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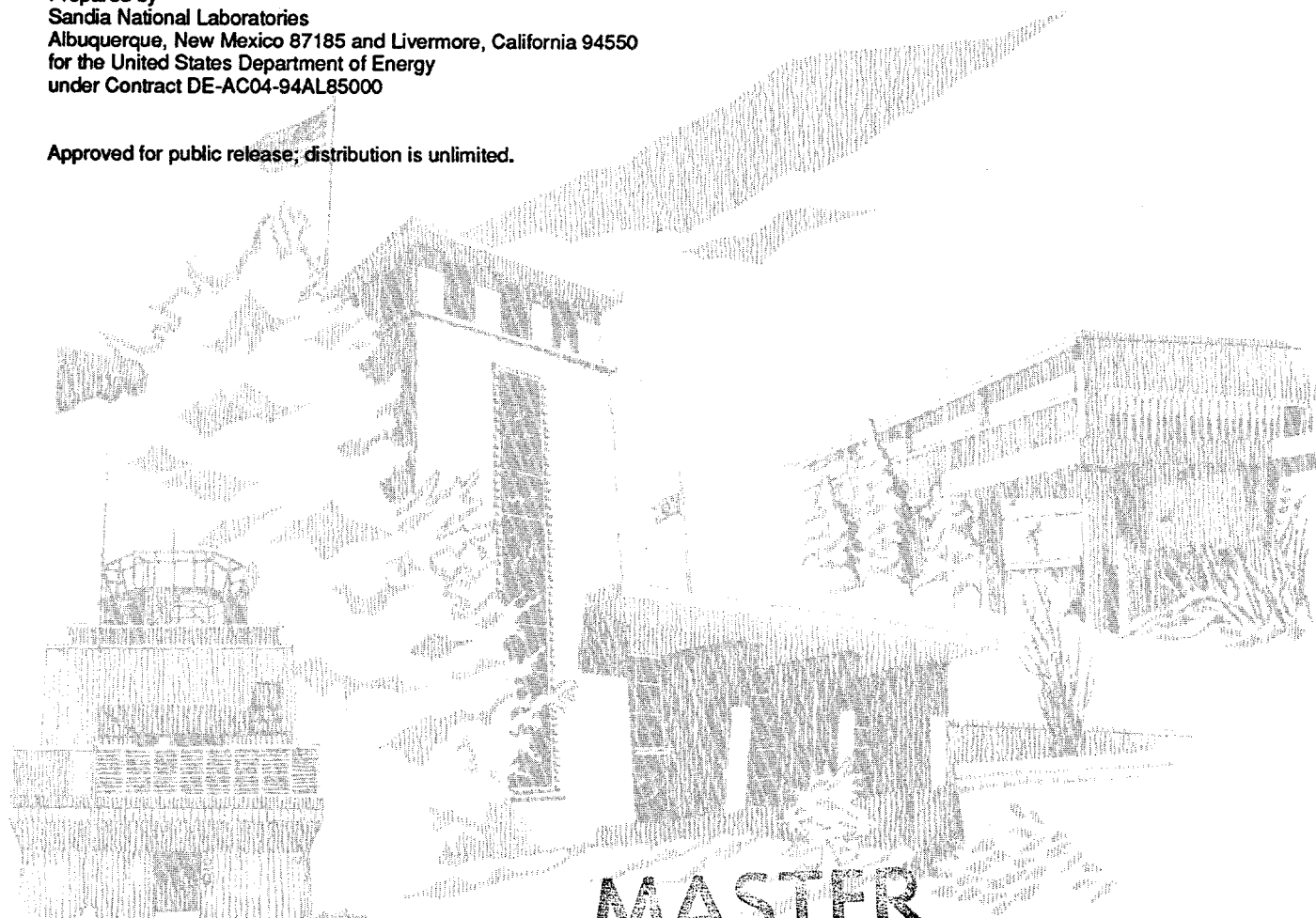
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## Dynamics of Nucleation in Chemical Vapor Deposition

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## Dynamics of Nucleation in Chemical Vapor Deposition

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### Abstract

We study the evolution of layer morphology during the early stages of metal chemical vapor deposition (CVD) onto Si(100) via pyrolysis of  $\text{Fe}(\text{CO})_5$  below  $250^\circ\text{C}$ . Scanning tunneling microscopy (STM) shows that nuclei formation is limited by precursor dissociation which occurs on terraces, not at step sites. Also, the average size of clusters formed during CVD is larger than for Fe growth by evaporation (a random deposition process). Based on STM data and Monte Carlo simulations, we conclude that the CVD-growth morphology is affected by preferential dissociation of  $\text{Fe}(\text{CO})_5$  molecules at existing Fe clusters - an autocatalytic effect. We demonstrate that nucleation kinetics can be used to control formation of metal nanostructures on chemically tailored surfaces. Reactive sites on Si (001) are first passivated by hydrogen. H atoms are locally removed by electron stimulated desorption using electrons emitted from the STM tip. Subsequent pyrolysis of  $\text{Fe}(\text{CO})_5$  leads to selective nucleation and growth of Fe films in the areas wher H has been removed.

