

5 Silicon for Photovoltaics

J.-C. Muller, P. Siffert

5.1 Introduction

Photovoltaics (PVs), which convert solar energy directly into electricity, are probably the most effective alternative energy source to conventional power supplies. The world demand for terrestrial applications of photovoltaic systems has increased, first for isolated areas and more recently in connection with grids in large PV roof programs, the most important cases being in Japan and Germany.

Many types of solar cell structures have been developed, based on various semiconductor materials, including II–VI or III–V compounds. Silicon was the first material used and is probably still the most appropriate candidate for the next 20 years of large-scale applications. The reasons for this success are the great availability, technical advance and environmental safety of silicon.

Solar cells produced from cast multicrystalline silicon (large-grain polycrystalline silicon, referred to as mc-Si) can be considered as the only present-day technology capable of achieving low-cost high-rate production without excessive loss of efficiency compared with single-crystal silicon (sc-Si). The global module-shipping market reached 530 MWp in 2002 [1] (Fig. 5.1), quite twice the market two years before, with 99% of the products still based on silicon and now more than 55% on cast mc-Si ingots [2] (see Fig. 5.2).

Polycrystalline silicon substrates contain grain boundaries and residual impurities in higher concentrations than in single-crystal silicon. For this reason, the photovoltaic efficiency was initially inferior to that obtained using conventional Czochralski (Cz) or float-zone (FZ) silicon.

However, 25 years of research on multicrystalline silicon have enriched the knowledge of the material and of its thermally activated defect reactions during cell processing, leading to a substantial improvement of the manufacturing techniques and to the capability to produce high-efficiency solar cells with this material. The difference between polycrystalline and single-crystal cells has since decreased continuously.

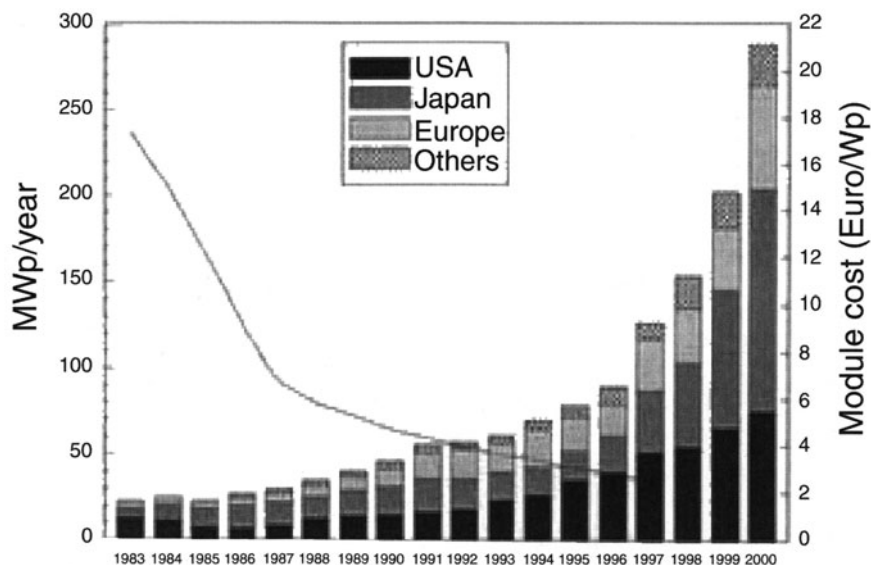


Fig. 5.1. Evolution of the cost per module and of the world module-shipping market. In 2001 the production reached 400 MWp, twice the 1999 market [1], and 530 MWp in 2002 [2] (also quite twice the year 2000 market), now with a shipping cost lower than 3 euros/Wp

5.2 Silicon Material for PVs

This part will be devoted to the material-growing technology. It covers the initially used Cz single crystals, cast multicrystalline ingots, silicon ribbons and the various thin-film deposition processes. It also addresses the emerging crucial problem of the preparation of cheap solar-grade feedstock material.

Although the ribbon and film processes can use electronic-grade silicon or pure SiH_4 gas as feedstock material, the cast-ingot technologies need cheap parent materials since about 60% of the material is lost during the slicing into 300 μm sheets. For these reasons, the PV industry has up to now used materials rejected by the microelectronics industry. However, since the PV market is growing faster than the microelectronic market, the cost of the parent material has multiplied by a factor of 3, now exceeding \$25 per kg. Consequently, the availability of silicon-grade feedstock material has become a crucial problem.

Today, simplified purification methods based on adapted chemical methods or derived from directional solidification processes that apply, for example, electron beam pulling methods [3] or plasma purification, are being investigated.

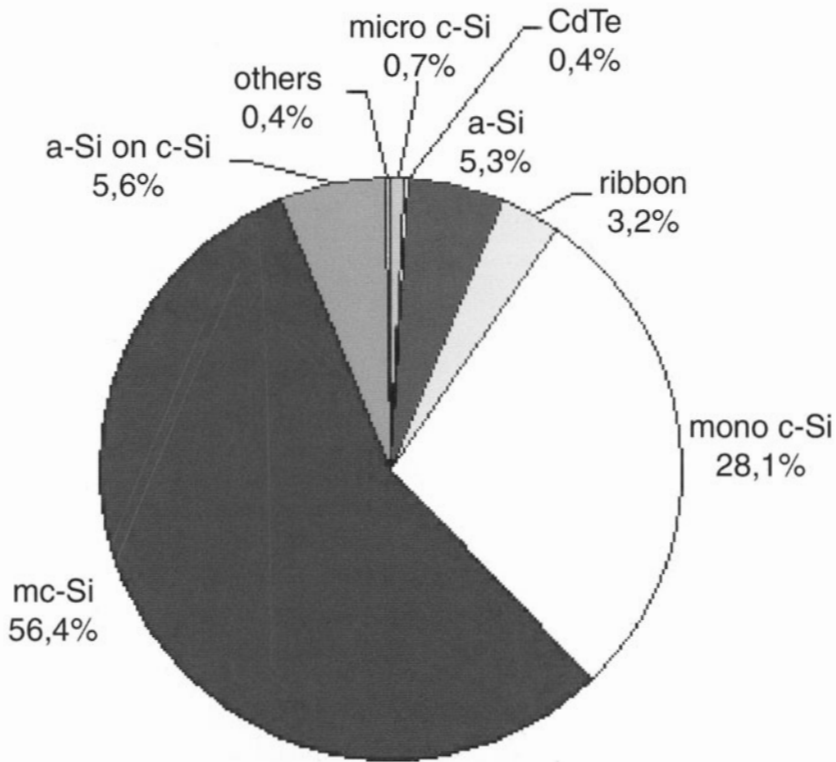


Fig. 5.2. Illustration of the percentage of the different type of materials in the total world production: The share of silicon in the total world product is still 99%, with more than 55% of it being cast mc-Si ingots [2]

5.2.1 History and State of the Art of Different Growing Processes

(a) Cast Ingots by Directional Solidification

In 1975, Wacker proposed a new method, called the Wacker ingot casting process (WICP) [4], to manufacture low-cost substrates for terrestrial solar cells, instead of the conventional Czochralski technology. Later on, many other casting processes were introduced by a large number of research groups and industrial companies throughout the world, such as Solarex (UCP) [5], Crystal-System (HEM) [6], CGE/Photowatt (Polix) [7], IBM (DS) [8], NEC/OTC (NMR) [9], Eurosolare [10], Crystalox [11] and Bayer [12].

