Chemical bond manipulation for nanostructure integration on wafer scale

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Abstract. In this paper, we have briefly summarized our activity in the area of chemical bond manipulation for the integration of nanostructures on a full wafer scale. Chemical bond manipulation involves a judicious combination of surface phenomena: reactions or diffusion, and growth process such as molecular beam epitaxy (MBE). Here, we present our results on oxidation, metallization and nitridation and their role in the formation of nanostructures. We find that oxygen changes the bonding partner from Ge to Si and this phenomenon can be controlled by controlling the annealing temperature. We have employed this phenomenon for the fabrication of novel, multiperiod Si/SiO₂/Ge layered structure which exhibits interesting light emitting properties. Further, by making use of selective diffusion of cobalt atoms through Ge layers it is possible to incorporate metallic features into Ge quantum dots. Moreover, it is possible to fabricate Si nanopillars through high temperature reaction of nitric oxide. NO molecules dissociate on the surface and nitrogen atoms thus produced form nitride islands. These islands act as protective masks for the etching of Si by the oxygen atoms, through the desorption of SiO species. Occurrence of these two simultaneous processes result in the formation of nanometre-sized Si pillars capped by silicon nitride. All these results emphasize the fact that we can extend information obtained through traditional surface science experiments for the fabrication of novel structures on a full wafer scale.

Keywords. Chemical bond manipulation; nanostructure integration; wafer scale.

1. Introduction

Semiconductor nanostructure formation in general and that of Si in particular have been gaining a lot of attention by the theoreticians as well as by the experimentalists. The main thrust is to explore new avenues of application, such as in optical devices and quantum effect devices (Fisher and Chow 1993; Shen *et al* 1995; Ashoori 1996). Further, these can also find application as photonic crystals (Yablonovitch *et al* 1991), if these structures could be fabricated in a three-dimensionally periodic fashion. In the area of fabrication of nanostructured materials, control of atomic processes is the key to achieve formation of the desired species. Several researchers have reported (Avouris *et al* 1993) nanostructure fabrication by atom manipulation, such as using scanning tunneling microscope tip. However, this is limited to only a very small area of the wafer and moreover it involves only a single type of material. Thus, this approach is technologically not feasible. In our laboratory, we have been working on various ways to control the formation of nanostructures by suitably modifying the substrates (Ogino *et al* 1996). Another approach to achieve formation of nanostructures is by controlling surface phenomena, i.e. chemical reactions.

We have termed this approach as chemical bond manipulation involving a judicious combination of surface phenomena and growth (Prabhakaran and Ogino 1997). This is schematically represented in figure 1. The main advantage of this is the possibility to achieve formation of the desired structures on the entire surface of the wafer. The important step is to identify the reaction pathways, any selectivity with regard to surface bond formation, bond cleavage and desorption, which are usually controlled by the thermodynamics and kinetics of the system. However, the main drawback is the poor controllability of the size and distribution which to some extent can be overcome by modifying the substrate (Ogino *et al* 1996) structure.

2. Experimental

We have used a variety of Si and Ge wafers, differing in conductivities. These are cleaned by thermally decomposing an ultrathin *ex situ* prepared chemical oxide (Ishizaka and Shiraki 1986; Prabhakaran *et al* 1994a), inside the UHV chamber. In order to ensure good quality starting substrates, we deposited buffer layers (~100–200 Å) on these samples. The UPS and XPS measurements were performed using an electrostatic analyzer (VG CLAM II, equipped with three channeltrons) with He I and Al/Mg K α sources, respectively, in the normal

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emission geometry. All the spectroscopic measurements were carried out at room temperature. NO gas was leaked on to the samples from a gas bottle (50 cc) attached to the UHV chamber through a needle valve. Ultra dry O_2 (moisture content less than 100 ppb, as monitored by a hygrometer attached on line) was leaked into the chamber from a main line through a needle valve. In the case of etching with oxygen, molecular oxygen was dosed onto the nitrated surface at a temperature of 850°C. Medium energy ion scattering (MEIS) measurements were carried out *in situ* to understand the atomic movements taking place during surface reactions (results not presented in this paper). Sample morphology was examined, outside the UHV chamber, using a DI AFM. In addition, we performed photoluminescence (PL), cross sectional transmission electron microscopy (XTEM), scanning electron microscope (SEM) and LEED measurements. Solid source MBE of Si and Ge was carried out in a separate chamber attached to the analysis chamber. We ensured quality of the epitaxial layers by monitoring the reflection high energy electron diffraction (RHEED). Metal deposition (cobalt) was performed from a high temperature Knudsen cell, from a BeO crucible operated at 1200°C. We also performed scanning Auger microscopic measurements (spatial resolution of the order of ~ 100 nm) to determine the atomic concentrations.

