CHAPTER 5

Tunable photoluminescence from nc-Si/a-SiN\textsubscript{x}:H quantum dot thin films on varying substrate temperature

5.1 Introduction

Nanocrystalline silicon (nc-Si) quantum dots (QDs) embedded in a silicon nitride (SiN\textsubscript{x}) matrix is currently attracting much interest as a candidate for efficient light-emitting devices [1–4]. Such devices are highly desirable for the integration of optical signals and electronic data-processing circuits on the same chip. Moreover, the fabrication process is compatible with present large-scale integration technologies [5,6]. Photoluminescence (PL) efficiency increases in nc-Si compared to bulk Si in a number of ways [7]. In the case of bulk Si, both the electrons and holes have long radiative lifetime that can diffuse large distances and thus have a high probability of encountering a defect acting as a non-radiative recombination centre; whereas in nanocrystalline Si this diffusion is limited by the crystal boundary, increasing the PL efficiency by reducing the probability of an electron encountering a defect. In addition,
the spatial confinement of electrons and holes also increases the PL efficiency of nanocrystalline Si [6,8]. The band gap of nanocrystalline Si is enlarged with respect to the bulk material and an intense visible PL at room temperature is observed. Consequently, it is possible, by using properly sized Si nanocrystals (Si-ncs), to cover the entire visible length of the electromagnetic spectrum.

A number of works have already been published on photoluminescence from nanostructured silicon, which includes crystalline or amorphous silicon quantum dots embedded in a matrix made of amorphous silicon or silicon dielectrics, e.g. SiO_x, SiC, SiN_x, etc., synthesized by various chemical and physical vapour deposition routes [9–14]. Interconnected silicon nanocrystals having wide-range PL have been prepared, using a pulsed PECVD technique under dusty plasma conditions [15]. In addition, freestanding luminescent Si nanocrystals were reported using different chemical synthesis processes [16–19]. Strong blue photoluminescent silicon nanoparticles were obtained by laser ablation of silicon wafer immersed in solution [20,21]. There is no doubt that the nanometer-sized silicon can emit light of different colors, which may arise from band-to-band recombination due to quantum confinement or recombination from defect centers, however, the search still continues for the elusive silicon light emitter which could make the prospect of silicon photonics a reality.

For the proper application of photoluminescent Si nanocrystals in devices, they need to be embedded in a dielectric matrix for which silicon nitride is a preferred one because of its low barrier height, high charge-trapping capacity and high carrier mobility. Recently, flat panel displays have evolved to flexible screens and electronic paper, and hence a low temperature process is required to produce the film on plastic or other soft substrates. In addition, enhancement of the PL intensity with narrow emission spectrum is desirable for light-emitting devices. In the quest to develop such a material we take the opportunity of using inductively coupled plasma-CVD as it can create high-density plasma at low temperature and pressure (in the milli-Torr range) and can produce overall film crystallinity much higher than for corresponding plasma parameters in glow discharge CVD. In this chapter, we demonstrate the formation of low-temperature inductively coupled plasma-synthesized nanocrystalline silicon quantum dots in hydrogenated amorphous silicon nitride (nc-Si/a-SiN_x:H) matrix, from which photoluminescence (PL) of strong intensity with relatively narrow width has been observed. In addition, the observed photoluminescence has been made tunable.
over the visible range 2.47–1.66 eV by controlling the substrate temperature within 150–400 °C. The quantum confinement effect (QCE) has been assigned as responsible for the origin of the dominant component of luminescence.

