3.12 a typical emission spectrum of solar cells
taken with the use of a germanium photodetector and the sensitivity of the silicon CCD camera is shown.

![Typical emission spectrum of a silicon solar cell and the spectral characteristics of a silicon CCD camera](image)

**Figure 3.12 Typical emission spectrum of a silicon solar cell and the spectral characteristics of a silicon CCD camera [52]**

The images, as obtained by the secondary electron microscope (SEM), the LBIC technique and the electroluminescence emission intensity mapping are compared in Figure 3.13.

![Images of silicon solar cells with SEM, LBIC and EL mapping technique](image)

**Figure 3.13 Images of silicon solar cells with SEM, LBIC and EL mapping technique [52]**

In the LBIC and in the EL mapping images it is possible to see specific defect features, that could not be observed in the SEM images. The difference between the two techniques, LBIC and EL mapping, is that in the first case there is a focused LASER beam required for scanning
the solar cells while in the second one there is only a CCD camera required. The emission intensity of the EL image has a one to one relation with the minority carrier diffusion length.

### 3.3 Electroluminescence measurements

Electroluminescence, excited by an electrical current through the samples, was used in very early work on luminescence from silicon [53,54], non-equilibrium carriers being injected through a p-n junction. Brown and Hall have shown that the efficiency of electroluminescence can be close to that of photoluminescence when carriers are injected through electrodes consisting simply of silver paint on the surface of the crystal [55]. Many defects in silicon (e.g. the divacancy [56]) are known to produce photoconductivity. A potentially important advance has reported by Kleverman et al. [57] using photothermal ionisation spectroscopy. In this technique (which is well established in studies of shallow donors and acceptors, see, e.g., ref. [57]) absorption of a photon excites a centre to a nearly ionised state. Ionisation is completed by thermal ejection of the particle into a band state. The weak optical transitions into the shallowest states are therefore enhanced, since these are the most easily thermally ionised. The setup that has been used for this measurement is shown in Figure 3.14. The solar cells have been dc forward biased using a Keithley model 2400 source meter in the current source mode. To measure the optical emission spectrum of the solar cells, the emitted light has been chopped at 230Hz and detected after passing a model CM110 monochromator from CVI, by measuring the photocurrent of a
calibrated large area InGaAs photodiode with a Stanford Research model SR380 digital lock-in amplifier.

Figure 3.14 Electroluminescence measurement setup

In the case of the non-wavelength resolved measurements of the optical emitted power, the measurement setup was basically the same. Only the monochromator has been removed. In this case the solar cell and the large area InGaAs detector have been positioned about 15mm apart with the chopper blade inserted in between them.
Chapter 4

High-energy particle radiation and its impact on solar cells

The effects of high-energy particle irradiation on materials and devices, used in space, are among the principal reasons for spacecraft operation failure and lifetime reduction. Main components of the radiation, found in space, are: galactic cosmic rays (GCRs), solar protons and the hot magnetosphere plasma (HP). The energy of electrons and ions of the space radiation is ranging from $10^3$eV up to $10^9$eV and various radiation effects - both on the spacecraft surface and inside the spacecraft - have to be taken into account, because shielding can decrease only the impact of the low energy fraction of the irradiation. It is possible to classify all appearing radiation effects into those related to the total dose and those related to the dose rate. Important components, besides the microelectronic circuitry, that are degraded by this radiation are the solar cells, the cover glass and the thermo-control coatings.
4.1 Radiation in space

The study of radiation damage in semiconductor devices and solar cells has been an important subject of research since the very beginning of modern transistor and solar cell technology. High-energy particle-induced defects, causing a permanent deterioration of the electrical characteristics of the semiconductor material, are one of the primary reasons for the failure of the crystalline silicon space solar cells. The fundamental understanding of the nature and role of different high-energy particle (electrons and protons) induced defects in silicon space solar cells is, therefore, extremely important to the further progress of the silicon space solar cell technology. Main sources of radiation affecting solar cells are protons and electrons trapped by the terrestrial magnetic field and protons coming from the Sun, the particle flux depending on the orbit of the mission.

The most important radiation damage in satellites at low altitude orbits (lower than 800 km) or in the high altitude ones (5000 km or higher) is due to protons (close to 90% of the whole damage). The permanent damage in the solar cell materials is caused by collisions of the incident particles with the atoms in the crystalline lattice, named PKA (Primary Knock-on Atom), which are displaced from their positions. Depending on the energy received, the PKA can generate a displacement cascade. The situation evolves dissipating the initial energy to form a stable structure of defects. These defects degrade the transport properties of the material and particularly the minority carrier lifetime. The damage will be a consequence of the production and/or migration of the defects produced, forming dislocation sets of diverse types. For silicon, the damage structure is extremely dependent on doping level and base
material. Recently, Alurralde [58] has proposed a method to calculate the equivalent dose for orbits in which the main part of the damage is due to protons. In this approach, a Monte Carlo binary collision approximation code, as TRIM [59], is used as a starting point to calculate the PKA (Primary Knock-on Atom) distribution taking into account the irradiation spectra and shielding in a more realistic way. In this paper, Alurralde compares the PKA distribution and the depth dependence of the damage for space proton irradiation and for laboratory irradiation of silicon solar cells. The radiation usually of interest in the study of the degradation of solar energy materials and devices consists of energetic particles (i.e. electrons, protons, neutrons or ions). The origin of these particles may be particle accelerators, nuclear reactors or the natural space radiation environment. Because they have mass, energy and possibly charge, these particles or other particles generated by them can interact in several ways with materials. The dominant interactions are:

• **Inelastic collisions with atomic electrons.** Inelastic collisions with bound atomic electrons are usually the predominant mechanism by which an energetic charged particle loses kinetic energy in an absorber.

• **Elastic collisions with atomic nuclei.** Energetic charged particles may have coulombic interactions with the positive charge of the atomic nucleus through Rutherford scattering. In some cases the amount of energy transferred to the atom will displace it from its original position in a crystalline lattice.

• **Inelastic collisions with atomic nuclei.** This general category of interactions includes several processes, which are important in radiation damage studies. Highly energetic protons undergo inelastic collisions
with the atomic nucleus. In this process, the energetic proton interacts with the nucleus and leaves the nucleus in an excited or activated state. There are currently two main approaches being used to model solar cell degradation in space. The first method was developed at the US Jet Propulsion Laboratory (JPL) more than 20 years ago and has been described in four extensive NASA publications [60,61,62,63]. The result of this approach is the determination of the normal-incidence 1MeV electron fluence, which produces the same level of damage to the cell as a specified space radiation environment. The second method was developed recently at the US Naval Research Laboratory (NRL) and has been described over the last 6 years in a series of publications and conference proceedings [64,65,66,67,68,69,70].

The core of the NRL method is the calculation of the displacement damage dose, using the energy dependence of the damage coefficients and the proton and electron spectra incident on a calibrated solar cell. The NRL approach is generally easier to implement than the JPL method, because it requires far fewer experimental measurements to specify the relative damage coefficients. Hence, it can be used when only limited experimental data are available, as is often the case with new and emerging solar cell technologies. However, many workers in the space community have used the JPL method successfully over a number of years. The relative damage coefficients are calculated from the non-ionizing energy loss (NIEL) for protons and electrons traversing the cell material. NIEL is a calculation of the rate of energy loss due to atomic displacements as an incident particle traverses a material, and the product of the NIEL and the particle fluence gives the displacement damage energy along the path. The NIEL can be calculated analytically from first principles based on differential cross
sections and interaction kinematics. NIEL is that part of the energy introduced via elastic (both Columbic and nuclear) and nuclear inelastic interactions that produces the initial vacancy-interstitial pairs and phonons (e.g. vibrational energy).