3. Results and discussions

3.1 Fabrication of novel nanometre sized multilayer structures through a combination of surface reaction and growth

There have been several attempts to improve the performance of Si-based devices, such as hetero-junction bipolar transistors and modulation-doped field effect transistors, by using SiGe alloys (People 1986; Iyer *et al*

1989; Park *et al* 1989). Since one of the most advantageous features in Si devices is the availability of good quality oxide, prepared by thermal oxidation and its interface with substrate, control of oxidation process is a key technology for future SiGe-based devices. It has been reported that during SiGe alloy oxidation, a SiO_2 layer is preferentially formed and Ge is rejected into the SiO_2 /SiGe interface. However, most of these studies have been carried out in furnaces and on thick layers of SiGe alloys (Fathy *et al* 1987; Patton *et al* 1988; LeGoues *et al* 1989; Nayak *et al* 1990; Liou *et al* 1991). In order to control the interface characteristics we believe that *in situ* studies are essential to understand the mechanism of the initial oxidation process. Sometime ago we reported our (Prabhakaran *et al* 1993, 1994b) study on the early stages of oxidation of a thin layer of Ge grown on Si(001) surface. We found that oxygen changes its bonding partner from Ge to Si on increasing the temperature. Further, surface sensitive experiments using synchrotron radiation revealed that such a reaction takes place in two different steps. The first step involves the breakage of Ge–O bonds and formation of Si–O bonds. This process continues until the germanium oxide is completely reduced to elemental Ge. The second step involves rearrangement of oxygen atoms around Si to form the stable stoichiometric SiO_2 . Figures 2a and b describe these results. The attenuation of the oxide signal in Ge 3d and its concomitant increase in Si 2p unequivocally establishes the occurrence of the bonding partner change reaction. The principle of manipulation of surface chemical bonds in this case can be explained on the basis of the large differences in the heats of formation (CRC Handbook of Chemistry and Physics 1989) of Ge oxides (for GeO and GeO_2 , $\Delta H_f = -62$ and -131 kcal/mol, respectively) compared to that of Si oxide ($\Delta H_f = -217$ kcal/mol). The same factor is reflected in the strength of the chemical bonds (Ge–O and Si–O, 151 and 191 kcal/mol, respectively).

The above-mentioned experiments were performed on Si(001) wafers and the occurrence of bonding partner change reaction resulted in considerable surface roughening as seen by the formation of nanometre sized voids. Recently, we carried out similar experiments on Si(111) surface, which has lower surface energies (Prabhakaran *et al* 1998). On a clean Si(111) sample, 4 monolayers (~ 6.5 Å) of Ge were deposited at room temperature (RT). The sample was oxidized by exposing to flowing oxygen, outside the ultrahigh vacuum (UHV) chamber, for one min. This process resulted in the formation of nearly two monolayers of Ge oxide. The sample was re-inserted into the UHV chamber immediately and characterized using photoemission measurements. Si (3 nm) was deposited onto this sample at RT. This procedure (Ge deposition, oxidation and Si deposition) was repeated seven times (in the case of the sample

Figure 1. Schematic description of the chemical bond manipulation and comparison with atom manipulation with STM tip.

discussed in this paper, hereafter called as '7-period sample'), each time characterizing the surface species. Finally, the sample was capped with Ge (6.5 Å) and Si (5 nm). The final structure of the sample consisted of

ultrathin silicon oxide layers sandwiched between ultrathin Si and Ge layers as described later on.

Figure 3a shows the XPS in the Ge 2*p* region after oxidizing the Ge covered Si(111) surface (spectrum (a)), and after depositing Si onto the Ge oxide covered surface (spectrum (b)). Spectrum (a) shows that oxidation results in the formation of a mixture of Ge oxides. On depositing Si onto this surface, Ge–O bonds are broken and Si–O bonds are formed and therefore the signal due to the oxides in Ge 2*p* spectrum is attenuated and returns to that of the sample prior to oxidation, implying complete

Figure 2. (a) SRPES wide scan of (a) air oxidized 5 Å Ge deposited on Si(100) and (b) annealed to 600°C. Inset shows the Si 2*p* regions recorded as a function of temperature. (b) Variation of the ratio of elemental Si 2*p* to that of the oxides. Right axis shows the energy separation between the peak positions due to elemental Si and that due to the combined oxides in Si 2*p* spectra. The plot shows two distinct steps in the bonding partner change reaction.

