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SILICON INGOT CASTING -
HEAT EXCHANGER METHOD
MULTI-WIRE SLICING -
FIXED ABRASIVE SLICING TECHNIQUE

Phase III

Silicon Sheet Growth Development for the
Large Area Sheet Task of the Low-Cost Solar Array Project
SILICON INGOT CASTING - HEAT EXCHANGER METHOD
MULTI-WIRE SLICING - FIXED ABRASIVE SLICING TECHNIQUE

(PHASE III)

Silicon Sheet Growth Development for the Large Area Sheet Task of the Low-Cost Solar Array Project

Quarterly Progress Report No. 3
by
Frederick Schmid and Chandra P. Khattak

Covering Period from April 1 through June 30, 1979
Report Issued: July 1979

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CRYSTAL SYSTEMS, INC.
35 Congress Street
Salem, MA 01970

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ACKNOWLEDGEMENT

The authors wish to acknowledge the valuable assistance and contributions from John A. Lesiczka, Barbara B. Nelson, and Maynard B. Smith.
ABSTRACT

Ingot casting has been scaled up to 16 cm x 16 cm square cross-section size weighing up to 8.1 kg. The high degree of crystallinity has been maintained in the large ingot. For large sizes, the non-uniformity of heat treatment causes "chipping" of the surface of the ingot. Significant effort and progress has been made in the development of a uniform graded structure in the silica crucibles.

The high speed slicer has been modified so that the blade-head weight is reduced to 37 pounds. This has allowed higher surface speeds of up to 500 feet per minute. Slicing of 10 cm diameter workpieces at these speeds has increased the throughput of the machine to 5.7 mils/min., 0.145 mm/min., 50 per cent more than the projected cutting rates used in the economic analysis.
SILICON INGOT CASTING -- HEAT EXCHANGER METHOD

The emphasis in silicon casting during the current quarter has been on improvement of the solidification cycle, crucible development and scale-up. Significant advances have been made and ingots up to 8.1 kg weight in 15 cm x 16 cm square cross-section have been cast.

Ingot Casting

It is intended to cast large silicon ingots by the Heat Exchanger Method (HEM). It has been found\(^1\) that the melt temperature during crystal growth has to be maintained very close to the melting point of silicon. This is associated with the higher conductivity of silicon in the liquid state as compared to the solid state. Under such conditions growth is impeded since heat is transferred rapidly from the crucible wall to the solid-liquid interface via the molten silicon. It is, therefore, necessary to keep the temperature of the melt as close to the melt point as the instrumentation will allow. In the HEM the submerged interface and uniform temperature gradients at the solid-liquid interface prevent spurious nucleation even at such low superheat.
Another feature observed during crystal growth of silicon by HEM was an increase in power levels as the solidification progressed. This was particularly noticeable towards the end of the growth cycle. The growth cycle was initiated by increasing the flow of helium gas through the heat exchanger. Solidification proceeded and the heat exchanger reduced the temperature of the melt. The furnace controller compensated for the heat extraction by increasing power to the heater to maintain a constant furnace temperature. When the size of ingot grown was small compared to the size of the heat zone, this effect was not very large. However, for large ingots, these mutually compensating effects of heat extraction by heat exchanger and increase in power to the furnace have to be minimized. During runs 317-C through 322-C (details in Table I) the furnace was controlled on the basis of power rather than temperature. The sensitivity of control on power is not as accurate as with temperature. Under power control constant power is maintained; therefore, the furnace temperature decreased during the run due to the cooling effect of the heat exchanger. While controlling with power the melt temperature is slightly higher; hence, a period of stabilization takes place before any growth occurs. This increases the growth time. This period will be the same for larger ingots; hence growth rates will improve with size of the ingots cast. Additionally power control was reproducible from experiment to experiment. In earlier runs when deposits formed on the windows of the furnace, it led to incorrect
### TABLE I. TABULATION OF HEAT-EXCHANGER AND FURNACE TEMPERATURES