5.2 Experimental Section

Inductively coupled radio frequency (13.56 MHz) plasma was employed to prepare thin films of nanocrystalline silicon quantum dots embedded in amorphous silicon-nitride matrix (nc-Si/a-SiNx:H) on p-type (100) single crystal silicon wafers, Corning® Eagle2000™ glasses and carbon coated copper microscope grids. Silane (SiH₄), ammonia (NH₃), and hydrogen (H₂) were used as reactant gases with individual flow rates of 2 sccm, 0.7 sccm, 30 sccm, respectively, to create a low-pressure plasma at 30 mTorr. The substrate temperature was varied from 150 to 400 °C. Other process parameter includes RF power 500 W. No post-deposition treatment e.g., thermal annealing was employed after growing the silicon nitride film. Room temperature PL spectra were obtained using a He-Cd laser source, with a uniform excitation of 325 nm. The luminescence spectrum was monitored using a TRIAX 320 monochromator fitted with a cooled Hamamatsu R928 photomultiplier detector. The optical density data of the samples were obtained from absorption and reflection measurements in the UV–visible region at room temperature using a double beam spectrophotometer (Hitachi 330, Japan) and the optical band gaps were estimated using Tauc’s plot. In order to confirm the size variation of the Si QDs, Raman measurement was performed by Renishaw Micro-Raman spectrophotometer at room temperature in a backscattering geometry, using 514 nm Ar⁺ Laser as the excitation source. The structure of the Si QDs was investigated by a JEOL-JSM2010 high-resolution transmission electron microscope (HR-TEM) operated at 200 kV. The micrograph images were obtained from samples ~30 nm thick, deposited on carbon coated copper microscope grids supplied by Pacific Grid-Tech. A comprehensive correlation between photoluminescence and structural properties has been obtained.

5.3 Results

Fig. 5.1 represents the photoluminescence (PL) spectra of six silicon-nitride thin films deposited at different substrate temperatures (T), varying from 400 to 150
CHAPTER 5 Tunable photoluminescence from nc-Si/a-SiNₓ:H quantum dot thin films on varying substrate temperature

Figure 5.1: Room temperature photoluminescence (PL) spectra of films, having nc-Si QDs embedded in a-SiNₓ:H matrix, prepared at different substrate temperatures varying from 400 to 150 °C with an interval of 50 °C.

°C with an interval of 50 °C. All the samples showed strong PL of individual color in the visible range under UV excitation. With the decrease of substrate temperature from 400 °C, PL peak energy was shifted from 1.66 eV towards the blue end of the visible electromagnetic spectrum and finally, it was observed at 2.47 eV, corresponding to T = 150 °C. The systematic shift of the PL peak energy has been well calibrated with the substrate temperature as shown in Fig. 5.2(a), while Fig. 5.2(b) demonstrates the variation of the integrated PL intensity and the corresponding full width at half maximum (FWHM) with T. With the decrease of substrate temperature from 400 to 150 °C the PL intensity initially increases, reaches its maximum at 250 °C and then decreases, whereas the FWHM increases monotonically from 0.16 to 0.45 eV with decrease of T. The overall high intensity and narrow FWHM are the two major PL characteristics that are significant for successful application of the material into light emitting devices. The tunable nature of PL peak energy with substrate temperature could arise because of the quantum confinement effect (QCE) within the material, however, the importance of several radiative defect states, coming from the interface and matrix, in luminescence cannot be neglected without proper justification. In this context, estimation of optical (Tauc) gap (from UV-vis spectroscopy), HR-TEM and
Figure 5.2: Variations of (a) photoluminescence (PL) peak energy and (b) integrated PL intensity as well as FWHM, corresponding to different substrate temperature, $T$.

Raman analysis of the material could be useful in exploring the origin of the PL spectra.

The optical gap ($E_g$) of nc-Si/a-SiNx:H films was estimated by measuring the absorbance and reflectance in the UV–vis region using Tauc’s equation [22]:

$$\alpha h\nu = A (h\nu - E_g)^n$$  \hspace{1cm} (5.1)

where $\alpha$ is the absorption coefficient, $\nu$ is the transmission frequency and the exponent $n$ characterizes the nature of band transition. Values of $n = 1/2$ and $3/2$ correspond to direct allowed and direct forbidden transitions, $n = 2$ and $3$ are related to indirect allowed and indirect forbidden transitions. As the nc-Si/a-SiNx:H thin film is a mixed
phase material, proper estimation of its Tauc gap is rather difficult, however, that may provide an approximate optical gap of the material. Using $n = 2$, considering an approximation of indirect transitions, the plot of $(a \nu)^{1/2}$ with $\nu$ determines the value of $E_g$. The variation of $E_g$ with substrate temperature, $T$ is shown in the Fig. 5.3. The gradual widening of optical gap, in the range 2.26–2.60 eV, has been observed as a result of decreasing $T$.