NIEL can be calculated for electrons, protons, neutrons and pions, using the following analytic expression that sums the elastic and inelastic contributions:

\[
NIEL = \left( \frac{N}{A} \right) \left[ \sigma_e T_e + \sigma_i T_i \right] \left[ \text{MeVcm}^2 / g \right]
\]

[1]

where \(\sigma_e\) and \(\sigma_i\) are total elastic and inelastic cross sections, respectively, \(T_e\) and \(T_i\) are elastic and inelastic effective average recoil energies corrected for the ionization loss, respectively, \(N\) is Avogadro’s number and \(A\) is the gram atomic weight of the target material. The NIEL has also been calculated by other means, including Monte Carlo programs such as TRIM. Figure 4.1 shows the energy dependence of the non-ionizing energy loss in silicon for protons, neutrons, pions and electrons, normalised to 1MeV-neutron in silicon.
The space environment surrounding the earth and encountered by satellites in orbit, are the primary focus of this section. The sources of particles here are four-fold: (1) protons and electrons residing in the Van Allen belts, (2) heavy ions trapped in the magnetosphere, (3) protons and heavy ions from solar flares and (4) protons and heavy ions from cosmic rays. The first three particle sources are strongly dependent on the solar cycle, while the origin of the most energetic part of the cosmic rays is still not clear. Recently it has been argued that it should be related to black holes within our galaxy. The radiation belts, encountered by satellites in orbit around the earth, are illustrated in Figure 4.2.
The Van Allen belts consist primarily of low energy electrons (a few MeV) and high-energy protons (several hundred MeV) in addition to some heavy ion particles trapped in the earth's magnetic field [73]. Galactic cosmic rays incorporate a range of elements with atomic numbers ranging from hydrogen to uranium, typically characterized via low flux levels and a large distribution of particle energies (a few MeV to several hundred GeV). The fractional composition of particles in these rays has been determined to be 85% protons, 14% alpha particles and 1% heavy ions [74]. The exact environment encountered by a satellite in orbit depends on many factors including orbital path or trajectory, solar activity, and mission length. IEEE Standard 1156.4 is used to classify these orbits as low, medium, geostationary and highly elliptical earth orbits [72]. In the case of low energy orbits (LEO), orbital paths take the satellite through the Van Allen belts several times over a 24-hour period with an altitude between 200 and 600km. In the case of highly elliptical orbits, the altitude is in the range of 30,000km.
thereby resulting in longer exposures to cosmic and solar flare environments [75]. The solar cells exposed to these environments are subject to particles and photons across a wide range of energies and flux. The observed effects on a device and circuit can be coupled to the fundamental interaction mechanisms between the semiconductor material and the incident radiation.

These can be classified as either displacement damage or ionization damage depending on the creation of electron-hole pairs and/or the displacement of atoms from their lattice site. It is important to classify an effect into one of these two categories, if possible, because the general behaviour of each phenomenon has been characterised largely. Figure 4.3 classifies the main effects that are produced by radiation in electronic devices.
4.1.1 Atomic displacement damage

Atomic displacement damage can occur ballistically through kinetic energy transfer, or by the conversion of radiation-induced excitation into atomic motion (i.e., recoil). As a charged particle passes through matter, the particle energy dissipates by exciting orbital electrons and by elastic collisions with the material nuclei. An elastic collision can expel an atom from its normal lattice position. The expelled atom is known as a primary knock-on atom, which, in turn, may cause a cascade of atomic displacements before eventually coming to rest. In Figure 4.4, it is shown, how the displaced atom becomes an interstitial, and the position
that the atom formerly occupied becomes a *vacancy*. Together the interstitial and vacancy are referred to as a *Frenkel pair*. Some displaced atoms can lead to secondary displacements. *Displacement damage* is the result of nuclear interactions, typically scattering, which cause lattice defects. Displacement damage can also be due to cumulative long-term non-ionizing damage from ionizing radiations. In space, the particles producing displacement damage include protons of all energies, electrons with energies above 150keV, and neutrons.

![Image of Frenkel pair generation](image_url)

**Figure 4.4 Mechanism of Frenkel pair generation [77]**

A single incident particle can cause a cascade of collisions to occur to a portion of the affected material lattice atoms. These collisions are produced either by incident “heavy” particles (p, n, ions) or by secondary particles. Defects (vacancies, interstitials, Frenkel pairs, dislocations) are produced along the tracks of the secondary particles and in clusters at the end of these tracks as shown in Figure 4.5.
Figure 4.5 Displacement cascade damage from the movement of a silicon atom after the primary collision. [77]

The production of vacancies and interstitials require a transfer of the particle kinetic energy to potential energy stored in the crystal lattice. Both vacancies and interstitials, especially the latter, are mobile at sufficiently high temperature and annealing facilitates their recombination. At higher temperatures, the vibrations of the atoms in the lattice increase, thereby providing a mechanism by which an interstitial atom can migrate to a vacancy; and hence, fix both defects.

Two convenient parameters have been introduced to describe the effects caused by a given particle at a given energy. These two parameters are the damage coefficient $K$ and the critical fluence $\phi$. The damage coefficient describes the rate of change of the lifetime with respect to the particle fluence:

$$\frac{1}{\tau} = \left(\frac{1}{\tau_0}\right) + K_\tau \phi$$

[2]
where $K_\tau$ is given by:

$$K_\tau = \sigma v_{th} F_e P_R$$

[3]

and $\tau_0$ is the initial lifetime, $\phi$ is the incident fluence (the total number of particles/cm$^2$), $\sigma$ is the capture cross section, $v_{th}$ the thermal velocity, $F_e$ the Fermi probability that the generated recombination center is occupied by a majority charge carrier, and $P_R$ is the number of centres per centimetre produced by each particle.

It has become more customary to describe the damage coefficient in terms of the minority carrier diffusion length. The relationship 2 then becomes:

$$\frac{1}{L^2} = \left(\frac{1}{L_0^2}\right) + K_L \phi$$

[4]

and the rate of change equation is:

$$\frac{d\left(\frac{1}{L^2}\right)}{d\phi} = K_L$$

[5]

The degradation of solar cells is generally described in terms of the variation of the short-circuit current ($I_{sc}$) and the open-circuit voltage ($V_{oc}$) versus the irradiation fluence ($\phi$).
• **Short circuit current**

The irradiation introduces non-radiative recombination centers, which decrease the minority carrier lifetime affecting only the diffusion current. The resulting rate of Shockley–Read recombination is very small and the generation current can be considered as constant until the width of the space charge region $W$ varies as a result of the compensation of the free carriers. Therefore, the degradation of $J_{SC}$ is that of the diffusion current density $J_d$, than can be written as:

$$
J_d = q \int_{\lambda_1}^{\lambda_2} \phi(\lambda) \alpha(\lambda) L \exp \left[ -\alpha(\lambda)(d + W) \right] d\lambda
$$

where $\phi(\lambda)$ is the illumination flux at the wavelength $\lambda$, extending from $\lambda_1$ to $\lambda_2$, and $\alpha(\lambda)$ is the absorption coefficient of the material at $\lambda$ ($\lambda_2$ corresponds to the limit of absorption, i.e. to the gap of the material, while $\lambda_1$ is the lowest wavelength for which $\phi(\lambda)$ is non-negligible), $d$ is the width of the emitter than is compared to the diffusion length $L$ of the minority carriers. The diffusion length $L$ is defined as $L = \sqrt{D \tau}$ where $D$ is the diffusion coefficient of the minority carriers and $\tau = (N \sigma v)^{-1}$ the minority carrier lifetime, $N$ the concentration of the recombination centers, $\sigma$ their capture cross section for minority carriers and $v$ the thermal velocity. The recombination centers being introduced by the irradiation with a rate $k$ can be written as $N = k \phi$. As a result we get that $J_d$ decrease linearly with $\ln \phi$:
\[ J_d = \frac{qA}{r} \left[ \ln \alpha(\lambda_1) - \ln \left( \frac{k\sigma_v}{D} \right)^{1/2} - \frac{1}{2} \ln \phi \right] \]  

[7]

where \( A = \phi(\lambda) \exp[-\alpha(\lambda)(d+W)] \). We can rewrite the expression for \( J_d \) as:

\[ J_d = \xi - \rho \log \phi \]  

[8]

the slope \( \rho \) of the plot \( J_d \) versus \( \log(\phi) \) is \( \rho = 2.3 \left( \frac{qA}{2r} \right) \) independent of the number of recombination centers generated by the radiation. It depends only on the illumination conditions, through \( \Lambda \), and on the material, through the constant \( r \). Therefore we can write:

\[ \xi = \frac{qA}{r} \left[ \ln \alpha(\lambda_1) - \frac{1}{2} \ln \left( \frac{k\sigma_v}{D} \right) \right] \]  

[9]

This result demonstrates that solar cells made of the same material exhibit identical slopes for the degradation, even if the recombination centers introduced by the irradiation are different. The nature and concentration of these centers are given by the value \( \xi \) of \( J_d \) extrapolated to the origin.

Degradation of \( J_d \) occurs for fluence larger than a minimum fluence \( \phi_m \) given by
\[
\log \phi_m = \frac{\xi - J_{SC}(0)}{\rho} \tag{10}
\]
is therefore related to the irradiation-induced recombination centers, through \(\xi\), and to the native recombination centers, through \(J_{SC}(0)\).