<table>
<thead>
<tr>
<th>RUN</th>
<th>PURPOSE</th>
<th>SEEDING</th>
<th>GROWTH CYCLE</th>
<th>REMARKS</th>
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</thead>
<tbody>
<tr>
<td></td>
<td>FURN. TEMP. ABOVE M.P.</td>
<td>H.E. TEMP.</td>
<td>RATE OF DECREASE</td>
<td>GROWTH TIME IN HOURS</td>
</tr>
<tr>
<td></td>
<td>OC</td>
<td>OC</td>
<td>H.E. TEMP.</td>
<td>FURN. TEMP.</td>
</tr>
<tr>
<td>317-C</td>
<td>Improve solidification cycle</td>
<td>8</td>
<td>160</td>
<td>161</td>
</tr>
<tr>
<td>318-C</td>
<td>Improve solidification cycle</td>
<td>&lt; 3</td>
<td>110</td>
<td>234</td>
</tr>
<tr>
<td>319-C</td>
<td>Improve solidification cycle</td>
<td>3</td>
<td>100</td>
<td>256</td>
</tr>
<tr>
<td>320-C</td>
<td>Improve solidification cycle</td>
<td>3</td>
<td>152</td>
<td>226</td>
</tr>
<tr>
<td>321-C</td>
<td>Improve solidification cycle</td>
<td>5</td>
<td>135</td>
<td>262</td>
</tr>
<tr>
<td>322-C</td>
<td>Improve solidification cycle</td>
<td>&lt; 3</td>
<td>165</td>
<td>258</td>
</tr>
<tr>
<td>323-C</td>
<td>Improve uniformity of graded structure of crucible</td>
<td>&lt; 3</td>
<td>213</td>
<td>237</td>
</tr>
<tr>
<td>324-C</td>
<td>Improve uniformity of graded structure of crucible</td>
<td>&lt; 3</td>
<td>145</td>
<td>333</td>
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</table>

(cont.)
<table>
<thead>
<tr>
<th>RUN</th>
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<th>SEEDING</th>
<th>GROWTH CYCLE</th>
<th>REMARKS</th>
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<tr>
<td></td>
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<td>FURN. TEMP.</td>
<td>H.E. TEMP.</td>
<td>RATE OF DECREASE</td>
</tr>
<tr>
<td></td>
<td></td>
<td>ABOVE M.P.</td>
<td>BELOW M.P.</td>
<td>H.E. TEMP.</td>
</tr>
<tr>
<td></td>
<td></td>
<td>°C</td>
<td>°C</td>
<td>°C/HR.</td>
</tr>
<tr>
<td>325-C</td>
<td>Crucible development using lathe</td>
<td>&lt; 3</td>
<td>145</td>
<td>416</td>
</tr>
<tr>
<td>326-C</td>
<td>Crucible development using lathe</td>
<td>&lt; 3</td>
<td>163</td>
<td>295</td>
</tr>
<tr>
<td>327-C</td>
<td>Crucible development using lathe</td>
<td>4</td>
<td>133</td>
<td>346</td>
</tr>
<tr>
<td>328-C</td>
<td>Crucible development using lathe</td>
<td>3</td>
<td>146</td>
<td>314</td>
</tr>
<tr>
<td>329-C</td>
<td>Cast 4.5 kg, 16 cm x 16 cm square ingot</td>
<td>&lt; 3</td>
<td>135</td>
<td>287</td>
</tr>
<tr>
<td>330-C</td>
<td>Cast 16 cm x 16 cm square ingot</td>
<td>&lt; 3</td>
<td>115</td>
<td>330</td>
</tr>
<tr>
<td>331-C</td>
<td>Cast 8.1 kg, 16 cm x 16 cm square ingot</td>
<td>4</td>
<td>110</td>
<td>360</td>
</tr>
<tr>
<td>332-C</td>
<td>Cast 4.3 kg, 16 cm x 16 cm square ingot</td>
<td>4</td>
<td>96</td>
<td>246</td>
</tr>
<tr>
<td>333-C</td>
<td>Cast 4.3 kg, 16 cm x 16 cm square ingot</td>
<td>&lt; 3</td>
<td>95</td>
<td>362</td>
</tr>
</tbody>
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(cont.)
## TABLE I. TABULATION OF HEAT-EXCHANGER AND FURNACE TEMPERATURES (cont.)