Keywords: Chemical vapor deposition, surface chemistry, nucleation, morphology, scanning tunneling microscopy, silicon, iron

The fabrication of thin-film structures often relies upon controlling or manipulating the processes of nucleation and growth. Therefore, it is critical to have an understanding of atomic-scale events taking place during the earliest stages of deposition. In the past, most atomistic studies of thin-film morphology have focused on growth by molecular beam epitaxy (MBE).<sup>1-5</sup> These investigations typically involve monitoring the temporal evolution of certain morphological features, such as islands in the submonolayer coverage regime, and kinetic processes affecting MBE growth are identified by comparing the morphology for different substrate temperatures and deposition rates. These studies have contributed greatly to the understanding of MBE. However, less emphasis has been directed at characterization of other growth techniques, in particular, the development of film morphology during chemical vapor deposition (CVD).<sup>6-11</sup>

It is not surprising that only a small amount of experimental work<sup>7-11</sup> has focused on the development of morphology during CVD, considering the complex nature of this growth technique. In conventional condensation theory, nucleation rates are limited by energetic barriers associated with the formation of critical-size nuclei, while morphology is controlled by adatom mobility.<sup>12</sup> In contrast, CVD may involve additional processes, which may control both formation of nuclei and the evolution of structure. Adsorption, desorption, site-specific chemical reactions, and transport can each affect film structure during CVD, depending on the deposition parameters. Furthermore, growth may involve several species including precursor molecules, adatoms, and dissociation fragments.

In this letter, we present a study of a model CVD system: Fe deposition on Si(100) by pyrolysis of  $\text{Fe}(\text{CO})_5$ . The goal of this work is to understand how various kinetic processes affect film morphology in a system controlled by chemical reactions. Using scanning tunneling microscopy (STM), we monitor the nucleation sites and make detailed measurements of the cluster density and the average cluster size as a function of coverage and substrate temperature. We identify the important kinetic processes affecting morphology by making comparisons of the CVD grown films to those grown by Fe evaporation and by use of kinetic Monte Carlo simulations.

We choose to study the pyrolysis of  $\text{Fe}(\text{CO})_5$  on Si(100) for several reasons. This reaction is known to be relatively clean, for no C or O has been detected using Auger electron spectroscopy<sup>13,14</sup> and x-ray photoelectron spectroscopy<sup>15</sup> after exposing  $\text{Fe}(\text{CO})_5$  to hot Si substrates fixed at the same temperatures used in this work. Furthermore, several key kinetic parameters have already been determined such as the activation energy for desorption of  $\text{Fe}(\text{CO})_5$  precursor molecules<sup>16,17</sup> and the activation energy for dissociation on an Fe surface.<sup>13</sup> This makes attempts to model the kinetics of growth tractable. Also, identifiable changes in the initial growth morphology are expected to develop over a range of low substrate temperatures, at which the relative rate of silicide formation may be small and the surface mobility of Fe atoms on Si should be low.

The experiments involve depositing Fe layers within an ultra-high vacuum (UHV) system equipped with a scanning tunneling microscope.<sup>18</sup> Si(100) samples (n-type (P), 0.1-1.0 W-cm) are annealed to 1250°C in-situ to remove the surface oxide and impurities. The substrates selected for subsequent deposition have less than ~10% "vacancy-type" surface defects with no evidence of extended (2 x n) defects. Before growth, the Si samples are equilibrated 1 hour at a temperature below 250°C.<sup>19</sup> A freshly distilled

$\text{Fe}(\text{CO})_5$  source is loaded onto the vacuum system during this temperature equilibration step. The source is freeze-pumped to remove impurities and then allowed to warm to room temperature. A controlled partial pressure of  $\text{Fe}(\text{CO})_5$  is introduced into the vacuum system using a UHV leak valve. In the current work, we used one of two different pressures:  $1 \times 10^{-8}$  Torr or  $5 \times 10^{-8}$  Torr.<sup>20</sup> Films are also deposited onto clean Si(100) substrates by Fe evaporation for comparison to CVD growth. The evaporation source consists of a W filament wrapped with an Fe ribbon.

Images are deemed suitable for measurement if Si dimer rows and steps are easily identified. All images analyzed in this work were taken using a sample - tip bias of - 2.5 V and a tunneling current of 0.15 nA. For the purpose of determining coverage, we assume that the bright features shown in STM images are clusters of Fe atoms. Individual atoms within a cluster are not clearly resolvable. The average cluster size measured from numerous STM images is expressed in  $\text{\AA}^2$ . The film coverages measured using STM in this way are within a factor of three of quantitative measurements using heavy ion backscattering spectrometry.<sup>21</sup>

The morphology characteristic of CVD growth at temperatures less than  $250^\circ\text{C}$  is displayed in Figs. 1.a. and 1.b. These images show Si surfaces after exposure to 0.5 and 5.0 Langmuirs of  $\text{Fe}(\text{CO})_5$  (substrate maintained at  $215^\circ\text{C}$ ). At low coverages, the morphology consists of isolated clusters distributed on Si terraces, showing no preference for nucleation at steps. The cluster nucleation rate is very sensitive to substrate temperature. In Fig. 2., we plot the cluster density versus exposure for different growth temperatures, covering a range of  $\sim 100^\circ\text{C}$ . In addition, no reaction is observed after lengthy exposures ( $\sim 100\text{L}$ ) at  $30^\circ\text{C}$ . This behavior is typical of thermally activated chemical reactions and indicates that formation of new nuclei is rate limited by dissociation of the precursor molecules on the surface. We also find that the average cluster size increases with coverage, as can be seen in a comparison of Figs. 1.a. and Fig. 1.b. The cluster density and average size remains unchanged after lengthy post-growth anneals for all coverages, i.e. no coarsening occurs at temperatures less than  $250^\circ\text{C}$ .

