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Thermomechanical residual stress evaluation in multi-crystalline silicon solar cells of photovoltaic modules with different encapsulation polymers using synchrotron X-ray microdiffraction

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\textbf{A B S T R A C T}
Photovoltaic (PV) module reliability issues, due to silicon cell cracking, are gaining more and more attention due to increasing demand for solar power and reduction of cell thickness to reduce cost. Recent reports show significant effect of encapsulation polymer material on cell cracks leading to the idea of tailoring encapsulation materials for more reliable PV modules. This paper investigates the effect of encapsulation modulus on the cell residual stress using Synchrotron scanning X-ray microdiffraction (\textmu SXRD), which has been proven to be an effective technique to probe the stress in silicon solar cells, especially once they are encapsulated. The post lamination residual stress in the encapsulated multi-crystalline silicon (mc-Si) solar cells was reported for the first time using \textmu SXRD in this manuscript and provide quantitative evaluation of the effect of encapsulation modulus on the cell residual stress. Further, simple approximate finite element (FE) model was also developed to evaluate the effect of the encapsulation polymer on the cell stress. The FE simulations predict the trend of the stress variation with encapsulation polymer modulus very well. Dynamic mechanical analysis and rheological testing of the encapsulation polymers was also performed to correlate the polymer behaviour with the experimental and simulated stresses. Both experimental and simulation results show a similar trend of significant cell stress variation with encapsulation polymer modulus. In the case of external loading, the temperature of load application is observed to be very significant as it dictates the elastic state of the encapsulant, leading to critical conclusion that the encapsulant needs to be selected based on elastic behaviour over the temperature history of the encapsulant during module fabrication and operation. The results and discussion presented are expected to be very useful for development of more reliable PV modules.

1. Introduction
Fracture of crystalline silicon solar cells in photovoltaic (PV) modules is widely studied and a well-known issue in the PV industry \cite{1,2}. The PV module is a multilayer laminate of dissimilar materials bonded thermally. In such laminates, after the thermal bonding process, there will be built-in stresses (residual stresses) in the individual layers which may increase upon external loading and cause fracture and failure. In the case of solar PV modules, the effect of soldering induced stresses was widely studied \cite{3,4}. The effect of lamination (encapsulation) and encapsulation polymers on cell cracks was not given enough importance as the encapsulation polymers are very soft and flexible materials compared to the other constituent materials such as silicon, copper and glass. However, recent reports have shown that the material properties and thickness of the encapsulation polymers strongly influence the stress and cracking of silicon cells in the modules \cite{5–10}. Mickiewicz et al. \cite{5} demonstrated strong dependence of cell cracks on the encapsulation polymer modulus at the loading temperature through static and dynamic load tests on modules made of ethylene vinyl acetate co polymer (EVA) and Silicone encapsulants. EVA is the most common commercial encapsulant but stiffer than Sillicone, which is a less common but soft and costly material. Finite element analysis (FEA) simulations by Dietrich et al. \cite{6} indicated that the elastic modulus variation during post lamination cooling of the PV module affected the module warpage and embedded cell gap, which they used as a metric for built-in stress in the module. Dietrich et al. \cite{7} also demonstrated that a soft EVA encapsulant would lead to lower failure probability of the module compared to a stiffer EVA. It should be noted that the properties of EVA can differ based on its manufacturer. Paggi et al. \cite{8} reported that the cracked cells upon loading will recover after unloading due to the binding effect of the encapsulation polymer. Li et al. \cite{9} used FEA to show that the stress in a Germanium solar cell PV module is linearly dependent on the modulus, thickness and curing temperature of the encapsulation polymer. Handara et al. \cite{10} compared mini PV modules made of monocristalline silicon interdigitated back contact (IBC) solar cells using Synchrotron scanning X-ray...
microdiffractometry (µSXRD) measurements of stress to show that a stiffer encapsulant caused more stress in the cell than a softer encapsulant. They also used fracture testing to prove the point further. Our recent work, Tippabhotla et al. [11] on residual stress evolution in IBC silicon solar cells using both µSXRD experiments and FEA, showed that the stress concentrates near the solder joint with the copper ribbon due to local deformation of the cell. This local deformation of the cell is a result of the vacuum pressure applied on the cell during lamination process. Due to the adhesive nature of the encapsulant, this local deformation of the cell is retained even after lamination process, which binds the cell to the stiffer front glass. Hence, the properties of the encapsulant effects the stress in the cell.

With the above review of literature, it is very clear that the encapsulant plays a vital role in the cracking of silicon cells, which means, a properly selected or tailored encapsulant can make the PV modules more resistant to cell cracking and associated degradation. However, the reported research is either based on probability of fracture or crack statistic, upon external loading, considering only a few encapsulants or simulations. Other reports [10,11] have used the µSXRD technique to investigate stress in encapsulated single crystalline silicon solar cells and demonstrated its effectiveness in probing mechanical stress in encapsulated silicon solar cells. In the current work, for the first time, an attempt was made to evaluate the effect of encapsulant modulus and thickness on the post lamination residual stress in multi-crystalline silicon (mc-Si) solar cells from REC Solar, using µSXRD technique. The unique capability of this approach is the quantitative evaluation of highly localized stress in the cell near the solder joint (cell to copper ribbon joint, Fig. 1(c)). This is where the stress is concentrated and cracks originate (or existing micro cracks propagate) in the solar cells [3,4]. Five different commercially available encapsulation polymers were used in this study to get a systematic understanding of the effects. Further, the effect of thickness of the encapsulant, EVA was also studied. A simple representative 2D FEA simulation was also performed to elucidate the mechanics of the residual stress evolution during the module integration process from soldering to lamination.