In order to study the details of internal microstructure, High-Resolution Transmission Electron Microscope (HR-TEM) was used on two samples, T350 and T250, prepared at 350 °C and 250 °C, respectively. The physical evidence of the presence of nano-crystalline Si quantum dots (nc-Si QDs) was obtained from HR-TEM imaging. Fig. 5.4(a) and (c) represent the magnified view of the microstructure of samples T350 and T250, respectively. In both figures, the crystal planes are noticeable in all the marked areas identified as QDs, which are spherical in shape. Quantum dots are denser material and relatively dark in appearance, randomly distributed within a less dense a-SiN$_x$:H matrix. The size distribution of nc-Si QDs, estimated from the statistical analysis in number density verses dot size plot, has been shown by the corresponding histograms in Fig. 5.4(b) and (d). The sample T350 possess nc-Si QDs having average size ~4.54 nm and density ~6.40x10$^{11}$ cm$^{-2}$ while, T250 contains nc-Si QDs of relatively small average size, ~2.67 nm but of higher
density, $\sim 1.12 \times 10^{12}$ cm$^{-2}$. Relatively wide size distribution and higher number density of nc-Si QDs in T250 sample may be accounted for its broader PL band and enhanced PL intensity, respectively, compared to those for T350 sample.

The Raman spectra of the samples shown in Fig. 5.5 identify a continuous broadening in line width with decreasing $T$. In addition, the Raman peak gradually shifts towards lower energy that has been pronounced from the enlarged view of the peak positions shown at the inset of Fig. 5.5. Both the line width broadening and the peak-shift indicate a spontaneous decrease in the size of the nc-Si QDs, arising out of the confinement of optical phonons within smaller nanostructures.

The average dot size has been estimated from first order Raman spectra using the following equation [23]:

$$
\left[\omega_c - \omega_0\right]^2 + \left(\frac{F_0}{2}\right)^2 \approx \frac{1}{3L}.exp\left(-\pi^2\right)
$$

(5.2)

Figure 5.4: HR-TEM image of the nc-Si/a-SiN$_x$:H films prepared at (a) $T = 350$ °C and (c) $T = 250$ °C, while, (b) and (d) represent the corresponding histograms demonstrating the variations of the number density, size and the FWHM of the distribution of the nc-Si QDs.
where $\omega_L$ is the frequency of the crystalline-like mode for a quantum dot of size $L$ and $\Gamma_0$ is the natural line width (inversely proportional to the intrinsic phonon lifetime). For crystalline silicon $\omega_0$ and $\Gamma_0$ are 520 and 3.5 cm$^{-1}$, respectively. The dot size is observed to decrease from 5.72 nm to 1.67 nm as $T$ goes from 400 °C to 150 °C and the variation of which has been shown in Fig. 5.6(a). Size estimated from HR-TEM image, shown in Fig. 5.4, have also been included in Fig. 5.6(a) and the Raman result is in good agreement with the result obtained from HR-TEM.

Quantitative estimation of the crystalline volume fraction ($X_C$) has been performed by the deconvolution of each Raman spectrum into three Gaussian components, the component on the higher frequency side represents nanocrystalline part of the material and the lower frequency component arises from amorphous part, whereas the intermediate one corresponds to the ultra-nanocrystalline and/or the grain boundary component [24]. The variation of crystalline volume fraction ($X_C$) and ultra-nanocrystalline fraction ($X_{unc}$) is shown in Fig. 5.6(b). $X_C$ is observed to decrease with the decrease of $T$ while, $X_{unc}$ increases for the same variation of $T$. With the lowering of substrate temperature, the film microstructure becomes more amorphous in nature that is monitored by the asymmetric broadening of the Raman line. The maximum

Figure 5.5: First order normalized Raman spectra of nc-Si/a-SiN$_x$:H films prepared at different $T$. Inset represents the enlarged view of the peak of Raman spectra.
crystallinity is obtained for T400 sample and the value of which is ~82% whereas, it decreases to the lowest magnitude, ~42% for T150.

The enhanced magnitude of ultra-nanocrystalline component indicates that with the lowering of substrate temperature, the network crystallinity becomes dominated by tiny crystallites that can be correlated with the consecutive blue shift of the PL band. The downshift of the Raman peaks could be correlated with the upshift of the photoluminescence peaks. The position of the TO crystalline Raman peak for T400, from which the red emission (~1.66 eV) is observed, is ~514.8 cm\(^{-1}\) and the same for T150, which emits near blue light (~2.47 eV), is ~510 cm\(^{-1}\). PL intensity is observed to be the maximum at T250, where the Raman peak occurs at 512.6 cm\(^{-1}\).
5.4 Discussion