- **Open-circuit voltage**

The forward current density in the dark of a junction is the sum of diffusion current and a recombination current density

\[
J(V) = J_1 \left[ \exp \left( \frac{qV}{kT} \right) - 1 \right] + J_2 \left[ \exp \left( \frac{qV}{2kT} \right) - 1 \right] \tag{11}
\]

where \(J_1\) and \(J_2\) are given by the following expressions [48]:

\[
J_1 = \frac{qn_i^2 \sqrt{D_n}}{N_A \sqrt{\tau_n}} + \frac{qn_i^2 \sqrt{D_h}}{N_D \sqrt{\tau_h}}
\]

\[
J_2 = \frac{\pi \ n_i kT \ W_0}{2 \sqrt{\tau_n \tau_h} \sqrt{V_d}} \tag{12}
\]

In these expressions \(n_i\) is the intrinsic carrier concentration, \(D_n\), and \(D_h\) are the electron and hole diffusion coefficients, \(\tau_n\), and \(\tau_h\) are the minority carrier lifetimes in the p and in the n regions, \(N_A\) and \(N_D\) are the acceptor and the donor concentrations on each side of the junction, \(V_d\) is the built-in voltage and \(W_0\) is the width of the space charge region.
with no applied bias. Introducing the rate $k$ we can rewrite the two expressions as [78]:

$$J_1 = J_1^* \phi^{\frac{1}{2}} \quad \text{and} \quad J_2 = J_2^* \phi$$

$$J_1^* = q n_1^2 \left[ \left( \frac{k \sigma_n \nu_n D_n}{N_A} \right)^{\frac{1}{2}} + \left( \frac{k \sigma_h \nu_h D_h}{N_D} \right)^{\frac{1}{2}} \right]$$

$$J_2^* = \frac{\pi}{2} n_i k T W_0 \left( \frac{k^2 \sigma_n \nu_n \sigma_h \nu_h}{V_d} \right)^{\frac{1}{2}}$$

Hence, we can deduce that in this case the dark current is essentially given by the diffusion term [78]:

$$J_{SC} = J_1^* \phi^{\frac{1}{2}} \exp \left( \frac{q V_{OC}}{2kT} \right)$$

In the case that recombination is dominating, we have:

$$J_{SC} = J_2^* \phi \exp \left( \frac{q V_{OC}}{2kT} \right)$$

Thus, in both cases, $V_{OC}$ can be written under the form

$$V_{OC} = \epsilon - \eta \log \phi$$

where the slope $\eta$ of the plot $V_{OC}(\log \phi)$ is [78]:
\[ \eta = \frac{2.3 \, KT}{2 \, q} \quad \text{for the diffusion region} \quad [16] \]
\[ \eta = 2.3 \frac{2KT}{q} \quad \text{for the recombination region} \]

The value of \( V_{OC} \) at the origin is:
\[ \varepsilon = \frac{KT}{q} \ln \left( \frac{J_{sc}}{J_1^*} \right) \quad \text{for the diffusion region} \quad [17] \]
\[ \varepsilon = \frac{2KT}{q} \ln \left( \frac{J_{sc}}{J_2^*} \right) \quad \text{for the recombination region} \]

thus provides \( J_1^* \) and \( J_2^* \), from which information on the recombination centre can be extracted using expression 13.

### 4.1.2 Ionization damage

*Ionization* is the process of removing or adding an electron to a neutral atom, thereby creating an ion. The term is also often used in connection with the removal of an electron from a partially ionized atom. Ionizing radiation tends to be increasingly damaging in the following order of molecular formation (largely due to the ability of ionization to disrupt the bonds):

1. **Metallic bonds**

Metallic bonding consists of positive ions with free valence electrons, which hold the ions together. Ionizing radiation increases the kinetic
energy of the electrons or excites the electron to a higher energy level, but they soon return to their normal energy level. In either case, there is no permanent damage from ionization, only temporary internal heat production.

2. Ionic bonds

Ionic bonds are less strong than metallic bonds regarding the impact of ionizing radiation. In general, the electrostatic attractive and repulsive forces between the ions lead to well-ordered, three-dimensional arrangements of the ions in crystals. Radiation causes only temporary ionization of the lattice atoms, which soon become neutral. Colour changes (e.g., in glass) may occur due to free electrons being trapped at lattice imperfections.

3. Covalent bonds

For a covalent bond, the outer electrons in a molecule are no longer uniquely associated with a particular atom, but instead are shared between all the atoms in the molecule. Covalent bonds are typical of gases, liquids and organic materials. Covalent bond energies are in the lower eV range. Radiation of sufficient energy to break the covalent bond can permanently separate the molecules into its constituent atoms or radicals. Therefore, the chemical composition of the material is fundamentally changed.

4.2 Radiation damage in silicon

Vacancies and interstitials in silicon are even at low temperatures highly mobile. In addition, depending on the position of the Fermi level, they may change their charge states. During the irradiation most of the
energy is transferred by ionisations, which can modify the charge state and the diffusion and reaction properties of the point defects. The degradation of silicon by high-energy particle irradiation and the creation of displacement damage is generally studied by a combination of macroscopic and microscopic techniques. An example for a microscopic technique is the deep level transient spectroscopy (DLTS), a transient capacitance technique, used to identify energy trap level energies [79]. DLTS has been successfully employed to identify the dominant radiation defect traps in silicon that are stable at 300K [80]. Experimental studies on both p and n-type proton irradiated silicon have identified the energy trap levels $E_T = 0.421eV$ and $E_T = 0.164eV$ as the dominant recombination centers for the low- and high-injection regimes of operation based on calculations of the respective recombination lifetimes [80,81,82]. Additionally, Electron Paramagnetic Resonance (EPR) and Localized Vibrational Mode (LVM) spectroscopy have been successfully employed to reveal the structures of intrinsic defects in silicon [83]. Ions with energy higher than a few MeV/amu dissipate their energy principally via electronic excitation and not via nuclear displacements (defect formation). We calculated the energy dependences of the electronic and of the nuclear stopping power, $S_e$ and $S_n$, the calculated results for protons in silicon are shown in Figure 4.6. In the case of protons in the whole energy range we observe a stronger electronic excitation, resulting in higher values of $S_e$ is much larger as compared to $S_n$. In contrast, in the case of heavier ions, $S_e$ is smaller than $S_n$ in the lower energy range. From TRIM calculations results that 99.96% of the incident energy dissipates via electronic excitation, and less than 0.01% is used for
defect formation including recoil events associated with cascade collisions.

![Figure 4.6 Calculated energy dependences of the electronic and the nuclear stopping power, $S_e$ and $S_n$, of proton in silicon](image)

However, the larger contribution of electronic excitation for protons does not imply that effects of nuclear displacements are completely negligible. In the energy range of 0 - 10 MeV, in which we are mainly interested, the energy loss per unit length, i.e. the stopping power, can be divided into nuclear stopping power and electronic stopping power [84]. The nuclear stopping governs the energy losses caused by elastic collisions between the ion and the nuclei of atoms in the target. The electronic stopping term governs the energy losses caused by the electronic interactions, which can be further divided into several different contributions depending on the nature of the interaction. Hence, the stopping power (or slowing down force) can be written as
\[
    S = \left( \frac{dE}{dx} \right) = \left( \frac{dE}{dx} \right)_{\text{nuclear}} + \left( \frac{dE}{dx} \right)_{\text{electronic}} \quad [18]
\]

For high energies (>10 MeV/amu), contributions coming from nuclear reactions and relativistic corrections have to be taken into account.

### 4.2.1 The impact of high-energy particles irradiation on c-Si based materials

The interaction of high-energy particles with crystalline semiconductor devices usually produces three types of effects:

- Displacement of atoms in the crystal lattice, removing atoms from their nominal positions: bulk damage.
- Creation of electron-hole pairs, used to detect the passage of high-energy particles.
- Creation of local charged regions due to long-term ionization effects. This is a surface damage.

Among them, the main damage contribution is caused in the silicon bulk and is produced by displacements of atoms.

The interaction of incident particles with the silicon can layer lead to:

- An increase of the diode reverse bias current, which is proportional to the equivalent 1 MeV neutron fluence.
- Inversion of the conduction type (i.e. a starting n-type material becomes p-type). This inversion produces a dramatic change of the depletion voltage (needed to maintain a full sensitivity over the whole detector thickness.
Creation of defects acting as trapping centers. This affects the charge collection efficiency of solar cells and photodiodes.

