Growth Mechanisms and Characterization of Hydrogenated Amorphous-Silicon-Alloy Films

**Final Subcontract Report
15 February 1991 – 14 April 1994**

A. Gallagher, D. Tanenbaum,
A. Laracuenta, P. Kalra
*National Institute of Standards and
Technology
Boulder, Colorado*

NREL technical monitor: W. Luft

National Renewable Energy Laboratory
1617 Cole Boulevard
Golden, Colorado 80401-3393
A national laboratory of the U.S. Department of Energy
Managed by Midwest Research Institute
for the U.S. Department of Energy
under contract No. DE-AC02-83CH10093

Prepared under Subcontract No. DD-1-11001-1

July 1994

MASTER

2B

NOTICE

NOTICE: This report was prepared as an account of work sponsored by an agency of the United States government. Neither the United States government nor any agency thereof, nor any of their employees, makes any warranty, express or implied, or assumes any legal liability or responsibility for the accuracy, completeness, or usefulness of any information, apparatus, product, or process disclosed, or represents that its use would not infringe privately owned rights. Reference herein to any specific commercial product, process, or service by trade name, trademark, manufacturer, or otherwise does not necessarily constitute or imply its endorsement, recommendation, or favoring by the United States government or any agency thereof. The views and opinions of authors expressed herein do not necessarily state or reflect those of the United States government or any agency thereof.

Printed in the United States of America

Available to DOE and DOE contractors from:

Office of Scientific and Technical Information (OSTI)

P.O. Box 62

Oak Ridge, TN 37831

Prices available by calling (615) 576-8401

Available to the public from:

National Technical Information Services (NTIS)

U.S. Department of Commerce

5285 Port Royal Road

Springfield, VA 22161

(703) 487-4650

DISCLAIMER

**Portions of this document may be illegible
electronic image products. Images are
produced from the best available original
document.**

PREFACE

This report describes the results of the third and final year of NREL contract DD-1-11001-1 with the National Institute of Standards and Technology (NIST): the contract covered a period of February 15, 1991 to April 14, 1994. The principal investigator is a NIST physicist, working at a research institute at the University of Colorado in Boulder. This institute, JILA, is jointly operated by NIST and the University of Colorado for the purpose of training students in conjunction with carrying out the NIST research mission. All other personnel working on the contract are employed by the University of Colorado through a subcontract of this NREL contract. The principal investigator, Alan Gallagher, has been primarily an atomic, molecular and plasma physicist for 30 years. This experience includes over a dozen years of studying the plasma chemistry associated with the production of hydrogenated amorphous silicon (a-Si:H) photovoltaics. These studies have led to a recognition that improved knowledge of the atomic scale character of the growing film surface is sorely needed to understand the causes of deficiencies in these photovoltaics. This includes the issue of voids and structural inhomogeneities in the final material. Consequently, the effort under this contract has been devoted largely toward this goal, based on use of a scanning tunneling microscope (STM) to obtain the desired atomic resolution of the film structure. It was not a given that one could force the high current density of a STM through these low conductivity films without damaging or altering the film. Nor was it certain that one could usefully study the surface of films grown at other facilities and exposed briefly to air during transfer. In addition to STM and deposition system construction and development, solutions were found to both of these problems during this contract and a variety of unexpected film surface morphologies were discovered.

Two post-doctoral research associates worked on the contract in its early stages. One of these, Robert Ostrom, had done his thesis research at UCLA in the area of MBE and surface science, and this expertise was very valuable in setting up the STM and related surface apparatus. The other post doctoral was Geoffrey Stutzin who had done his thesis research at Berkeley on discharge physics. His expertise and efforts were applied to the silane plasma and film deposition system. Two graduate students, David Tanenbaum and Arnaldo Laracuate, learned under these post doctorals, and then carried out most of the measurements during the remainder of the contract. A guest worker from the Hungarian Institute of Physics with enormous experience in plasma physics, Karoly Rozsa, has occasionally been available as a consultant to the contract. In the last year of the contract a new graduate student, Punit Kalra, has been constructing a new STM, which should have greatly improved characteristics and fewer problems than the commercial instrument we have had to rely on so far. On occasion, other graduate students normally working with the PI on other projects have also contributed when their assistance was needed; Edith Molenbroek in particular.

SUMMARY

Goal:

The overall objective of this work is to better understand the atomic-scale structure of glow-discharge-produced hydrogenated amorphous silicon and alloy (a-Si:H and a-Si:Ge:H) films, as well as its effect on film quality and its dependence on deposition discharge conditions. These films are known to be inhomogeneous, with many small regions of missing silicon, and we have been particularly interested in looking for how these microvoids appear at the growing film surface. The approach taken is to measure the morphology and chemical characteristics of the as-grown film surface with atomic-scale resolution using a scanning tunneling microscope (STM). Any deposition of discharge produced microparticulates should also be visible at the surface of the growing film with the STM.

Results:

We have measured the topology of the surfaces of device-quality, intrinsic and doped a-Si:H films deposited by rf discharge from silane, and by "hot-wire" CVD from a silane/He mixture. The films studied were either grown in our laboratory and observed without air exposure, or at NREL and viewed after <30 minutes of air exposure. Air exposure tests on our films indicated that this adds a small fraction of surface contaminants, but does not affect the character of the film topology. The surface topology of thick films (>100 nm) from different deposition systems is quite distinct. Many regions of the films grown in our laboratory are covered with low-slope hills, but atomically flat areas are also observed in some nearby regions. Most of these hills have surface slopes of less than 20 degrees to the horizontal, but some steep-sided valleys indicating growing voids are also seen. In contrast, the films grown at NREL were more uniform; they were completely covered with a low slope hills, without smooth areas or void forming regions. The hot wire deposited films also have hilly regions, but other areas contain steep surfaces and voids. For all films the surface roughness was typically ~1 nm rms and ~5 nm total relief in a 100 nm square region. These observations are discussed in the context of surface growth models and other data on the film surface and film voids.

