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Surface engineering in Cu(In,Ga)Se₂ solar cells

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Abstract

Surface modifications of 3-stage co-evaporated Cu(In,Ga)Se₂ (CIGS) thin films are investigated by finishing the evaporation with gallium-free (CuInSe₂, CIS) stages of various lengths. Secondary-ion mass spectrometry shows substantial interdiffusion of indium and gallium, smearing out the Ga/(Ga+In) profile so that the addition of a CIS layer merely lowers the gallium content at the surface. For the thinnest top layer, equivalent to 20 nm of pure CIS, X-ray photoelectron spectroscopy does not detect any compositional difference compared to the reference device. The modifications are evaluated electrical both by temperature-dependent characterisation of actual solar-cell devices and by modelling, using the latest version of SCAPS-1D. The best solar-cell device from this series is obtained for the 20 nm top layer, with an efficiency of 16.6 % after antireflective coating. However, we observe a trend of decreasing open-circuit voltage for increasingly thick top layers, and we do not find direct evidence that the lowering of the gallium concentration at the CIGS surface should generally be expected to improve the device performance. A simulated device with reduced bulk and interface defect levels achieves nearly 20 % efficiency, but the trends concerning the CIS top layer remain the same.

Introduction

One of the latest advances in efficiency of Cu(In,Ga)Se₂ (CIGS) solar-cell devices has been attributed to a thin layer of CuInSe₂ (CIS) on top of the CIGS absorber [repins2008], with the improvement over earlier record devices mainly being seen as an increase in the fill factor of the new record device. The CIS layer was fabricated as an additional stage in the commonly-used three-stage process by reducing the gallium evaporation rate to zero at the end of the evaporation. The nominal thickness of the layer, assuming no interdiffusion of indium and gallium, was estimated to be about 20 nm.

Our aim in this work was to use the CIS top layer concept to improve the efficiency of our solar-cell devices and to investigate the resulting gallium content at the surface of the absorber layer. An important part was the electrical modelling with SCAPS-1D [burgelman 2000, 2008] to model the impact of the modified gallium grading on the device performance. SCAPS-1D has recently been upgraded with features for modelling of graded profiles and the latest version 3.0 was used in this work.

There are several mechanisms that could conceivably account for the reported improvements yielded by the modified surface. The first two can be attributed to the interface between CIGS and CdS: The conduction-band offset (CBO) has an impact on the interface properties, and unless the Fermi level is pinned, the CBO will vary depending on the gallium content, from a positive value (spike) for pure CIS to a negative value (cliff) for CGS without indium. It has also been reported (for instance by Furlong et al. [furlong1998]) that the (112) planes of gallium-free CIS provide nearly ideal lattice matching with the (001) planes of hexagonal CdS, which consequently grow epitaxially on (112) CIS. CIGS films with gallium have smaller lattice constants and would thus be less beneficial for epitaxy. It is possible that a CIS surface therefore would be better than a CIGS surface from a lattice-match point of view.

Finally, the Ga/(Ga+In) ratio in CIGS films produced by the three-stage process normally exhibits a variation throughout the depth of the film. The position of the minimum of the Ga/(Ga+In) ratio in the CIGS is related to the length of stage three, but in our CIGS films, as in the films by NREL, it is situated between 0.4 μm and 0.6 μm into the film. This means that the bandgap will increase towards the surface, and there is a risk of forming an electron barrier that will cause voltage-dependent current collection. Lowering the conduction-band edge by means of the gallium concentration at the surface may reduce this barrier.

Experimental

Cu(In,Ga)Se₂ layers were prepared by co-evaporation of the elements onto Mo-coated soda-lime glass substrates. The deposition followed a three-stage process as described in an earlier publication [schleussner11pip-ra], with metals being evaporated from open boat sources and their rates controlled by means of mass-spectrometer feedback. The rates and temperatures are shown as functions of time in Figure 1. The net evaporation time was 52 minutes in all runs.

Besides a reference run without a top layer, samples were fabricated with CIS top layers of three different thicknesses, namely 20 nm, 50 nm and 100 nm. To obtain the top layers, the shutter was closed at a given time before the end of the CIGS deposition (35, 87 and 175 seconds, respectively). The Ga source was then turned off, and the In evaporation rate was increased to offset the missing Ga rate. When the new indium rate had stabilised for 90 seconds, the shutter was opened again and a CIS top layer was deposited onto the CIGS film.