2. Experimental

2.1. Experimental samples

Mini PV module samples with single mc-Si solar cell, as shown in Fig. 1 were fabricated at REC Module Technology Laboratory for this experimental study. This experiment was divided into two different case studies based on the literature reviewed.

1. Case Study 1: Effect of encapsulation polymer modulus on cell residual stress

2. Case Study 2: Effect of encapsulation polymer (EVA) thickness on cell residual stress

Hence two sets of single cell mini module samples were prepared as shown in the Table 1 below to cater the case studies 1 and 2. The cells used in this study were mc-Si cells of 156 × 156 mm² area and thickness 0.18 mm, with 4 busbars, soldered in the REC module fabrication line, which is fully automated. The front glass used in the mini modules was a typical mini-module glass plate of area 200 × 200 mm² with 3.2 mm thickness. A back sheet was not used in the mini modules as the cell needed to be visible from back side for the µSXRD experiments. The details of the encapsulants such as name, chemical formula etc. are proprietary information and cannot be disclosed, hence they are arbitrarily named as Encap1 to Encap5 as shown in Table 1. The lamination process was carried out at a temperature of 160 °C, under 0.1 MPa vacuum pressure for ~ 13 min, which is typical for REC PV modules. The detailed lamination recipe was given in the Table 2.

2.2. Synchrotron X-ray microdiffraction (µSXRD) experiments

According to Bragg’s law [12,13], a crystalline material sample subjected to X-rays, produce diffraction patterns, which can be analysed to evaluate crystal structure, orientation and deformation (strain). At the Advanced Light Source (ALS), a synchrotron facility at Lawrence Berkeley National Laboratory (Berkeley Lab, CA, USA), a microdiffraction beamline end station (BL 12.3.2) capable of producing a high energy (5–26 keV) polychromatic X-ray beam of 1 µm diameter is available [13–15]. The beamline utilizes highly focused, high flux X-rays generated by a superbend magnet source and specialized focusing optics by the Synchrotron facility [14–16]. Very high spatial resolution stress measurements are possible with this microdiffraction BL 12.3.2 [17] and such capability is essential for very local residual stress evaluation near soldered copper interconnects in encapsulated mc-Si cells. A detailed description of the microdiffraction beamline is given elsewhere [14–17]. The efficacy of synchrotron X-ray diffraction in
enabling technologically important innovations, including in micro-
electronics and nanotechnology industries, in additions to next gen-
eration silicon solar PV technologies, have recently been reported and
described elsewhere. [18–22]

Fig. 2(a) shows the schematic of the microdiffraction experiment
setup and an actual picture of the setup is shown in Fig. 2(b). The
sample was mounted on the stage with its back (Fig. 1(b)) facing the X-
ray beam, as X-rays cannot penetrate the front glass due to absorption.
To focus the X-ray beam on the cell, the back sheet was omitted in the
mini PV module samples. In this setup, the sample was mounted at 45°
to the X-ray beam, which is horizontal and a hybrid pixel area detector
(DECTRIS Pilatus 1 M), which is mounted above the sample records the
2D diffraction patterns from the sample. Hence this setup can record
multiple diffraction peaks satisfying Bragg’s condition in a single ex-
posure to polychromatic X-ray beam (white beam), also called Laue
diffraction [17] and the resulting diffraction pattern is called Laue
Pattern or Laue-gram. Further, the sample stage could move in X, Y and
Z directions, enabling raster scanning. The sample was placed in the XY
plane and the Z direction was normal to the sample plane [23] and the
stage was adjusted to focus the X-ray beam on the sample per the ex-
perimental requirement by moving in Z-direction, so that the move-
ment in the XY plane did not alter the focus of the X-ray beam. Then,
the sample was scanned using a 100 µm displacement step along the X
and Y directions. The actual scanned region is shown in Fig. 1(c), which is
an envelope of ~4000 scan points with 1.5 s exposure at each point,
requiring around 2.5–3 h for each scan.

The resulting Laue patterns can be analysed by the specialized
XMAS software [23], which uses a peak fitting routine to identify each
diffraction peak, and attribute the \( h k l \) Miller indices in a process called
indexing. The indexing provides the orientation of the crystal. The ac-
curacy of the crystal orientation depends on the number of peaks in-
dexed, which in turn depends on the position, area and sensitivity of the
area detector. In the present experiment, a threshold of 20 indexed
peaks, and attribute the Laue pattern analysis and stress evaluation. However, these previous
works adopt a curvature-based stress evaluation, which is more accu-
rate but possible due to single crystal cells. Since the entire scanned
region of the cell is one crystal, the change in the orientation of the
crystal within the scanned region can be used to estimate the local
curvature. The curvature can then be used to calculate the bending
strain and stress later. In the case of the current multi crystalline cell,
there can be several grains of different orientations within the scanned
region hence, local grain mis-orientation will lead to erroneous results
at grain boundaries. Hence a slightly different method was adopted to
evaluate the stress, as explained below [23].

A minimum of 5 peaks need to be indexed in the Laue pattern to
identify the crystal structure of the grain, but more peaks will improve
the orientation and strain accuracy. Once the crystal has been indexed,
the deformed and un-deformed crystal structures can be related by the
homogeneity property [23] i.e., the coordinates of an arbitrary point in
an un-deformed unit cell remain unaffected in the deformed unit cell.
Based on this condition, the deviatoric stress can be calculated as ex-
plained in [10,23,26].