TABLE OF CONTENTS

	<u>Page</u>
Introduction	1
Technical Approach	2
<i>Substrates</i>	2
<i>Deposition geometry</i>	4
<i>Tunneling parameters</i>	4
<i>Film transfer from other facilities</i>	5
Topology of a-Si:H surfaces	6
<i>PECVD NREL films</i>	6
<i>Intrinsic films deposited at NREL by the "hot wire" technique</i>	8
<i>PECVD films deposited in our chamber</i>	10
Discussion	14
<i>Comparisons</i>	14
<i>Mechanisms and causes</i>	16
<i>Comparison to ellipsometry measurements</i>	18
Conclusions	19
References	20
Figures	
Fig. 1. STM image of a H terminated Si(111) substrate	2
Fig. 2. Details of our film deposition chamber	3
Fig. 3. STM image of an intrinsic film deposited by PECVD at NREL	7
Fig. 4. High resolution image of the film in Fig. 3	7
Fig. 5. STM image of a n-doped film deposited by PECVD at NREL	9
Fig. 6. STM image of another region of the film in Fig. 5	9
Fig. 7. STM image of a film deposited at NREL by the hot wire technique	11
Fig. 8. STM image of a different region of the film in Fig. 7	11
Fig. 9. STM image of the film in Fig. 7, overcoated with 2 nm of gold	12
Fig. 10. Smooth region of a film deposited in our laboratory	12
Fig. 11. Void containing region of a film deposited in our laboratory	13
Fig. 12. Surface height range measured on various film surfaces	14

Introduction

The objectives of this project are to improve understanding of amorphous silicon and alloy film growth mechanisms, in bulk material and at interfaces, leading to improved deposition processes and improved film optoelectronic properties. In seeking the most effective way to achieve this, we note that the gas-phase plasma chemistry leading to glow-discharge a-Si:H film deposition has been studied for a long time, and the primary mechanisms and species responsible for film growth in a silane discharge have been established.¹⁻³ However, the ubiquitous character of particulate growth in the plasma is a relatively new discovery; some indications of its effect on film properties have been observed, but these are not well established.⁴ The electrical and related properties of a-Si:H have been extensively diagnosed and modeled, although the exact causes of light induced degradation are still ambiguous.⁵⁻⁷ In contrast there have been very few studies of the growing film surface where the structural properties of the film are produced. In particular, Si voids are thought to be critical to film quality, yet there have been no measurements capable of detecting if or how these arise at the growing surface. The present program is designed to obtain such information, using a scanning tunneling microscope (STM) to measure the surface characteristics of the as-grown film with atomic scale resolution.

Ellipsometry measurements^{8,9} of growing a-Si:H films have shown the existence of a ~ 2 nm thick, low Si-density layer at the growing surface, but such measurements do not determine whether this is a smooth, H-rich layer or a rough surface. We have previously shown that the 250°C growing film has a very thin (< 0.5 nm) hydrogen-rich surface,¹⁰ so most of the low density surface layer seen by ellipsometry can be attributed to a surface roughness of ~ 2 nm height range. However, this does not establish the shape of this roughness, i.e., whether it consists of narrow, steep, incipient voids or smooth low-angle undulations that do not lead to voids. This is a crucial difference, as it is commonly believed that the critical electrical properties of these films are largely determined by their inhomogeneities and voids, including those that occur rather infrequently.¹¹⁻¹⁵ Kazmerski¹⁶ and Wiesendanger et al.¹² have obtained STM measurements of a-Si:H film surfaces, showing that atomic-scale species and morphology information can indeed be detected on the surface of these amorphous materials. However, they did not study the as-deposited surface, or the intrinsic film whose properties are most critical in photovoltaics. The present program is carrying this capability toward much more extensive measurements of many surface regions -- to alloys, doped and intrinsic films and to surfaces deposited under a variety of deposition conditions.

In the first year of this contract we developed apparatus and procedures, and studied the initial deposition of a-Si:H films. This and the overall apparatus are described in our 1991-92 annual report.¹⁷ During the 1992-93 contract year we deposited intrinsic a-Si:H films by rf silane PECVD in our deposition chamber, then measured their surface topology with the STM. This is reported in our 1992-93 annual report.¹⁸ Since film properties can be highly dependent on deposition system and procedures, and our films have not been tested in devices, we have concentrated this year on measuring the surface topologies of films grown at other facilities. This required developing and testing procedures for transferring films between facilities without loss of surface information. Here we report on this development,

plus the properties of two types of films grown at the National Renewable Energy Laboratory (NREL). The first type was grown in a silane PECVD chamber that has grown high quality photovoltaics, and the second type was a "hot wire" deposition.

Technical Approach

Substrates

The initial atomic-scale morphology of a-Si:H deposited from a glow discharge will depend on the atomic-scale roughness and chemical composition of the substrate. To minimize the effect of such uncontrolled roughness and to optimize the connection to a large body of surface-science knowledge, we are currently studying a-Si:H and alloy-film deposition onto oriented, nearly atomically smooth, clean surfaces of crystal Si and GaAs. STM measurements also require passing a current through the a-Si:H film and into the substrate, so we cannot utilize silicon with a native oxide overlayer, as is often used for ellipsometry and infrared absorption studies. This may cause differences in initial growth and islanding, but for the thick films reported here this should not have a significant residual effect on the observed surface.

Fig. 1. The typical shape of an H-terminated Si(111) substrate surface produced by etching in buffered HF, as used for film deposition at NREL. The vertical scale is expanded a factor of 3 to emphasize height variations.

For depositions onto c-Si in our system, we prepare atomically flat crystal Si by heating an oxide covered Si substrate in ultra high vacuum to $>900^{\circ}\text{C}$. This is not practical for films grown at other facilities, so we have shipped thermally oxidized and annealed Si(111) samples, and these have been etched in buffered HF and rinsed in water just prior to insertion into the deposition chamber. This leaves an H-terminated surface that is atomically flat in many regions and has <0.2 nm rms of surface height variation (Fig. 1). This surface is also known to be exceptionally inert under air exposure,¹⁹ and this is an excellent mimic of the a-Si:H surface on which the film normally grows. Further verification of the topology was obtained with somewhat reduced resolution by overcoating an intrinsic a-Si:H film with ~ 4 nm of gold; this comparison will be shown below.

Fig. 2. Details of our rf discharge, film deposition chamber. All materials are stainless steel, except an alumina rf-feedthrough insulator and a molybdenum and tantalum substrate holder.

Deposition geometry

Our film-deposition region is designed and operated to mimic conditions and arrangements most commonly used to produce device-quality a-Si:H and alloy films by glow-discharge deposition. However, the special requirement of sample transfer into the STM necessitates some unusual arrangements, as seen in a detail of the deposition region shown in Fig. 2. As shown there, the sample is mounted on a holder that is transferred between the discharge region and the STM. This holder also contains separate connections to two sides of the crystal to allow resistive heating during sample preparation. Also, the substrate surface is slightly above the plane of the grounded electrode, with some slightly recessed and protruding regions nearby. When film is deposited, the entire discharge region, shower head, and sample are heated to the substrate temperature. The rf electrode is 5-cm square and 1.6 cm from the substrate electrode, defining the discharge volume and yielding a relatively uniform discharge region at the sample. Deposition rate is measured with laser interference fringes and is normally 1-2 Å/s. This discharge geometry, with grounded surfaces adjacent to all sides of the rf electrode, is "asymmetric"; the rf and grounded electrode areas differ by a factor of 2. This ensures a uniform, efficient gas flow, but lowers the sheath potential and substrate ion bombardment compared to a symmetric discharge. Industrial systems are normally symmetric, as they utilize large electrode areas compared to the surface area (electrode length/gap $\gg 1$).

