Metallization of silicon solar cells using electroplating is of interest recently as a cheaper replacement of silver screen printing. Light assisted electroplating of Nickel-Cobalt (NiCo) alloys were studied on crystalline silicon (Si) solar cells in this paper. The effect of illumination was investigated in conjunction with the light absorption properties of the plating solution, emission spectrum of the light source as well as the plating current. A solution dependent threshold light intensity was observed, beyond which the solar cells were under reverse bias during electroplating. The detailed compositional depth profiles of the silicide layer were studied along the silicidation reaction using microscopic energy dispersive X-ray spectrum (EDS). A significant improvement in the performance degradation at high temperature thermal stress was observed for solar cells metalized with the ternary silicide.

The cost of solar energy compared with fossil fuel is one of the key determining parameters in its adoption by the market. Therefore, ways of saving the manufacturing cost of silicon solar cells are of great interest. The current manufacturing process involves a screen-printing process to metalize the front side with silver paste. While the screen-printing process has been integrated into the process flow with standard tooling, it typically suffers from conversion loss due to the low conductivity of the paste and low line height-width aspect ratio. In addition, the high price of silver potentially limits the cost reduction of the solar cell.

Replacing the screen-printed Ag with electroplated high aspect ratio Cu grids has been proposed decades ago. This method has recently gained more attention due to the high price of Ag and the easy integration of Cu electroplating with laser patterning. The selective emitter formed by the in-situ doping during laser patterning further eases the process control for metallization and has the potential to improve the conversion efficiency of the solar cell.

Different Cu metallization schemes involving electroplating have been reported. Most of these schemes involve the formation of Ni silicide either from electroless or electrolytic deposited Ni. The electrolytic deposition process in these reports typically involves light, so called light induced or light assisted electrodeposition. The formation of Ni silicide lowers the contact resistance between the metal and Si and thus improves the solar cell performance.

Ni and Co silicidation has been extensively studied for the application in complementary metal oxide semiconductor (CMOS) contacts. The Ni silicidation starts at as low as 300 °C for metal rich phases, followed by monosilicide phases around 400 °C and silicon rich phases above 800 °C. Co silicidation follows a similar phase transition with the light illumination intensity and the energy spectrum. The grid pattern was created on SiN using photoresist and UV lithography. Cross-sectional composition profiles were investigated to understand the silicide formation from the electrodeposited thick alloy films. The effect of the ternary silicide on the solar cell's performance was also compared against the pure Ni silicide.

**Experimental**

Si solar cells (4 cm × 4 cm) were fabricated with p-type CMOS grade Si wafers adopting some of the CMOS fabrication processes. In brief, the fabrication process included POCl_3 diffusion, phosphorus silicon glass (PSG) removal, PECVD silicon nitride (SiN) deposition as the passivation and antireflection layer, PVD aluminium deposition followed by annealing to form the back surface field (BSF). The front grid pattern was created on SiN using photore sist and UV lithography. The front grid pattern comprises a 2 mm wide busbar and 38 fingers, 50 μm wide and 1 mm apart, with a total open area of about 1.52 cm². The photore sist pattern was transferred into SiN using hydrofluoric acid (HF) etching and then removed using acetone. Immediately before plating, the front surface of the cells were cleaned with 1:100 HF for 30 seconds to remove the native silicon oxide in the opened grid pattern.

Light assisted electroplating was carried out in a two-electrode setup in a stationary plating cell. A Princeton Research PAR 273 potentiostat was used as the power supply. The plating cell comprises a metal plate at the bottom as the contact to working electrode. The plating cell walls are sealed with the bottom plate using a gasket set. The solar cell was placed on the bottom metal plate before assembling the gasket, facing upward and in contact with the metal plate through the aluminum BSF. A flat Ni mesh placed above in parallel with the solar cell was used as the counter electrode. A halogen lamp with an adjustable power supply was used above the plating setup. The light intensity is calibrated with a light meter. A light intensity of 17000 lux was used during plating except for the study on the light intensity effect. This light intensity was measured in an empty plating setup but...
at the same height as the top surface of plating solution. The open area of the Ni mesh was about 90% and the plating solution height in the plating cell was 1 cm. Therefore the number of photons that reached the solar cell was kept at a constant.

The NiCo solutions contain various concentrations of NiSO$_4$ and CoSO$_4$, 0.2 M Na$_2$SO$_4$, 0.25 M H$_3$BO$_3$, and 1 g/L sodium dodecyl sulfate (SDS). A previous study on Ni plating showed a dependence of deposit uniformity on the solution concentration and an optimal uniformity with 0.2 M NiSO$_4$.