The deviatoric stress is given by the Eq. (1) and the same in matrix
form is given by Eq. (2) below.

\[
\sigma'_{ij} = \sigma_{ij} - \sigma_{ii} \delta_{ij}
\]  

(1)

or

\[
\begin{bmatrix}
\sigma_{11} & \sigma_{12} & \sigma_{13} \\
\sigma_{12} & \sigma_{22} & \sigma_{23} \\
\sigma_{13} & \sigma_{23} & \sigma_{33}
\end{bmatrix}
= \begin{bmatrix}
\sigma_{11} & \sigma_{12} & \sigma_{13} \\
\sigma_{12} & \sigma_{22} & \sigma_{23} \\
\sigma_{13} & \sigma_{23} & \sigma_{33}
\end{bmatrix}
- \sigma_{ii} \begin{bmatrix}
1 & 0 & 0 \\
0 & 1 & 0 \\
0 & 0 & 1
\end{bmatrix}
\]

(2)

where \( \sigma_{ii} \) is, the hydrostatic stress given by the Eq. (3) below.

\[
\sigma_{ii} = \frac{\sigma_{11} + \sigma_{22} + \sigma_{33}}{3}
\]  

(3)

Only deviatoric stress can be calculated by Laue diffraction and to
calculate the total stress, we need hydrostatic stress as shown by Eqs.
(1) or (2). Calculation of hydrostatic stress require to evaluate the en-
ergy of one of the diffraction peaks by performing an energy scan using a
monochromator [23]. The energy scan requires a few minutes at each
scan point, which means the time of total scan increases to days. Owing
to the limited beamtime allocation to each user group in a cycle, this is
rather impossible. However, the thin plate structure, of the mc-Si cell,
let us conveniently assume plane stress condition, leading to stress
along the thickness of the cell (along direction 3) to be zero, i.e.,

\[
\sigma_{33} = 0
\]  

(4)
Substituting Eq. (4) in Eq. (1), the hydrostatic stress can be calculated as shown in Eq. (5) below and substituting it in Eq. (1) will give the full stress tensor at each scan point. [23]

\[ \sigma_{ij} = \sigma_{ij} - \sigma_i n_i \] (5)

The indexing and stress calculation process of each Laue pattern of the scan can be automated using XMAS [23] and finally a map of output quantity such as grain orientation, strain, stress etc. over the scanned region can be generated. The stress maps thus generated are presented in the Section 4.

2.3. Encapsulation polymer characterisation

Encapsulation polymers (encapsulants) are soft materials, which generally exhibit viscoelastic behaviour [7]. In such materials, the elastic modulus is a strong function of temperature and time. Especially the modulus of the most common encapsulant, EVA exhibits glass transition between ~20 °C to ~30 °C and as a consequence, its modulus increases by nearly 100 times [5]. Such an increase in the encapsulant modulus affects the stress and fracture of the solar cells in the PV modules [5,6]. In this section we aim to characterize the elastic modulus of the encapsulants, as a function of temperature, to correlate with the experimentally measured residual stresses in the mc-Si cells. Later, the measured modulus of the encapsulants are used to simulate the stress in the cells, by finite element modelling.

The encapsulants were characterized by dynamic mechanical analysis (DMA) after curing. TA Q800 DMA tester was used in film tension mode for the testing [27]. The load was applied at 1 Hz frequency and the temperature was ramped from ~60 °C to 150 °C, with a ramp rate of 5 °C/min. The mechanical properties such as storage modulus (elastic modulus), loss modulus and tan(δ) were evaluated as a function of temperature. A more detailed description of the DMA experiment was presented in the Appendix-A. Additionally, the viscosity of the raw polymers was also measured using TA Discovery Hybird Rheometer (DHR-2) [28] from melting point (~80 °C) to the max lamination temperature (160 °C) to evaluate the tendency of low relaxation of the polymers during melting (or softening) and cross linking, which is difficult to measure with film tension mode in DMA. The results of the DMA and rheology testing were discussed in Section 4 and Appendix-A.

3. Finite Element (FE) model and simulations

A simple 2D FE model with plane-strain approximation was developed to represent a local portion of the mini PV module cross section as shown in the schematic in Fig. 3 below. Since the geometry and loading around the cell with soldered copper ribbon during the module integration is symmetric, this simple model is expected to simulate the localized residual stress evolution in the cell near the soldered copper (Cu) ribbon. Actual FE model and mesh details are shown in Fig. 4. The FE model was meshed with 8-node quadrilateral elements, with quadratic shape functions, using commercial FE code, ABAQUS V.6.14 [29]. The interfaces between cell and Cu ribbon, encapsulant and glass were simulated by tie constraints and the interfaces between cell/Cu ribbon and encapsulant are simulated by standard surface to surface contacts which don’t separate once in contact. Solder between interconnect and the cell was not modelled in this simulation.

All the materials used in the model were assumed to be isotropic and linearly elastic as shown in the Table 3 [11] except the copper ribbon. It was observed that the copper ribbon undergoes considerable yielding during soldering and hence elastic plastic properties were considered as shown in Table 4 [11]. The encapsulant is also assumed to be elastic and the measured storage modulus (Fig. 5) was used as the elastic modulus. Though this model doesn’t consider the melting of the encapsulant, the temperature varying storage modulus reduces to a negligible value (~0.6 MPa), which was shown to approximate the behaviour well [11]. Viscoelastic behaviour of the encapsulant was not modelled in this simple simulation as the primary focus of this study is to evaluate the fundamental interplay between the solar cell stress and encapsulant elastic modulus. A study considering the effect of viscoelasticity of the encapsulant is underway.

The coefficient of thermal expansion (CTE) of all the materials is given in the Table 3. All the encapsulants have the CTE value (0.00027 mm/mm °C) same as EVA.

The PV module fabrication process was simulated in 2 steps as explained below [11].

1. Soldering with Cu ribbons: CTE mismatch from soldering temperature, 210 °C to room temperature, 25 °C. Only the cell and the copper ribbons on the top and bottom of the cell were present in this step of the simulation. It should be noted that the actual soldering temperature is above 220 °C, where the solder is in molten state and it attains reasonable stiffness at 210 °C.
2. Lamination of PV Module: The lamination process was simulated in the following 5 sub-steps as shown below (ref. Fig. 3(b)).
   a. Preheating to 50 °C
   b. Hot Press - vacuum pressure (0.1 MPa) application to remove air bubbles
   c. Heating to lamination temperature, 160 °C.
   d. Removal of vacuum pressure
   e. Cooling to room temperature, 25 °C.