Tunneling parameters

Due to the large scale of film surface roughness compared to the STM tunneling gap of 5-10 Å, we always scanned the surface in the constant-current tunneling mode. However, we frequently interspersed this with current (i) versus probe-height (z) data at a point with constant voltage (V). If the probe penetrates the sample or is covered with a-Si:H debris, this produces a very gradual and often irregular decrease in i with increasing z . When the probe was clean and current and voltage conditions were optimized, we observed stable $z(t)$ at constant i , and $i(z) \propto \exp(z/z_0)$, where the scale height z_0 is < 1.5 Å. This rapid current variation with z indicates the desired tunneling through a small vacuum gap between sample and probe, and that the topography obtained from a constant-current scan closely represents the actual surface shape. Tunneling current and voltage were normally 40 pA and 2-9 V, where we believe that a portion of this voltage appears across the film, allowing it to carry the very high STM current density.

The conductivity of intrinsic or moderately doped a-Si:H is insufficient to pass the current density of the STM at reasonable voltages, and we believe that a strong electric field develops within the sample, causing the carriers to drift through the film to the conductive substrate. It is important that scanning over different atomic surface states should cause only minor changes in the sample voltage drop, and thereby a small change in the apparent surface height. The small z_0 value described above provides some assurance that any such changes do not greatly alter the detected surface topology. A variety of tests described in our 1993 Final Report indicated that this is not a major problem.¹⁸ The very similar

topology reported below for an intrinsic and for 100 Ω cm, n-doped a-Si:H film also strongly suggests that this effect does not contribute significantly to the observed topologies.

Film transfer from other facilities

We have previously reported measurements on intrinsic a-Si:H films grown by rf discharge in our laboratory. Here the deposition is done in an attached, turbomolecular pumped chamber and film-coated substrates are transferred through a gate valve into the UHV chamber after termination of the glow discharge and deposition-gas flows. The film is exposed to minimal impurity vapors during transfer, and these do not significantly alter the exceptionally inert a-Si:H surface. The primary thrust of our efforts this year has been to extend these STM measurements to films grown in other laboratories, especially films that have been incorporated into photovoltaics with state-of-the-art efficiencies. Transferring a film between laboratories without air exposure would require significant modification to those deposition systems, so we have developed and tested a much easier approach. As now described, this has been successful and has allowed us to study films grown at other facilities.

Previous reports suggest that the a-Si:H surface is very robust against oxidation, as expected for a hydrogen passivated silicon surface.²⁰ This passivation has been studied on both porous²¹ and crystalline¹⁹ silicon, where it is found to be relatively impervious to air exposure. To study if this passivation is adequate to allow air exposure of films during transfer between laboratories, a 60-nm a-Si:H film was grown in our deposition chamber on a clean crystal-silicon substrate. This film was very similar to the 100-nm films which we have previously reported, including some rolling hills (nanometer scale) and valleys with steep ($\sim 45^\circ$) slopes.²² Auger spectra were measured to determine the relative coverage of silicon, carbon, and oxygen on the surface. The sample was then exposed for 5 minutes to a 2:1 dry nitrogen and air mixture at 1 atm, after which Auger and STM data were again recorded. STM tunneling was stable at 40 pA and 5 V, showing similar features to the original a-Si:H surface. (It is not feasible to find the same 100 nm region.) These measurements were repeated after next exposing the sample to 1 atm of air for 20 minutes, and again after 3 days of air exposure. Bias voltages of 6 V and 8 V respectively were necessary for stable tunneling in these cases, but overall topology was consistent with the original surface. This suggests the a-Si:H surface is sufficiently passivated to withstand the <30 minutes of air exposure encountered in transferring samples between laboratories using the method outlined below. The Auger data show that contamination of the sample with time after deposition is dominated by carbon. The initial as-grown film shows no detectable carbon or oxygen (<1%) at the surface. Carbon contamination grows both in the UHV chamber and during the air exposures, up to levels of $\sim 20\%$ after several days. However, this does not seem to influence the topography of the surface, although it may be involved in the necessity for using a higher tunneling voltage. Oxygen contamination was only 5% after the three-day air exposure.

From these air-exposure studies, we have ascertained that we can transfer samples from a deposition chamber to our STM chamber with considerable air exposure, without loss of atomic-scale topology information. We have therefore made these transfers by taking the samples from the deposition system into air, then placing them in a small stainless steel transfer cell and evacuating it with an oil-free pump. After transport to our laboratory, the chamber is opened and the sample is again transferred through air into our load lock chamber. The total air exposure is <30 minutes. After this procedure, we have usually obtained relatively low voltage (3-5 V, 40 pA) tunneling with a small z_0 , similar to the conditions on our films that have not been exposed to air. This further supports the absence of significant surface deterioration due to excellent surface passivation. However, after hours or days in our UHV chamber, voltages up to 9 V are sometimes required, for reasons that are not yet understood. In a variety of instances we scanned a surface region at two very different sample voltages in the 2-9 V range; the detected topography was essentially independent of the voltage.

STM topography resolution on a rough surface is limited by the sharpness of the probe tip. Except where noted, for the present data the probe tip is sharper than any features on the surface and the actual surface shape is recorded. The methods used to prepare these sharp probe tips and to establish their shape are described in our 1993 Final Report.¹⁸ The use of these unusually sharp tips with a-Si:H does frequently cause rather noisy data; this can be seen in some of the topologies reported below as an irregular, small-scale roughness. This is attributed to atomic motion and changes in surface reconstruction on the end of the probe. These effects are enhanced by the very large electric fields that occur near the STM tunneling gap.