In earlier experiments, reported in the literature on crystalline silicon, proton energies and corresponding fluences of a large variety have been applied. For example, we find experiments with a proton energy of 1.3 MeV with fluences up to $5 \times 10^9 \text{ cm}^{-2}$ [85], experiments with $10^{-100}$ MeV proton energies with fluences up to $2 \times 10^{11} \text{ cm}^{-2}$ [86] and some experiments at a proton energy of 660 MeV [87]. There are also reports with energies in the GeV range [88], having a high-energy physics detector background. With emphasis on the change in crystalline silicon solar cell properties, for example, Yamaguchi et al. [89] and Morita et al. [90] report changes in the basic solar cell parameters with protons of energies smaller than 10 MeV, while Kishimoto et al. [91, 92] studied the properties of undoped and n-type as well as p-type crystalline silicon upon 17 MeV proton irradiation.

### 4.2.2 The impact of high-energy particles irradiation on a-Si:H based materials

In 1977 Staebler and Wronski found that the electrical and optical properties of hydrogenated amorphous silicon (a-Si:H) were affected by photon irradiation. They observed the creation of metastable defects, which was later ascribed to dangling bond generation due to breaking of weak silicon-silicon bonds [93].

Hydrogenated amorphous silicon solar cells are interesting for space applications, mainly because of their high radiation tolerance, favourable annealing properties [94], and the possibility to deposit them
on flexible and light weight substrates. In order to predict the end-of-life efficiency of an a-Si:H solar cell in a space environment, quantitative information on the defect generation in a-Si:H material due to proton irradiation is needed. Various measurement techniques have been used to study the degradation of a-Si:H due to electron irradiation, in particular the constant photocurrent method [95], electron-spin resonance, and electroluminescence.

In case of proton irradiation, a-Si:H tolerates fluences 10 times larger than Gallium Arsenide (GaAs) or more than 100 times larger than crystalline silicon for the same relative degradation of solar cells (at any given proton energy).

The radiation hardness of a-Si:H solar cells is almost comparable to that of solar cells based on Copper Indium Gallium Diselenide (CIGS) [96] or Cadmium Telluride (CdTe) [97]. Furthermore, defects created by proton irradiation can be partially removed by annealing even at low temperatures around 100°C [98]. Due to the relatively high operation temperature of the cell in space, an early saturation of the degradation of a-Si:H solar cells can therefore be expected.

### 4.2.3 The impact of high-energy particles irradiation on µc-Si based materials

The effect of irradiation on microcrystalline silicon (µc-Si) by photons or energetic particles is of fundamental interest because any increase in defect density allows studying influence of defects on transport and recombination in this semiconductor.
In contrast to amorphous silicon, microcrystalline silicon exhibits an insignificant degradation of the majority and minority carrier properties under light-soaking, but it has still a higher absorption coefficient than crystalline silicon in the infrared spectral region. Similar to crystalline silicon, however, degradation of the optical and electronic properties of microcrystalline silicon after irradiation with 1 MeV protons [99] and in another study after irradiation of either 1.7 MeV or 24 GeV protons has been reported [100].

The photoconductivity decrease after proton irradiation is due to the irradiation-induced increase in number of recombination states in the band gap, as evidenced for example by modulated photocurrent (MPC) measurements.

The MPC techniques determines the energetic distribution of states in the band gap of a photoconductor from an analysis of the phase shift between an ac photo-excitation and the ensuing ac photocurrent in function of the modulation frequency of the light [101]. Other techniques that indicate the creation of additional recombination centers upon proton irradiation are: the steady-state photoconductivity (SSPC) technique, that offers the possibility to determining the absorption coefficient in function of the energy of the incoming photons, and thus explore the electronic density of states around the band gap of a semiconductor [102].

In order to identify the silicon-dangling bond as the defect that gets created upon proton irradiation it is possible apply techniques like electron spin resonance (ESR). Also electrically-detected ESR has been useful for studying defect physics after proton irradiation [103,104].
4.2.4 The impact of high-energy particles irradiation on silicon based solar cell

Like all semiconductor, when exposed to particle irradiation, the solar cells are subject to electrical degradation. When protons particles incident upon the solar cell material, there are atomic collision with the atoms of the crystal lattice of the solar cells. In these atomic collisions, energy is transferred from the incident particle to the target atom in two ways. The primary energy transfer mechanism is through ionization of the target atom, while the second energy transfer mechanism is through non-ionizing events. It is the displacement damage induced by the non-ionizing interactions that causes the electrical degradation of the solar cells. Solar cell operation depends on the motion of photogenerated charge carriers through the material, but the generation of defects sites degrades the solar cell performance. These defects sites are created when an atom is displaced in a lattice, because the electron energy band is disturbed and are created localized energy levels, that can act to trap electrical charge carriers restricting their ability to move through the material.

The increment of these defect decreases the diffusion length of minority carriers in the substrate of the solar cell, and the electrical performance of the cell degrades logarithmically with increasing proton fluence. High energy proton irradiation causes atomic displacement in the solar cell materials. Lattice defects such as vacancies, interstitials and complex defects are a result of this atomic displacement. In particular recombination centres or majority and minority carrier trapping centres produce a decrease of the output power of solar cells,
because the internal resistances of the solar cells dissipate the power generated.

### 4.3 Calculation of the relative damage coefficients for solar cells

The Monte Carlo code TRIM (stopping and range of ions in matter), a Monte Carlo code developed in the 1980s that simulates the transport of energetic ions in materials [59], is also used to obtain information about the vacancy production rates. It can also be employed to calculate the NIEL. The TRIM output gives the vacancy production rate as a function of position as the incident proton slows down in the target material. Combining these data with the total energy loss data, the vacancy production rate as a function of proton energy can be found.

The vacancy concentration can be converted to damage energy using the modified Kinchin-Pease approximation. The vacancy concentration, calculated with the TRIM code, as a function of the depth for an incident proton on an amorphous silicon/crystalline silicon heterojunction, as can be seen in Figure 4.7.
Figure 4.7 TRIM-derived vacancy concentration as a function of the depth for an incident proton in silicon heterojunction

At proton energies $> 4\text{MeV}$ for silicon, TRIM results show, that for normal incidence the proton energy is essentially constant throughout the solar cell active region.

The width of the degraded region in the cells, investigated in this work, had been estimated to cover the range between 0.8 and 120$\mu\text{m}$ below the front surface.

The Table 4.1 reports the projected ranges of ions in matter over a wide band of ion energies.

These tables are useful to set up the full Monte Carlo TRIM program, so that the target thickness is adequate to contain all the ions. The ranges of ions are calculated using the transport equation approach developed by J. P. Biersack, which he called PRAL (Projection Range ALgorithm) [106,84].
<table>
<thead>
<tr>
<th>Ion Energy</th>
<th>dE/dx Elec.</th>
<th>dE/dx Nuclear</th>
<th>Projected Range</th>
<th>Longitudinal Straggling</th>
<th>Lateral Straggling</th>
</tr>
</thead>
<tbody>
<tr>
<td>800keV</td>
<td>1.99E-01</td>
<td>1.60E-04</td>
<td>11.74um</td>
<td>5719A</td>
<td>7542A</td>
</tr>
<tr>
<td>1.7MeV</td>
<td>1.25E-01</td>
<td>8.43E-05</td>
<td>36.85um</td>
<td>1.55um</td>
<td>2.03um</td>
</tr>
<tr>
<td>4MeV</td>
<td>6.91E-02</td>
<td>4.01E-05</td>
<td>148.36um</td>
<td>6.11um</td>
<td>7.24um</td>
</tr>
<tr>
<td>65MeV</td>
<td>8.09E-03</td>
<td>3.32E-06</td>
<td>19.42mm</td>
<td>819.57um</td>
<td>767.89um</td>
</tr>
</tbody>
</table>

**Table 4.1 Ranges of ions in matter over a wide band of ion energies**

This method allows the rapid calculation of ion ranges over a large band of ion energies. The ranges are quite accurate, usually within 5% of the ranges, calculated with TRIM (which is the most accurate method of calculating ranges).

### 4.4 Accelerator sites used for the proton irradiation

The proton radiation has been performed in two different accelerators: For the low energy (0.8MeV, 1.0MeV, 1.7MeV and 4.0MeV) protons we used the TIT-3 tandem accelerator at the physics department of Naples Federico II University, while for the high energy (65MeV) we used the accelerator at the Ion Beam Laboratory (ISL) of the Hahn-Meitner-Institut in Berlin. The homogeneity of the exposure has been checked by lifetime mapping of a silicon wafer using the scanning transient microwave reflection (TRMC) technique [107]. The resulting implantation depth profile and the expected vacancy distribution have been simulated by TRIM code Monte Carlo calculations [108].
Figure 4.8 the calculated projected range of the protons is shown as a function of the proton energy.