In order to find out the origin of such PL peaks, two types of luminescence mechanisms, e.g., through radiative defects in the films and by quantum confinement effect (QCE), have been proposed. In case of silicon-nitride four types of defects could exist, e.g., within Si–Si and N–N bonds, and by Si and N dangling bonds [25]. The Si–Si bond forms a bonding σ orbital and anti bonding σ* orbital that are separated by 4.6 eV for stoichiometric silicon nitride [26]. The silicon-dangling bond, which generates defects state about the mid-gap between σ and σ*, is being considered as a dominant trap and recombination center in silicon nitride and participates in the radiative transition giving rise to luminescence [27,28]. The nitrogen dangling bond (N₂⁺, N center), can also give rise to a level within the Si–Si gap. Another type of nitrogen defect state: namely N₄⁺ can also give rise to a level besides N₂⁰. The N₄⁺ can be formed by reaction between positively charged silicon–dangling bond and a bulk nitrogen [25]. For nanocrystalline silicon-nitride films, radiative defect centers at the interface between the Si nanocrystal and the a-SiNₓ:H matrix could be a source of luminescence for small Si nanocrystals [29]. In addition, oxide related defects also play vital role in the appearance of visible luminescence from silicon nanocrystals, because oxygen usually remains as an inherent residual incorporation in silicon based micro- and nano-structured films [30]. Six emission bands corresponding to 3.2, 2.8, 2.7, 2.4, 2.3, and 2.0 eV, originated from different radiative defect centers in a-SiNₓ:H, have been observed by Mo et al. [31] Deshpande et al. [28] reported another two radiative recombination centers related to nitrogen defect state around 1.8 and 3.0 eV where the 1.8 eV peak is due to recombination between N₄⁺ and N₂⁰ levels and the 3.0 eV peak is due to recombination either from the conduction band to N₄⁺ level. The radiative Si dangling bond (K⁰) center is responsible in appearing a PL peak at ~2.5 eV [32]. The PL peak due to neutral oxygen vacancy in the network normally appears at ~2.65 eV [33,34]. Besides this, a peak at ~2.1 eV may become visible consistent with emission from an oxygen-related defect such as the non-bridging oxygen hole center [35]. Based on the above discussion on the defect related recombination in silicon-nitride films; there are numerous possibilities of having several defect origins of the photoluminescence spectra. However, defects being the origin of PL, the peak
positions should appear only near the defect energy levels. Rather, the tunable nature of the PL peaks with substrate temperature and the corresponding systematic downshift of the peak frequency of Raman line that identify the gradual reduction in the size of Si QDs which was further supported by TEM studies, altogether suggest the possibility of the photoluminescence from quantum confinement effect (QCE). The blue-shift in photoluminescence spectra on changing the process parameter is, therefore, mostly attributed to the quantum size effect. Fig. 5.7 represents the schematic of radiative recombination process between conduction and valence band inside Si QD where the recombination is quasi-direct, following the principle of uncertainty between the position and momentum. Possibilities of band-to-band transition and formation of excitons by direct transition within Si nanocrystals as the plausible origin of the blue photoluminescence from silicon nano-particle powder product obtained by laser ablation have been discussed in an earlier report [36].

However, it has been identified that the PL spectrum corresponding to T400, T350 and T300 are nearly symmetric whereas, the PL spectrum for the lower temperature samples gradually becomes asymmetric. Asymmetric broadening of the PL spectrum for the lower temperature samples may indicate the possibility of some defect related origins to co-exist with band-to-band quasi-direct recombination. For detail understanding regarding the defect related issues, the spectra corresponding to the samples T250, T200, and T150 have been deconvoluted into possible satellite
components and those have been shown in Fig. 5.8. The component (a1) at ~1.79 eV for T250 can be attributed to the recombination processes between conduction edge of the intrinsic band and $N_4^+$ level, and between the $N_4^+$ and $N_3^0$ states whereas, the component (c1) at ~2.37 eV may be originated from the electronic transition of $E_c \rightarrow Si^0$ where $E_c$ is the conduction band edge and $=Si^0$ is the Si donor energy level [31]. The peak (b1) at ~2.05 eV may result from band-to-band recombination in nc-Si QDs which appears as (b2) at ~2.19 eV for T200 and as (b3) at ~2.32 eV for T150 sample. The component (c2) at ~2.37 eV has come out with relatively higher magnitude than the corresponding component (c1) because of increasing defect related luminescent centers at lower substrate temperature. The component (a2) at ~2 eV corresponding to T200 may have the origin of the electronic transition between $E_c$ and $=Si^-$, the Si acceptor energy level [31]. The ~2.1 eV component (a3) for sample T150 can be considered as the luminescence from the oxygen-related defect such as the nonbridging oxygen hole center due to the higher (appreciable) amount of oxygen incorporation in the network at lower substrate temperature. The ~2.66 eV PL peak (d3) could be attributed either to the electronic transitions of $=Si^- \rightarrow N^-$ defect states [31], or neutral oxygen vacancy in the network whereas, the peak (c3) at ~2.49 eV may be originated from the radiative $K^0$ center, the Si dangling bond back bonded
Figure 5.9: The variation of peak energy of the typical photoluminescence spectrum obtained experimentally and the specific component of PL identified as a contribution from band-to-band recombination in nc-Si QDs, originated due to quantum confinement effect plotted separately as a function of estimated size of the quantum dots.