The total metal concentration, NiSO$_4$ and CoSO$_4$, is therefore kept constant at 0.2 M. As shown in Table I, the ratio between the two metal species varies for different solutions in this study. A rapid thermal annealing (RTA) furnace with a thermal couple embedded in the carrier was used to anneal the NiCo plated solar cells to form silicide. The RTA was carried out at a variety of temperature for different periods of time. After the annealing, the excessive un-reacted metal was removed with piranha solution (H$_2$SO$_4$ : H$_2$O$_2$ for different periods of time. After the annealing, the excessive un-reacted metal was removed with piranha solution (H$_2$SO$_4$ : H$_2$O$_2$ = 1:10) at about 65 °C for 10 minutes.

Further metallization of the solar cells were processed with a 1:100 HF cleaning for 30 seconds immediately followed by Ni and Cu plating. Both plating processes were carried out with the assistance of light. Ni was plated with the same chemistry as the NiCo solution except that the metal in the solution is pure Ni at 0.2 M. A typical damascene plating solution for interconnect fabrication was used to plate the Cu grid. While studies with other chemistries showed the damascene plating solution was not necessary, the studies presented in this paper used a same damascene Cu chemistry.

The plated and annealed samples were inspected with a Zeiss Leo scanning electron microscope (SEM). For cross sectional SEM analysis, the samples were cleaved, coated with a carbon-based ink, and then polished with a FEI focused ion beam (FIB) after a portion of the cleaved edge was protected with Pt (in the FIB tool). The composition depth profiles of the silicide layers were analyzed through cross-sectional lamellas. The lamellas were prepared using FIB and manually lifted out onto a gold grid with membrane support. The final thickness of the lamella estimated by top down FIB image was up to 200 nm. However, the lamellas were prepared in such a way that a lower cell potential. Regardless of the plating solution, the electron cell potential measured thereof included the potentials for cathode reaction, solution resistance, and anode reaction. In this particular set of experiments, the solution conductivity and the cathodic potentials are believed to be similar among the solutions. The oxidation reaction of Co$^{3+}$ (to Co$^{4+}$) was believed to occur on the anode when significant amount of Co$^{2+}$ was present in the solution and contributed to a lower cell potential. Regardless of the plating solution, the electron...
generation rate is so high at high light intensity that only a small portion of the electrons needs to flow to the deposition area. As the light intensity decreases, fewer electrons are generated in the n-type emitter and more significant portion of the photo-generated electrons is used for electrodeposition. When the light intensity falls below a critical value, the photo-generated electrons are not produced fast enough to support the plating current anymore. In this case, not only almost all the photo-generated electrons will be forced to flow to the busbar and fingers to support the electrodeposition, but also the external power forces an additional reverse current to flow across the p-n diode. In other words, the diode is under reverse bias and a much higher negative plating potential is observed in this case.

As the light intensity further decreases and falls far below the critical value, even fewer carriers are generated, most of which recombine around the p-n junction. The plating is completely enabled by the reverse current. A constant potential of about –26 V is observed and further dimming the light does not increase the potential anymore. Because the reverse bias only occurs when the photo-generated carriers are insufficient to carry the current for plating, the critical light intensity value depends on the plating current as well as the light absorption by the solution.

Among all the mixture solutions tested, 0.05 M NiSO4 + 0.15 M CoSO4 (25% Ni, 75% Co) showed the lowest critical light intensity, 2370 lux. As the Ni/Co ratio increases, this critical light intensity value increases as well, suggesting a stronger absorption by the solution. Interestingly, this critical value also increases as well when the Ni/Co ratio further decreases as shown in Figure 1b. But the increase is much less pronounced as compared the effect of increase of Ni/Co ratio. In detail, the critical light intensity were 2990, 3560, 4300 and 5370 lux for the 0.2 M nickel cobalt sulfate mixture solutions with 50%, 75%, 90% and 100% Ni, respectively. On the other hand, this value only increased to 2600 and 2950 lux with 90% and 100% Co, respectively.