<table>
<thead>
<tr>
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<th>PURPOSE</th>
<th>FURN. TEMP. ABT. M.P.</th>
<th>H.E. TEMP. ABOVE M.P.</th>
<th>RATE OF DECREASE H.E. TEMP. (^\circC/HR.)</th>
<th>FURN. TEMP. (^\circC)</th>
<th>GROWTH TIME IN HOURS</th>
<th>REMARKS</th>
</tr>
</thead>
<tbody>
<tr>
<td>334-C</td>
<td>Cast 4.3 kg, 16 cm x 16 cm square ingot</td>
<td>4</td>
<td>93</td>
<td>274</td>
<td>4</td>
<td>6.5</td>
<td>Minor chipping of ingot</td>
</tr>
<tr>
<td>335-C</td>
<td>Cast 4.3 kg, 16 cm x 16 cm square ingot</td>
<td>15</td>
<td>109</td>
<td>308</td>
<td>15</td>
<td>6.75</td>
<td>Very good crystallinity</td>
</tr>
<tr>
<td>336-C</td>
<td>Cast 4.3 kg, 16 cm x 16 cm square ingot</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>Run terminated; reaction of Si with graphite pipe</td>
</tr>
<tr>
<td>337-C</td>
<td>Cast 8.2 kg ingot</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>Run terminated; Heat Exchanger failure</td>
</tr>
<tr>
<td>338-C</td>
<td>Cast 7.7 kg ingot</td>
<td>11</td>
<td>205</td>
<td>400</td>
<td>11</td>
<td>16.5</td>
<td>Minor chipping of ingot</td>
</tr>
</tbody>
</table>
temperature measurement and thereby prolonged growth cycles.

Very good crystallinity of the silicon ingots has been maintained.

A major emphasis of the present program is to cast large, square cross-section ingots. During Phase II 10 cm x 10 cm ingots were cast. It was intended to scale up the size to 15 cm x 15 cm during the next step. The crucibles fabricated have an inside dimension which is close to about 17 cm x 17 cm. A 4.5 kg ingot cast in run 329-C is shown in Figure 1. This was the first ingot of this size that was cast.

During run 331-C the size of the ingot was almost doubled to 8.1 kg. No problems were encountered during crystal growth of this ingot. The as-cast surface of this ingot is shown in Figure 2. One of the major advantages of the Heat Exchanger Method (HEM) is its simplicity. This allows easy scale-up in size and this has been demonstrated again. However, the crystal growth from the bottom means that as the size is increased the interface is further away from the heat exchanger. So far the height to diameter/width of the ingot has always been slightly less than unity. In run 331-C the ingot cast was such that the height to width ratio was greater than one and no problems were encountered. A similar size ingot cast in run 338-C is shown in Figure 3.
Figure 1. Two views of the 16 cm x 16 cm square ingot cast by HEM in run 329-C.
Figure 2. A view of 8.1 kg square ingot as removed from crucible
Figure 3. A view of 8.1 kg square ingot cast in run 338-C
**Crucible Development**

In order to cast crack-free silicon ingots in silica crucibles it is necessary to develop a graded structure in the crucible. This structure has been developed by heat treatment manually and has worked satisfactorily. With larger crucible sizes it has been very difficult to achieve uniformity in the graded structure. A glass-working lathe has been procured. The crucibles are rotated at a constant rate and heated to develop a uniform graded structure. It is intended to mechanize this approach further to achieve a higher degree of uniformity. Once the conditions are standardized, it could be extended to still larger crucibles. Initially crucible development and optimization was conducted with 15 cm diameter crucibles. Once some experience was gained with the glass-working lathe for heat treatment of the 15 cm diameter crucibles, larger size 16 cm x 16 cm square crucibles were heat treated. The first such size was used in run 329-C and a 4.5 kg crack-free ingot was cast. During runs 330-C and 331-C minor attachment of the crucible to the bottom of the ingot was observed. This problem has since been corrected by heat treating the bottom area a little differently to correct for the different temperature of the bottom of the crucible as compared to the temperature profile of the sides of the crucible during crystal growth.
MULTI-WIRE SLICING--FIXED ABRASIVE SLICING TECHNIQUE

The emphasis during the current quarter in silicon slicing was on machine and blade development. Efforts were pursued to reduce the weight of the bladehead so that still higher speeds can be achieved.

Machine Development

It has been established\(^2\) that for effective slicing with fixed diamond abrasive it is desirable to operate at high surface speeds. At lower speeds the diamond is in a dragging mode rather than in a cutting mode. In order to achieve higher speeds it is necessary to lighten the reciprocating bladehead. The weight of the modified Varian slicer was about 200 pounds and a surface speed of 100 feet per minute was achieved. The high speed slicer designed was such that the blade carriage was twice as long and weighed half as much. Slicing tests with this bladehead were carried out at surface speed of about 200 feet per minute.\(^2\) Significant improvement of the cutting performance was seen at the higher speeds. To further increase speeds it was necessary to lighten the reciprocating carriage still further. A new 37 pound bladehead has been designed and
fabricated. A dry run test for speed has shown that 500 feet per minute surface speeds can be achieved. This lightweight blade carriage together with higher speeds is expected to show a big improvement in cutting performance.