These processes use quartz crucibles to melt silicon pieces from Cz ingots rejected by the microelectronics industry and subsequently apply a directional solidification. The solidification process is carried out with a controlled vertical temperature gradient of the silicon melt in the mould. This leads to a silicon block with vertical lines of crystalline silicon with a columnar structure

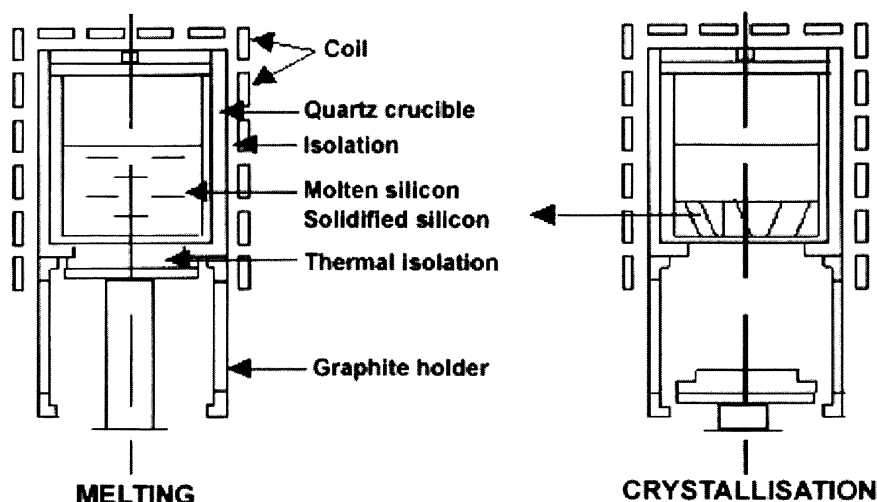


Fig. 5.3. Illustration of the solidification process carried out with a controlled vertical temperature gradient of the silicon melt in the mould to obtain a block of crystalline silicon with a columnar structure of vertical lines from the bottom to the top of the block, e.g. Polix pulling. From Photowatt International [13]

extending from the bottom to the top (see Fig. 5.3). The casting methods were rapidly proven to be cheaper than conventional Cz methods, but their main disadvantage was that they yielded materials containing stresses, impurities and grain boundaries.

Many methods have been investigated to improve the heat flux control during solidification, in order to yield interfaces as planar as possible and hence perfect columnar crystal growth. The manufacturers have gradually reduced the internal stresses and have been able to suppress the generation of cracks by a more careful control of the temperature gradient.

The manufacturers have also tried to avoid interaction of the molten silicon with the mould. Graphite moulds, which can be reused with an additional coating of the inner walls, induce significant contamination. They have been progressively replaced by single-use quartz moulds, as these allow manufacturers to reduce considerably the oxygen and carbon content of the material.

Moreover, in order to reduce the manufacturing costs, efforts to obtain larger blocks have been made. Ingots with cross-sectional areas larger than $50 \times 50 \text{ cm}^2$, a height of 18 or 28 cm, and 250 kg in weight are now currently available.

After contouring and removal of the bottom and top regions, which do not satisfy the conditions for use, the ingot is partitioned into 9, 16 or 25 elementary blocks having a base area between $10 \times 10 \text{ cm}^2$ and $15 \times 15 \text{ cm}^2$, using bandsaws. The final slicing into wafers with a thickness of about $300 \mu\text{m}$ is performed with multi-wire saws.

Both the progress in multi-wire sawing of very thin slices – Photowatt [13] was the first to cut at about $200\text{ }\mu\text{m}$ – and the further development of continuous casting are the main reasons for the success of cast ingots as compared with the ribbon technology.

(b) Electromagnetic Casting (EMC)

In 1981–1985, OTC [14] and SERI [15] developed a cold-crucible casting technique based on induction heating. Neither a crucible nor a mould is used in this method. The molten silicon is heated and confined electromagnetically by induction power in order to avoid impurity contamination from the furnace

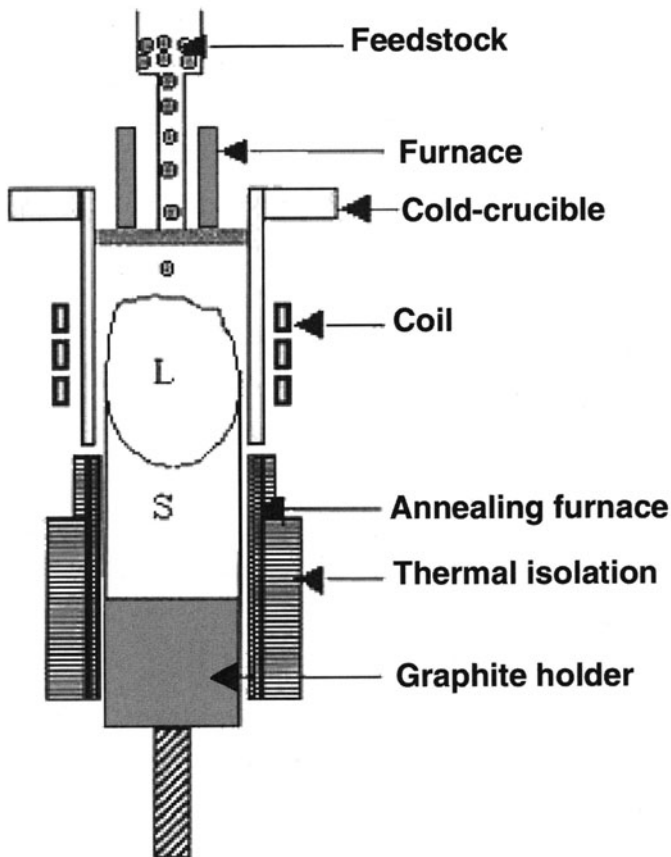


Fig. 5.4. Illustration of the cold-crucible casting technique based on induction heating. In this method, no mould is used. The molten silicon is heated and confined electromagnetically, for example in the Emix process, initially developed by EPM-Madlylam with Photowatt [17] and now used industrially by EMIX S.A. [18]

(Fig. 5.4). The EMC vertical continuous-pulling method was developed by Sumitomo SiTiX Corp. [16], formerly OTC, as well as by EPM-Madylam [17] together with Photowatt, and will be used by EMIX SA in the future [18]. Crystallox [19], on the other hand, has developed a horizontal-zone refining method for silicon applying induction heating.

Ingots of a cross-sectional area of approx. $35 \times 35 \text{ cm}^2$, 3 m long and with a weight of 850 kg can be obtained by these continuous-pulling techniques using plasma torch melting for better control of the solid/liquid interface and of contamination [20].