For a photon to be absorbed and generate carriers, its energy has to be above the Si bandgap, 1.1 eV, or its wavelength shorter than 1100 nm. In addition, due to the high reflectance of UV light by Si solar cell, photons below 400 nm wavelength produce only a few carriers. However, the quantum efficiency (the ratio between the number of the photo-generated carriers and the number of photons of a certain wavelength) of the silicon solar cell stays roughly constant for photons between 400 nm and 1100 nm.28 In this study, a tungsten halogen lamp was used as the light source, which has an emission peak at about 900 nm. The majority of the emission energy is between 700 nm to 1200 nm. This emission decreases rapidly from 900 nm to 300 nm.29 Considering the quantum efficiency of silicon solar cell, the majority of the photo-generated carriers were generated by the long wavelength photons from the halogen lamp, 700 nm to 1100 nm.

Figure 1c shows the UV-vis absorption spectra of four different solutions. NiSO4 solution showed two absorption regions, a strong absorption peak between 350 to 450 nm and a lower but much broader absorption peak between 600 to 800 nm. On the other hand, CoSO4 solution showed a single strong absorption between 430 and 570 nm. It is clear that the second absorption peak of Ni solution between 600 to 800 nm would have the most pronounced effect for the light emitted by the halogen lamp. The lower the NiSO4 concentration, the lower the light absorption, the higher the intensity of the light that reaches the solar cell, and therefore the lower the critical light intensity required for the plating current. When the NiSO4 concentration is below certain value, the absorption by Ni becomes negligible and a weak effect of CoSO4 concentration was observed.

Because the emission spectrum of the tungsten halogen lamp peaks at a long wavelength of 900 nm, most of the energy emitted is dissipated as infrared heat wave. This not only wastes energy for the lamp, but also heats the solution and solar cell during experiment and changes the electrodeposition kinetics. From the absorption spectra of Ni and Co solutions in Figure 1c, it is clear that a monochromatic light source at 500 nm and 600 nm would be optimal for electrodepositing pure Ni and NiCo alloy, respectively. In this study, the light intensity was kept at constant of 17000 lux to ensure the forward bias of the plating of solar cells.

Silicidation.—The silicidation of the electrodeposited NiCo alloys were studied using a blanket n-type emitter formed on p-Si. Electrodeposition was conducted at a current density of –16 mA/cm2 for 50 seconds, equivalent to 280 nm Ni at 100% efficiency. The composition of the deposited film was analyzed with EDS on top down SEM mode. The X-ray signals were averaged across a 11 um × 8 um area and quantified for Ni and Co. Table I shows the compositions of alloy films deposited from different solutions. While the Co content in the film increases with the Co content in the solution, the deposit is much Ni-richer than the solution. This so-called anomalous codeposition has been well known to inhibit the deposition rate of the more noble elements and, in some cases, enhance the deposition rate of the less noble element.30 In the extreme case, where the Ni accounts for 25% of the total metal in the solution, the Ni content in the film was only 3%. Therefore, solutions with further lower Ni portion were expected to result in Co films with trace amount of Ni and therefore were not further studied.

The deposited alloy films were annealed in N2 ambient. A variety of annealing conditions from 300 to 500°C and from 1 to 30 minutes were used. The study was focused on the general trend of the different annealing conditions for NiCo alloys with different compositions. In addition, a silicide sheet resistance between 10 to 20 ohm/sq was often found optimal for the solar cell performance.25 Therefore some of the annealing time was not applied for all the temperatures used in this study. Sheet resistance of the silicide was presented in Table I as well. The excessive untreated metal was removed by dilute piranha etching before the measurement. Significant silicidation was observed for pure Ni only after 2-minute annealing at 300°C. On the other hand, no silicide was formed from pure Co for all the annealing conditions used. When a small amount of Ni was present in the Co film, silicide formation became easier than for pure Co. While no silicidation occurred for Ni49Co51 and Ni36Co64 during annealing at 450°C for 30 min, certain degree of silicidation was observed after annealing at 500°C for 30 min. Increase of the Ni content in the alloy further facilitated the silicidation. Ni49Co51 and Ni36Co64 both showed silicide formation at 400°C for only 1 min. Annealing at higher temperature or for longer time results in a lower sheet resistance or a thicker silicide film.