One of the problems identified during slicing tests of 10 cm x 10 cm workpiece was the hysteresis of the feed mechanism and the shape of the cut profile. These problems have been solved by using weights to control feed forces and by altering the rocking mechanism so that a smooth cut profile is achieved.

Figure 4 shows two views of the new lightweight bladehead assembled in the slicer. Figure 5 shows the new rocking mechanism incorporated.

The first slicing test with the modified slicer was carried out during run 325-SX (details in Table II) to slice a 10 cm diameter silicon workpiece. The surface speed was about 400 feet per minute with a 16-inch stroke length. Very good cutting rates were achieved; such high cutting rates have never been achieved with commercially impregnated wires. The total slicing time was 11 hours, 41 minutes, giving an average cutting rate of 5.7 mils/min., 0.145 mm/min. This compares to 2.33 mils/min., 0.059 mm/min. obtained with similar wires and 10 cm diameter workpiece before the present modification. This means that about 250% cutting rates have been achieved with increase in surface speed. These cutting rates are about 50% better than the projected cutting rates (0.1 mm/min.) in the
Figure 4. Two views of the high-speed slicer with lightweight bladehead.
Figure 5. A view of the rocking mechanism incorporated in the high-speed slicer
<table>
<thead>
<tr>
<th>RUN</th>
<th>PURPOSE</th>
<th>FEED FORCE/BLADE</th>
<th>AVERAGE CUTTING RATE</th>
<th>WIRE TYPE</th>
<th>REMARKS</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>lb</td>
<td>gm</td>
<td>mil/min</td>
<td>mm/min</td>
</tr>
<tr>
<td>318-S</td>
<td>Life test continuation after dressing wires</td>
<td>0.070</td>
<td>31.9</td>
<td>2.38</td>
<td>0.060</td>
</tr>
<tr>
<td>319-S</td>
<td>Test plating from another source</td>
<td>N/A</td>
<td>N/A</td>
<td>N/A</td>
<td>N/A</td>
</tr>
<tr>
<td>320-S</td>
<td>Test modified plating</td>
<td>0.070</td>
<td>31.9</td>
<td>2.22</td>
<td>0.056</td>
</tr>
<tr>
<td>321-S</td>
<td>Life test</td>
<td>N/A</td>
<td>N/A</td>
<td>N/A</td>
<td>N/A</td>
</tr>
<tr>
<td>322-S</td>
<td>Test 30 μm diamond electroplated wires</td>
<td>0.056</td>
<td>25.4</td>
<td>3.00</td>
<td>0.076</td>
</tr>
<tr>
<td>323-S</td>
<td>Life test</td>
<td>0.056</td>
<td>25.4</td>
<td>1.15</td>
<td>0.029</td>
</tr>
<tr>
<td>324-S</td>
<td>Life test continuation</td>
<td>0.056</td>
<td>25.4</td>
<td>1.04</td>
<td>0.026</td>
</tr>
<tr>
<td>325-SX</td>
<td>Slicing of 4&quot; Ø ingot with high speed slicer</td>
<td>0.091</td>
<td>41.3</td>
<td>5.70</td>
<td>0.145</td>
</tr>
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</table>

Very high cutting rates. Roller degradation observed; 21% yield (cont.)
<table>
<thead>
<tr>
<th>RUN</th>
<th>PURPOSE</th>
<th>FEED FORCE/BLADE</th>
<th>AVERAGE CUTTING RATE</th>
<th>WIRE TYPE</th>
<th>REMARKS</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>lb</td>
<td>gm</td>
<td>mil/min</td>
<td>mm/min</td>
</tr>
<tr>
<td>326-S</td>
<td>Test impregnated wire</td>
<td>N/A</td>
<td>N/A</td>
<td>N/A</td>
<td>N/A</td>
</tr>
<tr>
<td>327-SX</td>
<td>Test 30 μm diamond electroplated wire</td>
<td>-</td>
<td>-</td>
<td>5.19</td>
<td>0.13</td>
</tr>
</tbody>
</table>
economic analysis for this technique.\textsuperscript{3}

The yield obtained during run 325-SX was only 21%. It was found that during first half of the ingot, a 100% yield was maintained; however, from then on, breakage of wafers was observed. Examination after completion of the run showed that the guide rollers degraded; i.e., wires cut deep grooves into the rollers. The heavy mass of these rollers had too much inertia to change directions at such high speeds and the wire cut into them.