![Figure 4.8 Projected range of the protons as a function of the proton energy (SRIM Simulation)](image)

At some proton energies (1.7MeV and 65MeV) we irradiated the samples with different proton doses in order to calculate the damage constant.

### 4.4.1 TTT-3 Tandem accelerator (INFN)

A particle accelerator is a device that uses electric fields to propel electrically charged particles to high speeds. In a tandem accelerator, the negatively charged ion gains energy by attraction to the very high positive voltage at the geometric centre of the pressure vessel. When it arrives at the centre region, called the high voltage terminal, some
electrons are stripped from the ion, which becomes positive. The ion is then accelerated away by the high positive voltage. Thus this type of accelerator is called a tandem accelerator. It has two stages of acceleration, first pulling and then pushing the charged particles. For the presented irradiations, we used typically beam currents below 10nA in order to avoid excessive sample heating during irradiation. The beam was focused to a diameter of about 1mm and then scanned over the irradiated area (typically about 1cm²) using incommensurable x- and y-deflection frequencies in order to avoid inhomogeneous irradiation by Lissajous figure formation. The irradiations at medium energies (below 10MeV) have been all performed with the irradiated solar cells under high vacuum. The schematically setup of this accelerator is shown in Figure 4.9

**Proton irradiation at TANDEM TTT-3 accelerator (0.5-4MeV)**  
(Physics Dep., Univ. Federico II, Naples, Italy)

Figure 4.9 Tandem TTT-3 accelerator at the physics department of Naples Federico II University
CHAPTER 4 – HIGH ENERGY PARTICLE RADIATION AND IMPACT ON SOLAR CELLS

4.4.2 ISL Accelerator (HMI)

The irradiation facility at the ISL offers various types of ions accelerated in various accelerator combinations with energies covering several orders of magnitude: from a few eV up to over 100 MeV. Figure 4.10 shows a schematic picture of the ISL accelerator.

![Figure 4.10 High energy particle accelerator at the ISL (Hahn-Meitner-Institut)](image)

The irradiations of all samples at the ISL, as presented here, have been performed in air, which was feasible at the used energy of 65 MeV. For the dose calibration the beam has been checked before sample irradiation with a commercial dosimeter.
Chapter 5

Results and discussion

For all solar cells the current-voltage characteristics with and without illumination under AM1.5 conditions and the quantum yield characteristics have been measured. From these measurements the fundamental parameters of a solar cell have been extracted. Three different types of solar cells: two types of an a-Si:H/c-Si heterojunction, with and without the insertion of a thin intrinsic a-Si:H buffer layer, and for comparison a conventional p-n homojunction crystalline silicon solar cell, have been irradiated. Additionally the proton irradiation damage has been simulated, using the TRIM code, and the fundamental characteristics of the heterojunction solar cells before and after irradiation has been simulated, using the AFORS-HET simulation program.
5.1 Monte Carlo simulation of the proton impact on silicon based solar cells

In order to evaluate the effect of the proton irradiation induced damage of the solar cells, a simplified inhomogeneous proton induced damage profile, similar to the calculated damage profile using the TRIM code [109], has been assumed. In Figure 5.1 the depth distribution of the produced vacancy density in silicon after irradiation with protons at various energies are shown.

![Vacancy concentration depth profiles in silicon after irradiation with protons at different energies](image)

**Figure 5.1** Vacancy concentration depth profiles in silicon after irradiation with protons at different energies

In order to understand the effect of the proton damage on the solar cell performance, a numerical computer simulation program for amorphous silicon/crystalline heterojunctions (AFORS-HET) has been used, that has been developed at the Hahn-Meitner-Institut in Berlin [110]. The simulation results for a proton energy of 1.7MeV are reported in the
following Figure 5.2. The damage profile has been considered constant in the amorphous silicon layer, while different trap densities \((N_{tr})\) after irradiation have been assumed in three different regions of the crystalline silicon wafer. The region A is the region close to the heterointerface, followed by a thin region C at a depth of about the projected range of the implanted protons with an elevated trap density and finally a region B beyond the projected range of the protons with a rather low value of electronic traps. In Figure 5.2 this distribution, as used as input for the AFORS-HET simulation, is shown and compared to the vacancy concentration, as calculated by TRIM.

![Figure 5.2](image)

**Figure 5.2** The calculated trap depth distribution, used for the AFORS-HET simulation (straight lines), compared to the vacancy depth distribution, as simulated by the TRIM code (fillers squares)

Region B, which is not affected by the irradiation, bears the native defect density, which is determined from the simulated sample without
proton irradiation. The density of region B has little impact on the simulation results. For simplicity, for the trap density in region C a value 10 times higher than that of region A has been assumed. The trap densities of the three regions, used for the AFORS-HET simulation, are given in Table 5.1.

<table>
<thead>
<tr>
<th>Dose [p/cm²]</th>
<th>Region A [cm⁻³]</th>
<th>Region C [cm⁻³]</th>
<th>Region B [cm⁻³]</th>
</tr>
</thead>
<tbody>
<tr>
<td>0</td>
<td>2.6·10¹²</td>
<td>2.6·10¹²</td>
<td>2.6·10¹²</td>
</tr>
<tr>
<td>6.5·10¹⁰</td>
<td>1.1·10¹⁴</td>
<td>1.1·10¹⁵</td>
<td>2.6·10¹²</td>
</tr>
<tr>
<td>2.0·10¹¹</td>
<td>1.8·10¹⁴</td>
<td>1.8·10¹⁵</td>
<td>2.6·10¹²</td>
</tr>
<tr>
<td>1.0·10¹²</td>
<td>1.6·10¹⁵</td>
<td>1.6·10¹⁶</td>
<td>2.6·10¹²</td>
</tr>
<tr>
<td>5.0·10¹²</td>
<td>6.4·10¹⁵</td>
<td>6.4·10¹⁶</td>
<td>2.6·10¹²</td>
</tr>
</tbody>
</table>

Table 5.1 Substrate trap density of the non-irradiate samples and trap density of region A, B and C of the samples irradiated with different doses at 1.7MeV

The effect of the region C can be seen also in the band diagram, realized with AFORS-HET program, where the valence band and conduction band are not flat in the defect region. In the Figure 5.3 the band diagram, before and after irradiation with 2.0·10¹¹ protons/cm² at 1.7MeV energy, for an a-Si:H/c-Si heterojunction solar cells without i-layer are compared. In the inset of band diagrams for sample irradiated (right plot) is possible to see a zoom of interface between defect layer and peak layer region.
Figure 5.3 Comparison of Band diagrams, as calculated with the help of the AFORS-HET program, of the a-Si:H/c-Si heterojunction solar cell, without i-layer, before (left plot) and after irradiation (right plot) with $2.0 \times 10^{11}$ protons/cm$^2$ at 1.7MeV energy. The two vertical lines indicate the limits of region C. The inset on the right side shows region C more in detail.

For the whole energy range of interest in Figure 5.4 the value of the projected range, the defect density in the peak region and the defect density in the base region, as calculated by TRIM code, have been plotted.
It can be observed, that there is a constant factor of about 40 between the values of the peak and in the base region defect density. For the experiment only five proton energies (0.8MeV, 1MeV, 1.7MeV, 4MeV and 65MeV) and at two energies four different doses (between $5 \times 10^{10}$ and $5 \times 10^{12}$ protons/cm$^2$) have been chosen. The 65MeV proton irradiation, resulting for all samples in a constant damage profile, has been done at the ISL facility of the Hahn-Meitner-Institut in Berlin (Germany), while all other irradiations have been done at the TANDEM facility of the Physics Department of Naples University Federico II (Italy). At an energy of 65MeV only two type of solar cells have been irradiated: the c-Si homojunction (SCH1) and the a-Si:H/c-Si heterojunction (1461) with intrinsic buffer layer. Based on this trap profile due to proton irradiation, the quantum-yield and the current-voltage characteristics of the heterojunction solar cells have been
simulated for all applied proton doses and energies. In the simulation program local electrical shunts or other inhomogeneities are not taken into account. It should be mentioned, that the future version of the simulation program is also including this type of defects. The actual version, however, permits to add external parasitic elements, as the series resistance and the parallel resistance. Other important factors that are not considered in the simulation program are the contact resistance and the grid metallization, that come into play at the high voltages, indeed we can see, in the Figure 5.5, that we get a better fit for the I-V characteristics under illumination for a non irradiated a-Si:H/c-Si heterojunction solar cell without intrinsic buffer layer, when adding an additional external series resistance of 3.5Ω/cm².