In order to elucidate the role of processes that involve the chemistry of the precursor, we performed experiments in which Fe was simply evaporated onto Si(100) substrates at the same growth temperatures. In evaporation, Fe atoms are deposited randomly on the surface. As can be seen in Fig. 1., the morphology developed during evaporation is different from CVD growth. Namely, a smaller average cluster size is consistently found on samples grown by evaporation - summarized in Fig. 3. The slow increase in the cluster size of evaporated films with dose and the fact that the islands do not coarsen, suggests that Fe adatoms are immobile at these temperatures.

The difference between the CVD and evaporation morphologies arise from thermally activated processes specific to the chemistry of the precursor. In particular, we find that there is a differential dissociation probability such that the precursor dissociates preferentially at an existing nucleus (an autocatalytic effect).

We have employed kinetic Monte Carlo simulations of the growth process to show that a differential reaction probability is necessary to produce large islands for CVD growth.<sup>22</sup>  $\text{Fe}(\text{CO})_5$  molecules arriving on top or at the edge of an existing island, either by diffusion of a mobile precursor or by direct impingement, must have a higher dissociation probability than those precursor molecules arriving at Si substrate sites to obtain the

experimentally observed structure. This is true regardless of whether or not we allow the weakly-bound precursor molecules to be mobile. In the simulations we assume that the activation energy for dissociation on Si is  $\sim 3\times$  greater<sup>23</sup> than the activation energy for dissociation onto a preexistent Fe cluster, previously measured to be 0.14 eV.<sup>13</sup> The barrier to dissociation on the Si surface used in the simulations is a rough estimate taken from the STM data. The difference in island morphology generated by preferential dissociation is shown in Fig. 4. In this figure, the island morphology at similar coverages is contrasted for the case of differential dissociation probability (Fig. 4.a. and 4.b.) with that assuming uniform dissociation probability (Fig. 4.c. and 4.d.). The autocatalytic behavior results in an incubation period followed by rapid linear growth, as observed in bulk CVD experiments.<sup>24</sup> Uniform dissociation probability produces a larger density of small islands, similar to the experimental observations for Fe evaporation.

We also use Monte Carlo simulations to test for more subtle kinetic effects involving the precursor transport mechanism(s). Two different mechanisms that may affect morphology are considered. This includes transport from the gas phase and the process of molecular surface diffusion. In the first case, the precursor either dissociates upon impact, or does not stick and is simply reflected from the surface. In the second, a large fraction of physisorbed  $\text{Fe}(\text{CO})_5$  molecules diffuse and either dissociate or desorb. We consider the case for having mobile carbonyl molecules because previous thermal desorption measurements of  $\text{Fe}(\text{CO})_5$  from Si indicate an activation barrier for desorption of only 0.35 eV,<sup>17</sup> suggesting a very weakly-bound short-lived molecular precursor, with very low steady-state coverage during exposure.

In the simulations, we assume that the activation energy for surface diffusion of the precursor is about 1/5 that for desorption and calculate the average number of hops the molecule takes before it either desorbs or dissociates. The simulations demonstrate that the precursor residence time and average diffusion distance *increase* with decreasing temperature. This result arises from the fact that although the diffusion rate decreases at lower temperatures the desorption rate decreases more rapidly so that the precursor makes on average more hops before desorbing. This leads to a larger island size at lower temperatures, which is opposite to the usual temperature dependence of island size found during MBE growth. We also investigated the effect of direct impingement and dissociation of the precursor without physisorption and diffusion. In this case, the differential dissociation probability due to the nature of the landing site is responsible for island growth. The simulations indicate that this mechanism should also lead to larger islands at lower temperatures, due to the autocatalytic effect.

At this time, the experimental evidence is insufficient to determine which of these two transport mechanisms is affecting morphology. It is difficult to identify the presence of a mobile precursor species because the same general trends are predicted from simulations (i.e. a larger average cluster size for autocatalytic growth and an increase in cluster size with decreasing temperature are predicted to occur with or without a mobile precursor species present). In addition, the STM data in this work show no obvious temperature dependence to the island morphology when comparing images at similar coverage.(Fig. 3.) There is a marked change in the nucleation rate with temperature, as shown in Fig. 2. However, this is due to the activated precursor dissociation process, rather than an adsorption-diffusion-dissociation process. The lack of a temperature

dependence for the island morphology in these experiments could be due to the rather large errors associated with measuring island size. Perhaps the mobile precursor is so weakly bound and the activation energy for diffusion sufficiently small that over the limited temperature range studied here, effects due to precursor mobility are not resolvable.

Preliminary analysis of the simulated island size distributions suggests that there may be differences for the mobile precursor versus the direct impingement mechanism. Direct impingement results in continuous nucleation in uncovered areas of the substrate, while a mobile precursor mechanism results in a denuded zone around existing islands due to incorporation at island edges. This leads to a larger number of small islands for direct impingement and is the subject of future work.