Similar high cutting rates were also achieved in run 327-SX even though the non-uniformity of the nickel plating on this bladepack is not conducive to effective slicing.

**Blade Development**

During this quarter it was found that very good cutting performance was achieved with 30 µm diamond electroplated wire—high yields and good wafer quality were obtained for four runs. In run 318-C these wires were used for a fifth run after dressing. It was found that the cutting rates achieved during this run were higher; however, only 64% yield was obtained. Visual examination showed absence of diamonds in some areas of the cutting edge. SEM examination of these wires revealed that swarf had dried up and formed a layer over the diamonds in some areas (Figure 6). These wires have given a good cutting performance and it has been maintained for five runs without much degradation. The cutting rates revived after dressing of wires.
Figure 6. Sections of wire showing swarf is covering the diamonds
A bladepack of commercially impregnated wire was plated with 7.5 μm nickel. During plating the solution was modified to see if the bond between the nickel and diamonds can be improved. These wires were used for slicing tests in run 320-S. Good cutting performance was achieved; however, during the second test (run 321-S) the wires stopped cutting towards the end of the run due to diamond pull-out.

Two significant developments have occurred in impregnation of diamonds into copper sheath. It has been found that it was possible to impregnate 45 μm size diamonds into 7.5 μm copper sheath. However, the copper was distorted. A 12.5 μm sheath was not distorted and the concentration of diamonds was very high. However, after plating the diamond concentration was not as high. Impregnation of 45 μm diamonds into 12.5 μm sheath was not deep enough and diamonds fell out during plating. Impregnation of 30 μm into a 15.5 μm copper sheath under high forces produced satisfactory impregnation and retention of diamonds. This showed that the diamonds have to be impregnated deep into the sheath with high pressure. However, as shown in run 326-S a 7.5 μm electroless nickel buries the diamonds.

So far the plated length of the wires has been equal to the stroke length of the slicer. It was found that at the end of the stroke a support roller and the workpiece passed over the transition area. Better performance was achieved when the diamond plated length was longer than the stroke length.
A similar blade pack used for runs 314-S through 318-S but with diamonds only in the cutting edge was used in runs 322-S through 324-S. Very effective slicing, good quality wafers, and 100% yield was achieved during the first run. Subsequently a deterioration in performance was observed. Examination of the blade pack after the third run showed that considerable diamond pull-out was observed.

A comparison of performance of this blade pack with a similar earlier one shows two significant differences which may explain the difference in their lives. Firstly, the present blade pack was electroplated on one side of the wire only. Secondly, the feed forces used in recent runs were lower, 25.7 gms. vs. 31.9 gms. used earlier. Both sets of wires were used at low surface speeds which is not conducive to effective slicing with fixed diamonds. These conditions put the diamond in a dragging mode. Further, it has been established that higher feed forces result in enhanced cutting rates. Therefore, lower feed forces may be less conducive conditions for diamond slicing in addition to dragging mode at lower surface speeds.
CONCLUSIONS

1. Ingot casting has been extended to 16 cm x 16 cm square cross-section size. Ingots up to 8.1 kg have been cast.
2. Good crystallinity has been maintained for ingots using simpler power control.
3. A glass-working lathe makes it easier to achieve a uniform and reproducible graded structure in the crucibles.
4. Surface speeds of 500 feet per minute have been achieved with the 37 pound bladehead.
5. Higher forces used in impregnation helped retain diamonds during plating.
6. First slicing test on 10 cm diameter silicon after modifications have given average cutting rates of 5.7 mils/min., 0.145 mm/min. with impregnated wires.
REFERENCES


MILESTONE CHART

1. Design HEM modification package
2. Fabrication of new HEM furnace
3. Operate HEM furnace 20 cm diameter; 8 kg, 20 cm cube, 18.5 kg
4. Blade development
5. Operate FAST slicer 150 parallel slices; 20 slices/cm
6. Slicer modification 250 parallel slices; 25 slices/cm
7. Design and performance review
8. Reports

(Modified milestones)
APPENDIX

Origin of SiC Impurities in Silicon Crystals
Grown from the Melt in Vacuum

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T. G. Digges, Jr.
Larry Kaufman

Journal of the
Electrochemical Society

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Origin of SiC Impurities in Silicon Crystals Grown from the Melt in Vacuum

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ABSTRACT

A main source of high carbon levels in silicon crystals grown from melt under reduced pressures and contained in silica crucibles supported by graphite—silicon carbide/retainer has been identified by thermodynamic analysis. The calculations have been verified by experimental results and the carbon level can be reduced by approximately 50% with the use of molybdenum retainers.