Figure 5.5 Current-voltage characteristics for an heterojunction solar cell without i-layer (sample 1458), measured and simulated with and without additional series resistance.
The measured and simulated device parameters are compared in the Table 5.3. A very good agreement has been obtained for the open circuit voltage and the short circuit current, but still considerable differences remain between the measured and simulated data regarding the fill factor and the efficiency. This difference is principally due to the grid metallization, which has not been considered in the version of the AFORS-HET program, used in this thesis.

<table>
<thead>
<tr>
<th></th>
<th>Voc [mV]</th>
<th>Jsc [mA/cm²]</th>
<th>FF%</th>
<th>η%</th>
</tr>
</thead>
<tbody>
<tr>
<td>Measured</td>
<td>608,9</td>
<td>30,80</td>
<td>69,46</td>
<td>12,40</td>
</tr>
<tr>
<td>Simulated with Rseries</td>
<td>607,82</td>
<td>30,69</td>
<td>73,92</td>
<td>13,87</td>
</tr>
<tr>
<td>Simulated without Rseries</td>
<td>607,82</td>
<td>30,69</td>
<td>82,92</td>
<td>15,56</td>
</tr>
</tbody>
</table>

Table 5.2 Characteristics parameters measure and simulated of an a-Si/c-Si heterojunction without i-layer

The simulation and measured results for an heterojunction without i-layer (sample 1458), before and after irradiation with $2.0 \cdot 10^{11}$ protons/cm² at 1.7MeV energy, are shown in Figure 5.6. In Figure 5.6a the current-voltage characteristics are compared and it is possible to see that there are of the difference between the data simulated and the data measured, while in the Figure 5.6b are compared the Quantum yield simulated and measured.
Figure 5.6 Current-voltage and quantum yield characteristics, for an heterojunction without i-layer (sample 1458), before and after irradiation with $2.0 \cdot 10^{11}$ protons/cm$^2$ at an energy of 1.7MeV

Apart from a minor displacement of about 20 nm, the simulated and experimental curves are almost identical. In order to properly adjust this shift, one should adapt the recombination velocity at the rear contact and optimize the discretization of the mesh, used in the region of the c-Si absorber.
5.2 Influence of the proton energy on the solar cell degradation

The influence of the proton energy on the cell performance has been exclusively investigated on a-Si:H/c-Si heterojunctions without intrinsic buffer layer. In this case devices has been used with the same technology as the samples type P1458, but with a small variation: the substrate temperature during deposition of the amorphous silicon emitter layer was 150°C instead of 230°C for the sample type P1458. The resulting properties of the non-irradiated samples were comparable, but with slightly lower $I_{sc}$ and FF values. In Figure 5.7 the current-voltage characteristics under AM1.5 irradiation is shown for the non irradiated heterojunction and four other cells irradiated respectively with 0.8MeV, 1.0MeV 1.7MeV and 4MeV, all with a dose of $2 \times 10^{11}$ protons/cm$^2$. 
Figure 5.7 Current-voltage characteristics (AM1.5) of an a-Si/c-Si heterojunction irradiated with $2 \cdot 10^{11}$ proton/cm$^2$ and different irradiation energy

It can be seen that the maximum degradation occurs after irradiation with an energy of 1.7MeV.

Furthermore we can see that proton irradiation results in a reduction of the open circuit voltage and a strong reduction of the short circuit current. In the Table 5.3 the characteristic parameters before and after irradiation are reported with the values of maximum damage evidenced in bold.

<table>
<thead>
<tr>
<th>Proton energy</th>
<th>Voc [V]</th>
<th>Isc [A]</th>
<th>FF</th>
<th>$\eta$</th>
</tr>
</thead>
<tbody>
<tr>
<td>not irradiated</td>
<td>0.61</td>
<td>-0.030789</td>
<td>69.46%</td>
<td>12.4%</td>
</tr>
<tr>
<td>0.8MeV</td>
<td>0.57</td>
<td>-0.027538</td>
<td>67.53%</td>
<td>10.6%</td>
</tr>
<tr>
<td>1.0MeV</td>
<td>0.56</td>
<td>-0.023873</td>
<td>59.30%</td>
<td>7.93%</td>
</tr>
<tr>
<td>1.7MeV</td>
<td>0.55</td>
<td>-0.022360</td>
<td>63.28%</td>
<td>7.78%</td>
</tr>
<tr>
<td>4.0MeV</td>
<td>0.58</td>
<td>-0.026140</td>
<td>64.96%</td>
<td>9.85%</td>
</tr>
</tbody>
</table>

Table 5.3 Characteristics parameters before and after irradiation of an a-Si/c-Si heterojunction

The corresponding quantum yield spectra are shown in Figure 5.8. It can be seen, that in the whole wavelength range the highest quantum yield values belong to the non-irradiated sample and the lowest values to the sample that has been irradiated with 1.7MeV protons.
Besides the classical measurement techniques like quantum yield and current-voltage characteristics under illumination, additionally also the electroluminescence of the forward biased solar cell has been measured. In the Figure 5.9 the emitted optical power as a function of the injected forward bias current of the heterojunction solar cells is shown. Also in this case, it can be see that the maximum degradation happens for an energy equal to 1.7MeV. Indeed, the emitted optical decreases monotonically with increasing proton energy up to a value of 1.7MeV and then increase again.
Figure 5.9 Emitted optical power of an a-Si/c-Si heterojunction
irradiated with $2 \times 10^{11}$ proton/cm$^2$ and different
irradiation energy

In the Figure 5.9 a threshold current in the emitted optical power can be observed. This threshold is due to the shunts in the solar cell. In order to check for the homogeneity of the irradiation, maps of the transient microwave conductivity (TRMC) amplitude have been done with pulsed laser illumination at a wavelength of 630nm on a crystalline silicon wafer, irradiated at 1.7MeV. One observes in Figure 5.10 an inversion of the TRMC signal in the central irradiated region. The inversion indicates a strong change of the wafer conductivity with irradiation. The rather constant TRMC amplitude value of about -1mV within the whole irradiated region confirms homogeneity of the proton irradiation
Figure 5.10 Imagine of the (a) TRMC amplitude of a proton irradiated crystalline silicon sample and (b) the histogram of the measured amplitude values

Also the LBIC-image (Figure 5.11) of a c-Si homojunction, irradiated in this case in a circular region in the center with protons of 1.7MeV energy shows a rather homogeneous about 20% decrease of the photocurrent, as compared to the non irradiated regions at the borders.
5.3 Solar cell modification with protons with an energy of 1.7MeV

Subsequently two types of an a-Si:H/c-Si heterojunction with and without i-layer and for c-Si homojunction have been irradiated, maintaining a constant energy of 1.7MeV, with four different proton doses.

5.3.1 Influence of irradiation dose and solar cell structure on the degradation

The influence of irradiation dose has been investigated through principal characteristics for the solar cells. The current- voltage
The characteristics of solar cells investigated has been measured before and after irradiation and are reported in Figure 5.12.

Figure 5.12 Current-Voltage characteristics under AM1.5 conditions of a two different a-Si:H/c-Si heterojunctions solar cells and of a c-Si homojunction before and after irradiation with \(5 \times 10^{12} \text{ protons/cm}^2\) at 1.7MeV.

The arrows indicate the effect of the irradiation. In this Figure is possible to see clearly the construction differences of the solar cells. Both heterojunction solar cells show an open circuit voltage of about 600mV while the homojunction cell has a higher value about of 625mV, partly due to higher quality of base. The heterojunction without TCO has a higher short circuit current density \((J_{SC})\), due to textured silicon surface (inverted pyramid structure) and due to the low back contact recombination losses (back surface field). Instead the heterojunction show a short circuit current density higher respect of the short circuit current of homojunction solar cells, this is due to the
properties of amorphous silicon layer. The homojunction cell has shown the highest value of open circuit voltage ($V_{OC}$) and fill factor (FF) but a rather low short circuit current density about 25mA due to high reflection because no anti-reflection coating has been deposited on the top cell. In order, the quantum yield spectra of both heterojunction solar cells and homojunction, before and after irradiation has been compared in Figure 5.12.

Figure 5.13  Quantum yield spectra of (a and b) two different a-Si:H/c-Si heterostructure solar cell with i-layer, (c) a a-Si:H/c-Si heterojunction solar cell without i-layer and (d) a p-n homojunction solar cell, before and after irradiation with various doses of 1.7MeV protons.
All heterojunction solar cells (Figure 5.13a, Figure 5.13b and Figure 5.13c) show a similar degradation in the long wavelength range, again due to the enhancement of the effective optical absorption length achieved by surface texturing, while they show a remarkable difference in the short wavelength range. For all doses that have been used for irradiate the solar cells with intrinsic layer and without BSF, the quantum yields increase after irradiation for the short wavelength.