The silicidation of vacuum deposited Co thin film (8 nm) and the alloying effect of Ni was previously studied using in-situ X-ray diffraction techniques along with the annealing.24 Co rich silicide phase and monosilicide phase start to form at 430 and 490°C, respectively, for

<table>
<thead>
<tr>
<th>Solution</th>
<th>Deposit</th>
<th>Silicide Rs (ohm / sq)</th>
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</thead>
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<tr>
<td>pure Ni</td>
<td>Ni</td>
<td>9.1 1.3 0.5 0.3 0.2 0.2</td>
</tr>
<tr>
<td>Ni:Co = 9:1</td>
<td>Ni49Co51</td>
<td>84 21 4.1 1.1 0.6 0.4</td>
</tr>
<tr>
<td>Ni:Co = 3:1</td>
<td>Ni36Co64</td>
<td>92 77 41 16 2.2 1.3</td>
</tr>
<tr>
<td>Ni:Co = 1:1</td>
<td>Ni36Co54</td>
<td>96 97 92 94 101 95 20</td>
</tr>
<tr>
<td>Ni:Co = 1:3</td>
<td>Ni36Co97</td>
<td>88 90 97 88 93 91 50</td>
</tr>
<tr>
<td>pure Co</td>
<td>Co</td>
<td>91 92 101 100 103 NA 100</td>
</tr>
</tbody>
</table>
bulk sample as discussed in Figure 2b. The lamellas were separately prepared using FIB with a final thickness of less than 100 nm at the silicide layer. While an EDS detector coupled with a transmission electron microscope (TEM) typically allows a resolution of 1 nm, this high resolution is enabled by the much higher electron energy in TEM. The SEM was operated at 20 kV in this study, much lower than TEM. Therefore the spatial resolution was expected to be poorer than TEM-EDS analysis. However, because of the limited thickness of the lamellas, the electron scattering was much limited compared with bulk sample. A quick estimation of the electron scattering was obtained using a Monte Carlo simulator software.35,36 For a 100 nm thick lamella, the trajectories of electrons at 20 kV scattered across $+/−25$ nm for NiSi and $+/−15$ nm for Si. As compared with the $+/−350$ nm and $+/−750$ nm for bulk NiSi and bulk Si, the interaction volume was much limited and the spatial resolution was improved using lamella.

Figures 3a to 3c show the X-ray signal profiles across the thickness of three silicide layers formed at 400° C for 1, 5, and 30 minutes, respectively. Figures 3d to 3f show the corresponding atomic composition profiles across the thickness of these three silicide layers. Two dashed lines were used in the figures to label the width of the Ni peak at its half maximum. The regions between two dashed lines, called full width at half maximum (FWHM) of the Ni peaks, were found 58, 135 and 340 nm for these three annealing time, consistent with the silicide thickness obtained from SEM analysis in Figure 2. Due to the removal of excessive metals before the analysis, the absence of materials beyond the silicide layer resulted in insufficient X-ray signals and poor quantification above the dashed line on the right in the composition profile.

As discussed in the earlier paragraph, the electron scattering may be up to 50 nm in NiSi and 30 nm in Si for a 100 nm thick lamella. While the majority of the electron trajectories are located in a much narrower range, each X-ray signal or composition value in the figures is indeed an averaged value among neighboring position. The dashed lines in the Figure 3 were drawn to label some big compositional changes. Detailed change below the electron scattering range, if any, will be lost. For example, the FWHM of the Ni peak in Figures 3a, 58 nm, was comparable with the electron scattering range. Therefore, the peak Ni content was a significantly underestimated average value. This could be also true for thicker silicide layers in Figures 3b and 3c due to possible finer features. However, the comparisons between different samples or between different elements in the same sample were believed to be still valid to show an general trend.

The comparison between Ni and Co profiles clearly showed a greater silicidation depth for Ni than Co. After 1 min annealing at 400° C, the Ni and Co profiles were about the same, resulting in an average silicide composition of Ni$_{4}$Co$_{3}$Si$_{2}$. However, Ni was 15 nm deeper into the substrate than Co. This depth difference was so small that the composition of the Ni silicide could not be determined clearly. This depth difference increased to 65 nm after 5 minutes of annealing. While the top 70 nm was a ternary silicide alloy, Ni$_{4}$Co$_{3}$Si$_{20}$ in average, the bottom 65 nm layer had an average composition of Ni$_{30}$Si$_{70}$. However, the double layered structure was not obvious in the cross sectional SEM in Figure 2b. After annealing for a much longer time, 30 min, this layered structure was more magnified and was clearly observed in the SEM in Figure 3c. The bottom layer kept the same average composition of Ni$_{30}$Si$_{70}$, but grew thicker, to 160 nm. At the same time, the ternary silicide alloy also grew thicker to 180 nm. A finer sub-layered structure with two layers with different NiCo ratios might be present in this top ternary layer. Both layers have an average stoichiometry of Me$_{2}$Si where Me stands for Ni or Co. In addition, the ternary silicide alloy observed in the 5-min annealed silicide also matched the Me$_{2}$Si stoichiometry. As discussed earlier, the spatial resolution of the analysis was compromised due to the electron scattering across the thickness of lamella. However, the relative trends in the composition and thickness are believed to be valid for the silicides formed at different conditions.