The heat exchanger method (HEM) developed to grow large sapphire crystals (1-3) has been extended to the growth of silicon crystals (4). In this method the seed is placed at the bottom of the crucible and the temperature in the melt increases upwards. This suppresses convection that causes temperature and concentration fluctuations at the solid-liquid interface (5). Early experiments indicated that SiC particles were found in crystals solidified by the HEM (6). However, even with the presence of SiC particles, large grains have been grown with limited interface breakdown during solidification. This observation was contradistinct to Czochralski (CZ) growth where interface breakdown due to SiC is followed by twin/polycrystalline growth (7).

The basic elements of the HEM and the CZ growth processes are quite similar (heaters, crucibles, insulation, etc.). A silica crucible loaded with the charge and set in a graphite retainer is placed in the furnace. The chamber is evacuated, and after melting the charge crystal growth is achieved. In the HEM process, the chamber is typically evacuated during growth to 0.1 Torr. For CZ growth an argon blanket is used and the chamber pressure can vary from 10 Torr to 1 atm (8). In early experiments with HEM growth it was quite surprising that high carbon concentrations were found in the silicon ingots. The purpose of this paper is to explain the origin of this carbon concentration and to present thermodynamic and experimental results that support the conclusions.

Reactions between Graphite and Silica

Silicon, oxygen, and carbon are used in the HEM in the form of silicon melt stock, silica crucibles and graphite furnace parts, and retainers. A thermodynamic evaluation has considered possible reactions between these three reactants and the salient reactions are reported.

Carbon has been observed in silicon crystals as silicon carbide. Silicon melt stock and graphite parts are not in direct contact; therefore, the following reactions between graphite and silica crucible were studied:

\[ 2C + SiO_2 \rightarrow SiC + CO_2 \]  [1]
\[ 3C + 3SiO_2 \rightarrow SiC + 2CO + 2SiO \]  [2]
\[ 4C + 2SiO_2 \rightarrow SiC + 3CO + SiO \]  [3]

In addition to the reactions forming SiC two additional reactions were considered:

\[ C + SiO_2 \rightarrow SiC + CO \]  [4]
\[ C + 2SiO_2 \rightarrow 2SiO + CO_2 \]  [5]

Using the data in Table I, reactions [1]-[6] have positive free energy changes at atmospheric pressure and in the temperature range 1800°-1750°K (Table II) with those reactions that produce CO2 tending to be more positive. The least positive free energies at atmospheric pressure are given by Eq. [4] and [5]. For reaction [5] the standard free energy of formation, \( \Delta G^* \), using data in Table I, is

\[ \Delta G^* = 162900 - 81T \]

Also

\[ \Delta G = \Delta G^* + RT \ln K \]

where \( R \) is the gas constant, \( T \) is temperature in degrees Kelvin, and \( K \) is the equilibrium constant for the reaction. Since equimolar ratios of SiO and CO are formed, their partial pressures are

\[ p_{CO} = p_{CO_2} = \frac{1}{2} p \]

where \( p \) is the total pressure. Under equilibrium conditions, \( \Delta G = 0 \), hence

\[ \log p \text{(atm)} = 9.15 - 1.78 \times 10^4/T \]

or

\[ \log p \text{(Torr)} = 12.03 - 1.78 \times 10^4/T \]

These equations are used to plot the curve (line) along with \( \Delta G = 0 \), i.e., \( C + SiO_2 \) is in equilibrium with \( SiO + CO \). The line separates the range of \( T \); \( p \) space in which \( C + SiO_2 \) has a lower free energy than

<table>
<thead>
<tr>
<th>Compound</th>
<th>( \Delta G^* ) (cal/mol)</th>
</tr>
</thead>
<tbody>
<tr>
<td>SiC (solid)</td>
<td>-16200 + 37T</td>
</tr>
<tr>
<td>CO (gas)</td>
<td>-22000 - 207T</td>
</tr>
<tr>
<td>( CO_2 ) (gas)</td>
<td>-13300 - 427T</td>
</tr>
<tr>
<td>( SiO_2 ) (solid)</td>
<td>-21800 - 167T</td>
</tr>
</tbody>
</table>

* Note: \( T = ^\circ K \).