Topology of a-Si:H surfaces

PECVD NREL films

Samples of a-Si:H film grown at the National Renewable Energy Laboratory (NREL) were produced in the rf discharge (PECVD) chamber of Yeuqin Xu, under conditions that achieve 10% initial efficiency, single i-layer solar cells. An intrinsic sample was grown at 210°C with 45 sccm silane, a pressure of ~83 Pa, and a rf discharge power that yielded ~0.2 nm/s growth rate. A n-doped film was grown by adding an additional 18 sccm of 5% phosphine in helium. Examples of surface topologies measured on the 300 nm thick, intrinsic a-Si:H film are shown in Figs. 3 and 4, for two different size regions. Relatively few regions were measured on this film, due to a tunneling instability that occurred after a few hours and was not correctable with a new probe. As already noted this may be due to occasional hydrocarbons on the surface, but this has not been verified. Very little topology variation was observed for any of the regions that were scanned. As seen in these examples, the scanned regions were uniformly covered with "hills", and had a typical full height range of ~6 nm on a 100 nm × 100 nm region. The surfaces of these hills are gently sloping, generally not more than 30° from horizontal. Figure 3 uses a 3 times expanded vertical

Fig. 3. The STM image of a 300 nm thick, $T_s = 210$ C, intrinsic a-Si:H film deposited at NREL from an rf silane discharge. The vertical scale is expanded 3 times.

Fig. 4. A higher resolution image of the film in Fig. 3. The vertical scale is not expanded.

scale, to best show the network of randomly positioned hills, of typically ~ 20 nm width. Note that each edge of each figure provides a *de facto* scan line through the figure, but with slopes exaggerated 3 times since the vertical scales have been 3 times expanded. Occasional narrow, sharp spikes of ~ 0.2 nm height occur in random positions in Fig. 3; these result from probe-surface instabilities, and are not deemed a part of the actual surface. The vertical scale in Fig. 4 is not expanded, so that it is more apparent that the hills have low slopes and the valleys between the hills are relatively smooth and do not contain growing voids.

Two regions of the n-doped, 500 nm thick NREL prepared film are shown in Figs. 5 and 6, for two different size regions. In both of these figures the vertical scale has been expanded a factor of 3 to make features more apparent. A large number of different regions were scanned on this film, and all looked equivalent to these. In essence, all were completely covered with low-angle ($< 20^\circ$) hills, at essentially random positions and with characteristic heights of 2-4 nm and widths of 20-30 nm. The full relief on all $100 \text{ nm} \times 100 \text{ nm}$ regions was 6-8 nm, while the rms height variation was 1-1.2 nm; a larger relief occurs on larger areas, as seen in Fig. 5. This is essentially the same topology observed on the intrinsic film shown in Figs. 3 and 4. It is much easier to tunnel into the ($100 \Omega \text{ cm}$) doped film, so this similarity provides additional verification that we are detecting the actual surface topology of the intrinsic films, without significant bias from different conducting regions or varying surface states. The finer scale roughness in Figs. 5 and 6 results from an unusually large vibrational coupling at the time of these data, and is not representative of the film surface.

Intrinsic films deposited at NREL by the "hot wire" technique

Films grown by decomposition of silane on a hot tungsten wire suspended directly above the substrate have been shown to yield unique, low H-content a-Si:H films with excellent electrical properties.²³ We have studied the surface topology of these "hot wire" deposited films, provided by Edith Molenbroek and A. H. Mahan, but so far only for films of $\sim 10\%$ H content that were deposited at $\sim 250^\circ\text{C}$ substrate temperature. Due to gas-safety issues at NREL in 1993, these were also deposited from SiH_4 diluted in He, rather than the pure silane deposition conditions reported in Ref. 23. The deposition conditions were optimized for this H content and gas mixture, and infrared absorption and conductivity measurements indicate that these films are of the same general quality as PECVD films of the same H content. However, SAXS measurements indicate that these hot wire films contain a much larger void fraction than PECVD films optimized by photovoltaic manufacturers.²⁴ We show the surface topology of three different regions of a 200 nm thick, hot wire deposited film in Figs. 7-9. The data in Fig. 7 were taken relatively soon after placing the substrate into the STM, when 4 V bias achieved proper tunneling behavior. Figure 8 was taken after 3 days, many intervening data sets, and reforming the probe in the STM; at that time ~ 5 V was needed for proper tunneling behavior. A 4 nm gold coating was then deposited on top of the a-Si:H film, in the same vacuum chamber, and Fig. 9 represents the type of surface topologies that were observed with this overcoating. We have deposited a similar gold coating directly onto a H-terminated crystal substrate, and found that this introduced less than

Fig. 5. A STM image of a phosphorous doped, $\sim 100 \Omega \text{ cm}$, a-Si:H film deposited by rf discharge at NREL. The vertical scale is expanded 3 times.

Fig. 6. A STM image of another region of the film in Fig. 5. The vertical scale is expanded 3 times.

0.5 nm of surface roughness. The gold coating spreads the STM current across a large area before it enters the a-Si:H film, so low voltage, quieter tunneling is obtained.

The total and rms height range of these hot wire film surfaces are similar to that on the PECVD films in Figs. 3-6. The surface in Fig. 9 also looks quite a bit like those in Figs. 3-6; it is covered with low sloping hills of similar heights and widths. But some very different topology is seen in Fig. 8. This region, and a number of other areas, were found to contain some very steep sided ridges. The ridges seen in outline in the left edge of Fig. 8 have such steep sides (45-70 degrees) that the observed slopes probably represent the probe surface rather than a steeper film surface. This suggests a long void growing below the ridge, and it is also likely that the dip between these two ridges is another long void that the probe is too wide to penetrate into. A scan of the adjacent region found that these ridges coalesced ~ 10 nm beyond the edge of this figure. Figure 7 also contains ridges, although these do not have such steep sides. Repeating features in STM scans, such as the approximately horizontal ridges in Figs. 7 and 8, often represent the probe shape as it is repeatedly reproduced above sharper surface features. There is a component of that occurring here, as established by scanning a narrow column on a Pt substrate, but most of what is seen in the figures represents the substrate surface. The probe is somewhat elongated at a $\sim 15^\circ$ angle to these ridges, so sharp features in the ridge direction are underrepresented. Figure 9 is somewhat smoother than typically obtained on the bare surface. Not many areas were scanned for this case, so this may be an accident or, in contrast to the slight roughening of the crystal Si surface, the 4 nm gold coating may have smoothed the a-Si:H surface. Nonetheless, Fig. 9 further supports the assumption that the above surface topologies are the film surface, in spite of the low conductivity of intrinsic a-Si:H.