to three N atoms. The less intense component (e3) at ~2.79 eV may be related to the transition between $=\text{Si}^0$ and $=\text{Si} - \text{Si} =$, the Si-Si energy level [31].

Hydrogen passivation of dangling bonds in nanoscale silicon is generally known to increase photoluminescence intensity [37]. Proper surface passivation of the Si QDs could be the cause for obtaining high intensity of photoluminescence for T250. Surface passivation remains incomplete both at lower and higher temperatures; at low temperature e.g., 150 °C the available atomic H in the plasma may not be sufficient for satisfying the huge number of available dangling bonds in the network, while at 400 °C out diffusion of H from the network could create additional dangling bonds. The variation of peak energy of the typical photoluminescence spectrum obtained experimentally and the specific component of that identified as a contribution from band-to-band recombination in nc-Si QDs, originated due to quantum confinement effect, have been plotted separately in Fig. 5.9, as a function of estimated size of the quantum dots. Widening of the optical gap, $E_g$ (2.26–2.60 eV) of nc-Si/a-SiN$_x$:H quantum dots thin films on lowering in the substrate temperature is the result of mostly the quantum confinement effect arising due to decreasing size of the nc-Si QDs and the simultaneous minor changes due to increasing bonded hydrogen content and varying nitrogenation of the network.
The plasma-assisted development of nc-Si QDs within a-SiNₓ:H matrix depends entirely on the optimization of process parameters. Deposition was performed at low temperature and low pressure while the power delivered in the plasma was relatively high, facilitating the production of high density radicals and ions from the precursor gases. The thin film growth processes is mainly controlled by two separate phases, gas phase reaction and solid phase reaction. Hydrogen atoms are produced in abundance by inductively coupled radio frequency (rf) discharge of H₂ through the reaction, \( H_2 \rightarrow H + H \) and mixed with ammonia (NH₃) and silane (SiH₄) in the plasma chamber [38]. Silyl (SiH₃), and amidogen (NH₂) radicals are produced by atomic hydrogen induced gas phase reactions [39]:

\[
\begin{align*}
H + NH_3 & \rightarrow NH_2 + H_2 \\
H + SiH_4 & \rightarrow SiH_3 + H_2
\end{align*}
\]

(5.3)

(5.4)

The NH₂ radicals are unreactive with SiH₄ however, readily insert into SiH₃ and eliminate hydrogen, forming amino-silate complexes. Successive insertions of NH₂ into the complex continues to add amino groups and eliminate H, until reaching Si(NH₂)₄ [40]:

\[
\begin{align*}
NH_2 + SiH_3 & \rightarrow H + SiH_2(NH_2) \xrightarrow{NH_2} H + SiH(NH_2)_2 \\
NH_2 \xrightarrow{NH_2} H + Si(NH_2)_3 \xrightarrow{NH_2} H + Si(NH_2)_4
\end{align*}
\]

(5.5)

Heterogeneous reactions of higher silanes and the aminosilanes are considered as the viable pathway for film growth. The structural network is primarily formed by the energy relaxation process of adsorbed precursors on the growing surface and is principally determined by their surface diffusion length [41,42]. A huge amount of atomic hydrogen flux generated from large-scale dissociation of all the three gas molecules (SiH₄, NH₃ and H₂) in high-density inductively coupled plasma (ICPs) ensures a very high surface coverage, which increases the surface diffusion length of radicals because of the lower surface diffusion activation energy [43,44].