Intimate knowledge of the surface kinetics of nucleation allows us to tailor the substrate for precise control of film growth and structure. As an example we have demonstrated selective area growth of metal nanostructures by control of nucleation kinetics. From the above studies we have shown that reaction of the  $\text{Fe}(\text{CO})_5$  molecular precursor occurs primarily at dangling bond sites associated with Si dimers on the (100) surface, and that reaction on existing Fe clusters is even more favorable. Then nucleation can be controlled by controlling the availability of Si dangling bond sites. Dangling bond sites are first passivated by adsorption of atomic hydrogen. H atoms are then desorbed locally by electron stimulated desorption. In this process, electrons field emitted from the STM tip in close proximity to the sample can define a feature with linewidth less than 10 nm. This exposes reactive dangling bond sites on the surface which can serve as Fe nucleation sites. Following H stripping the sample is heated and exposed to  $\text{Fe}(\text{CO})_5$  at 275 C, resulting in localized nucleation of Fe, shown in Fig. 5.

## References:

- <sup>1</sup> Y.W. Mo, J. Kleiner, M.B. Webb, and M.G. Lagally, *Phys. Rev. Lett.* **66**, 1998 (1991).  
Y.W. Mo, B.S. Swartzentruber, R. Kariotis, M.B. Webb and M.G. Lagally, *Phys. Rev. Lett.* **63**, 2393 (1989).
- <sup>2</sup> M.D. Johnson, C. Orme, A.W. Hunt, D. Graff, J. Sudijono, L.M. Sander, and B.G. Orr, *Phys. Rev. Lett.* **72**, 116 (1994).
- <sup>3</sup> J.A. Stroscio and D.T. Pierce, *Phys. Rev. B* **49**, 8522 (1994). J.A. Stroscio, D.T. Pierce and R.A. Dragoset, *Phys. Rev. Lett.* **70**, 3615 (1993).
- <sup>4</sup> D.D. Chambliss and K.E. Johnson, *Phys. Rev. B Rapid Comm.* **50**, 5012 (1994).
- <sup>5</sup> R.Q. Hwang, J. Schroder, G. Gunther and R.J. Behm *Phys. Rev. Lett.* **67**, 3279 (1991).
- <sup>6</sup> G.S. Bales, A.C. Redfield, and A. Zangwill, *Phys. Rev. Lett.* **62**, 776 (1989) and references therein.
- <sup>7</sup> W.L. Gladfelter, *Chem. Mater.* **5**, 1372 (1993).
- <sup>8</sup> L. Kipp, R.D. Bringans, D.K. Biegelsen, L.-E. Swartz and R.F. Hicks, *Phys. Rev. B* **50**, 5448 (1994).
- <sup>9</sup> J.J. Boland, *Phys. Rev. Lett.* **67**, 1539 (1991); *Surf. Sci.* **261**, 17 (1992). *Phys. Rev. B* **44**, 1383 (1991).



- 10 M.J. Bronikowski, Y. Wang and R.J. Hamers, Abstracts of the Amer. Chem. Soc. **207**, 284 (1994). M.J. Bronikowski, Y.J. Wang, M.T. Mcellistrem and R.J. Hamers, Surf. Sci. **298**, 50 (1993).
- 11 D.S. Lin, E.S. Hirschorn, T.C. Chiang, R. Tsu, D. Lubben and J.E. Greene, Phys. Rev. B **45**, 3494 (1992).
- 12 J.A. Venables and G.L. Price in *Epitaxial Growth*, Part B edited by J.W. Matthews (Academic Press, New York, 1975).
- 13 R.R. Kunz and T.M. Mayer, J. Vac. Sci. Technol. B **6**, 1557 (1988).
- 14 for analysis of various carbonyl systems see F.A. Houle and K.A. Singmaster, J. Phys. Chem. **96**, 10425 (1992).
- 15 D.P. Adams, E. Chason, T.M. Mayer and B. Kellerman (unpublished).
- 16 J.R. Swanson, C.M. Friend and Y.J. Chabal, J. Chem. Phys. **87**, 5028 (1987).
- 17 J.S. Foord and R. B. Jackman, Surf. Sci. **171**, 197 (1986); Chem. Phys. Lett. **112**, 190 (1984).
- 18 B.S. Swartzentruber, Y.W. Mo, M.B. Webb, and M.G. Lagally, J. Vac. Sci. Technol. A **8**, 210 (1990).
- 19 Temperature calibration involved sinking a W/WRe thermocouple junction into an 0.008" diameter hole (~0.040" deep) drilled in the side of the Si samples.
- 20 The pressure was measured by an uncalibrated ion gauge that was turned on only for the brief amount of time required for pressure equilibration (~ 30 sec).
- 21 for a description of the HIBS technique see J.A. Knapp and B.L. Doyle, Nucl. Inst. and Meth. B **45**, 143 (1990) or B.L. Doyle, J.A. Knapp, and D.L. Buller, Nucl. Inst. and Meth. B **42**, 295 (1989).
- 22 E. Chason and B.W. Dodson J. Vac. Sci. Technol. A **9**, 1545 (1991).
- 23 The energy for dissociation onto Fe was taken to be different for various n-Fe nearest neighbors (nn).  $E_{\text{act}}(0 \text{ Fe nn}) = 0.4 \text{ eV}$ ,  $E_{\text{act}}(1 \text{ Fe nn}) = 0.25 \text{ eV}$ ,  $E_{\text{act}}(2 \text{ Fe nn}) = 0.19 \text{ eV}$ ,  $E_{\text{act}}(3 \text{ Fe nn}) = 0.13 \text{ eV}$  and  $E_{\text{act}}(4 \text{ Fe nn}) = 0.07 \text{ eV}$ .
- 24 R.R. Kunz, T.E. Allen, and T.M. Mayer, J. Vac. Sci. Technol. B **5**, 1427 (1987).