Figure 3. XPS in the Ge 2*p* region from a thin (4 ML) Ge layer deposited on Si(111) and oxidized (spectrum (a)). Spectrum (b) is obtained after depositing 20 Å Si onto this sample at RT, indicating the disappearance of germanium oxide peaks. The height has been normalized to make the comparison clearer. Inset shows the corresponding Si 2*p* spectra. Signal due to silicon oxide formation on depositing Si onto the germanium oxide covered surface is indicated.

reduction (see spectrum (b)). Corresponding Si 2*p* spectra are plotted in figure 3b and show signal due to silicon oxides after the deposition of Si onto the Ge oxide covered Si(111) surface. The signal due to the oxides in Si 2*p* is weak compared to the Ge oxides because the main peak is dominated by bulk Si. Spectral changes in O 1*s* as well as UPS corroborate such chemical bond manipulation reaction.

Figure 4 is the XTEM image from the 7-period sample, indicating the formation of continuous layers with an interface roughness of the order of atomic dimensions. Figure 4 also shows the Auger depth profiling data. The sharp dips observed in the Si LVV signal due to the depletion of elemental Si and the corresponding peaks in O KLL signal, show the formation of silicon oxide layers with abrupt interfaces. Figure 5 shows the morphology of the clean substrate and that of the 7-period sample, as observed by AFM, (a) and (b), respectively. It is interesting to note that despite the overlayer growth (total thickness, ~ 30 nm) and chemical bond manipulation in each period, the original wafer morphology is preserved. This is because the chemical bond manipulation is achieved at RT where the atom diffusion rates responsible for interface roughening are considerably reduced (Grossman and Schubert *et al* 1993). This helps to maintain a smooth interface, reduce the strain in each layer and thereby maintain surface planarity. This is a unique aspect of the chemical bond manipulation and permits the formation of the new structure on a full wafer scale.

Figure 6 shows the PL spectrum from this sample, measured using 325 nm excitation light from a He–Cd laser. Blue–green luminescence is observed (visible to the naked eye) at RT with a spectral peak at ~ 2.6 eV (full width at half maximum of the order of 1 eV). We recorded the PL intensity as a function of temperature and found that the intensity decreases when the sample is brought to room temperature from 13 K. Additionally,

we fabricated samples with different periods (4 and 2) and found that the PL intensity from those samples was significantly low compared to the 7-period sample. We fabricated samples with varying thicknesses of Si and Ge in a similar way and performed PL measurements, in order to examine the effect of quantum confinement on the blue–green emission. We did not observe any shift in the PL peak energy and therefore, rule out the quantum confinement effect as a possible explanation for the light emission. Other possibilities for the emission are defects created at the interface or SiO₂ layer (Skuja 1994). Further, we extended this work to fabricate ultrathin SiO₂ layer buried between Si and Ge layers by oxygen implantation into Si/Ge multilayer structures (Ogino *et al* 1997).

3.2 Fabrication of Ge/CoSi₂/Si nanodiscs and incorporation of metallic feature into Ge quantum dot network

Since it is important to have metallic features on semiconductors for contacts and interconnection in devices, we investigated the behaviour of cobalt atoms on Si and Ge wafers. In the literature, there is a vast amount of work reported in the area of Si/metal systems (von Kanel 1992). However, interaction of metals with systems involving Ge are not many (Ashburn *et al* 1992) and deserves a thorough investigation. From our experiments we find a distinct selectivity with regard to the onset temperature of inward diffusion as shown in figure 7. Co atoms begin to diffuse into bulk Ge at lower temperatures than that in Si. The explanation of this difference in behaviour is mainly attributed to the equilibrium defect concentration and the mechanism of diffusion. By employing this phenomenon we fabricated epitaxial cobalt silicide at the interface between a thin layer of Ge deposited on Si. 2ML of Co is deposited on a

Figure 4. XTEM image from the 7-period sample showing the formation of continuous layers and sharp interfaces. A schematic description of the sample is also given. Auger depth profile data from the 7-period sample is also shown. The sharp dips in the Si LVV signal (due to depletion of elemental Si) and the corresponding peaks in O KLL signal (as indicated by the arrows), confirm the formation of silicon oxide layers, as a result of the chemical bond manipulation.