Subsequently, an external load of 2.4 MPa was applied on the module at different temperatures, as shown in the Fig. 3(c), to evaluate the effect of encapsulants on external load transfer.

All the simulation steps were carried out assuming steady state conditions. The soldering step is important in the simulation as it captures the initial residual stress in the silicon cell before lamination and the Cu ribbon undergoes considerable yielding during lamination [30] and its stiffness decreases in subsequent simulation steps. During the lamination process, the silicon cell undergoes bending under the vacuum pressure due to the front Cu ribbon height, leading to higher bending stresses in the cell near the ribbon. This stress relaxes to a lower value upon heating to lamination temperature due to softening and expansion of the encapsulant but increases to a higher value as the laminate cools down to room temperature.

4. Results and discussion

4.1. Characterization of encapsulants

Fig. 5 shows the variation of the storage modulus with temperature, of the cured encapsulation polymers considered in this study. Referring to Appendix-A, storage modulus is the elastic modulus of the encapsulants, responsible for the load bearing and transfer. All the polymers have a similar storage modulus value at high temperature above 100 °C, but as the temperature decreases, the modulus values tend to increase differently for different polymers. Encapsulant 5 is stiffer in the temperature range from 20 °C to 100 °C and Encapsulant 3 is softer in the same range (ref Fig. 5(b)). However, it can be noticed, from the Fig. 5(b), that the Encapsulants 2, 3 and 4 have very small differences in their storage modulus from 20 °C to 100 °C. But, below 20 °C (ref Fig. 5(c)), the storage modulus of all the encapsulants tend to increase steeply due to glass transition except Encapsulant 4, which has distinctly low storage modulus throughout the range of temperatures considered. From the Fig. 5(c) it can also be noticed that the Encapsulant 2 has the highest storage modulus from 0 °C to ~20 °C and all the encapsulants have significantly different storage modulus values with in this temperature range. Hence, different encapsulants have
different modulus variation with the temperature in the range from −20 °C to 100 °C. This makes the encapsulant choice very complex, because an encapsulant which is soft at low temperatures may behave stiffer at higher temperatures (like encapsulant 3) and vice-versa.

Fig. 5(d) shows the variation of the tan(δ) of the cured polymers, over the temperature range considered. Tan(δ) is the ratio of loss modulus to storage modulus of the encapsulant and it gives the relative measure of the elastic (or viscous) behaviour of the encapsulant. For tan (δ) < 1, the behaviour is more elastic and hence all the cured encapsulants behave more elastically. Except the encapsulant 4 (at very low and very high temperatures), all the encapsulants have comparable tan(δ) values. With the knowledge of storage modulus and tan(δ), the other mechanical properties of the encapsulants such as the loss modulus and the complex modulus can be easily evaluated. Comparison of the loss modulus and detailed results of DMA for each encapsulant were presented in the Appendix-A.

Storage modulus gives insight into the elastic behaviour of the encapsulant during post lamination cooling and thereafter. However, it is also essential to understand the behaviour of the encapsulant during the lamination heating process, as the encapsulant melts (or softens) under heat and start to cross-link (in case of thermosetting polymers such as EVA). Due to these changes in the encapsulant, the vacuum pressure load transfer to the cell will be affected. The strained silicon cell, under the vacuum pressure, tends to relieve upon melting (or softening) of the encapsulant. The amount of the strain relief is inversely proportional to the viscosity of the polymer. Fig. 6 shows the viscosity of the uncured encapsulants during curing from 80 °C to 160 °C. The viscosity of the Encapsulant 5 was not shown as it did not melt even at 160 °C. It only softened above 130 °C, enabling adhesive bonding under pressure, and it exhibited a stiff thermoplastic behaviour below 120 °C. Encapsulant 4 also showed thermoplastic behaviour as its viscosity decreased till 150 °C and only a marginal increase is noticed beyond that. It has the lowest viscosity among the 5 polymers considered. The viscosity of Encapsulants 1 and 2 dropped from 80 °C to 130 °C and at 140 °C, it increased steeply, which is characteristic of cross-linking polymers. They solidified around 150 °C and beyond which the viscosity dropped abruptly due to shearing of the solidified encapsulant layer between the rheometer plates. Encapsulant 3 also showed thermosetting behaviour but it melted around 100 °C and showed much higher viscosity compared to Encapsulants 1 and 2.
4.2. Residual stress distribution measured using μSXRD

The residual stress maps (pertaining to the scan region in the Fig. 1c), obtained from μSXRD experiments on case study 1 (effect of encapsulant modulus) are shown in the Fig. 7. The magnitude of the cell residual stress in the Y-direction (across the copper ribbon) is different for different encapsulation polymers. Further, it can also be seen that

Table 3
Elastic material properties used in FE simulations.

<table>
<thead>
<tr>
<th>Material</th>
<th>Young’s modulus, E (GPa)</th>
<th>Poisson’s ratio</th>
<th>CTE (mm/mm/°C)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>Value</td>
<td>Temperature (°C)</td>
<td>Value</td>
</tr>
<tr>
<td>Silicon</td>
<td>130</td>
<td>–</td>
<td>0.28</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td></td>
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<td></td>
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<tr>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Copper (Interconnect)</td>
<td>91.5</td>
<td>−40</td>
<td>0.3</td>
</tr>
<tr>
<td></td>
<td>85.7</td>
<td>25</td>
<td></td>
</tr>
<tr>
<td></td>
<td>82</td>
<td>125</td>
<td></td>
</tr>
<tr>
<td></td>
<td>79.2</td>
<td>225</td>
<td></td>
</tr>
<tr>
<td>Glass</td>
<td>73</td>
<td>–</td>
<td>0.235</td>
</tr>
<tr>
<td>Encapsulant</td>
<td>Storage modulus from DMA was used (refer Section 4)</td>
<td>0.4</td>
<td>2.70E−04</td>
</tr>
</tbody>
</table>

Table 4
Elastic-plastic material properties of copper used in FE simulations.