The cold-crucible casting technique, which reduces the energy consumption for crystal pulling by almost a factor of 4, constitutes a major opportunity for the future evolution of multicrystalline silicon, which is in constant competition with silicon ribbons and, more recently, with silicon thin films.

(c) Ribbon Silicon for PV

Silicon ribbon technology, another promising method that is in competition with both ingot casting and thin-film technologies, also produces multicrystalline material. The production of the first silicon ribbons was done simultaneously in 1975 by Westinghouse [21] with the dendritic WEB process called

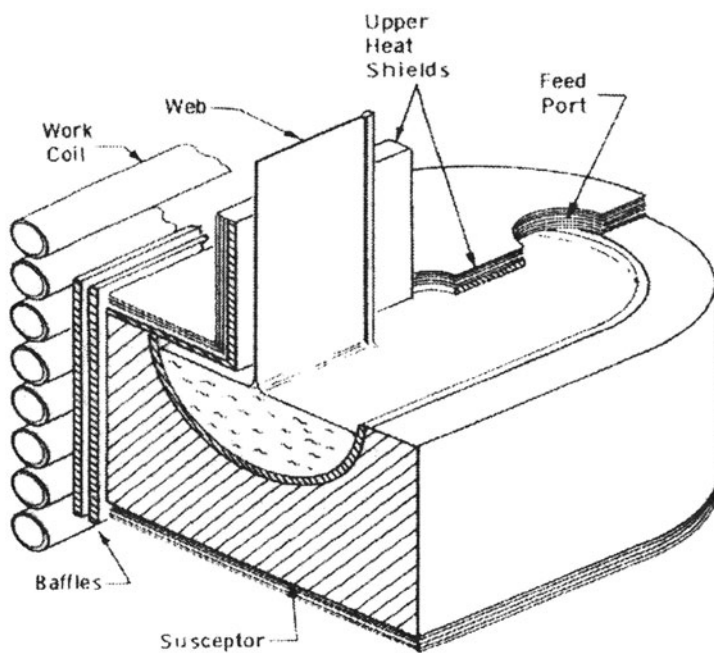


Fig. 5.5. Illustration of the WEB process for growing silicon sheets [21]

“WEB” ribbon process (Fig. 5.5), by ASE-Americas [22] (formerly Mobil-Tyco/Mobil-Solar) with the EFG, edge defined film fed grown [23] process using two graphite dies, and by Motorola [24] with the RTR, ribbon-to-ribbon crystal growth process using laser melting.

A process of two-sided film deposition on carbon sheets (RAD, ribbon against drop pulling process ribbon) was developed later by CGE [25] in France. The silicon grain structure obtained by all these processes shows more longitudinal grains compared with cast ingots. The crystallization quality depends strongly on the pulling velocity (generally around a few cm per minute) and on the pulling material, such as graphite or carbon sheets.

At present, the ribbon material represents 3.7% of the world market [26]. Research is going towards thinner substrates by developing 50 cm diameter hollow cylindrical tubes with a thickness of 75–80 μm [27]. In Europe, RWE-Schott-Solar, Germany, has recently started a 2×30 MWp factory based on the EFG technology [28].

5.2.2 The Thin-Film Deposition Processes

(a) Amorphous, Microcrystalline or Quasi-crystalline Silicon Thin Films for PVs

The thin-film technologies represent today about 9% of the world PV market, and this percentage is mainly accounted for by amorphous silicon. Figure 5.6 [29] reports all laboratory cell efficiency results reported for the var-

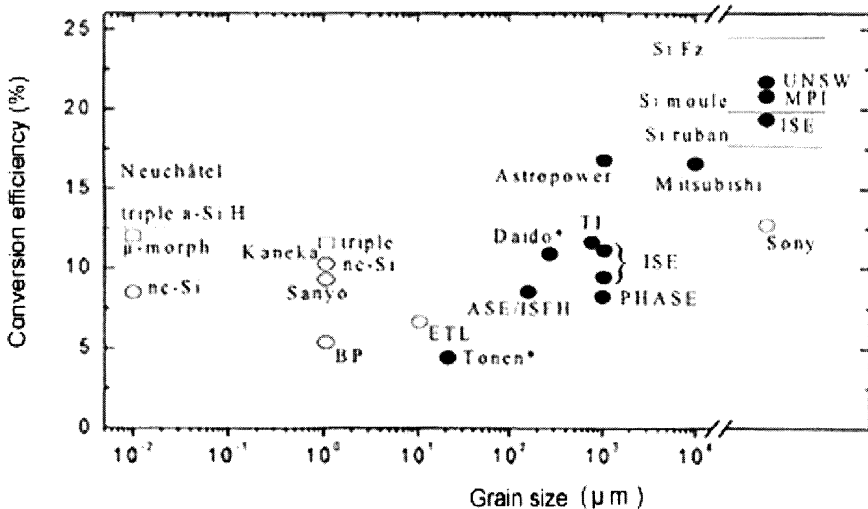


Fig. 5.6. Illustration of the laboratory cell efficiencies of thin-silicon-film technologies as a function of the grain size (from amorphous to epilayers) [29]

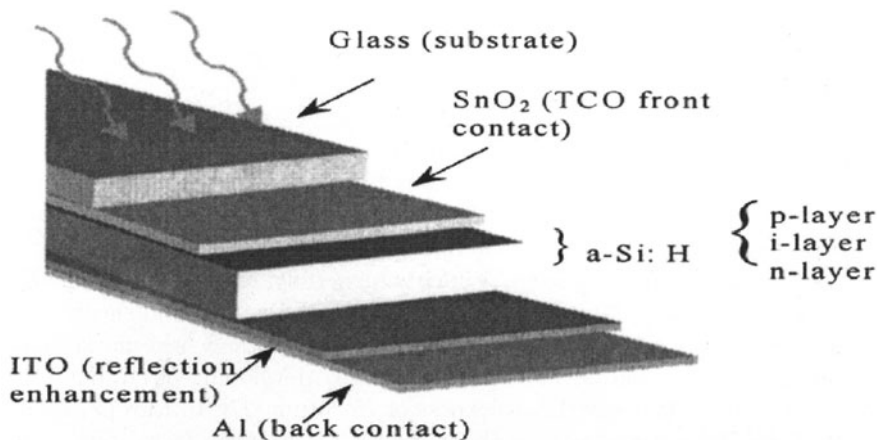


Fig. 5.7. Illustration of the amorphous-thin-silicon-film technology [30]

ious silicon film technologies, from amorphous to almost single-crystal films (quasi-monocrystalline epitaxial layers).

For amorphous silicon, which can be prepared at low temperatures, the substrate is usually glass. The collecting structure consists generally of three layers: a boron-doped and a phosphorus-doped layer with an intrinsic layer in between for improved charge transport in the internal electric field (p-i-n); see Fig. 5.7 [30].