PECVD films deposited in our chamber

The surface topology of films deposited in our laboratory from a rf silane discharge have been reported in the 1993 Annual Report. Most areas of these films were also covered with gently rolling hills of similar height and width to those reported here for PECVD and HW films. But a number of very different surface topologies were also occasionally found on those films, although our deposition conditions (250°C, ~ 70 Pa, ~ 10 sccm, 0.15 nm/s) were very similar to those used for the NREL PECVD films. The doped NREL film in particular was measured over many different areas looking for such regions, and none were found. Here we repeat, in Figs. 10 and 11, examples of these two very different surface characteristics that were observed on the films from our deposition chamber. Figure 10 shows a very large smooth area on a 400 nm thick, intrinsic film. The total height range for this $260 \text{ nm} \times 260 \text{ nm}$ region is only 4 nm, and most of that is due to a broad depression in the foreground and ~ 1 nm roughness due to scanning noise. The latter is easily discernable from actual film topology by streaks in the scan direction. This indicates a region of extremely homogeneous, void and discontinuity-free film. The other type of region that was occasionally detected is shown in Fig. 11, where a deep, steep-sided canyon, is seen. This is clearly a growing void; the cusp-shaped scan line shown in (b) results from the inability of the probe to penetrate further into the narrow canyon.

Fig. 7. A STM image of an intrinsic, $T_s \sim 250^\circ\text{C}$, 200 nm thick a-Si:H film deposited at NREL by the hot wire method using 1% silane in He. The vertical scale is expanded 3

Fig. 8. A STM image of a different region of the hot wire deposited film in Fig. 7. The vertical scale is expanded 3 times.

Fig. 9. A STM image of the hot wire deposited film in Fig. 7, after evaporating a 4 nm thick gold coating on top. The tunneling conditions are 0.4 V and 40 pA. The vertical scale is expanded 3 times.

Fig. 10. A STM image of a smooth region on a 400 nm thick, intrinsic a-Si:H film deposited by rf silane discharge in our deposition chamber. The vertical scale is expanded 3 times.

Fig. 11. A STM image of a void in another region of the a-Si:H film in Fig. 10, deposited in our laboratory. The vertical scale is not expanded.

Fig. 12. Total height difference, Δz , from the maximum to minimum on $100 \text{ nm} \times 100 \text{ nm}$ regions of various films. The range of values from films grown in our PECVD chamber are shown as solid lines; those from the NREL PECVD chamber as dashed lines.

The range of full height ranges measured on 100 nm square areas of the different films is shown in Fig. 12. Here one sees that the solid lines, representing films grown in our laboratory, have a large range of values, reflecting these very different types of surface regions. In contrast, the dashed lines that represent the NREL deposited PECVD films have a small variation in height range, as these surface regions were uniformly hilly. However, relatively few regions were scanned on the 300 nm , intrinsic film and this is currently somewhat uncertain.

Discussion

Comparisons

Intrinsic amorphous a-Si:H films are comprised mainly of 4-fold coordinated Si atoms with bond lengths and angles very close to crystalline values. However, the density is typically 5% below the crystalline density and the film typically contains $\sim 10\%$ hydrogen atoms. The simplest way to reconcile these properties is by postulating many small Si voids in the material, with H atoms bonded to the Si on the surfaces of these voids. Alternatively, the voids could be lower Si density regions, with H on dangling Si bonds within the region.

NMR and SAXS studies confirm that we are dealing with such an inhomogeneous material,¹¹⁻¹⁵ although it is not clear if these are actual Si voids or merely low Si density regions. The character of these voids may be crucial to the electronic properties and photovoltaic stability of these films, and finding the causes of void formation during film growth is a major concern. Here we report a search for the exact character of these voids as they form, through observations of the atomic-scale topography of the growing surface. We previously found growing void features on intrinsic films grown in our laboratory by rf PECVD, under conditions which yield conductivities and infrared absorption indicating good quality film, and which mimic typical device preparation conditions. Thus, we were surprised to find no such features on films prepared by rf PECVD at NREL. We searched many regions of the NREL films, particularly on the n-doped film, for the downward cusps that indicate a surface void, as seen in Fig. 11. We found only a few, in contrast to the films deposited in our chamber. Interestingly, we also found no smooth regions on the NREL films, again in contrast to the films produced in our laboratory (Fig. 10). It is possible that air exposure of the films deposited at NREL could have somehow obscured surface voids, but we know of no mechanism that would cause this. Any deposits must be reasonably conducting to alter the surface detected by the STM, and it appears unlikely that such material will deposit from air, especially not preferentially into voids. Thus, it seems that there is a real difference between void growth at the surface of these different PECVD films.

There are many subtle differences between the NREL deposition and that in our laboratory. The NREL chamber is a 4" cross with the substrate above and the rf electrode below the center. Only the substrate is directly heated to T_s , although the rf electrode generally rises to $\sim 1/2$ of the substrate temperature in such an arrangement. In our chamber (Fig. 2) all surfaces around and upstream of the substrate are heated to T_s and the substrate and rf electrode are vertical. The edges of the rf electrode in the NREL chamber are some distance from the grounded walls, so that it forms a moderately asymmetric discharge. Grounded surfaces are very close to all but the downstream electrode gap in our chamber, so it is a more asymmetric discharge. As a result, lower substrate ion-bombardment energies occur in our system, for a given discharge power density. The substrate in our system is also on a removable holder, producing a local irregularity in the substrate surface. Such irregularities affect where discharge-produced silicon particulates reside in the discharge, since these are negatively charged.⁴ The discharge voltage and power density that produces a particular film deposition rate also depends on asymmetry, as well as on the electrode gap and other factors; it is probably different in the two depositions. The silane pressure in our depositions varied from 70-100% of that used for the NREL depositions, while the gap between substrate and rf electrode is almost identical in the two chambers. The total silane mass flow in the NREL deposition is 3 times what we use, but the flow speed between the electrodes is probably less than in our chamber. It is not known which, if any, of these differences is critical, but the fact that very different film properties are obtained from these very similar depositions shows that these poorly understood, subtle differences can be very important.

A low surface-slope, hilly topology (Fig. 3) was the most common surface on the PECVD films from our laboratory. This topology was observed everywhere on the NREL PECVD

films and it occurred frequently on the hot wire (HW) deposited film. However, other areas of the HW film exhibited ridges, sometimes with very steep sides, that were unique to this film. These ridge walls in Fig. 8 could even be overhanging, since the measured slope will not exceed that on the sides of the probe. This implies that a unique form of voids are growing on this film. However, we must reiterate that these are not the same deposition conditions or T_s used in Ref. 23 to obtain exceptionally good, low H content films.