The large surface diffusion length facilitates the adsorbed radicals to reach the energetically convenient positions on the growing surface and that enables the formation of controlled and well-oriented crystalline QD structure of spherical shape [45]. All these events occurring on the growth surface account for the presence of enough crystallinity (more than 40% in volume) in the nc-Si/a-SiNₓ:H film obtained
even at a temperature as low as 150 °C. At relatively higher value of the substrate temperature, the large surface diffusion length of the adsorbates induces the dots to grow larger in size. Whereas, smaller surface diffusion length at lower substrate temperature confines the growth of the Si QDs to small size, however, with wide size distribution and in larger number density, leading to ~27% volume fraction of ultra-nanocrystalline component at 150 °C.

The nc-Si/a-SiNₓ:H thin films with intense PL tunable over a wide visible range (1.66–2.47 eV) described in the present work show great promise towards the fabrication of light-emitting devices (LEDs). Previously wide-range visible photoluminescence from crystalline Si QDs within SiNₓ matrix prepared with the variation of chamber pressure and NH₃ flow rate in the plasma was reported using PECVD [1,3] and also by sputtering technique [2], with or without the assistance of post-deposition annealing at higher temperatures. Later, visible PL at different wavelengths was reported from materials prepared by catalytic CVD [46], while plasmon-induced enhancement of PL has also been reported from similar material [47]. In connection with the above, the novelty of the present work is in obtaining intense visible photoluminescence of narrow width from one-step plasma-synthesized nc-Si/a-SiNₓ:H quantum dot thin films by down-scaling the substrate temperature down to 150 °C; simultaneously, it provides the nitrogenated dielectric matrix, which is uniquely preferred for device fabrication. Lower substrate temperature enhances the device feasibility on inexpensive soft substrates that in turn reduces the device cost and simultaneously improves the flexibility, lightness, and durability of the device.

5.5 Conclusions

Tunable visible range (1.66 to 2.47 eV) photoluminescence (PL), having high enough intensity and relatively narrow width (0.16-0.45 eV), have been observed from nanocrystalline silicon quantum dots embedded in amorphous silicon-nitride matrix (nc-Si/a-SiNₓ:H). Quasi-direct band-to-band recombination due to quantum confinement effect (QCE) in the nanocrystalline silicon QDs of appropriate size seems to be the origin of the dominant PL component, described in the present work. Planer ICP-CVD of 13.56 MHz RF plasma was employed to prepare silicon nitride films using a mixture of silane (SiH₄), ammonia (NH₃), and hydrogen (H₂) gases, by changing the substrate temperature from 400 to 150 °C, keeping plasma power and
chamber pressure remaining constant at 500 W and 30 mTorr, respectively. A large amount of atomic hydrogen flux that originates due to large scale dissociation of the gas molecules in high-density inductively coupled plasma (ICPs), efficiently passivates the nonradiative dangling bonds by providing a very high coverage on the growing surface, and thereby favors to increase the PL intensity from the nc-Si/a-SiNₓ:H network. Nearly spherical Si QDs of average size ~4.54 and 2.67 nm with particle density ~ 6.40x10¹¹ and 1.1x10¹² cm⁻² for the materials prepared at 350 and 250 °C, respectively, were observed from HR-TEM images and the corresponding histogram plots. The downshifts of Raman peak and the corresponding asymmetric broadening of the Raman line due to the confinement of optical phonons indicate the size reduction of Si QDs with amorphous dominated nature of the network during the decrease of substrate temperature (T). Size estimated from the Raman spectra varies from 5.72 nm to 1.67 nm as T reduces from 400 to 150 °C and it agrees well with TEM results. A number of groups previously reported on the successful emission of PL with different colors from nc-Si QDs embedded in SiNₓ matrix, however, creation of tunable multicolor emission with narrow band width, from similar material prepared by one-step plasma processing described in the present work, would be a great promise towards the fabrication of light emitting devices (LEDs). Moreover, photoluminescence emerging from nc-Si/SiNₓ:H films prepared at a low temperature around 150 °C would enable the fabrication of devices on soft inexpensive substrates e.g., flexible flat panel displays.
CHAPTER 5 Tunable photoluminescence from nc-Si/a-SiN$_x$:H quantum dot thin films on varying substrate temperature

Bibliography