In order to yield higher efficiencies, double and triple p-i-n structures have been developed with thinner layers, resulting in improved minority carrier diffusion length values. Also, small grain sizes in an otherwise amorphous matrix improve the charge transport properties, so that micro- or polymorphous materials have been developed successfully [31] (left part of Fig. 5.6).

The right side of Fig. 5.6 represents the results of research efforts either to growth by CVD processes epitaxial silicon on quasi-single crystals obtained from thermal crystallization of a double porous layer on a Si wafer [32] (see Fig. 5.8), or to transfer a silicon film from a silicon ingot by various techniques, including fragility enhancement of an intermediate layer and ion beam processing [33]. The data in the central region of Fig. 5.6, with grain sizes corresponding to polycrystalline silicon films, are not discussed in this chapter. See Chap. 4 on “Polycrystalline Silicon Films for Electronic Devices” for further information.

From an industrial point of view, the thin-film technologies are important both for reducing substantially the amount of silicon used and for being compatible with roll-to-roll processes [34] (Fig. 5.9) with flexible substrates such as stainless steel or plastic films.

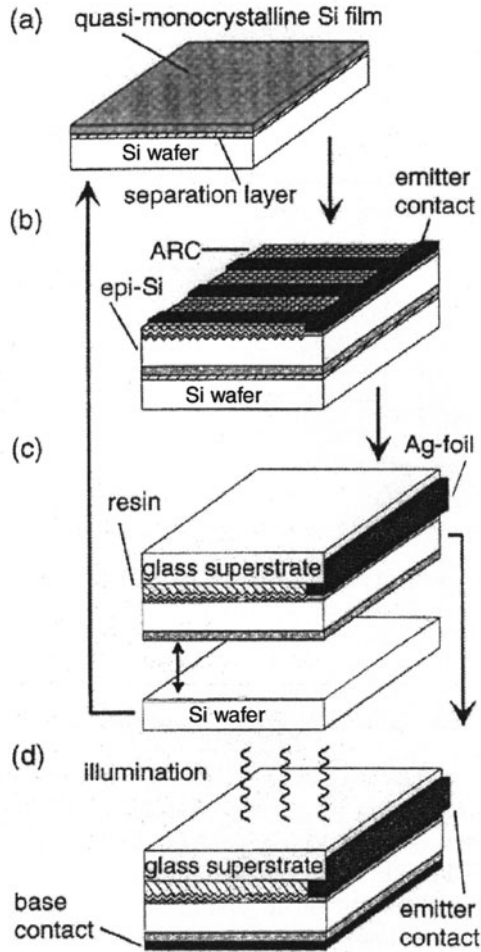


Fig. 5.8. Illustration of the transfer technology of quasi-monocrystalline silicon films [32]

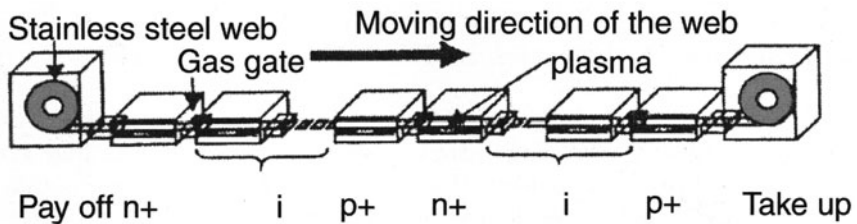


Fig. 5.9. Illustration of the roll-to-roll process adapted to thin-film technologies [34]

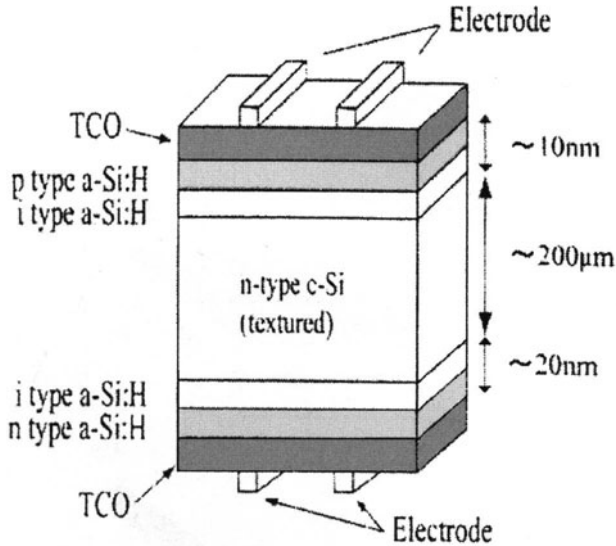


Fig. 5.10. Newly developed HIT structure, which combines the advantage of crystalline bulk transport properties with low-temperature amorphous heterostructures [35,36]. TCO, transparent conductive oxide

(b) Wending of Amorphous with Crystalline Silicon for PV

Sanyo has developed a new structure (HIT) [35] (Fig. 5.10) which combines the efficient charge transport properties of bulk crystalline silicon with the low-temperature process of amorphous silicon deposition. Two heterostructures of amorphous silicon were produced on both sides of the cell. Efficiencies up to 21% have been obtained on n-type crystalline silicon with two-sided illumination [36]. If the industrial feasibility is clearly demonstrated for single crystals, the feasibility of the process on p-type multicrystalline substrates has to be confirmed next, before the process is used industrially on a larger scale.

5.3 Transport Properties in PV Silicon

The crystallographic properties of mc-Si materials are now quite satisfactory: a columnar structure with large grains more than 1 cm^2 in size, and few dislocations and intragrain defects. However, multicrystalline silicon contains larger quantities of impurities than does single-crystal silicon, which can have detrimental effects on the bulk minority carrier diffusion length ($L_{n,p}$). These impurities, which include metals as well as high concentrations of carbon and oxygen, can degrade the photovoltaic properties of solar cells and require specific treatments, presented in Sect. 5.3.2.

5.3.1 Effects of Defects and Impurities on the Transport Properties of Silicon

Impurities and residual defects can have detrimental effects on charge carrier transport and degrade the photovoltaic properties of a solar cell. Specific treatments such as gettering, applied separately or in conjunction with doping steps, can limit or avoid the degradation of the bulk diffusion length. The efficiency of these treatments depends strongly on the type of impurities present in the bulk.

Hydrogen passivation of electrically active defects is another efficient way to improve the transport properties, especially for materials having high densities of intragrain defects and smaller grains. It can be advantageously coupled with gettering processes, as well as with processes used to provide antireflection coatings such as SiN:H plasma deposition.

(a) Defects in Single-Crystal Silicon (sc-Si)

In high-efficiency cells fabricated using single-crystal silicon, a small degradation of the bulk minority carrier diffusion length $L_{n,p}$ can result from the presence of residual defects and impurities. It was found that for cells manufactured on boron-doped Cz silicon material, the efficiency decreases with time and with illumination. This effect, which does not appear in FZ silicon or in Ga-doped Cz materials, is now attributed to a vacancy–oxygen–boron complex [37,38].