Mechanisms and causes

It is well established that a-Si:H film growth from silane glow discharge results predominantly from neutral radical reactions with the film surface.¹⁻³ These radicals, primarily SiH_3 , arrive at the surface in a cosine distribution, so they strike protruding regions of the surface more frequently than low regions. If they reacted and remained at the point of arrival, this microscopic shadowing combined with the incident angular distribution would inevitably lead to a highly porous, filamentary and useless film.²⁵ Since this is not observed, either radical surface diffusion before incorporation occurs or energetic surface ion bombardment erodes away protruding features of the growing film. Although energetic bombardment can occur for some deposition conditions, the highest (electronic) quality films are grown in low-power or triode discharges that yield minimal energetic bombardment of the film. Thus, it is clear that considerable radical surface diffusion must occur, and moreover that this diffusion must be accompanied by some degree of preferential incorporation in valleys as opposed to hilltops.²⁵ Such a preference is logical since the concave character of the valleys offers more bond sharing or multiple bonding than on the convex hilltops. The necessary radical diffusion, as well as a dangling-bond "radical" diffusion, is also reasonably explained by the inability of SiH_3 to insert exothermically into an H passivated surface.^{26,27}

A surface diffusion-reaction model can produce a surface covered with hills of a particular steepness by adjusting the difference in incorporation probability between the valleys and hilltops; this varies the strength of the leveling which competes with the roughening effect of the incident angular distribution. The characteristic hill width is then related to the diffusion distance before incorporation. This could be postulated as the cause of the hilly regions seen most frequently on all films, but it does not explain the inhomogenities we observe on all but the NREL PECVD films (especially the very smooth regions as in Fig. 10). Nor does it explain the many voids seen by SAXS in all films; the radical diffusion and preferential incorporation apparently fails to fill in many atomic scale gaps before they are covered over. Alternately, voids might form below the surface, perhaps due to strain relaxation or residual hydrogen elimination. Our present data have not detected the many small (< 1 nm) voids seen by SAXS. It is not yet clear if they do not occur on the surface, or are just not visible in our present images. It is possible that a low Si density region could conduct well enough to look the same as the rest of the film in our STM images. However, frequent tunneling instabilities from using a sharp probe on these comparatively weak films makes it difficult to detect small surface voids; a small pit must be distinguished from a sudden jump in height due to a change in the probe tip. This necessitates rescanning regions, and searching for the

most stable bias voltage. To further clarify this issue, we are planning studies on thinner, doped films that are more robust, as well as point-by-point comparisons of scans at different tunneling voltages and currents (i.e. barrier height spectroscopy).

There are many possible causes of the film surface inhomogeneities and voids we have observed. There are also many possible causes of the uniformly hilly regions besides the one described in the previous paragraph. We will describe several of these here.

1) Silicon microparticulates produced in the discharge might deposit on the surface, and if these can conduct the STM current or be covered by a conducting film they will contribute to the surface roughness as well as to film voids. It is known that such microparticulates grow under many silane discharge conditions, particularly as power is increased or gas flow is decreased.⁴ As these particles grow in size, to a few nm or so, they become negatively charged due to the much higher electron versus ion velocities in the discharge. The plasma is positively biased relative to the surfaces, so these negative ions are trapped in the discharge and can grow to micron dimensions. These larger particles generally escape, if at all, at the downstream edge of the plasma where the sheath fields are weaker and the gas flow drags them away. When the discharge is turned off they will move under a combination of gravitational, gas drag, and thermophoretic forces, which may allow some to land on the substrate. The smaller particles may occasionally be neutralized and escape to the substrate during the discharge, so these are a possible cause of film quality deterioration. Since our STM detects only the upper surface of deposited particles, these would initially appear as a steep sided hill in our images, and perhaps later as a more rounded protrusion as they are covered with additional a-Si:H. If they are not electrically well connected to the substrate, the STM probe will "crash" into them, probably followed by scanning instabilities and other problems. We have so far been unable to determine if these cause any of our observed surface irregularities or scanning instabilities.

2) Any region that can sustain more (or less) rapid than average growth, through a number of layers, will produce a hill (or trough). But to produce the structural inhomogeneities observed here an irregularity at one surface atomic site, or a small surface region, must influence the growth of many overlying atomic layers; it cannot be buried and restored to normal by the next Si-layer. This might be achieved by a surface impurity that has enhanced (or decreased) reactivity with SiH_3 , but only if the incorporation reaction frequently transfers the impurity up to the new surface. This might be possible for surface-bonded OH or NH_2 radicals, for example. The average impurity concentration in intrinsic a-Si:H films is $\sim 10^{18}/\text{cm}^3$, which is sufficient to thus induce reported void densities. However, most studies of the electronic properties of intrinsic a-Si:H suggest minimal correlation with impurity concentrations, for reasonably low concentrations, arguing against such a mechanism.

3) Crystal growth rates vary considerably in different directions and at step edges, and such bond orientation effects could also occur for SiH_3 reactions or dangling bond residence on the H covered a-Si:H surface. The short-range order in a-Si:H may be sufficient to maintain effective orientations across nm scale distances, and this could lead to unequal growth rates

over a similar or larger width range. This could combine with incomplete diffusional compensation of shadowing effects to yield many small voids.

4) A less reactive surface site can induce a small void. (A CH_3 attached to a Si surface atom is a well-known example in a-Si:C:H alloy films; the CH_3 is often buried intact.²⁸) Shadowing will tend to expand and propagate this void upward through successive Si layers, while SiH_3 and dangling bond diffusion counters this. In addition to the impurities mentioned above, a less reactive surface site might be caused in pure Si films, as opposed to alloys, by a particularly strong Si-H-Si bonding arrangement at the surface.

5) The canyon-like incipient voids we have detected (e.g. Figs. 8 and 11) suggest an entire void surface reaching down into the film. Although this looks somewhat like a finer-scale form of "columnar" films,²⁹ we do not think these are directly related. Those columnar films were grown in very different discharges under "physical deposition" conditions, where shadowing effects dominate and minimal SiH_3 deposition occurred. Perhaps, in the present case, impurities tend to coalesce into such regions, preserving diminished growth rates. SiH_3 incorporation on surfaces facing the plasma might be enhanced relative to inside deep canyons, causing initially small voids to grow laterally as well as upward as additional layers of film deposit.

Comparison to ellipsometry measurements

Spectroscopic ellipsometry provides a very sensitive method of obtaining thin film growth and surface layer information, particularly at interfaces which are very important in photovoltaics. It is capable of thickness accuracy of ~ 0.1 nm and can also establish the dielectric constants of film layers. It provides only averages over large areas, and cannot distinguish the shapes of surface irregularities as the STM can, but it can give a continuous picture of the growing film throughout the growth process as well as chemical bond information. However, it is not clear if ellipsometry can isolate surface roughness from other effects for thicker films (> 10 nm). It would be highly desirable to combine complementary information with the present data, which provides a very detailed picture of the film surface shape, but for only a few film thicknesses and deposition conditions. Reference 9 and a number of other studies have used infrared spectroscopic ellipsometry to establish the chemical bonding characteristics of growing a-Si:H films. As these do not involve the surface topology, they are not directly relevant to the present study and will not be discussed here.