<table>
<thead>
<tr>
<th></th>
<th>Yield stress (MPa)</th>
<th>Tangent modulus (MPa)</th>
<th>Temperature (°C)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>116.2</td>
<td>1000</td>
<td>−40</td>
</tr>
<tr>
<td></td>
<td>95.1</td>
<td>1000</td>
<td>25</td>
</tr>
<tr>
<td></td>
<td>82</td>
<td>1000</td>
<td>125</td>
</tr>
<tr>
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<td>79.2</td>
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4.2. Residual stress distribution measured using μSXRD

The residual stress maps (pertaining to the scan region in the Fig. 1c), obtained from μSXRD experiments on case study 1 (effect of encapsulant modulus) are shown in the Fig. 7. The magnitude of the cell residual stress in the Y-direction (across the copper ribbon) is different for different encapsulation polymers. Further, it can also be seen that

Fig. 5. (a) Storage modulus variation of the cured encapsulants with temperature, (b) Zoomed in plot from 20 °C to 150 °C, (c) Zoomed in plot from −20 °C to 20 °C, (d) Variation of tan(δ) of the cured encapsulants with temperature.

Fig. 6. Viscosity of the uncured encapsulation polymers with temperature (Encapsulation polymer 5 did not melt at 160 °C).
the sample #1–4, with Encapsulant 5 shows the highest cell residual stress, while the sample #1–2, with Encapsulant 3 shows the lowest residual stress. Cell residual stress in the sample #1–3, with front Encapsulant 4 and back Encapsulant 2 falls in between. The residual stress in the cells correlates well with the storage modulus of the encapsulants at room temperature, i.e., the sample with stiffer polymer (high storage modulus) has high residual stress in the cell and vice versa. The effect of encapsulant modulus on the cell residual stress is evident from this result and proves that the effect is significant.

Fig. 8 shows stress maps obtained from the µSXRD experiments on case study 2 (effect of encapsulant (EVA) thickness). The sample #2-1 with thicker front encapsulant has lower cell residual stress compared to sample #2-2 with thinner (regular) front encapsulant. This result is in good agreement with the electroluminescence (EL) images of full-size modules with different EVA thicknesses after mechanical load testing (MLT) as shown in Fig. 9.

The plot of max. cell residual stress vs. encapsulant modulus, at room temperature (RT), in the Fig. 10, clearly shows a linear correlation between the residual stress and the front encapsulant storage modulus. Hence, the cell residual stress increases with increase of the encapsulant modulus leading to higher probability of fracture or lower fracture load (the load at which the solar cell cracks) for the cell.

Experimental evaluation of the stresses in the encapsulated mc-Si solar cells was not reported so far and we report them for the first time using µSXRD technique. Hence, no direct comparison is available. However, the reported magnitudes of stress are comparable to that reported in the literature, for encapsulated monocrystalline silicon solar cells [10,25,26]. Further, the µSXRD technique is a well-established method for evaluation of the residual and thermo-mechanical stresses in the silicon and metallic interconnects in the micro/nano systems such as TSVs, Nano-layered interconnects etc. [17,22,23].

Fracture tests and simulations in the reviewed literature [5–7,9] and our own earlier experiments [10], reported reduced fracture load of the cells, with increased encapsulant modulus and vice-versa. Since stress drives the fracture, our results are in good agreement with the reported literature. Recently, Rowell et al. [31] show that the fracture load of cells decreases with the increase of encapsulant modulus and decrease of encapsulant thickness, which correlates very well with our findings.

It should be noted that the experimental residual stress magnitudes may be subjected to systematic and calibration errors from the µSXRD beamline hardware [32], which are often difficult to evaluate accurately. However, these errors do not affect a comparative evaluation, provided the experimental conditions for all the samples remain unchanged. In our experiments, we made sure that all the samples were subjected to identical experimental conditions and settings, such as the beamline setup, calibration, the X-ray beam size and energy. Hence, we assure that the systematic errors (if any) do not invalidate the findings of our comparative evaluation.

Further, the stress maps in Figs. 7 and 8 show that the high stress in the cell is concentrated along the lower edge of the interconnect ribbon.
and no high stress along the upper edge is shown. This is due to the μSXRD experimental setup used. In reality, there will be high residual stress, in the cell, along both the edges of the Cu interconnect ribbon. In our μSXRD experiments, the cell region near the upper edge of the interconnect ribbon was shadowed by the thickness of the Cu interconnect [10]. Hence, the X-ray beam was unable to reach there. Therefore, no diffraction signal from the cell region near the upper edge of the interconnect ribbon was recorded and hence, the stress maps show high stress, only at the lower edge of the interconnect ribbon.

Since the solar cells used in this study were made of mc-Si, the stress maps were influenced by the orientation of the individual grains in the μSXRD scan region. The grain orientation maps are shown in the Appendix-A. Effect of the grain orientation on the stress is also discussed in it.