This effect is absent for the other heterojunction solar cells, therefore it is possible to conclude, that it is strongly correlated to a irradiation induced modification of the intrinsic amorphous layer in the heterojunction solar cells without BSF. Before it should be mentioned, that it has been assumed in the AFORS-HET simulation, that the proton-induced modifications of the amorphous silicon emitter can be neglected. This is reasonable for the highly doped a-Si:H layer at a proton dose below $5 \times 10^{12}$ protons/cm$^2$ and a proton energy of 1.7MeV. It has, however, been found, that under this conditions, doped microcrystalline films and intrinsic amorphous silicon layer already exhibit strongly modified values of the dark conductivity activation energies [112]. The degradation of homojunction solar cells follows another degradation kinetics, indeed we can see a stronger degradation for low irradiation doses and subsequently a saturation of the damage.

The last dose irradiation is similar at the case of the heterojunction solar cell. Indeed it is possible to conclude, that the dominant degradation for this solar cells is due to the degradation of the monocrystalline silicon base layer transport properties. In Figure 5.14 the irradiation dose dependence of the effective charge carrier diffusion length, $L_{\text{eff}}$, as obtained by a fit of these characteristics in the long wavelength range (procedure as described in) has been shown. $L_{\text{eff}}$
decreases monotonically from values of about 160µm before irradiation down to values of about 6µm for the highest dose of $5 \times 10^{12}$ protons/cm$^2$ for both types of heterojunctions, while for the homojunction is not possible to individuate a good liner fitting. For this solar cells it is possible to individuate two linear fitting (red lines), after an initial degrade there is a saturation in the effective diffusion length.

![Comparison of the effective minority carrier diffusion length extracted from the Quantum Yield measurements of heterojunction and homojunction solar cells, after and before irradiation](image)

**Figure 5.14** Comparison of the effective minority carrier diffusion length extracted from the Quantum Yield measurements of heterojunction and homojunction solar cells, after and before irradiation

### 5.3.2 Change of electroluminescence properties of silicon solar cells after irradiation

For the evaluation of the bulk and interface modification of the solar cells due to irradiation has been used a room temperature
electroluminescence measurements under forward bias. This technique has recently been used to follow the device degradation and it is a very sensitive tool for the characterization of heterointerface in solar cells. The experimental setup that has been realized is shown in the chapter 6. In the Figure 5.15a emission spectra of homojunction solar cells before and after irradiation, for an energy of 1.7MeV and for a doses of $2 \times 10^{11}$, are compared and in the Figure 5.15b emission spectra of a-Si:H/c-Si heterojunction solar cells before and after irradiation, in the same condition of homojunction solar cells, are compared.

![Figure 5.15 Comparison of the electroluminescence spectrum of (a) a c-Si homojunction solar cells and (b) of a-Si/c-Si heterojunction solar cells before and after irradiation](image)

For both solar cells it is possible to see a single emission peak around 1150nm that is due to silicon band-to-band recombination. The spectrum of irradiated solar cells is obviously smaller because the emission is still dominated by silicon band-to-band recombination and no defect related emission has been detected. For the full width half maximum (FWHM) of the emission spectrum a value of 95nm has been measured for the homojunction solar cells, that
is slightly lower than the value of 110nm, measured for the heterojunction silicon solar cells. This small difference is due to an increased re-absorption of the emitted light in the shorter wavelength range in the case of the homojunction solar cell, because the amorphous silicon emitter in the heterojunction with its higher band gap value acts as a window layer. Subsequently the emitted optical power (P) as a function of the dc bias current (I) for a series of five solar cells has been measured, irradiated all with the same proton energy, but with different proton doses. In the Figure 5.16 we compare the emitted optical power for an a-Si:H/c-Si heterojunction irradiated with 1.7MeV at different doses.

![Figure 5.16 Emitted optical power of an a-Si/c-Si heterojunction irradiated with 1.7MeV and different doses energy](image)

From the electroluminescence intensity versus bias current dependence, can be determined the differential electroluminescence quantum efficiency.
A plot of these value as a function of irradiation doses is reported in Figure 5.17 for an a-Si:H/c-Si heterojunction with and without intrinsic layer.

![Figure 5.17](image)

**Figure 5.17** Comparison of the differential electroluminescence efficiency as a function of the proton irradiation doses for an a-Si:H/c-Si heterojunction with and without intrinsic layer.

It can be clearly seen that the insertion of the thin intrinsic layer results in a faster degradation with increasing proton doses. In particular in Figure 5.18 the differential electroluminescence efficiency for the same solar cells irradiated with $2 \cdot 10^{11}$ protons/cm$^2$ has been plotted. A non-linear degradation with a maximum degradation at about 1.7MeV is clearly visible.
Figure 5.18 Comparison of the differential electroluminescence efficiency as a function of the irradiation energy for $2 \cdot 10^{11}$ protons/cm$^2$.

5.4 Solar cell modification with protons with an energy of 65MeV

This energy has been chosen, because in this way has been obtained an homogeneous degradation profile throughout the silicon wafer, which simplifies the interpretation of the results.

5.4.1 Influence of irradiation dose and solar cell structure on the degradation
In this paragraph the effect of proton radiation at 65MeV has been investigated for different technology of solar cells and for different doses of irradiation. The solar cells irradiated with this proton energy are an c-Si homojunction and an a-Si:H/c-Si heterojunction solar cells with i-layer. In Figure 5.19a the quantum yield spectra measured for c-Si homojunctions, irradiated with four different doses, are compared. While in Figure 5.19b we the quantum yield spectra for an a-Si:H/c-Si heterojunction solar cells with i-layer are shown.

For c-Si homojunction solar cells a constant degradation for all wavelength has been found, while for a-Si:H/c-Si heterojunction solar cells has been found a constant degradation for higher wavelength and a not substantial degradation for blue range wavelength. This indicates that the degradation is mainly due to defect creation in the crystalline silicon substrate, as evidenced by the large decrease in quantum yield for wavelengths above 500nm. For this solar cell has been measured also the current-voltage characteristics that is reported in Figure 5.20.
In Figure 5.20a the current – voltage characteristics for an c-Si homojunction solar cells irradiated with a different doses are shown, while in Figure 5.20b the same characteristics but for an a-Si/c-Si heterojunction with i-layer are reported.

![Figure 5.20 Current – Voltage characteristics under AM1.5 conditions for a c-Si homojunction (a) and for an a-Si:H/c-Si Heterojunction with i-layer (b) with various doses of 65MeV](image)

In case of the c-Si homojunction solar cells the differences between the non-irradiated sample and the sample irradiated with $5 \times 10^{10}$ proton/cm$^2$ are very small, the similar series resistance ($R_s$) and the similar open circuit voltages ($V_{OC}$) but a small decreases of short circuit current density ($J_{SC}$) can be seen. For the higher doses a constant degradation has been found. Subsequently, the variation of fundamentals parameters of the c-Si homojunction solar cells as a function of the proton irradiation has been extracted. These parameters are reported in Figure 5.21 and 5.22 In the Figure 5.21 the short circuit current density and the open circuit voltage are compared and it can see that both decreases whit increasing proton dose, but the short circuit current has an higher decrease of the open circuit voltage.
Figure 5.21 Short circuit current density and open circuit voltage for n c-Si homojunction solar cells, irradiated with 65MeV, as a function of proton doses

In the Figure 5.22 instead the diffusion length and the efficiency as a function of proton dose for the c-Si homojunction solar cells has been plotted.

Figure 5.22 Diffusion length and efficiency, for c-Si homojunction solar cells, irradiated with 65MeV, as a function of proton dose
For the diffusion length a higher decrease for low doses, of about 75%, and a slower decrease for high dose are clearly visible, while for the efficiency has been noted an inverse situation, a slow degradation for a low doses and a higher degradation for high doses. For an a-Si:H/c-Si heterojunction solar cells have been extracted the fundamentals parameters, too. In Figure 5.23 the normalized efficiency and normalized short circuit current, as a function of proton dose, are shown.

**Figure 5.23** Normalized efficiency and normalized short circuit current density for a-Si:H/c-Si heterojunction irradiated with 65MeV

Normalized efficiency and normalized short circuit current decreases linearly with increasing of proton dose. In Figure 5.24 the normalized open circuit voltage and normalized fill factor as a function of proton dose are plotted. The Voc and the FF also decrease with increasing of proton dose.
Figure 5.24 Normalized open circuit voltage and normalized Fill Factor for a-Si:H/c-Si heterojunction irradiated with 65MeV

In order, have been extracted the diffusion length for the a-Si:H/c-Si heterojunction solar cells. The measured quantum efficiency is converted to the bulk diffusion length, $L_D$, according to a commonly applied analysis using a plot of the IQE$^{-1}(\alpha-1)$ versus the inverse of the absorption coefficient, $\alpha^{-1}(\lambda)$.