Collins and his collaborators have studied the surface roughness of a-Si:H and alloy films produced by a variety of deposition techniques. In Ref. 8 the effective a-Si:H surface roughness during the initial 6.5 nm of film growth was studied, under ostensibly the same PECVD deposition conditions as those used by ourselves and by NREL. The films were deposited onto crystal silicon native oxide, or on a thin Cr overlayer which yielded equivalent film growth results. The data indicated an initial island formation and nucleation period, during which a low average-density surface layer grew. This low density layer is

considered to be due to a rough surface, so that the layer volume is partly filled with a-Si:H. After island coalescence at ~ 2 nm thickness, a flat bulk film grew linearly in thickness with the low density layer on top. If the low density layer was assumed to contain 50% film and the rest vacuum, then a layer thickness of ~ 1.2 nm was obtained, varying little for a variety of T_s and deposition rates. In Ref. 30 a "roughness overlayer" thickness, equivalent to the low density layer thickness, was found to increase from ~ 0.8 nm on a 10 nm thick film to ~ 1.2 nm on a 600 nm thick film. We assume that the corresponding rms roughness is $\sim 1/3$ as large, or ~ 0.4 nm on the thick film.

We cannot pass a tunneling current through native oxide or into glass, so we cannot study the initial growth nucleation on native oxide. We have studied the surface of a 10 nm thick a-Si:H film grown on a crystal GaAs surface and found it to be very flat,²² with ~ 0.5 nm full height range (0.1 nm rms) over $100\text{ nm} \times 100\text{ nm}$ areas. This implies that island growth is much less on this surface, and in addition, this a-Si:H film grew very smoothly to this 10 nm thickness. The low density surface layer of Ref. 8 implies ~ 0.4 nm rms height variation over a very large area. This is a factor of 4 greater roughness than we observed on GaAs, and may be due to the different substrates if the film growth to 6.5 nm on SiO_x was not sufficient to smooth out the initial island surfaces. For thicknesses greater than 300 nm, we typically observe ~ 1.1 nm rms roughness on $100\text{ nm} \times 100\text{ nm}$ areas, increasing to ~ 1.7 nm on $400\text{ nm} \times 400\text{ nm}$ areas. The ellipsometry data of Ref. 30 yielded an rms surface roughness of ~ 0.4 nm for 600 nm thick films, assuming a factor of $1/3$ compared to total thickness. This represents a very large surface area, for which the 1.7 nm value from our 400 nm regions would be a lower limit. Thus, we detect a 4 times lower roughness on the 10 nm film and at least a 4 times greater roughness on the thick films. It would be useful to study the same films, to better understand the causes of these differences.

Conclusions

This STM-based investigation of a-Si:H surface topographies has found clear differences between a-Si:H films produced in different deposition systems and by different techniques. The biggest difference is that the NREL PECVD film, which has been optimized for photovoltaic efficiency, is uniformly hilly across the entire film, whereas those prepared in our laboratory and by the hot wire method have different surface characteristics in different regions. While film uniformity appears highly desirable, PECVD films deposited in our chamber contain some regions with much smoother surfaces than any on the NREL PECVD films, suggesting that a more homogeneous and void free material exists, although we do not know if it could be produced throughout an entire film. Another difference is that relatively low sloping surfaces were found almost everywhere on the NREL PECVD films, with very few indications of void forming regions. In contrast, a significant fraction of the scanned surfaces on the PECVD films from our chamber and on the hot wire film had steeply sloping and cusped regions, indicating growing voids.

From the comparisons in the previous paragraph, it is apparent that STM images can distinguish unique aspects of and differences between a-Si:H films, and can discern both

large scale and small scale film inhomogeneities. However, this does not indicate how to make a better film for a photovoltaic cell, nor is STM imaging a simple or readily available diagnostic technique. What is needed is to determine the causes of these film differences, and how the surface and bulk properties can be improved. For this purpose we are now investigating the dependence of these surface characteristics on deposition methods and parameters. (Of course, more traditional film quality parameters are also measured for the same films.) As implied by our lengthy discussion in the previous section, we are very suspicious of the role played by nm to μm size Si particles that are formed in the plasma under almost all discharge conditions. Our present investigations are therefore concentrated on how these affect the film, and on how it may be possible to decrease their accumulation in an rf discharge. In essence, we have now developed a film diagnostic that can obtain unique information about these inhomogeneous films, but its value will only be known once it is applied to developing better film growth techniques.

References

1. D. A. Doughty and A. Gallagher, J. Appl. Phys. 67, 139 (1990).
2. J. R. Doyle, D. A. Doughty, and A. Gallagher, J. Appl. Phys. 68, 4375 (1990).
3. N. Itabashi, N. Nishiwaki, M. Magane, S. Naito, T. Goto, A. Matsuda, C. Yamada, and E. Hirota, Jpn. J. Appl. Phys. 29, L505 (1990).
4. A. A. Howling, L. Sansonnens, J.-L. Dorier, and Ch. Hollenstein, J. Appl. Phys. 75, 1340 (1994) and references therein.
5. J. D. Joannopoulos and G. Lucovsky, Eds., *The Physics of Hydrogenated Amorphous Silicon, I and II, Topics in Applied Physics* (Springer, Berlin, 1984), Vols. 55, 56.
6. E. F. Demichelis, Ed., *Physics and Application of Amorphous Semiconductors* (World Scientific, Singapore, 1988).
7. R. A. Street, *Hydrogenated Amorphous Silicon* (Cambridge solid state science series), (Cambridge University Press, 1991).
8. Y. M. Li, I. An, H. V. Nguyen, C. R. Wronski, and R. W. Collins, Phys. Rev. Lett. 68, 2814 (1992).
9. N. Blayo and B. Drevillon, Appl. Phys. Lett. 57, 786 (1990).
10. G. H. Lin, J. R. Doyle, M. He, and A. Gallagher, J. Appl. Phys. 64, 188 (1988).
11. A. H. Mahan, D. L. Williamson, B. P. Nelson, and R. S. Crandall, Phys. Rev. B 40, 12024 (1989); Solar Cells 27, 465 (1989); Appl. Phys. Lett. 55, 783 (1989).
12. R. Wiesendanger, L. Rosenthaler, H. R. Hidber, H. J. Güntherodt, A. W. McKinnon, and W. E. Spear, J. Appl. Phys. 63, 4515 (1988).
13. H. Wagner and W. Beyer, Solid State Commun. 48, 585 (1983).
14. P. C. Tayler, in *Semiconductors and Semimetals, Vol. 21C: Hydrogenated Amorphous Silicon*, J. I. Pankove, Ed. (Academic Press, Orlando, 1984), p. 99.
15. V. P. Bork, P. A. Fedders, D. J. Leopold, R. E. Norberg, J. B. Boye, and J. C. Knights, Phys. Rev. B 36 9351 (1987).
16. L. L. Kazmerski, Ninth E. C. PV Solar Energy Conf., Freiburg, Germany, September 25, 1989.
17. NREL/TP-411-5326, Feb. 1993, Acc. No. DE93000071.