The silicidation of CoNi alloys has been previously studied for applications in CMOS contacts.24,27,34,44 Although the interest was
Figure 3. (a-c) X-ray counts and (d-f) atomic compositions for (diamond) Si, (square) Ni, and (triangle) Co acquired from cross sectional lamellas of silicide formed by annealing Ni$_{49}$Co$_{51}$ alloy film electroplated on Si at 400°C for (a, d) 1, (b, e) 5 and (c, f) 30 min.

Later shifted to the nickel monosilicide for that application, the ternary silicide was then of interest as alternative to CoSi$_2$ because the addition of Ni improves the nucleation of Co disilicide and reduces the silicide roughness and Si consumption. Therefore, not only all those studies$^{21,24,37-44}$ were based on vacuum deposited extremely thin films but also most of them focused on the disilicide formation from Co rich alloys at high temperatures above 500°C. One article,$^{41}$ however, reported the silicidation of Ni$_{50}$Co$_{50}$ films. While a much thinner film, 10 nm, and a much higher temperature, 700 to 1100°C, were used, the Ni concentration peak was found slightly deeper than Co, consistent with the observation in Figure 3. However, because an extremely thin metal film was used in their study, the Ni and Co peaks broadened and the difference in their depth disappeared as the temperature increased.

In this study, electrodeposited NiCo layers of hundreds of nanometers were annealed. A large portion of the metal films was unreacted and acted as metal reservoir for the silicidation. When extremely thin alloy films are annealed, the films can be consumed completely and converted into silicide layer with uniform composition at high temperature. In addition, before that happens, as the silicidation continues the composition of remaining alloy film changes due to the different silicidation rates of different metals. That will also change the silicidation from the remaining metal layer. In this study, while the alloy composition in the vicinity of the top surface the silicide can also
deviate from the bulk alloy composition due to the different silicidation rates, the excessive thick alloy film is believed to mitigate this change.

Considering the growth sequence of the three profiles in Figure 3, the depth of Ni silicide, i.e., 58, 135 and 340 after 1, 5, and 30 minute annealing, follows a square root relation with the time. This suggests that the Ni silicide formation rate is limited by diffusion, consistent with the literature. On the other hand, the depth of Co penetration, i.e., 43, 65 and 180 nm, showed a faster reaction at the beginning and slower afterwards, not a perfectly diffusion limited process.

A previous study on the silicidation of electroplated Ni showed the silicide formed at 340°C was Ni2Si. In this study, Ni3Si2 was formed at the growth front of silicidation at 400°C. The slightly Si-poorer nature (than Ni2Si) in this case is consistent with the slightly higher annealing temperature (than 340°C). However, due to the lower spatial resolution of the SEM-EDS technique used in this study, the small differences between these two compositions can be negligible. On the other hand, the tailing side of the ternary silicide growth resulted in a metal richer composition. While the 30 min annealed sample showed two layers within the ternary layer with slightly different composition, the average composition of the ternary silicide contains Si for 5 or 30 min annealed samples. Alloying 15% Ni into Co thin film was reported previously not to alter the formation temperature of Co3Si. However, ternary silicide reported in this study was even richer in metals, probably due to a lower annealing temperature and a thicker metal film.

Figure 4 shows the same compositional profiles for silicide annealed at 350°C for 60 min and 450°C for 5 min. The leading side of silicidation shows a composition of Ni50Co10Si40, a metal rich composition Me3Si2, similar to the results in Figure 3. In addition, the trailing side has a composition of Ni40Si40Si20, also same as in Figure 3, i.e., Me4Si.

On the other hand, when annealed at 450°C for 5 min, Ni depth of 410 nm and Co depth of 230 nm were observed. The 180 nm thick silicide layer at the leading side showed a Ni3Si2 composition, still same as in Figure 3. However, the trailing side of the silicide, Ni30Co40Si30, showed slightly higher Si content than in Figure 3, consistent with the higher annealing temperature. The Ni content is slightly lower in the ternary silicide region at 450°C. The reason is unclear, but it is expected to relate to a reaction between the ternary Ni-Co-silicide with Si to form Ni silicide and a slightly Si richer Ni-Co-silicide.