Solar cells were finalised using the standard baseline process for devices at the Ångström laboratory, starting with a buffer layer of CBD-deposited CdS and a sputtered bilayer of ZnO and ZnO:Al. An electrode grid pattern consisting of Ni, Al and Ni was evaporated through a shadow mask. The solar cell areas were defined using a hard metal tip for scribing, and indium soldering was used to ensure good electrical contact for the probes to the Mo back-contact layer after removal of a strip of CIGS. A MgF₂ antireflective coating was added to all devices in order to avoid the interference effects resulting from the fairly flat surface of the CIGS.

The average CIGS film composition was measured by means of energy-dispersive X-ray fluorescence (XRF) using a *Spectro X-Lab 2000* spectrometer, with calibration against a standard of known composition. The resulting values of the average Ga/(Ga+In) and Cu/(Ga+In) ratios are listed in Table 1.

X-ray diffraction (XRD) was used to check the crystal quality of the film and to detect possible differences in film texture. The θ – 2θ scans were performed with Cu–K α radiation in a non-focusing geometry in a *Philips X'pert MRD*, equipped with an X-ray mirror and a parallel-plate collimator.

Cross-sections of samples were investigated by scanning electron microscopy (SEM) in a *Zeiss 1550* instrument at 5 kV acceleration voltage and 60k-fold magnification.

Elemental depth profiles of the two extreme cases, the reference without CIS and the sample with 100 nm of CIS, were analysed by secondary-ion mass spectrometry (SIMS), recorded in a *Cameca IMS 4f* instrument. The profiles were calibrated with the average compositions obtained by XRF, assuming linear relationships between elemental contents and SIMS count rates, that is to say, that the sputtering and ionisation yields in the SIMS analyses are independent from composition. The first 1–2 % of the data points were discarded as unreliable due to oxidation and other surface effects invalidating the above assumption.

The surface composition of all samples in terms of Ga/(Ga+In) was studied by ex-situ X-ray photoelectron spectroscopy (XPS) in a *Quantum 2000 Phi* instrument, using monochromatic Al-K α radiation for sample excitation. The analysis was performed on CIGS surfaces of samples where the window layer had been removed by chemical etching in dilute (3 %) hydrochloric acid for 2 min. After etching, the samples were rinsed in deionised water, blown dry in N₂ and transported to the XPS chamber as quickly as possible to minimise the effects of air exposure on the clean sample surfaces. In order to avoid interference with other surface peaks, the data were collected from the Ga-L₃M₄₅M₄₅ Auger peak and the In-3d^{5/2} peak at 421 eV and 445 eV, respectively. Based on absorption and cross-section data for the peaks at hand, the analysis-information depth is estimated to be a few nm. For quantitative analysis, the peak areas were determined, and sensitivity factors were calculated such that for the reference, XPS yielded the same surface Ga/(Ga+In) values as did SIMS.

Electrical parameters were determined by means of current–voltage (IV) and quantum-efficiency (QE) measurements. A halogen lamp equipped with cold mirrors was used for the IV measurements, while the QE measurements were performed under ambient light and were calibrated by means of reference cells with known spectral response. QE-measurements were performed at zero bias and at reverse bias down to –1 V. In order to correct for the spectral mismatch, the IV curves were shifted along the *J* axis to agree with the *J*_{SC} values calculated from the QE measurements, and the parameters were calculated after this operation.

In order to obtain information about the dominant recombination path and the approximate absorber doping, we also collected temperature-dependent current–voltage (IVT) and capacitance–voltage (CVT) characteristics from the reference and from the sample with the 100 nm thick top layer. The measurements were performed in temperature steps of 20 K from 160 K through 200 K and then in steps of 10 K up to 330 K. White illumination from a *Newport Oriel Sol2A* solar simulator was used for the light IVT measurements.

Simulations

Simulations were performed using version 3.0 of the software *SCAPS-1D* [burgelman2000, 2008]. As a starting point for establishing a model of the reference solar cell in this work, the baseline parameter set defined in [pettersson2010] was used. Some important electrical parameters can be found in Table 2. As the CIGS absorber layers in the present work were prepared in another system than the ones on which the original model is based, some important adjustments to the model were needed, which we describe in the following.