## List of Figures

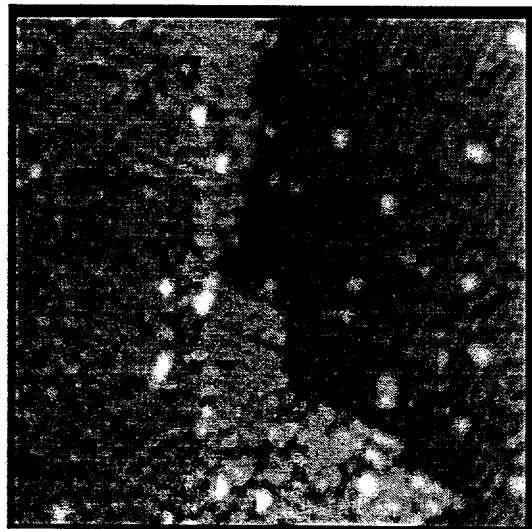
Fig. 1.  $310\text{\AA} \times 310\text{\AA}$  STM images of Fe layers grown at  $215^\circ\text{C}$  onto clean, reconstructed Si(100) by chemical vapor deposition (a. and b.) and Fe evaporation (c. and d.). All images were taken using a bias of  $-2.5\text{ V}$  and a tunneling current of  $0.15\text{ nA}$ .

Fig. 2. Plot of cluster density as a function of exposure for three different substrate temperatures. The nucleation rate is very sensitive to substrate temperature.

Fig. 3. Plot of the average cluster size that develops during growth by CVD vs. evaporation. For equal areal coverages, a larger average cluster size develops during growth by CVD. The lines are guides to the eye. Each point is the average measured from 10 images. A few error bars (characteristic of all data) are shown in the plot, which represent the scatter in average cluster size taken from different areas on the same sample.

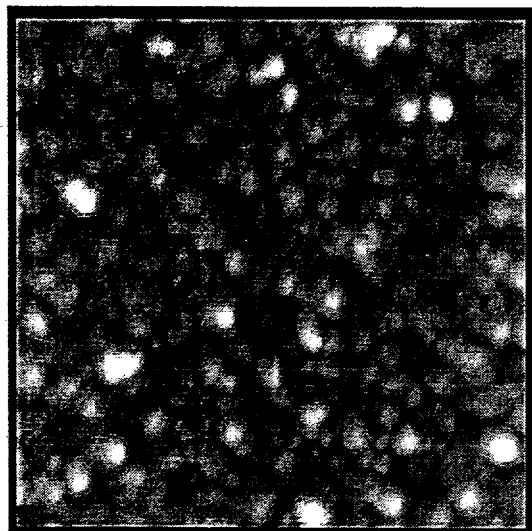
Fig. 4. Simulated CVD cluster morphology for coverages of  $\sim 0.07$  and  $0.25\text{ ML}$  showing the effect of differential dissociation probability (a. and b.) versus the effect of a uniform dissociation probability (c. and d.). We assume that physisorbed precursor molecules are mobile in each. Note, Figs. c. and d. are equivalent to the simulated morphology for growth by Fe evaporation.

Fig. 5. Selected area growth of Fe on H-passivated Si(100), where adsorbed H atoms have been locally removed by electron stimulated desorption. Nucleation is confined to areas where H has been removed.

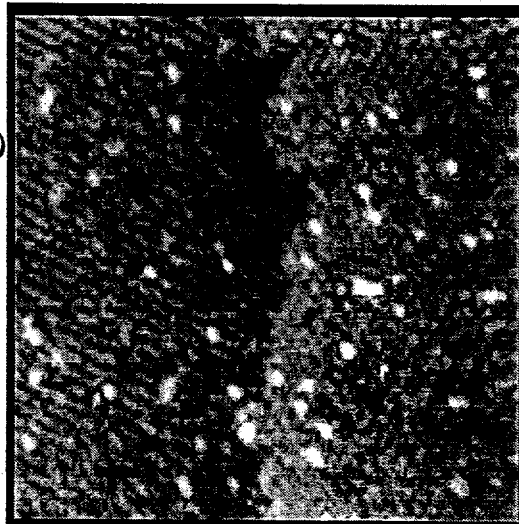


Fe CVD

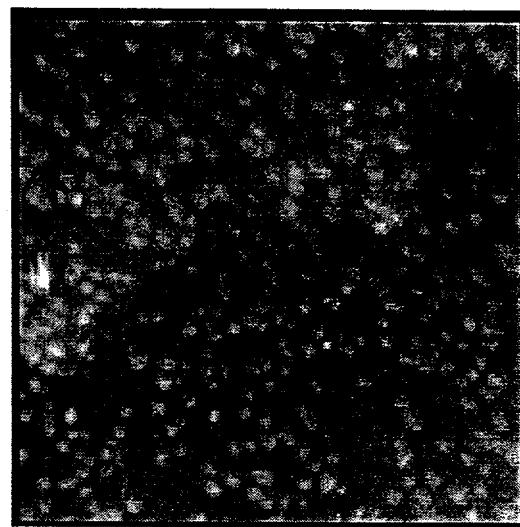
low coverage



high coverage



Fe  
evaporation



10 nm

Figure 1.