### 4.3. Residual and mechanical stress from finite element simulations

The experimental results clearly establish the significant effect of encapsulant material on the post-lamination residual stress in the mc-Si cell. However, it cannot give information about the origin of the stress or the reason for such a high stress and how it gets affected by the encapsulant material. FE simulations are used to clarify this. Fig. 11 below shows simulated residual stresses in the silicon cell at different stages of the module-making process for the model with Encapsulant 1. The stress plots in Fig. 11 show that the maximum stress in the cell varies with each step of the lamination process. Due to mismatch of the coefficient of thermal expansion (CTE) of copper and silicon, residual stress develops in the cell upon cooling from soldering temperature to room temperature (RT), 25 °C, as shown in Fig. 11(a). Upon preheating to 50 °C during lamination, the maximum stress in the cell is slightly reduced (Fig. 11(b)). Hot pressing (application of vacuum pressure, 0.1 MPa) leads to bending of the cell over the edge of the front ribbon (glass-side (bottom) ribbon in the stress plots) as shown in Fig. 11(c), which introduces high bending stress in the cell. Upon further heating to lamination temperature, 160 °C, the encapsulant expands, softens (or melts and start to crosslink), relieving stress in the cell as shown in Fig. 11(d). At the end of the lamination process, the vacuum pressure is removed but, by then the encapsulant adhesively bonds with the interfaces and hence no difference in the cell stress is seen (Fig. 11(e)). The final step in the process is cooling of the hot laminate to RT, during which the encapsulant stiffens, regaining its solid state and in the process, induces high bending stress in the cell as shown in Fig. 11(f).

Upon external loading (uniform pressure of 2.4 MPa, as shown in the Fig. 3(c)) at room temperature, the bending stress tend to further increase as shown in Fig. 11(g).

It should be noted that the absolute magnitude of the simulated stresses may not be very accurate due to the simplified 2D FE model used in the analysis, especially during lamination heating, the encapsulant melts (or softens) filing the gaps at the corners of the Cu ribbon and silicon cell joints (ref. Figs. 11(d) to 11(g)) and this will relieve the stress in the cell partially. Hence the actual post-lamination stress values may be lower. If the encapsulant melts and flows like a liquid at lamination temperature, the bending stress induced in the cell due to hot pressing (Fig. 11(c)) will be completely relieved. But the viscosity of the molten (or softened) encapsulant (ref. Fig. 6) is very high (> 3000 Pa. s) even at max lamination temperature compared to that of liquids like oil (~0.1 Pa. s). Hence, the encapsulant acts as a viscous gel and allow only partial relaxation of the cell stress. In such a case, the simulated stress values from the present FE model are reasonably well expected to give the worst-case scenario. However, as we used a local portion around the Cu interconnect, in the mini PV module, for the standalone FE model used in this simulation (ref. Fig. 3), magnitude of the stress is overestimated. Also, the simplifications such as ignoring the solder layer (between the cell and the Cu interconnect) and plane strain model will add to this. Hence, the results of this FE model are suitable for a comparative evaluation of different encapsulants and also help to look at the sensitivities of various design and lamination parameters in a systematic manner. Further, it will also help to look at the different stages of the manufacturing process to identify the critical step which requires optimization.

The behaviour of the simulated stresses in the silicon cell at different stages of lamination process, as shown in the Fig. 11, was observed to be similar for all the encapsulants considered. The magnitude of the simulated cell stresses at different process stages for different encapsulants is shown in Fig. 12. It is observed from the plot that the

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**Fig. 9.** EL images of full size modules, (a) with thick EVA – before MLT, (b) with thick EVA – after MLT, (c) with thin EVA – before MLT, (d) with thin EVA – after MLT.

**Fig. 10.** Variation of the max. cell residual stress with the front encapsulant storage modulus at RT.

**Fig. 11.** Stress plots in different stages of the module-making process for the model with Encapsulant 1.
Fig. 11. Silicon cell stresses (for the model with Encapsulant 1) along direction 1 (ref. Fig. 3), from FEA at, (a) Post soldering, (b) Lamination preheat to 50 °C, (c) Lamination Hot Press, (d) Lamination heating to 160 °C, (e) Vacuum pressure removal, (f) Post lamination, (g) External load 2.4 MPa @ room temperature (RT).
post lamination cell residual stress is significantly different for different encapsulants. The cell stresses at the remaining process stages are similar, showing no dependence on the encapsulant. This result clearly shows that the cell residual stress upon cooling to RT is a function of encapsulant modulus.

To further reinforce the effect of the encapsulant on the cell stress, a plot of the post-lamination cell residual stress as a function of the encapsulant storage modulus at RT is presented in the Fig. 13. A linear dependence of the cell residual stress on the encapsulant modulus is noticed from Fig. 13, which matches very well with the experimental prediction, shown in the Fig. 10. The Fig. 13 also gives a relative order of encapsulants, in terms of the increasing cell residual stress at room temperature.

It would also be very important to evaluate the effect of the front and back encapsulants (ref. Fig. 3(a)) on the cell stress separately to address the optimization of module lamination more effectively. The post-lamination residual stress in the cell was simulated for different front encapsulants by keeping the back encapsulant to be Encapsulant 5. As shown in Fig. 14, the post lamination cell residual stress is a function of front encapsulant storage modulus at RT. The stress variation is almost same as the case with same front and back encapsulants (Fig. 13).