The most prominent difference from the baseline model lies in the compositional grading profiles of the group-III elements. As visualised in Figure 4, they are modelled as parabolic functions for the bulk of the absorber layer and as linear functions for the top layer. The linear regimes are defined such that they extend over twice the nominal thickness of the respective top layer, in accordance with the SIMS profiles in Figure 3, and reach the surface Ga/(Ga+In) values that were determined by XPS. In addition to emulating the processed devices, we also simulate cells with even thicker (200 nm and 300 nm) deposited CIS-layers. For these modelled cells, the slope of the linear grading is assumed to be the same as for the 100 nm sample. The reference model does not contain the linearly graded top layer.

From the Ga content $x = \text{Ga}/(\text{Ga}+\text{In})$, the bandgap is calculated using the following expression [alonso2002, shafarman2003]:

$$E_g = 1.01 + 0.626 x - 0.167 x (1 - x).$$

It is important to note that the model was adjusted with respect to a good match to the measured data in the *reference* case, and that the other models were derived from the reference model solely by changing the bandgap profile at the top.

In the current work, no surface defect layer is included in the model. The Fermi level is not pinned, and instead we introduce positive conduction-band offsets (CBO) between the absorber and buffer layers. Since the surface bandgap of the Cu(In,Ga)Se_2 films decreases with decreasing surface $\text{Ga}/(\text{Ga}+\text{In})$ ratio, the CBO values used in the model are adjusted to increase accordingly, as shown in Table 3. Furthermore, the doping in the CIGS layer is lowered relative to the baseline model, and also the compensating donor density and the interface defect density are slightly decreased. Finally, while in our baseline model the defects at the CIGS/CdS interface were considered to be neutral, they are now defined as being of the donor type. All deep bulk defects are single level and positioned at the intrinsic level close to midgap of the corresponding layer.

A front surface reflectivity of 6% across the whole wavelength range is introduced in order to reproduce our QE measurements.

Results and Discussion

Material:

As seen from the deposition profile in Figure 1, the change in surface composition from CIGS with approximately 50 % gallium to 0 % gallium is abrupt and may therefore induce a strain in the crystal lattice. SEM analyses were performed in order to investigate if the abrupt change of $\text{Ga}/(\text{Ga}+\text{In})$ ratio results in any formation of new grains or a visible top layer. The SEM images in Figure 2 show the interface regions of the reference sample and of the sample with a 100 nm thick indium-rich top layer. While the CdS and intrinsic ZnO layers can be distinguished, there is no obvious interface between the CIGS layer and the top layer, even in this case with the thickest top layer. At a wider scale, both samples furthermore display large grains that extend all the way from the back to the front of the CIGS film. We conclude that the Ga gradient exists *within* the grains and that the lattice strain is relaxed by the interdiffusion of indium and gallium during the deposition of the top layer, as will be discussed further down.

X-ray diffraction θ – 2θ scans were carried out in order to detect possible changes between the different Cu(In,Ga)Se_2 runs which could result in changes in film texture. All Cu(In,Ga)Se_2 deposition runs were made within two days of each other and used process sequences that were identical to each other except for the top layers. This timing should ensure very similar growth conditions, and our XRD results confirm that the addition of the In-rich top layers has not resulted in any noticeable changes in the diffractograms, with the CIGS peaks being similar for all samples in the study. Thus, the only measurable difference between the samples lies in the surface modification of the Cu(In,Ga)Se_2 grains. This modification is not expected to be detectable by XRD since some peak broadening will already result from the $\text{Ga}/(\text{Ga}+\text{In})$ in-depth variation inherent to the three-stage process.

In Figure 3, the SIMS depth profiles of the gallium content of the reference as well as the sample with the thickest (100 nm) In-rich top layer show that the films have notch-like $\text{Ga}/(\text{Ga}+\text{In})$ variations, as is usual in films resulting from the three-stage process. The distance between the bottom of the notch and the surface is about 600 nm for the reference sample and about 500 nm for the sample with the top layer, and the $\text{Ga}/(\text{Ga}+\text{In})$ ratio in the bottom of the notch is 0.35 for both samples. The difference in the distance between the surface and the notch can be explained by the small variation in total thickness between the samples. As is seen in Figure 3, the $\text{Ga}/(\text{Ga}+\text{In})$ ratio at the front surface differs between the reference and the 100 nm top-layer sample, with a significantly lower gallium content for the sample with the top layer. However, it is clear that

significant interdiffusion of indium and gallium has smeared out the profile and that the surface composition of the film is far from gallium-free. The SIMS analyses suggest that the surface compositions correspond to Ga/(Ga+In) ratios of 0.44 and 0.32 for the reference and the 100 nm top layer samples, respectively. The formation of the notch-like in-depth variation of Ga/(Ga+In) is discussed more in detail in another study [schleussner11pip-ra], where we observe that copper migration in one direction leads to preferential indium migration (as compared to gallium migration) in the opposite direction. As the final part in the three stage process is copper-free, we believe that out-diffusion of copper to the growth front may have enhanced indium–gallium intermixing between the bulk and the top layer.