The effective diffusion length is determined from the slope of this curve. A typical $1/L_D^2$ versus $\Phi$ is shown in Figure 5.25.
In order to get also information on the effective lifetime degradation, the TRMC technique has been used. Besides the c-Si homojunction solar cells have been irradiated with boron doped references wafers, and measured them after excitation with 8ns pulses of a Nd:YAG laser (10264nm, 600nJ cm\(^{-2}\)).

As it can observe in Figure 5.26, do we find for low excitation density before and after irradiation an almost single exponential decay of the excess charge carrier concentration with time. We find a monotonic decrease of the effective lifetime from the initial lifetime before irradiation of 2.74µs down to a value of 60ns after irradiation with 5\(\times 10^{12}\) protons/cm\(^2\) [114].
Figure 5.26 TRMC transients measured with a pulsed optical excitation on boron doped p-type c-Si substrates (5–10 Ω cm) without and for different fluences of proton irradiation at 65 MeV. [114]

5.4.2 Change of electroluminescence properties of silicon solar cells after irradiation

The setup described in the chapter 6 for to measure the electroluminescence of c-Si homojunctions irradiated with 65MeV has been used. In Figure 5.27 the emitted optical power for c-Si homojunction solar cells and for a-Si:H/c-Si heterojunction, non-irradiated and irradiated with different doses, are compared.
Figure 5.27 Emitter optical power for a c-Si homojunction (a) and for an a-Si:H/c-Si heterojunction (b) irradiated with 65MeV at different doses.

In particular, in the Figure 5.27a are compared the data for a c-Si homojunction solar cell and in the Figure 5.27b are compared the data for a a-Si:H/c-Si heterojunction solar cell.

The c-Si homojunction solar cell is degraded strongly until $1 \cdot 10^{12}$ protons/cm². Subsequently has been extracted the differential electroluminescence efficiency (dP/dI) and has been plotted it in the Figure 5.28. It can be note the strong degradation for these solar cells.
5.5 Influence of electron irradiation at 1MeV on the silicon heterojunction solar cell degradation

The relationship between proton and electron damage is especially important and it is usual to compare them. For this reason, for one type of an a-Si:H/c-Si heterojunction solar cell, with the intrinsic a-Si:H buffer layer on top of the c-Si absorber, the influence of 1MeV electron irradiation on the device characteristics has been investigated at different doses of $1 \cdot 10^{12}$, $1 \cdot 10^{13}$ and $1 \cdot 10^{14}$ electrons/cm$^2$.

For these solar cells the spectral response is measured and plotted in Figure 5.29.
Comparing this result with proton irradiation, it can be seen that also in this case a photocurrent enhancement has been observed. The difference is that this effect happens in the wavelength range between 600nm and 1000nm for the cells irradiated with intermediate doses. In figure 5.30 the current-voltage characteristic under AM1.5 illumination before and after irradiation with different doses of 1MeV electrons is shown.
Figure 5.30 Solar cell current-voltage characteristics under AM1.5 illumination of a-Si:H/c-Si heterojunction solar cells with intrinsic buffer layer before and after irradiation with different doses of 1MeV electrons

Additionally the characteristics parameters are been extracted and are reported in the Figure 5.31.

Figure 5.31 Comparison of the solar cell parameters under AM1.5 illumination of a-Si:H/c-Si heterojunction solar cells with intrinsic buffer layer before and after irradiation with different doses of 1MeV electrons, normalized to the values before irradiation
Defect related photocurrent enhancement has been discussed for solar cells, exploiting the subbandgap absorption on defect levels, but to the best of our knowledge, no such enhancement has been observed after particle irradiation of homojunction silicon solar cells. We propose therefore another mechanism for the observed photocurrent enhancement, related to the defect enhanced tunnelling through the spike in the conduction band at the heterointerface between the amorphous and crystalline silicon.
Conclusions

The crystalline silicon as absorber material will certainly continue for a while to dominate the market for space applications of solar cells. In literature, there is a large number of studies regarding the stability of crystalline silicon solar cells under space conditions, but there has been so far almost no study regarding the stability of silicon heterojunction solar cells, based on the growth of an amorphous emitter on top of the crystalline silicon absorber. The reason of an eventual future utilization of this kind of solar cells for electrical power generation in space, is their simple fabrication technology, the use of only low-temperature processes during fabrication and the higher efficiency at elevated temperatures, as compared to homojunction crystalline silicon solar cells, due to the efficient silicon surface passivation by the amorphous layers.

In this thesis the applicability of a-Si:H/c-Si heterojunction solar cells in space has been tested by the investigation of the cell modification by high energy protons and comparing the results to the degradation of homojunction crystalline silicon reference cells. The investigated solar cells have been irradiated with protons of different energies and doses. It has been found that for all investigated solar cells the maximum damage happens for an energy of about 1.7MeV and is mainly due to the decrease of the effective minority carrier diffusion length in the
crystalline silicon absorber. Simulations carried out by AFORS-HET, a heterojunction simulation program, also confirmed this result, when assuming the same inhomogeneous defect distribution, as the damage profile, calculated using the TRIM code. The main degradation mechanism for all types of devices is the monotonically decreasing charge carrier diffusion length in the p-type monocrystalline silicon absorber layer. Despite the monotonic decrease of the solar cell performance with increasing doses for proton irradiation, for the heterojunction solar cell an enhancement of the photocurrent in the blue wavelength region has been observed, but only in the case of the heterojunction solar cell with intrinsic a-Si:H buffer layer. In the case of the electron irradiation at 1MeV of the same solar cells with intrinsic buffer layer, a photocurrent enhancement in the wavelength range between 600nm and 1000nm for the cells irradiated with intermediate doses has been observed. In this case, for the first time for a silicon based solar cell, also the solar cell efficiency increased upon irradiation at moderate doses.

Additionally to the traditional characterization techniques the electroluminescence technique used for monitoring the modifications of the heterointerface between amorphous silicon and crystalline silicon in solar cells after proton irradiation. For all investigated samples, the electroluminescence spectrum, measured at room temperature, was dominated by the IR emission around 1150nm and no defect related emission peak has been observed under these measurement conditions. A direct relation between minority carrier diffusion length and electroluminescence quantum efficiency has been observed, but also details of the interface modification could be monitored by this technique.
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Acknowledgements

First of all I would like to thank Prof. Wolfgang Fahrner, for believing in me and giving me the opportunity to work on this dissertation. Very special thanks go to Prof. Heinz-Christoph Neitzert for enthusiasm and the support given during the years, spent in the optoelectronics laboratory at Salerno University. Also I would like to thank Nicola Giovanni Rainone for performing the LBIC measurements on the silicon homojunction solar cells. Thanks are due to Prof. M. Romano, Dr. Lucio Gialannella and Dr. Benedetta Limata from the Physics department at the Federico II University in Naples and the Naples chapter of the INFN for the possibility to use the accelerator facility and the continuous support during proton irradiation. I appreciated very much their availability and their professionalism. Another thanks are due to Dr. Marinus Kunst and Dr. Frank Wuensch from the Hahn-Meitner-Institut in Berlin for the TRMC measurement, and Dr. Andrea Denker and Dr. Zsofia Kertész from the ISL facility of the same institute for the proton irradiation at 65MeV. I wish to thank Dr. Eugenia Bobeico, Dr. Pasquale Morvillo, Franco Roca and Dr. Mario Tucci, at the ENEA photovoltaic research center in Portici for the realization of a part of the characterized solar cells and stimulating discussions. I would also like to thank all members of Prof. Fahrner's group at Hagen University for their hospitality and their patience during my introduction into silicon heterojunction technology. Just to mention one name: I want to thank Maximilian Scherff for the deposition and characterization of myriads of a-Si:H/c-Si
heterojunction solar cells and his introduction into the simulation of the heterojunction solar cells.

Finally I would like to thank my mother and my brother for all. A special thank to Mauro for believing in me and for the support given during the years.
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List of publications

Publication related to this thesis

Journals


Conferences


Other publications

**Journals**


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- M. Ferrara, N. Roseigno, D. Cutolo, H.C. Neitzert and F. De Santis, “VCSEL’s and LED’s as wavelength sensitive optical sensors operating near avalanche breakdown”, European Optical Society Topical Meeting on Optical Microsystems (OµS), Capri, Italy, 2005, p. 51