18. NREL/TP-411-5749, Aug. 1993, Acc. No. DE93018202.
19. G. S. Higashi, R. S. Becker, Y. J. Chabal, and A. J. Becker, *Appl. Phys. Lett.* 58, 1656 (1991).
20. R. W. Collins, C. J. Tuckerman, C. Y. Huang, and H. Windischmann, *J. Vac. Sci. Technol. A* 3, 2077 (1985).
21. C. Tsai, K. H. Li, J. Sarathy, S. Shih, J. C. Campbell, B. K. Hance, and J. M. White, *Appl. Phys. Lett.* 59, 2814 (1991).
22. G. C. Stutzin, R. M. Ostrom, A. Gallagher, and D. M. Tanenbaum, *J. Appl. Phys.* 74, 91 (1993).
23. A. H. Mahan, J. Carapella, B. P. Nelson, R. S. Crandall, and I. Balberg, *J. Appl. Phys.* 69, 6728 (1991).
24. D. L. Williamson, private communication.
25. K. K. Gleason, K. S. Wang, M. K. Chen, and J. A. Reimer, *J. Appl. Phys.* 61, 2866 (1987).
26. R. Robertson and A. Gallagher, *J. Appl. Phys.* 59, 3402 (1986).
27. D. A. Doughty, J. R. Doyle, G. H. Lin, and A. Gallagher, *J. Appl. Phys.* 67, 6220 (1990).
28. J. Sotiropoulos and G. Weiser, *J. Non-Cryst. Solids* 92, 95 (1987), and references therein.
29. J. C. Knights and R. A. Lujan, *Appl. Phys. Lett.* 35, 244 (1979); J. Knights, *J. Non-Cryst. Solids* 35/36, 159 (1980).
30. R. W. Collins and J. M. Cavese, *J. Appl. Phys.* 60, 4169 (1986); R. W. Collins, *J. Vac. Sci. Technol. A* 4, 514 (1986), relatively few regions were scanned on the 300 nm, intrinsic film and this is currently somewhat uncertain.

REPORT DOCUMENTATION PAGE			Form Approved OMB NO. 0704-0188	
Public reporting burden for this collection of information is estimated to average 1 hour per response, including the time for reviewing instructions, searching existing data sources, gathering and maintaining the data needed, and completing and reviewing the collection of information. Send comments regarding this burden estimate or any other aspect of this collection of information, including suggestions for reducing this burden, to Washington Headquarters Services, Directorate for Information Operations and Reports, 1215 Jefferson Davis Highway, Suite 1204, Arlington, VA 22202-4302, and to the Office of Management and Budget, Paperwork Reduction Project (0704-0188), Washington, DC 20503.				
1. AGENCY USE ONLY (Leave blank)	2. REPORT DATE July 1994	3. REPORT TYPE AND DATES COVERED Final Subcontract Report-15 February 1991 - 14 April 1994		
4. TITLE AND SUBTITLE Growth Mechanisms and Characterization of Hydrogenated Amorphous-Silicon-Alloy Films		5. FUNDING NUMBERS C: DD-1-11001-1 TA: PV431101		
6. AUTHOR(S) A. Gallagher, D. Tanenbaum, A. Laracuente, P. Kalra				
7. PERFORMING ORGANIZATION NAME(S) AND ADDRESS(ES) National Institute of Standards and Technology Boulder, Colorado		8. PERFORMING ORGANIZATION REPORT NUMBER		
9. SPONSORING/MONITORING AGENCY NAME(S) AND ADDRESS(ES) National Renewable Energy Laboratory 1617 Cole Blvd. Golden, CO 80401-3393		10. SPONSORING/MONITORING AGENCY REPORT NUMBER TP-411-6864 DE94011842		
11. SUPPLEMENTARY NOTES NREL Technical Monitor: W. Luft				
12a. DISTRIBUTION/AVAILABILITY STATEMENT		12b. DISTRIBUTION CODE UC-271		
13. ABSTRACT (Maximum 200 words) This report describes work performed to better understand the atomic-scale structure of glow-discharge-produced a-Si:H, a-Ge:H, and a-Si:Ge:H films; its effect on film quality; and its dependence on deposition discharge conditions. Hydrogenated a-Si films are deposited from a silane rf discharge onto atomically flat crystal Si and GaAs substrates. The substrates are then transferred in vacuum to a scanning tunneling microscope, where the atomic-scale surface morphology is measured. The films were deposited using device-quality deposition conditions; IR absorption, σ_i , and σ_o indicate high-quality intrinsic films. From the thickness dependence of the surface morphology, we determined that the films initially conform smoothly to an atomically flat Si or GaAs substrate, but as the thickness increases the roughness steadily increases to approximately 10% of the length of the scanned region. The surface of 100-400-nm-thick films is highly inhomogeneous, with steep hills and canyons in some areas and large atomically smooth regions in others. These unexpectedly large surface irregularities indicate severe and often connected void structures in the growing film, as well as relatively limited-range surface diffusion of the incorporating SiH ₃ radicals. On the other hand, large areas of atomically flat surface were occasionally found, indicating the possibility of growing a homogeneous and compact amorphous film if appropriate growth conditions could be discovered.				
14. SUBJECT TERMS characterization ; amorphous silicon ; alloys ; photovoltaics ; solar cells		15. NUMBER OF PAGES 28		
		16. PRICE CODE A03		
17. SECURITY CLASSIFICATION OF REPORT Unclassified	18. SECURITY CLASSIFICATION OF THIS PAGE Unclassified	19. SECURITY CLASSIFICATION OF ABSTRACT Unclassified	20. LIMITATION OF ABSTRACT UL	