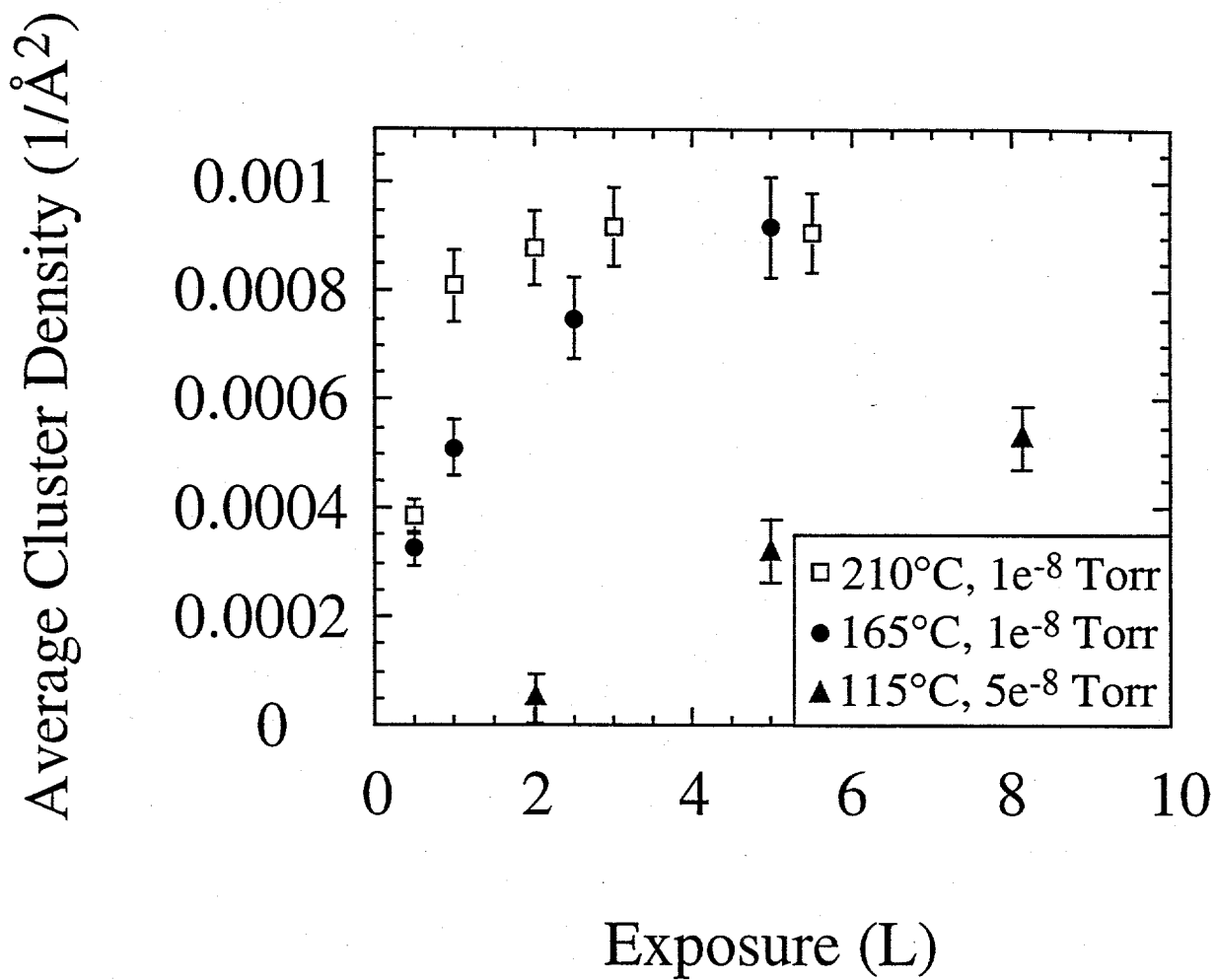


Figure 2.