(b) Defects in Multicrystalline Silicon (mc-Si)

Multi-crystalline silicon wafers are characterized by a large grain size (larger than a few mm) and by a high density and large variety of intragrain defects, independently of the method used to grow the ingots, i.e. for different crucibles, liners, thermal steps during the solidification, etc. Dislocations, dislocation lineages, dislocation tangles, twins (decorated or not by dislocations) and subgrain boundaries are found. These defects can interact with grain boundaries [39].

Point defects such as vacancies and self-interstitials must also be taken into account, as well as impurities such as metal, oxygen and carbon atoms. These point defects and impurities can segregate at extended defects and form precipitates at defects sites or in the grains. The simultaneous presence of all these imperfections is the main characteristic of this type of materials, which appear to be very complex systems.

The interactions between defects and impurities are driven by chemical and mechanical potential gradients, since oxygen and carbon, at least, are frequently present in supersaturation concentrations [39]. The rate of interaction increases with temperature. Therefore the interaction is strongly influenced

by high-temperature processing steps during solar cell manufacture and explains the large variations in conversion efficiency; see the section devoted to solar cells (Sect. 5.4).

5.3.2 Improvement of the Material by Gettering

The main part of the recombination strength of extended defects is due to segregation of impurities. Impurity precipitates are also sources of recombination centres. Thus mc-Si wafers could be markedly improved by removing impurities from the bulk, especially metallic impurities.

The “gettering” effect can be used to clean the material. The basic mechanisms were divided into three categories by Schröter et al. [40]: (i) relaxation-induced gettering, i.e. trapping of supersaturated impurities by crystallographic defects; (ii) segregation-induced gettering by enhancement of the solubility in a certain region of the wafer; and (iii) injection-induced gettering, i.e. injection of self-interstitials and interaction with impurities.

(a) External Gettering

Multicrystalline silicon wafers are used exclusively to produce solar cells. Such wafers have the advantage of being cost-effective compared with single-crystal wafers. Hence inexpensive gettering techniques must be employed. The best solutions will be those which correspond to processing steps already included in the industrial cell fabrication process, such as phosphorus, boron or aluminium diffusion near the surfaces, and oxidation. “External” gettering of impurities is based on the creation of gettering sites at or near the external surfaces, and on the extraction, fast diffusion, and capture of impurities.

A heavy phosphorus diffusion near the surface is able to create crystallographic defects which behave as gettering sites. The phosphorus concentration is frequently beyond the solubility, and precipitates of SiP are formed. Ourmazd and Schröter [41] have shown that dislocations which trap fast diffusers such as Fe and Ni are generated at the border of these orthorhombic precipitates.

Gettering also results from the existence of a region where the impurity solubility is enhanced by the high phosphorus doping level, owing to the formation of complexes between impurities and doping atoms or to the shift of the Fermi energy level towards the conduction band. Highly N^+ phosphorous-doped regions, for instance, contain a high density of vacancies due to the phosphorus diffusion, which can trap interstitial atoms and increase their solubility.

Phosphorus diffusion has a further beneficial effect on gettering. External gettering requires the extraction of impurities from their substitutional positions, from precipitates, from segregation sites at extended defects or from the impurity cloud which surrounds these defects. The extracted impurities

must reach interstitial positions in order to diffuse rapidly through the crystal to the gettering sites. The extraction is facilitated when an excess of Si_i is injected into the bulk during phosphorus diffusion and participates in the well-known kick-out mechanism and in the shrinkage of precipitates [42]. This is exactly what phosphorus diffusion can do when SiP precipitates are formed near the surface, because this formation involves a molar-volume expansion which generates an excess of Si_i .

Very large minority carrier diffusion lengths could be obtained when thin wafers ($\approx 200\ \mu\text{m}$ thick) of multicrystalline material were phosphorus diffused in a classical furnace at 900°C for several hours, using POCl_3 as a diffusion source [43]. The effect of phosphorus diffusion was applied successfully to various multicrystalline materials such as Silso, Polix, Eurosil and Solarex [44], increasing solar cell efficiencies [45–47] by the removal of metallic impurities such as Fe, Cu and Ni from the active part of the cell and by their accumulation in the front N^+ layer [43, 45].

A large number of reports show that the minority carrier diffusion length can also be enhanced if the silicon wafers are annealed in the presence of an aluminium layer. Aluminium is an acceptor in silicon, and its diffusion leads to P^+ -type regions, which can be used as ohmic contacts or to realize a back surface field (BSF). This diffusion length improvement, which can be as large as several hundred micrometers [48–50], suggests the existence of a gettering mechanism: metallic impurities could be trapped at the aluminium/silicon interface by an aluminium–silicon alloy. This process was found to be enhanced by dislocations.

Aluminium gettering is also an efficient means to improve multicrystalline silicon wafers. Large values of the effective diffusion length L_{eff} (frequently larger than $200\ \mu\text{m}$) are obtained after the deposition of a $1\ \mu\text{m}$ thick layer of aluminium and annealing at 950°C for 30 min, especially when the wafers have been previously gettered by phosphorus diffusion [51].

These improvements have been applied to solar cells to obtain high efficiencies [51, 52]. In spite of the complexity of the measurements (of the back surface field effect), a large part of the improvements of the cell efficiency have been attributed to gettering. Note that phosphorus and aluminium gettering can advantageously be applied simultaneously at the same temperature [49, 53].

(b) The Particular Case of Rapid Thermal Gettering

Despite its success in the microelectronics industry, rapid thermal processing (RTP) had to overcome a major difficulty for PV devices. RTP was revealed to be much more sensitive to contamination than were classical thermal treatments with slow cooling rates. In multicrystalline materials the higher concentrations of impurities are in supersaturation during the high-temperature step, and in the presence of a large density of crystallographic defects, some of

them can be frozen into electrically active sites during the quenching step [54] which is inherently associated with the fast cooling rate of an RTP cycle.

Fast-diffusing impurities (e.g. Co, Cu and Ni) generally have enough time to form electrically inactive precipitates, whereas the slower-diffusing impurities (e.g. Mo, Ti, V, Cr and Fe) mostly remain interstitially dissolved, depending on the quenching speed, and introduce recombination centres. For this reason the second group of metals is harmful to the performance of solar cells at much lower concentrations than the first group [55].

It has been shown that rapid thermal gettering can be just as efficient in improving the bulk minority carrier lifetime as conventional gettering [56]. The precise mechanism which allows gettering of metallic impurities within seconds is not yet well established. Some suggestions can be found in [57], where rapid thermal diffusion of phosphorus from a spin-on deposited source was applied to gold-contaminated monocrystalline silicon.

For mc-Si materials, simultaneous rapid thermal co-diffusion of phosphorus and aluminium [58] is required. Only under this condition does the cumulative effect of both gettering mechanisms lead to an improved minority diffusion length. The optimal process temperature depends on the mc-Si material, owing to the permanent competition between gettering and the quenching-induced formation of recombination centres; see also [59].