Similarly, the post-lamination residual stress in the cell was simulated for different back encapsulants by keeping the front encapsulant to be Encapsulant 5. Fig. 15 shows that the cell residual stress decreases negligibly, with increase of the back encapsulant modulus at RT. Hence, the front encapsulant effect on the cell residual stress is more significant, which matches well with our experiments (ref. Fig. 10). This is because, in our experimental samples, no back sheet was used (as explained in the Section 2.1) and the same condition was modelled in our simulations as well. Hence, the back encapsulant do not transfer any constraints to the cell, unlike the front encapsulant, which transfer the constraints of CTE mismatch from the rigid front glass to the cell during cooling to RT. In the case of a module with polymeric back-sheet, no significant difference would be expected, as the back sheet is a thin (~0.5 mm thick) polymer of stiffness much lower than the front glass. However, in the case of a glass -glass module, both the encapsulants transfer the constraints to the cell equally and the module cross-section would be symmetric about the mid plane of the cell. Hence, stress in the cells reduce and the effect of the front and back encapsulants nullify each other, causing less significant overall effect on the cell stress.
The experimental and simulation results shown so far discuss the effect of the encapsulant modulus on the cell residual stress, at room temperature. But, the PV modules under operation would be subjected to external loads and colder climates. In such a scenario, it would be even more important to evaluate the effect of the encapsulant modulus on the cell stress at lower temperatures, as the encapsulant modulus increases significantly with the lowering temperature (ref. Fig. 5). Hence, we simulated the cell stresses in the laminate subjected to an external uniform load of 2.4 MPa, at two different temperatures, RT and −10 °C for different encapsulants. The results are shown in the Figs. 16 and 17 respectively. Note that the cell stress here is the combination of the residual stress and the mechanical stress due to the external load.

Both the plots, in the Figs. 16 and 17, show that the cell stress is directly proportional to the encapsulant modulus. However, something more important to be noticed is, the relative ranking of the encapsulants, in the order of increasing cell stress. Which is different for different load application temperatures. This is because the variation in the elastic modulus with temperature (below RT) is different for different encapsulants (Fig. 5(c)). This scenario indicates the complexity of encapsulant selection. The encapsulant which is good at RT may not be good at lower temperatures. In the current study, Encapsulant 3 appears to be better at RT however Encapsulant 4 gives lower cell stress, both at RT and below RT.

Further, from Fig. 17, it can be noticed that the effect of the encapsulant modulus on the cell stress becomes less significant with the increasing encapsulant modulus, beyond a limiting value. This is because, with the lowering temperature below RT, the modulus of the encapsulant increases steeply to a higher value, losing its soft nature (glass transition). At this point, the increase in the encapsulant modulus significantly increases the module bending stiffness, along with the cell stress. As the bending stiffness is inversely proportional to the stress, the increase in the module bending stiffness tends to reduce the cell stress. Hence, the overall effect of the encapsulant modulus on the cell stress reduces as it continues to increase at low temperatures.

In summary, we report the post lamination residual stress in the mc-Si solar cells, encapsulated into PV modules with different encapsulants for the first time, using the µSXRD technique. We also performed characterization of the elastic modulus, of different encapsulants, to correlate with the cell residual stress. We further simulated the post lamination residual stress and mechanical stress in a PV module, with different encapsulants, using finite element simulations.

Both our µSXRD experiments (Fig. 10) and simulations (Fig. 13) show that the cell residual stress is directly proportional to the encapsulant elastic modulus, which means, a stiffer encapsulant increases cell stress or in other words, it reduces the fracture load ability of the cells. Similarly, a softer encapsulant reduces the cell stress and hence, leads to increased fracture load ability of the cells. These findings are in good agreement with those in the reported literature [5–9,10,31].

The µSXRD experimental (Fig. 8) and REC mechanical load testing (Fig. 9) results show that the cell stress is lower for a thicker encapsulant and vice-versa, which is again in good agreement with the reported literature [6,7,9,31].

The simulations further show that the effect of the front encapsulant on the cell residual stress (Fig. 14) is more significant and that of the back encapsulant (Fig. 15) is negligible. Which is a new observation in our study and need to be researched further. In the case of a glass-glass module, due to symmetry about the cell mid plane and transfer of equal load transfer through the front and back encapsulants to the cell during cooling, we expect that both encapsulants show equal and opposite effect on the cell stress. Due to this, the overall effect of the encapsulant on the cell may become less significant or negligible. In the case of a module with a thin polymeric backsheet (like the conventional modules), due to the lower stiffness of the back sheet compared to the front glass, we expect to see no much difference from our results.

The simulations of externally loaded modules at RT (Fig. 16) and −10 °C (Fig. 17), reinforce that the cell stress is a strong function of the encapsulant modulus. But, more importantly they show that the encapsulant need to be selected based on the entire manufacturing (max. lamination temperature) and operational range of temperatures. For example, encapsulant 3 and encapsulant 5 caused the lowest and the highest stresses, respectively at RT. However, at −10 °C, the lowest cell stress is caused by encapsulant 4 and the highest stress is caused by the encapsulant 2. In the overall temperature range, encapsulant 4 causes low stress and appears to be the best encapsulant. This observation may be used as guideline by the polymer scientists and engineers to synthesise polymers to suite the requirements of the processing and operation of the PV module.

The reported µSXRD results in conjunction with the simulations can be used, as a guideline for the selection of encapsulants, to reduce the cell stresses during manufacturing and operation. Especially in the case of thin crystalline silicon cells, as they are more fragile than the current standard silicon cells (thickness 0.18 mm), a proper choice of the encapsulant may help to sustain the thin silicon technology. The µSXRD technique can be applied to evaluate sensitivities of other constituent materials and operational parameters of the PV module and help to optimize the PV modules for higher yield and reliability. Currently, synchrotron X-ray sources occupy large area (comparable to size of a football stadium) and not portable. However, research is being done to develop portable synchrotron X-ray sources [33–37], which may be used for production quality control in the future. But, synchrotron X-
rays are dangerous as they pose severe health hazards to the living and hence, implementing them in a production-line need to be carefully evaluated. Since µSXRD gives quantitative evaluation of stresses, it can also be used as a tool to validate a much simpler stress/deformation measurement system, based on Laser technology, Raman spectroscopy etc., which can be implemented as a quality control tool in the production-line.