XPS is a more surface-sensitive method than SIMS and more suitable for analysing the outermost surface of the samples. In all cases, also for the thickest top layer, we found gallium at the surface by XPS analysis. The difference between the 20 nm sample and the references is small in the XPS data and may fall within the error margin of the measurements. Using the SIMS quantification for the reference sample for determining the sensitivity factors as described in the experimental section, we find that the Ga/(Ga+In) value decreases down to 0.31 for the sample with the 100 nm top layer, which is in good agreement with the SIMS results for the same sample. The results of the XPS surface quantification are shown together with the XRF values in Table 1.

We also found that while the C1s peak had a similar magnitude for all of the measured samples, the O1s peak indicated higher oxygen contents at the surface for the two reference samples than for the samples with a CIS top layer despite identical times between etching of CdS and loading into the XPS vacuum system. We explain this difference in oxygen content with a more rapid oxidation of gallium relative to indium at the Cu(In,Ga)Se₂ surface [guillemoles2000].

Electrical Simulations and Measurements:

Table 3 shows the results of one-dimensional simulations obtained using SCAPS-1D. The simulations use simplifications of the in-depth Ga/(Ga+In) variations, as is clear from a comparison of Figures 3 and 4, but include what we believe are the main features, namely the distance between the notch and the surface, and for the samples with In-rich top layers, the largely linear slope near the surface. In the simulations two more extreme samples with top-layer thicknesses of 200 and 300 nm are also included, going beyond the thicknesses used in the experimental series.

Comparing the measurement and simulation results in Tables 1 and 3, one finds that the V_{OC} , FF and J_{SC} values resulting from the modelling of the reference case agree reasonably well with measurements of the reference sample. The same judgement can be reached looking at the top left graph in Figure 5, which shows light IV data of the real devices and of their equivalent models.

IV curves measured at different temperatures were fitted to the one-diode model using the procedure outlined by Hegedus and Shafarman [hegedus2004] in order to extract the ideality factor and the saturation current density. These fits were acceptable for temperatures between 250 K and 330 K. Two established techniques to evaluate the activation energy of the recombination path limiting V_{OC} were used [rau2001]. One method is to plot the product of the ideality factor and the saturation current density against $1/kT$, yielding the activation energy as the slope of a linear fit to the high-temperature data. This was done both for measurements in the dark and under illumination. The second method is to extrapolate V_{OC} at high temperatures linearly to 0 K, which yields the activation energy at that temperature. According to both evaluation methods the activation energies were close to the bandgap for both the reference and the sample with the 100 nm top layer. This fact suggests that the dominant recombination path lies in the bulk, and this is also the case in our SCAPS models of these samples. Also, our choice of a relatively low absorber doping is supported by CVT measurements, which at low temperatures around 150 K imply an apparent doping in the range of 10^{15} cm^{-3} to 10^{16} cm^{-3} .

The material parameters obtained from the reference model were used for the models with the top layers. Only the CIGS compositional profiles varied between the models. These profiles are shown in Figure 4.

Figure 6 shows the QE data corresponding to the IV curves in Figure 5. There were no significant differences between the QE curves of the various devices, all showing similar features. At wavelengths around 900 nm the QE-curves of the measured cells are slightly lower than the simulated curves. One explanation of this could be free carrier reflection and absorption in the highly doped ZnO window layer. These effects are not accounted for in the simulations. The bandgaps as obtained from the QE measurements, which correspond to the minimum bandgaps of the devices, were between 1.18 and 1.19 eV in all cases. These values are equivalent with a Ga/(Ga+In) minimum ratio in the notch of 0.33–0.35, in good agreement with the SIMS measurements. No significant voltage-dependent current collection was found by measuring QE at reverse bias.

Both simulations and measurements show a trend of V_{OC} decreasing with increasing thickness of the In-rich top layer. According to model, this effect can be attributed to increased interface recombination. In the model, the voltage drop becomes more pronounced with thicker top layers of 200 nm and 300 nm, due to the low band gap at the CIGS/CdS interface in these cells. In these extreme cases we also get an increase of J_{SC} from the modelling since the surface band gap gets lower than the minimum band gap in the notch. The trends remains largely the same both when all the conduction-band offsets are shifted to higher or lower values, and when the Fermi level is pinned by donor defects close to the conduction band at the CIGS/CdS interface.