Solar cell thermal degradation.— Solar cells metalized with silicide and Cu were tested and compared with respect to their resistance to thermal stress. Figure 5 shows light I-V curves of two solar cells measured under one-sun illumination on solar simulator. The two cells were processed in the identical way except for the electroplating and annealing to form the silicide. Both cells were plated at −16 mA/cm² for 50 seconds. However, the cell in Figure 5a was plated with 0.2 M NiSO4 solution and the cell in Figure 5b NiCo solution with 0.18 M NiSO4 and 0.02 M CoSO4. The former cell was annealed at 350°C for 2 min Ni silicide and the latter at 400°C for 5 min. After the excessive metal was removed the silicide sheet resistance in the 60 min annealing at 350°C forms thinner silicide layer than 30 min annealing at 400°C in Figure 3. The silicide thereof showed a similar profile for Ni and Co with the depth of 130 nm and 100 nm respectively. The leading side of silicidation shows a composition of Ni80Co10Si40, a metal rich composition Me3Si2, similar to the results in Figure 3. In addition, the trailing side has a composition of Ni20Si60Si20, also same as in Figure 3, i.e., Me4Si.

On the other hand, when annealed at 450°C for 5 min, Ni depth of 410 nm and Co depth of 230 nm were observed. The 180 nm thick silicide layer at the leading side showed a Ni3Si2 composition, still same as in Figure 3. However, the trailing side of the silicide, Ni30Co40Si30, showed slightly higher Si content than in Figure 3, consistent with the higher annealing temperature. The Ni content is slightly lower in the ternary silicide region at 450°C. The reason is unclear, but it is expected to relate to a reaction between the ternary Ni-Co-silicide with Si to form Ni silicide and a slightly Si richer Ni-Co-silicide.
bushar was measured as 14.5 and 13.7 ohm/sq, respectively. A 40 nm pure Ni layer followed by 2 um Cu was plated on the silicide to finish the metallization. The two cells were stressed at 300 °C for up to 2 hours in a N₂ ambient. Light I-V measurements were conducted before and during the thermal stress. The two cells showed very similar performance before the stress, with the conversion efficiency of 16% and 15.9%, respectively. However, the composition of the silicide has a strong impact on the thermal stress on the solar cell. The cell with Ni silicide showed significant performance degradation only after 30 min stress. While the Voc only decreased by 12 mV, the fill factor and cell efficiency dropped from 76.3% to 62.5% and from 16% to 12.7%. On the other hand, the cell with ternary Ni-Co silicide showed no degradation in Voc, fill factor, and cell efficiency after 120 min stress.

While the solar cells in the field will be stressed by many other factors such as humidity and UV irradiation, thermal stress is believed to be one of the main factors. The operation temperature of solar cells in the field might not go well beyond 80 °C, thermal stress test have been used to accelerate the degradation and make the test practical. The study here was conducted at 300 °C. While the degradation mechanisms at different temperatures can be different, the test results suggest an advantage of using ternary silicide to improve the thermal stability and the lifetime of silicon solar cells. Detailed studies were carried out on the thermal degradation of different cells up to a few years and the results will be published separately. While the degradation mechanism will be discussed in detail there, the improvement in the thermal stability of solar cells with the ternary silicide is believed to relate to the higher temperature needed for further silicidation.

Conclusions

NiCo electrodeposition and silicidation were studied on crystalline silicon photovoltaic cells fabricated using CMOS silicon. A threshold light intensity was observed during the light assisted plating, below which electroplasting occurs with the solar cell under reverse bias. This threshold intensity was found more sensitive to the Ni concentration in the solution than Co because the light absorption of the halogen lamp was dominated by the long wavelength absorption by NiSO₄. The electroposited NiCo alloys were annealed at different temperatures for different time to form the silicide. The silicide growth front was found to be Ni₃Si and the trailing side a ternary silicide with a stoichiometry of Me₅Si. While the 80% metal in this ternary silicide can be composed of different ratio of Co and Ni, the ratio is at about 1:1. The silicide formed from NiCo still have a Ni silicide region at the leading edge, this pure Ni silicide region is extremely thin at the beginning of growth. An improvement in the reliability against thermal stress was observed for solar cells metalized with NiCo electropolishing and silicidation.

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