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Key words: carbon, thermodynamics, transport, pressure.
Table II. Standard free energy change for various reactions as a function of temperature

<table>
<thead>
<tr>
<th>Reaction</th>
<th>1000°K</th>
<th>1500°K</th>
<th>1700°K</th>
<th>1750°K</th>
</tr>
</thead>
<tbody>
<tr>
<td>(4C + 3SiO \rightarrow SiC + CO)</td>
<td>-17,000</td>
<td>-17,750</td>
<td>-17,700</td>
<td>-17,550</td>
</tr>
<tr>
<td>(3C + SiO \rightarrow SiC + CO)</td>
<td>-9,700</td>
<td>-7,600</td>
<td>-5,500</td>
<td>-3,400</td>
</tr>
<tr>
<td>(C + SiO \rightarrow Si + CO)</td>
<td>-4,400</td>
<td>-4,000</td>
<td>-3,600</td>
<td>-3,200</td>
</tr>
</tbody>
</table>

The free energy of these reactions at atmospheric pressure would be further emphasized if graphite were to react with oxygen in place of CO in Eq. [7].

Another vapor species expected in the HEM furnace is SiO. This is formed by the reaction of molten silicon with the silic carbide (9, 10)

\[Si + SiO_2 \rightarrow 2SiO\]

The equilibrium vapor pressure of gaseous SiO as a function of temperature for this reaction is plotted in Fig. 2.

The reaction involving the two dominant vapor species, CO and SiO, has been examined

\[CO + 3SiO \rightarrow SiC + 2SiO_2\]

and the results plotted as a function of pressure and temperature in Fig. 3. It can be seen that under the operating experimental conditions, this reaction does not proceed to the right.

Examination of the following reactions between graphite and SiO show that in two cases SiC is formed

\[2C + SiO \rightarrow SiC + CO\] \[3C + 2SiO \rightarrow 2SiC + CO_2\] \[C + SiO \rightarrow Si + CO\]

The free energy of these reactions at atmospheric pressure would be further emphasized if graphite were to react with oxygen in place of CO in Eq. [7].
pressure is negative (Table I). This will result in coating of the furnace parts with SiC, yet no contamination of the silicon.

The following reactions involving CO and Si were studied:

\[ \text{Si} + \text{CO} \rightarrow \text{SiC} + \frac{1}{2} \text{O}_2 \]  \[ 13 \]
\[ 3\text{Si} + 2\text{CO} \rightarrow 3\text{SiC} + \text{SiO}_2 \]  \[ 14 \]
\[ 2\text{Si} + \text{CO} \rightarrow \text{SiC} + \text{SiO} \]  \[ 15 \]

Reaction [13] yields a positive free energy at atmospheric pressure. Considering the pressure and temperature dependence of this reaction, the reactants will be stable in the experimental operating conditions (10). The stability range of reaction [14] shows that the reactants are stable (Fig. 4). The standard free energy of reaction [15], \( \Delta G^* \), using data in Table I can be represented by Eq. [16] where \( T = \text{K} \):

\[ \Delta G^* = -17,000 + 37' \text{calories} \]  \[ 16 \]

This reaction proceeds to the right at 1700°K. The free energy of the reaction, \( \Delta G \), utilizing the usual free energy relationship, can be calculated from:

\[ \Delta G = \Delta G^* + RT \ln K \]  \[ 17 \]

Assuming unit activity in the condensed phase and \( P_{\text{SiO}} \), \( P_{\text{CO}} \) (the partial pressures of \( \text{SiO} \) and \( \text{CO} \) respectively), Eq. [17] can be written as:

\[ \Delta G = -17,000 + 37' + 1.0077 \text{ln } (P_{\text{SiO}}/P_{\text{CO}}) \]  \[ 18 \]

If the equilibrium fraction of \( \text{SiO} \) can be represented by \( x \), then:

\[ x = \exp \left( \frac{17,000 - 37'}{1.0077} \right) \left[ 1 + \exp \left( \frac{17,000 - 37'}{1.0077} \right) \right] \]  \[ 19 \]

The corresponding fractions of the vapor species at different temperatures are shown in Table III. Therefore, a major source of silicon carbide formation in silicon can come from reaction [15]. The vapor pressure of carbon over graphite is less than 10^-11 Torr at the melting point of silicon and can, therefore, be ignored. The formation of CO, besides a vacuum leak and reaction of absorbed gases on graphite parts, is from reactions [4] and [5]. This appears to be the most likely source for transport of C to the silicon.