Figure 5. Surface morphology (AFM) of (a) clean Si(111) substrate, and (b) 7-period sample. The original surface morphology is preserved, as indicated by the clear observation of surface steps.

Si(111) surface, which is covered with ~ 4 ML of Ge, both at room temperature. On annealing the sample at 460°C , Co diffuses through the Ge layer and reacts with the Si atoms underneath. This results in the formation of a buried CoSi_2 layer. XTEM pictures as well as the blocking dip pattern in MEIS suggest that the CoSi_2 formed at the interface between Si and Ge, is epitaxial, twinned at the interface (Prabhakaran *et al* 1996, 1997a). There is significant atomic movement associated with the diffusion and reaction resulting in the formation of nanometre-sized discs. Figure 8 shows the XTEM showing the lattice image as well as the nanodiscs formed in this reaction. Co layer undergoes agglomeration and form Ge/ CoSi_2 /Si dot structures and Ge/Si terraces. A mixture of 1×1 and 5×5 patterns originating respectively from Ge/ CoSi_2 /Si and Ge/Si regions are observed in RHEED and LEED. These results suggest a new way of fabrication of buried epitaxial metallic layers and semiconductor/metal nanostructures.

It is well known that growth of Ge on Si follows the Stranski–Krastanow mode in which about 4 monolayers of Ge form a wetting layer, and if deposited in excess

they form islands. In our group, we have developed a technique by which we can control the spatial distribution of these Ge quantum dots and it is possible to have an ordered arrangement (Hibino and Ogino 1995). By utilizing the selective diffusion of Co atoms through these dots, we can incorporate metallic features into these quantum dots. Figure 9a shows the SEM picture of such a Ge quantum dot network. Figure 9b shows the variation of atomic concentration of Ge and Co on the wetting layer as well as on the dots monitored using SAM as a function of Ar ion sputtering. As can be seen, the Co and Ge signals from the wetting layer decrease and finally disappear as the sputtering proceeds. The Ge signal from the top of the quantum dot also decreases because of the removal of the top few layers. However, the Co signal shows only a marginal decrease and maintains a constant value. These results in combination with XPS, UPS and XTEM results clearly indicate that the dots contain metal atoms in the form of silicide. Photoemission as well as XTEM support this view. It is to be remembered that traditional methods such as ion implantation for metallization, fail in the nanoregime

as the beam damage itself is of the order of several hundred nanometres. The process described here is schematically represented in figure 10.

Figure 6. RT PL spectrum from the 7-period sample excited by He-Cd laser (325 nm, ~ 2 mW)

Figure 7. Temperature dependence of Co concentration, as obtained from the XPS core level spectra, for 1.2 ML of Co deposited at room temperature on Si(100) and Ge(100). The plot shows a distinct selectivity with regard to the onset temperature of inward diffusion.

3.3 Fabrication of Si nanopillars through high temperature reaction of nitric oxide with Si wafer

Several researchers have reported oxynitridation of Si surfaces using NO gas (Hegde *et al* 1995; Lu *et al* 1996; Kamath *et al* 1997). We have employed this reaction in a novel way to fabricate Si nanostructures. Here, we describe the fabrication of Si nanopillars through the high temperature reaction of NO with Si surfaces. Figure 11 gives a schematic representation of the principle behind the process of Si nanopillar formation. When NO molecule approaches the surface, it dissociates and nitrogen atoms stick to the surface forming nitride islands. Since the substrate temperature is high (~ 850–900°C), oxygen atoms etch off the surface Si atoms through the desorption of SiO species and nitride islands act as protective masks for this etching process. Silicon nitride is stable on the surface upto relatively higher temperatures (~ 1200°C). It is to be noted that both the island formation and the etching processes take place simultaneously. Continuation of both the processes by long exposure to NO gas, result in the formation of nanometre-sized Si pillars capped by nitride islands. The advantage with

Figure 8. Cross sectional TEM of the Co(1.2 ML)/Ge(4ML)/Si(111) after annealing at 460°C, showing the formation of flat terraces and nanodisc/dot structures. The bottom figure is the expansion of the disc structure showing the lattice images and indicate the epitaxial cobalt silicide layer formation.

this single step process is that the nitride capping also grows thicker and therefore its protective action is significantly enhanced and even becomes resistant to chemical etching performed outside the UHV. The region between the pillars seem to be preferentially etched by the oxygen rather than get nitridized during the later stages of the reaction. This selectivity may be linked to the thermodynamics and kinetics of the two competitive processes. Figure 11 also shows the AFM in the initial stages of the etching mode reaction on Si(111). Corroded step edges can be observed where the step edges are irregular (due to partial etching) and this phenomenon