(c) Improvement by Passivation of the Material

Besides the remarkable efforts in research on impurity gettering as described above, many kinds of hydrogen passivation techniques have been developed in order to neutralize the activity of dangling bonds and of residual metallic impurities. It has been shown, in particular, that passivation by atomic hydrogen is very effective in improving the performance of multicrystalline-silicon solar cells [60, 61].

Hydrogenation can be performed by various techniques, including RF, remote-plasma and ion beam processes. These treatments need to be performed on the finished cells since the bonds of hydrogen with defects or impurities can be broken at temperatures above 400°C [62], where hydrogen diffuses out of the wafer.

Hydrogen passivation can be applied to the front side and/or to the back. However, for all techniques involving energetic beams [63], back-surface irradiation is preferred in order to avoid damaging the active part of the cell. The most effective hydrogenation procedure was performed by Kyocera [64] during silicon nitride deposition (by PECVD) and is now widely used in the industry in order to form the antireflective coating simultaneously. By this method the cell is encapsulated and passivation can be performed during a short-time high-temperature sintering of the screen-printed contacts, with an additional positive effect on the bulk diffusion length if aluminium is present on the back surface [65].

(d) Role of Hydrogen in Amorphous Silicon

It has been proved that atomic hydrogen is very effective in improving the transport properties of amorphous materials by saturating the dangling bonds, and similarly in most materials with a low degree of crystallization. The hydrogenation is generally done during plasma or CVD deposition processes. The stability of this passivation process depends mostly on the deposition temperature. At 200°C, Si-H(n) bonds are formed, which are less stable than Si-H bonds formed close to 400°C [62] (for more details see Chap. 7, “Amorphous Hydrogenated Silicon”).

5.4 Silicon Solar Cells

5.4.1 Silicon Solar Cell Technology in Comparison with Other Technologies

Silicon solar cell technology has now reached almost its theoretical efficiency limit with the development of the PERL, passivated emitter rear locally-diffused cell by Green at the University of New South Wales (see Fig. 5.11) [66]. The technology using amorphous material which started in the 1960s has now been overtaken by the newly developed CIS, Cu-In-Se compounds. The Si thin-film technology now has to improve its crystal quality and efficiency. Organic cells, which have stability problems, still have a long way to go before industrial production. The new concepts which have been suggested to circumvent the thermodynamic efficiency limit of silicon cells are still lacking experimental confirmation (see Fig. 5.12) [67].

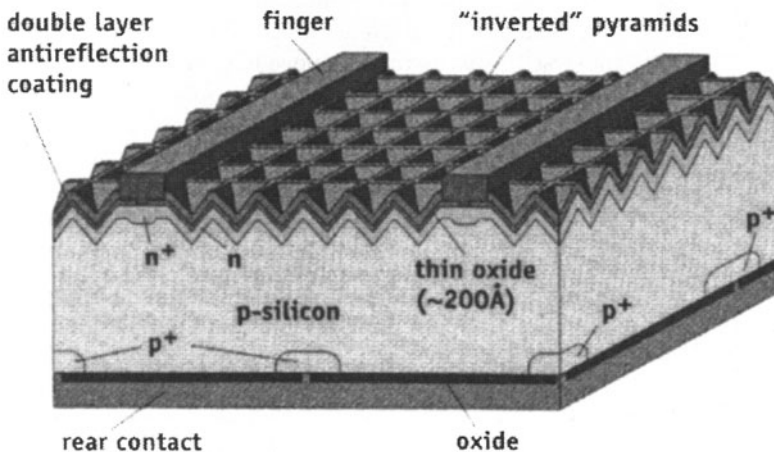


Fig. 5.11. Illustration of the best cell structure that can be formed on an FZ silicon substrate, confirming that silicon solar cell technology has now reached almost its efficiency limit with the development of the PERL cell by Green [66]

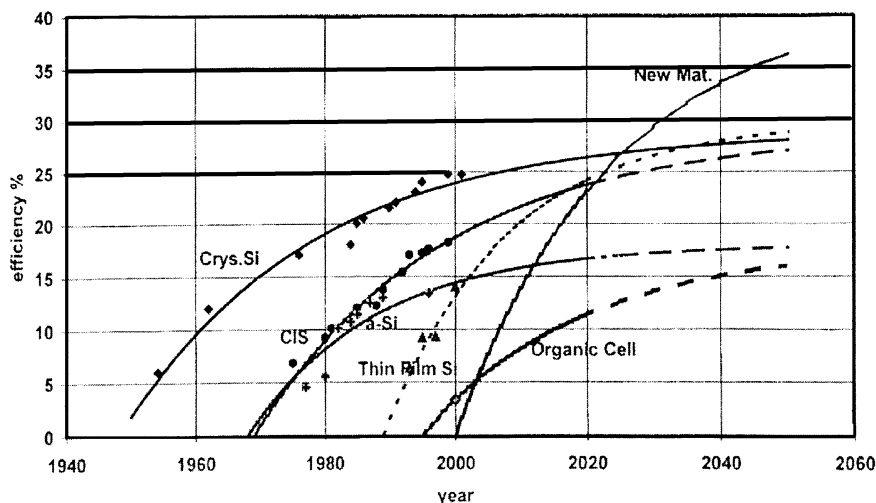


Fig. 5.12. Evolution of the laboratory conversion efficiencies for the three generations of silicon solar cells: the first generation, represented by crystalline silicon; the second one, ranging from amorphous silicon, CIS compounds, which have rapidly surpassed a-Si, and silicon thin films to the more recent organic cells; and the third one, with new materials and new concepts which may improve the efficiency over the present limit [67]

5.4.2 Multicrystalline Silicon Solar Cell Technology

Many organizations have been involved in research and development on multicrystalline silicon solar cells processed using one of the above-mentioned casting methods. The University of New South Wales [68] and Georgia Tech. [47, 52] were the first to successfully achieve high conversion efficiencies in 1 to 4 cm² laboratory-scale multicrystalline-silicon solar cells.

The conversion efficiency of large, 100 cm² pre-manufacturable multicrystalline-silicon solar cells has rapidly reached a maximum of 17.2% by use of a mechanically grooved front electrode [69], screen-printing technology [70] and selective emitters [71]. Moreover, on larger-area (225 cm²) solar cells, efficiencies up to 16.7% [72] have been achieved.

However, the industrial conversion efficiencies of 100 cm² cells in production are between 12.5 and 14.5% (see Fig. 5.13 [59]). For industrial cells in which SiN(H) has been introduced in production as an antireflective coating and as bulk and surface passivation, the efficiencies are closer to 16%.

For rapid thermal processing with fast cooling, it has been shown that the efficiencies of multicrystalline-silicon solar cells can be of the same order as those obtained by classical thermal processes [73]. The highest value, of 16.7% on a 25 cm² cell on Polix material from Photowatt, was obtained at the CNRS-PHASE Laboratory [74].