Reactions [1]-[15] and Table I cover 9 of the 12 reactions cited by Caldwell (11). The three remaining reactions discussed in Caldwell's paper and ignored here \( n = 12 \) with the reduction of \( \text{SiO}_2 \) and/or \( \text{SiC} \) by carbon to form silicon. These reactions are unlikely in the \( T, P \) range under consideration here. The present analysis is concerned with a wide range of pressure in contrast to Caldwell's discussion which is confined to 1 atm conditions.

Experimental Verification

The above thermodynamic analysis indicates that SiC in HEM grown silicon is attributable to the reaction of CO and Si. A major source of CO is from reactions [4] and [5], interaction of silicon crucible with graphite retainer. This was verified indirectly by experiment. A graphite piece was sandwiched between two pieces of single crystal silicon with polycrystalline silicon around them. The furnace was heated close to the melting point of silicon and cooled. It was found that a silicon carbide layer was formed in areas where the graphite was in contact with the silicon crystals and also a light layer was found in the immediate vicinity. The pieces far removed from the graphite piece.

### Table III. Equilibrium fraction of CO and SiO at various temperatures in reaction [15]

<table>
<thead>
<tr>
<th>( T (°K) )</th>
<th>( x = P_{\text{CO}}/P_{\text{SiO}} )</th>
<th>% CO</th>
<th>% SiO</th>
</tr>
</thead>
<tbody>
<tr>
<td>1650</td>
<td>0.920</td>
<td>6.5</td>
<td>93.5</td>
</tr>
<tr>
<td>1670</td>
<td>0.921</td>
<td>6.5</td>
<td>93.5</td>
</tr>
<tr>
<td>1700</td>
<td>0.925</td>
<td>7.5</td>
<td>92.5</td>
</tr>
<tr>
<td>1750</td>
<td>0.935</td>
<td>6.5</td>
<td>93.5</td>
</tr>
</tbody>
</table>

\( P_{\text{SiO}} = P_{\text{SiO}} + P_{\text{CO}} \).
showed no evidence of SiC. In this experiment the graphite piece was not in contact with silica, hence reactions [4] and [5] were suppressed. In another run when the charge was heated in a silica crucible held in a graphite retainer and cooled, the entire melt stock was coated with SiC.

Experimental evidence of SiC formed in HEM grown silicon is shown in Figs. 5 and 6. Silicon carbide has accumulated at the solid liquid interface. The sample has been etched with CP-4. The groove represents a boundary where spontaneous nucleation and freezing from the top of the liquid has met the advancing interface solidifying from the bottom towards the top. In this growth a graphite retainer was used and the operating pressure was 0.1 Torr. The precipitates were identified by x-ray dispersive analysis. Infrared measurements indicated a concentration of carbon in the (3.5-4.1) \times 10^{17} \text{ atoms/cm}^3 range which is close to the solid solubility limit near the melting point of silicon (12). The entire top surface of the solidified ingot was covered with SiC. A similar experiment carried out with a molybdenum retainer indicated a carbon concentration of (1.5-3.1) \times 10^{17} \text{ atoms/cm}^3 for twelve samples. The top surface of the ingot was clean and shiny with no evidence of the SiC layers. The replacement of graphite retainer with molybdenum resulted in lower carbon levels in silicon. This eliminated the contact of graphite with the silica crucible. As far as reactions involving O_2, H_2 and water absorption in graphite, they will not be changed significantly because the HEM is carried out in a resistance heated furnace which has all graphite parts (e.g., heater, liner, insulation, etc.).

Conclusions

Evidence of high carbon levels in HEM grown silicon have led to a proposed mechanism for the cause of silicon carbide formation. It is associated with the use of graphite retainers in contact with silica crucibles under reduced pressures. High carbon levels in silicon have been reported to cause breakdown in crystallinity in CZ (7, 13) as well as in ribbon growth (14). The electrical effects of SiC inclusions have been detrimental in solar cell devices (15). The origin of high carbon levels in silicon processed in vacuum is because of reactions between silica crucibles and graphite retainers. If the graphite retainers are replaced by molybdenum, these reactions do not take place, and thereby the carbon levels in the silicon can be considerably reduced.

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