Average Cluster Size ( $\text{\AA}^2$ )

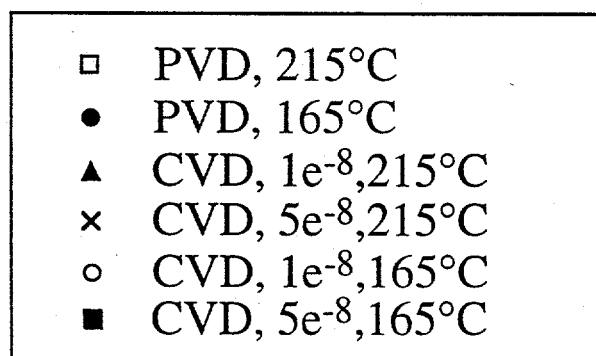
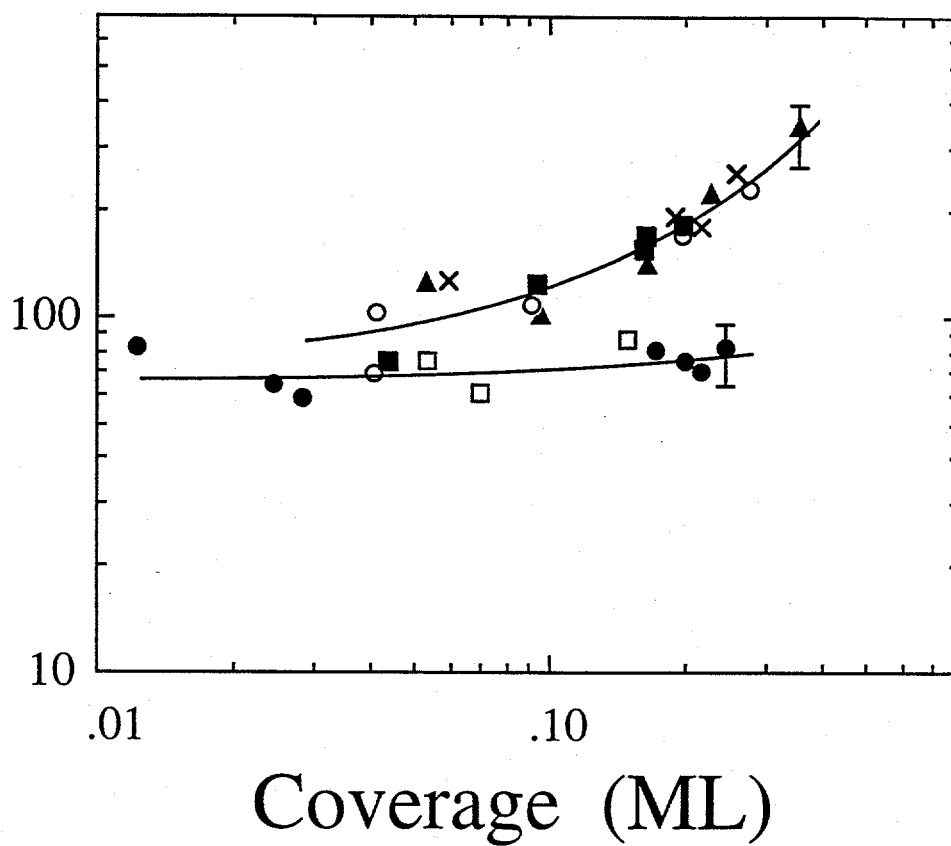
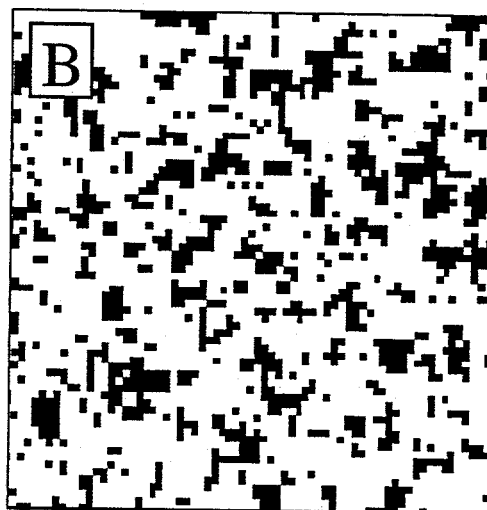
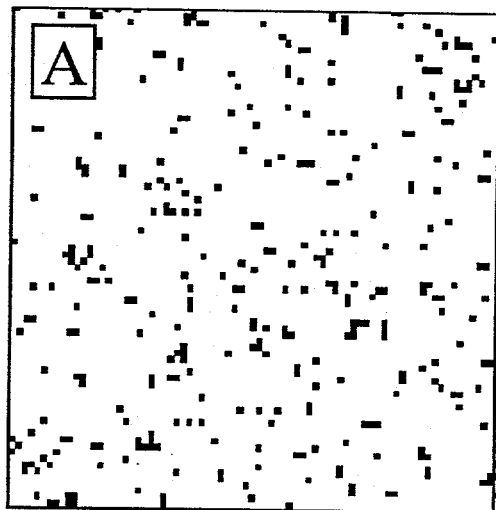


Figure 3.