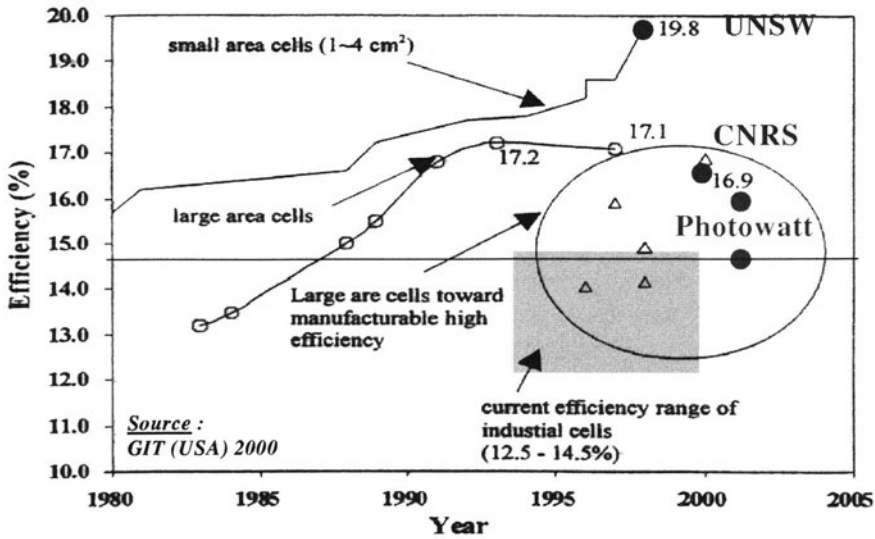


Fig. 5.13. Evolution of the laboratory conversion efficiencies for multicrystalline-silicon solar cells from small laboratory cells to the current industrial cells [59]

5.5 Conclusion

The strategy of future research in industrial silicon solar cell technology has to be concentrated upon the fabrication of larger cells (more than the present $15 \times 15 \text{ cm}^2$), cut from larger cast multicrystalline ingots or from continuous-pulling electromagnetic-casting materials. Oxygen, carbon and metallic impurity concentrations need to be reduced and, consequently, bulk transport properties enhanced. Classical and rapid thermal gettering as well as hydrogenation processes can still be improved, reducing the difference in the conversion efficiency between multi- and monocrystalline silicon cells.

Solar cells have also to be made thinner ($150 \mu\text{m}$ slices for the near future and down to $50 \mu\text{m}$ if possible) in order to reduce the silicon losses during ingot slicing from 40 to 20% and to remain competitive with the ribbon and thin-film technologies for another 50 years.

References

1. *Eurec Position Paper*, Eurec Agency, Brussels, BE-1040 (2000)
2. *World Energy Outlook 2003*, IEA Annual Report
3. N. Yuge, M. Abe, K. Hanazawa, H. Babe, N. Nakamura, Y. Kato, Y. Sakaguchi, S. Hiwasa, M. Obashi: 11th Int. Photovoltaic Science & Engineering Conf. (Sept. 20–24, 1999), Sapporo, Japan, p. 115
4. B.H. Authier: German Patent (DOS) No. 25 0883 (1975); H. Fischer: Proc. 1st EC PV SEC, Luxembourg (1977) p. 52

5. G.M. Storti: Proc. 15th IEEE PVSC, Kissimmee (1981) p. 442
6. E. Schmid, C.P. Khattak: Proc. 5th EC PV SEC, Athens (1983) p. 1019
7. J. Fally, D. Guignot, L. Goeffron: Proc. 7th EC PV SEC, Seville (1986) p. 754
8. T.E. Ciszek, G.H. Schwuttke, K.H. Young: J. Cryst. Growth **46**, 527 (1979)
9. T. Saito, A. Shimura, S. Ichikawa: Proc. 15th IEEE PVSC, Kissimmee (1981) p. 576
10. D. Margadonna, F. Ferrazza, R. Peruzzi: Proc. 10th EC PV SEC, Lisbon (1991) p. 678
11. I.A. Dorrity, B.J. Garrard, D.A. Hukin: Proc. 10th EC PV SEC, Lisbon (1991) p. 317
12. W. Koch, W. Krumbe, H. Lange, W. Schmidt, F. Schomann, G. Wahl: Proc. 12th EC PV SEC, Amsterdam (1994) p. 797
13. Photowatt International S.A., FR-38300 Bourgoin-Jallieu
14. K. Kaneko, T. Misawa, K. Tabata: Proc. 21st IEEE PVSC, Kissimmee (1990) p. 674
15. T.F. Ciszek: J. Electrochem. Soc. **132**, 963 (1985)
16. K. Kaneko, R. Kawamura, T. Misawa: Proc. 1st WCPEC, Hawaii (1994) p. 30
17. I. Perichaud, G. Dour, F. Durand, D. Sarti, G. Goar, Q.N. Le, F. Floret, S. Martinuzzi: Proc. 13th EC PV SEC, Nice (1995)
18. EMIX S.A., FR-23300 La Souterraine
19. D.A. Hukin: Proc. 4th PVSEC, Sydney (1989) p. 719
20. K. Kaneko, K. Sasatani, M. Ohnishi, N. Kimura: Proc. 11th Int. PVSEC Conf., Sapporo, Japan (1999) p. 119
21. R.G. Seidensticker, L. Scudder, H.W. Brandhorst: Proc. 11th IEEE PVSC, Scottsdale, AZ (1975) p. 299
22. ASE-Americas, Inc., Billerica, MA, 01821-3980, USA
23. K.V. Ravi, H.B. Serreze, H.E. Bates, A.D. Morrison, D.N. Jewett, J.C.T. Ho: Proc. 11th IEEE PVSC, Scottsdale, AZ (1975) p. 281
24. A. Lesk, A. Baghdadi, R.W. Gurtier, R.J. Ellis, J.A. Wise, M.G. Coleman: Proc. 12th IEEE PVSC, Baton Rouge (1976) p. 173
25. C. Belouet, J.J. Brissot, R. Martres, N.T. Phuoc: Proc. 1st EC PV SEC, Luxembourg (1977) p. 164
26. EurObserv'ER, Systèmes Solaires, No. 136 (2000)
27. B.H. Mackhintoich, M.P. Ouellette, M.D. Rosenblum, J.P. Kaleijs, B.P. Piwczyk: 28th IEEE PVSC, Anchorage, AK (2000) p. 46
28. RWE-Schott-Solar, Alzenau, Germany
29. R.B. Bergmann: Appl. Phys. A **69**, 187 (1999)
30. P. Roca i Cabarrocas, Ecole Polytechnique, Palaiseau, France, personal communication
31. P. Roca i Cabarrocas, A. Fontcuberta i Morral, Y. Poissant: Thin Solid Film **403–404**, 39 (2002)
32. R.B. Bergmann, T.J. Rinke, C. Berge, J. Schmidt, J.H. Werner: 12th Int. PVSEC Conf., Jeju, Korea (2001) published in: Solar Energy Materials and Solar Cells, **74**, 213 (2002)
33. P.J. Riberon, A. Beaumont, A. Laugier, A. Kaminski, A. Fave, M. Lemiti, G. Fantosi: ADEME-CNRS Annual Seminar on PV Materials and Processes, Sophia-Antipolis (2002) p. 111
34. M. Sano, K. Saito, S. Okabe, S. Sugiyama, A.K. Ogawa: Proc. 12th Int. PVSEC Conf., Jeju, Korea (2001) p. 29