Comparison with NREL devices:

Comparing our Cu(In,Ga)Se₂ material with the material presented by Repins et al. [repins2008], we find that CIGS layers with a smaller total Ga/(Ga+In) ratio were used for the NREL's 20.0 % device than in this paper. From the Auger depth analyses in the NREL paper we can calculate a surface bandgap of about 1.19 eV for the NREL reference and 1.15 eV for the NREL sample with a indium-enhanced top layer, while a minimum bandgap in the order of 1.13 eV (equivalent with a Ga/(Ga+In) ratio of 0.24) can be extracted from their QE data. The NREL profiles were thus shifted by about 0.05 eV to lower bandgap values relative to our devices. The notch position was stated to be 0.5 μm from the surface, similar to the results shown here.

With this 50 meV bandgap difference in mind, it is clear that the devices presented in this work suffer from both current and voltage losses in addition to fill-factor losses compared to the 20.0 % device. In principle our devices should gain up to 50 mV in open-circuit voltage, since the bandgap is larger by that amount, but our best device was just similar to the NREL device in terms of V_{OC} . A noticeable difference is found in the QE data of the longer wavelengths, which in our devices drop earlier and more gradually than in the NREL devices. It seems likely that these losses are related to shorter diffusion lengths, possibly along with optical losses in the window as mentioned above.

By reducing the number of defects in the bulk of the Cu(In,Ga)Se₂ layers and at the Cu(In,Ga)Se₂/CdS interface, our reference-device model can be improved to match the NREL record device. Reducing the number of compensating donors from $9 \times 10^{13} \text{ cm}^{-3}$ to $5 \times 10^{12} \text{ cm}^{-3}$ increases the diffusion length from 0.86 μm to 3.6 μm , which results in an increase particularly of V_{OC} and FF , yielding an efficiency of 19.2 %. Changing the number of defects at the interface from $9 \times 10^{12} \text{ cm}^{-2}$ to $1 \times 10^{11} \text{ cm}^{-2}$ gives rise to a slight additional improvement of all the parameters by reducing the interface recombination. To reproduce the maximum level of 96 % in the QE curve from NREL, a lower front-contact reflection of 4 % is also assumed. In total these modifications yield an efficiency of 19.9 %, on par with the NREL device. The complete set of simulated solar-cell parameters is given in Table 3. When indium-rich top layers (as presented in Figure 4) up to a

nominal thickness of 100 nm are introduced to this improved model device, the solar-cell parameters are virtually not influenced at all. With thicknesses above that value a similar trend with increasing J_{SC} as seen before appears, while the decrease in V_{OC} is very small in this case due to the low interface defect density.

Conclusion

We have found that changing the surface by ending the deposition with a gallium-free stage does not improve significantly the quality of the Cu(In,Ga)Se₂/CdS interface of our devices. A high degree of intermixing of gallium and indium will tend to smear out the profile for the thinnest top layer, so that the resulting surface profile is only slightly changed relative to references without this final stage. Furthermore, while modelling suggests that layers of 200 and 300 nm thickness will decrease the voltage and increase the current, we observe both experimentally and by modelling that for a nominal thickness of up to 100 nm, the indium-rich top layer does not have any significant impact on the device performance. Specifically, neither does the reduction of the surface electron barrier prove beneficial, nor can we find a detrimental effect of the lower bandgap at the surface or of the change in the conduction-band offset. However, we cannot rule out the possibility that an indium-enhanced surface might affect the Cu(In,Ga)Se₂/CdS interface beneficially, insofar as lower gallium content may lead to improved lattice matching and thereby a less defected interface. We also observed less surface oxidation with a higher concentration of indium at the surface.

We have also used modelling to investigate the shortcomings of our three-stage CIGS devices compared to record devices made by NREL. The modelling suggests that the diffusion length of our devices is the primary factor limiting the performance, although improvements of the Cu(In,Ga)Se₂/CdS interface may be able to contribute some additional performance gain.

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Table 1

Data for the samples with different thicknesses of CIS top layers. IV data are for best devices and calibrated with QE.