## CVD Simulation



## Evaporation Simulation

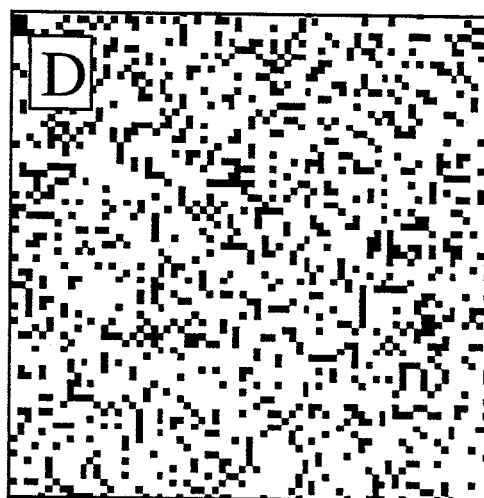
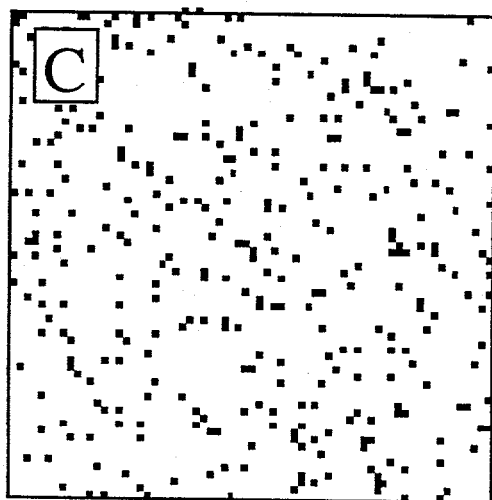


Figure 4.

Pyrolysis at ~275C:  
75 L, Fe(CO)<sub>5</sub>

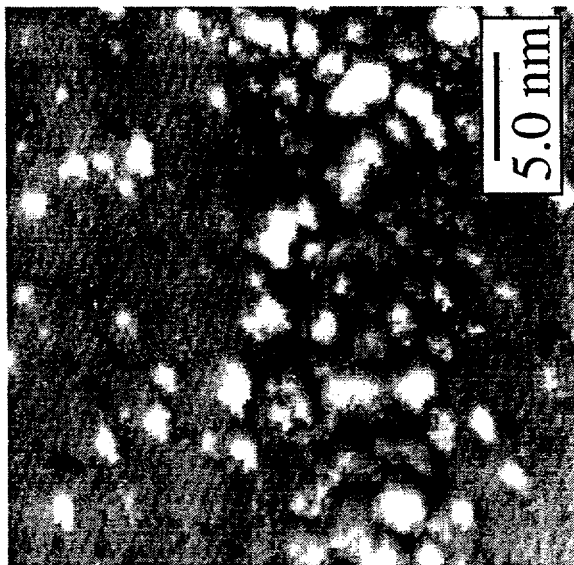


Figure 5.

## Appendix

### List of Publications and presentations resulting from this project

- D. P. Adams, L.L. Tedder, T. M. Mayer, B. S. Swartzentruber, E. Chason, "Initial Stages of Fe Chemical Vapor Deposition onto Si (100)", Phys Rev. Lett. 74, 5088 (1995). SAND94-3275J
- D. P. Adams, T. M. Mayer, B. S. Swartzentruber, E. Chason, "Island Structure Evolution During Chemical Vapor Deposition", Surface Science, in preparation.
- D. P. Adams, T. M. Mayer, B. S. Swartzentruber, "Selective Area Growth of Metal Nanostructures", Appl. Phys. Lett., in preparation.
- T. M. Mayer, D. P. Adams, L. L. Tedder, B. S. Swartzentruber, E. Chason, "Nucleation Phenomena in the Chemical Vapor Deposition of Fe on Si(001)", 10/26/94, American Vacuum Society National Symposium, Denver, CO. SAND94-1250A
- D. P. Adams, T. M. Mayer, B. S. Swartzentruber, "Development of Surface Morphology During the Initial Stages of CVD Growth", 2/10/95, Microphysics of Surfaces: Nanoscale Processing Topical Meeting, Santa Fe, NM. SAND94-2481A
- T. M. Mayer, , "Development of Surface Morphology During the Initial Stages of CVD Growth", 2/7/95, Annual Symposium of the American Vacuum Society Florida Chapter, Clearwater Beach, FL. SAND94-3153A
- D. P. Adams, T. M. Mayer, B. S. Swartzentruber, "Selective Area Growth of Metal Nanostructures using STM-based Lithography and CVD", 10/17/95, American Vacuum Society National Symposium, Minneapolis, MN. SAND95-1032A
- B. K. Kellerman, E. Chason, D. P. Adams, T. M. Mayer, "In-situ X-ray Reflectivity Investigation of Fe Chemical Vapor Deposition on Si(001)", 10/18/95, American Vacuum Society National Symposium, Minneapolis, MN. SAND95-1058A
- D. P. Adams, T. M. Mayer, B. S. Swartzentruber, E. Chason, S. G. Bales, D. C. Chrzan, "Early Stages of Metal Film Growth by CVD", 11/29/95, Materials Research Soc., Fall Meeting, Boston MA. SAND95-2161A
- B. K. Kellerman, E. Chason, D. P. Adams, T. M. Mayer, "In-situ Investigation of the Evolution of Growth Rate and Surface Morphology During Fe Chemical Vapor Deposition on Si(001)", 11/29/95, Materials Research Soc., Fall Meeting, Boston MA. SAND95-1574A

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List of patents: none

List of copyrights: none

Information regarding employee recruitment: NA



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Information regarding involvement of students in the project: This work partially supported a student from the University of Texas, Bruce K. Kellerman. Mr. Kellerman performed in-situ measurements of surface morphology evolution by x-ray reflectivity.

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