35. T. Sawada, N. Terada, S. Tsuge, T. Baba, T. Takahama, K. Wakisaka, S. Tsuda, S. Nakano: 1st WCPEC, Hawaii (1994) p. 1219
36. M. Taguchi et al: Prog. Photovolt. Res. Appl. **8**, 503 (2000)
37. S.W. Glunz, S. Rein, W. Warta, J. Knobloch, W. Wettling: Proc. 11th Int. PVSEC Conf., Sapporo, Japan (1999) p. 549
38. J.Y. Lee, S. Peters, S. Rein, S.W. Glunz: Proc. 12th Int. PVSEC Conf., Jeju, Korea (2001) p. 27
39. J.C. Muller, S. Martinuzzi: J. Mater. Res. **13**, 2721 (1998)
40. W. Schröter, M. Seibt, D. Gilles: Mater. Sci. Technol. **4**, 540 (1992)
41. A. Ourmazd, W. Schröter: Appl. Phys. Lett. **45**, 781 (1984)
42. J.S. Kang, D.K. Schroder: J. Appl. Phys. **65**, 2974 (1989)
43. I. Périchaud, S. Martinuzzi: J. de Phys. III **2**, 313 (1992)
44. I. Périchaud, F. Floret, M. Stemmer, S. Martinuzzi: Solid State Phenomena **32**, 77 (1993)
45. B. Sopori, L. Jastrzebski, T. Tan, S. Narayanan: Proc. 12th EC PV SEC, Amsterdam (1994) p. 1003
46. L.Q. Nam, M. Rodot, M. Ghannam, D. Sarti, I. Périchaud, S. Martinuzzi: Int. J. Solar Energy **11**, 273 (1992)
47. A. Rohatgi: Proc. 23rd IEEE PV SEC, Louisville (1993) p. 52
48. R. Sundaresan, D.E. Burk, J.G. Fossum: J. Appl. Phys. **55**, 1162 (1984)
49. L.A. Verhoef, S. Roorda: Proc. 20th IEEE PVSC, Las Vegas (1988) p. 1551
50. O. Porre, M. Pasquinelli, S. Martinuzzi, I. Périchaud: Proc. 11th EC PV SEC, Montreux (1992) p. 1053
51. M. Pasquinelli, S. Martinuzzi, J.Y. Natoli, F. Floret: Proc. 22nd IEEE PVSC, Las Vegas (1991) p. 1035
52. A. Rohatgi, P. Sana, J. Salami: Proc. 11th EC PV SEC, Montreux (1992) p. 159
53. L.A. Verhoef, P.P. Michiels, S. Roorda, R. Sinke, R.J. Van Zolingen: Mater. Sci. Eng. B **7**, 49 (1990)
54. W. Eichhammer, Vu-Thuong-Quat, P. Siffert: J. Appl. Phys. **66**, 3857 (1989)
55. E. Weber. In: *Impurity Diffusion and Gettering in Silicon*, Mater. Res. Soc. Proc. **36**, 3 (1985)
56. B. Hartiti, A. Slaoui, J.C. Muller, R. Stuck, P. Siffert: J. Appl. Phys. **71**, 5474 (1992)
57. B. Hartiti, A. Slaoui, J.C. Muller, P. Siffert: Appl. Phys. Lett. **63**, 1249 (1993)
58. B. Hartiti, A. Slaoui, J.C. Muller, P. Siffert, R. Schindler, I. Reis, B. Wagner, A. Eyer: Proc. 23rd IEEE PV SEC, Louisville, (1993) p. 224
59. A. Rohatgi, V. Yelundur, J. Jeong, A. Ristow, A. Ebong: 10th Workshop on Crystalline Silicon Solar Cell Materials and Process, (Copper Mountain, CO August 2000), p. 12
60. J.C. Muller, E. Hussian, P. Siffert, D. Sarti: Proc. 9th EC PV SEC, Freiburg (1989) p. 407
61. B. Sopori, K. Jones, X. Dung, R. Matson, M. Al-Jassin, S. Tsuo, A. Doolittle, A. Rohatgi: Proc. 22nd IEEE PVSC, Las Vegas (1991) p. 833
62. J.C. Muller, B. Hartiti, E. Hussian, J.P. Schunck, P. Siffert, D. Sarti: Proc. 22nd IEEE PVSC, Las Vegas (1991) p. 883
63. S. Sivothythaman, M. Rodot, J.C. Muller, B. Hartiti, M. Ghannam, H.E. El-gamel, J. Nijs, D. Sarti: Appl. Phys. Lett. **62**, 3172 (1993)
64. H. Watanabe: 4th Int. PVSEC Conf., Sydney (1989) p. 103

65. V. Yelundur, A. Rohatgi, J.W. Jeong, A.M. Gabor, J.I Hanoka, R.L. Wallace: Proc. 28th IEEE PVSC, Anchorage, AK (2000) p. 91
66. M.A. Green, A. Wang, G.F. Zheng, Z. Zhang, S.R. Wenham, J. Zhao, Z. Shi, C.B. Honsberg: Proc. 12th EC PVSEC, Amsterdam (1994) p. 776
67. A. Goetzberger, J. Luther, G. Willeke: Proc. 12th Int. PVSEC Conf., Jeju, Korea (2001) p. 5
68. S. Narayanan, S.R. Wenham, M.A. Green: Proc. 4th PV SEC, Sidney (1989) p. 111
69. H. Nakoya, M. Nishida, Y. Takeda, S. Moriuchi, T. Tonegawa, T. Machida, T. Nunoi: 7th PV SEC, Nagoya (1993) p. 91
70. J. Coppye, J. Szlufcik, H. Elgamel, M. Ghannam, P. De Schepper, J. Nijs, R. Mertens: Proc. 22nd IEEE PVSC, Las Vegas (1991) p. 873
71. J. Szlucik, F. Duerinckx, J. Horzel, E. Van Kerschaver, S de Wolf, P. Choulart, H. Dekkers, J. Nijs: Proc. 12th Int. PVSEC Conf., Jeju, Korea (2001) p. 305
72. K. Shirasawa, H. Takahashi, Y. Inomata, K. Fukui, K. Okache, M. Takayama, H. Watanabe: Proc. 12th EC PV SEC, Amsterdam (1994) p. 757
73. S. Sivothythaman, B. Hartiti, J. Nijs, A. Barhdadi, M. Rodot, J.C. Muller, W. Laurey, D. Sarti, Proc. 12th EC PV SEC Amsterdam (1994) p. 47
74. S. Noël, H. Lautenschlager, J.C. Muller: Prog. Photovolt. Res. Appl. **9**, 41 (2001)