	Reference	20 nm	50 nm	100 nm
$\langle \text{Ga}/(\text{Ga}+\text{In}) \rangle$ (XRF)	0.42	0.43	0.43	0.41
$\text{Ga}/(\text{Ga}+\text{In}) _{\text{surface}}$ (XPS)	0.44	0.43	0.35	0.31
$\text{Cu}/(\text{Ga}+\text{In})$	0.79	0.83	0.85	0.79
CIGS thickness [μm]	1.85	1.75	1.68	1.60
V_{OC} [mV]	685	687	674	679
J_{SC} [mA/cm ²]	31.4	31.5	32.0	31.8
FF [%]	75.2	76.8	75.4	75.0
η [%]	16.2	16.6	16.3	16.2

Table 2

Some important electrical parameter used in the solar cell simulations.

Layers properties: thickness d , bandgap E_g , conduction- and valence-band carrier densities N_C and N_V , thermal velocities v_{th} , mobilities μ , doping concentrations N_B .

Defects (single-level traps): concentrations N_t , cross-sections σ , energy positions E_t . Indices e and p stand for electrons and holes; (a) and (d) denote *shallow* acceptor and donor defects, while (A) and (D) denote *deep* acceptor and donor defects. Energetic positions of deep defects are indicated.

Layer Properties			Bulk Defect Properties		
	CIGS	CdS		CIGS	CdS
d [μm]	1.85	0.07	N_t [cm ⁻³]	9×10^{13} (D)	5×10^{16} (A)
E_g [eV]	Graded	2.4	σ_e [cm ²]	10^{-13}	10^{-15}
N_C [cm ⁻³]	6.8×10^{17}	1.3×10^{18}	σ_p [cm ²]	10^{-15}	5×10^{-13}
N_V [cm ⁻³]	1.5×10^{19}	9.1×10^{18}	E_t	Midgap	Midgap
$v_{\text{th,e}}$ [cm/s]	3.9×10^7	3.1×10^7	Interface Properties		
$v_{\text{th,p}}$ [cm/s]	1.4×10^7	1.6×10^7	N_t [cm ⁻²]	9×10^{12} (D)	
μ_e [cm ² /Vs]	100	72	σ_e [cm ²]	10^{-13}	
μ_p [cm ² /Vs]	12.5	20	σ_p [cm ²]	10^{-15}	
N_B [cm ⁻³]	4×10^{15} (a)	5×10^{17} (d)	E_t	CIGS midgap	

Table 3

CBO values used in the particular simulations, and simulated IV parameters. The listed CIS-film thicknesses correspond to the *nominal* thicknesses of the experiments. As seen in Figure 4, the actually simulated thicknesses were twice the stated value.

‘Improved’ refers to the simulated device with reduced bulk and interface defect densities.

	Ref	20 nm	50 nm	100 nm	200 nm	300 nm	Improved
CBO [eV]	0.10	0.10	0.15	0.17	0.24	0.28	0.10
V_{OC} [mV]	688	688	688	687	673	643	732
J_{SC} [mA/cm ²]	31.7	31.7	31.7	31.7	32.7	34.1	33.2
FF [%]	77.4	77.4	77.2	77.2	77.7	77.1	82.0
η [%]	16.9	16.9	16.8	16.9	17.1	17.0	19.9

Figure 1

The evaporation rates, substrate temperature, and selenium-source temperature used for one of the experiments in this paper, plotted versus process time. Shaded areas indicate times for stabilisation of boat sources with the shutter closed. The final stage in this process was 175 s long and produced a nominally 100 nm thick In-rich layer.

Figure 2

SEM cross-sections of complete solar-cell devices including AR coating. The CIGS layers are without a CIS top layer (top) and with a 100 nm thick top layer (bottom).

Figure 3

SIMS Ga/(Ga+In) profiles of the reference (solid black line) and of the sample with a CuInSe₂ top layer of nominally 100 nm thickness (dotted green line). The surface of the film is to the right. The minimum Ga/(Ga+In) ratio in the notch is 0.6 and 0.5 μm from the surface for the reference and the 100 nm top layer film, respectively. (Colour online.)

Figure 4

Cu(In,Ga)Se₂ composition profiles used in simulations of the different cell structures. The profiles consist of parabolas which are truncated and followed by straight lines which extend to the surface. The total thickness for all modelled layers is 1.8 μm . The position of the minimum Ga/(Ga+In) ratio is 0.5 μm from the surface of the film. (Colour online.)

Figure 5
Light IV measurements on typical devices from the series (symbols) and corresponding simulated curves (solid lines).

Figure 6
QE measurements on typical devices from the series (symbols) and corresponding simulated curves (solid lines).