5. Conclusions

Residual stress in the encapsulated mc-Si cells measured using µSXRD was reported in this work for the first time. The elastic modulus and viscosity of different encapsulants was evaluated using dynamic mechanical analysis and rheological testing. Later, the effect of encapsulants on the cell residual stress was evaluated using µSXRD experiments and simplified finite element simulations. The encapsulant characterization results show that the elastic modulus of different encapsulants vary significantly with temperature, which need to be considered for module stress evaluation.

The experimental cell residual stress after lamination, as shown by the µSXRD based stress maps, is directly proportional to the encapsulant modulus at RT. The simulated cell residual stress, from the FE analyses, also follow the same trend. Both the experimental results and FE simulations show that the encapsulant 3, which is the softest at RT, causes the lowest stress at RT. The encapsulant 5, which is the stiffest at RT, causes the highest stress at RT.

The encapsulant characterization show that the elastic modulus, of all the encapsulants (except encapsulant 4), raise steeply with the decrease of temperature below RT. FE simulations of an externally loaded module with these encapsulants, at −10 °C, show a very high cell stress. Encapsulant 2 is the stiffest at −10 °C, causes the highest stress at −10 °C. Encapsulant 4 is soft with relatively moderate raise of modulus below RT, causing lower cell stress all throughout the temperature range considered. Hence, it can be concluded that the selection of the encapsulant should be based on the overall temperature range, that the module would be subjected to, during its manufacture and operation, to achieve a lower cell stress.

The experimental stress maps further show that the thickness of EVA encapsulant has significant effect on the cell residual stress. Thicker the front encapsulant (EVA), smaller are the residual stresses and vice-versa. Which is confirmed by the mechanical load tests at REC.

It was further shown by FE analysis that the front encapsulant has significant effect on cell stress while the back encapsulant effect is negligible. However, this may not be true for a glass-glass module.

Appendix A

A.1. Grain orientation maps of the scanned mc-Si cell PV modules

Silicon is an anisotropic crystalline material and hence its elastic properties depend on the grain orientation. The mc-Si solar cells, used in this study have several grains in the µSXRD scanned region (shown in Fig. 1(c)). The orientation of the silicon grains of all the scanned samples, with respect to the reference Silicon (001) plane (aligned with Z direction), are shown in the Fig. A1. It can be noticed that the crystal plane normals of the grains have different orientations starting from 0° to 45° with respect to the normal to the Si (001) plane. Comparison of the grain orientation maps with the stress maps in Figs. 7 and 8 clearly show that the stress contours are influenced by the grain orientations. However, the high stress in the cell in all the samples is aligned along the lower edge of the interconnect ribbon. Further it can also be noticed that the crystals with different orientations adjacent to the edge of the interconnect show the same level of stress, irrespective of the grain orientation. This was prominently observed in the stress maps of samples #1-1, #1-2, #1-4 and #2-2. In the samples #1-3 and #2-1, there is mostly one crystal along the lower edge of the ribbon. This is a clear indication that the high stress is governed by the local bending of the cell near the edge of the interconnect ribbon, which is also in line with the predictions of our finite element simulations, shown in the Fig. 11(f).

A.2. Properties of encapsulation polymers

The encapsulants are viscoelastic materials, whose elastic properties vary significantly with temperature, especially in the glass transition region...
as shown by the schematic in the Fig. A2(a) below. Dynamic mechanical analysis can be used to evaluate the properties of the viscoelastic materials. In the current study, the properties of the encapsulants used in this work were evaluated using the TA Q800 DMA Analyzer [27]. As the encapsulants are thin films of thickness 0.45 mm, a few (3–4) layers were cured together in the laminator using the same lamination recipe used for making the mini PV modules (Table 2) to make the samples for DMA. The sample was loaded in the film tension mode as shown by the schematic in the Fig. A2(b). A dynamic load of constant amplitude was applied at the movable clamp (Fig. A2(b)), at 1 Hz frequency, while the temperature of the test chamber was ramped from −60 °C to 150 °C at a rate of 5 °C/min. In the viscoelastic materials, the strain lags the stress by a phase angle (δ), giving rise to the complex response, represented by the complex modulus (E*). The elastic response, represented by the storage modulus (E’), and the viscous response, represented by the loss modulus (E’’) are related to the E* by the relation shown in the Fig. A2(c).

The storage modulus, E’ represents the elastic energy stored in the material, which can be recovered. The loss modulus, E’’ represents the dissipated energy. The ratio of the loss and storage moduli, (E’’/E’) is given by the ‘tan(δ)’, which is an important parameter to estimate the dominating response of the material. If tan(δ) > 1, the material behaviour is more viscous (fluid like) and, if tan(δ) < 1, the material behaviour is more elastic (solid like).

The results of the DMA of all the encapsulants are presented individually in the Fig. A3. Comparison of storage modulus and tan(δ) of the encapsulants was already presented in the main manuscript (Fig. 5). Comparison of loss modulus is presented in Fig. A4. Salient properties of the encapsulants, evaluated based on the DMA (Fig. A3) and Rheometric analysis (Fig. 6 in the main manuscript) are presented in the Table A1.
Fig. A2. (a) Schematic of elastic modulus variation of the encapsulant with temperature, (b) Schematic of the DMA test setup - film tension mode, (c) Schematic of the relationship among the complex modulus (E*), storage modulus (E'), loss modulus (E'') and the phase angle, δ.

Fig. A3. Results of Dynamic Mechanical Analysis (DMA) of the encapsulation polymers. (a) Encap1, (b) Encap2, (c) Encap3, (d) Encap4, (e) Encap5, (f) EVA.
Fig. A4. (a) Loss modulus variation of the cured encapsulation polymers with temperature, (b) Zoomed in plot from -20 °C to 20 °C, (c) Zoomed in plot from 20 °C to 150 °C.

Table A1

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<th>No.</th>
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<th>Cross-linking onset temperature/°C</th>
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<td></td>
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<td>From loss modulus peak</td>
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References