Co/Ge(QD)/Si(111)

Figure 9. (a) SEM picture of Ge quantum dot network showing the dots and wetting layer and (b) Auger signals were recorded selectively from these areas and plotted in the figure below. The plot shows that Co atoms can be incorporated into the Ge dots through selective diffusion.

seems to be similar to the step flow growth occurring in the case of homoepitaxy of Si. This means that atom attachment/detachment process takes place selectively at step edges. The AFM image clearly indicates that the etching process starts at the step edges and is sporadically pinned by the formation of nitride islands at step edges.

AES, UPS, XPS and LEED measurements were performed to ascertain the nature of the surface species. The clean surfaces are characterized by the presence of sharp emission from the dangling bond states of the atoms in the termination layer. After the reaction surface states continue to be observed with diminished intensity, and clear peaks (~ 6.4 eV due to N $2p$ -derived orbitals) due to surface nitride species are also observed. This clearly indicates the formation of two different domains consisting of bare Si area and the Si nanopillars capped by the nitride species. In the AES, strong signals due to surface nitride species are observed at kinetic energies 84 eV and 387 eV. Corresponding XPS spectra in the Si $2p$ region shows a clear signal chemically shifted by ~ 2.8 eV typical of the formation of Si_3N_4 type of nitride and the N $1s$ is observed at 398 eV (Tabe and Yamamoto 1996). One important difference between (111) and (100) surfaces is the nature of the N $2p$ -derived peak in the He I UPS. The peak is rather narrow in the case of (111) surface and this, we believe, is due to the better ordering in the nitride layer. For the same reason, the signal due to nitride in the Si $2p$ spectrum is broad in the case of (100) compared to that in (111). The LEED observation also substantiates this view. In the case of (111) surface, during the early stages of the reaction with NO, the 7×7 pattern becomes diffuse. However, as nitrogen coverage increases (~ 1 ML) a mixture of 7×7 and a quadruplet structure emerges. There is some evidence for an 8×8 pattern also at this stage. At higher coverages (2–3 ML), we observe sharp pattern consisting of 1×1 and quadruplet structures. In the case of (100) surface, at higher nitrogen coverages, a diffuse 1×1 pattern is observed.

Figure 12 shows the typical AFM image after the high temperature reaction with NO in the case of (111) and (100) surfaces. The images clearly show high density pillar formation in both the cases. The nature of the pillars and their interaction between one another appears to be quite sensitive to the surface temperature. For example, images (b) and (c) in figure 13 are recorded from different portion of the same sample, with the difference that the attained temperature was slightly higher ($\sim 25^\circ\text{C}$) for the portion imaged in (c) compared to that in image (b). As seen in the AFM images, the pillars seem to undergo coalescence giving rise to a kind of nanometer-sized Si walls. We are performing other experiments to understand this phenomenon.

Figure 13 depicts that the height of these pillars can be increased by performing further etching of Si sepa-

Figure 10. Schematic description of the process to incorporate metal into Ge quantum dots.

Figure 11. (a) Schematic of the high temperature reaction of NO with Si surface, which results in the formation of Si nanopillars capped by nitride islands and (b) is an AFM image showing the morphology in the early stage of the reaction (10 L NO at 850°C). Surface steps become irregular in the etching process.

Figure 12. AFM image after dosing (a) the Si(111), (b). The Si(100) surfaces with NO (1 h at 1×10^{-8} torr) at a substrate temperature of 850°C. (c) is recorded from a different portion of the sample, same as in b, where, the temperature is slightly higher ($\sim 25^\circ\text{C}$).

Figure 13. Experimental demonstration of the fact that the pillar height can be increased through *in situ* and *ex situ* etching processes. (a) image is from a Si(100) exposed to NO at 850°C (1×10^{-8} torr for 1 h) and subsequently etched with O₂ at 850°C (1×10^{-8} torr for 1 h). The comparison of UPS shown as an inset to figure 2b correspond to this stage; (b) the same sample was taken out of the UHV chamber and etched with very dilute (~1–2 %) HF for 25 s; (c) after 3 cycles of oxidation in very dilute (~5%) HNO₃ and HF etch; and (d) is a plot of the typical pillar height after treatments a, b and c.

rately. For example, the pillars formed by reaction of NO at 850°C was subjected to an etching reaction with oxygen (1×10^{-8} torr for 1 h) at ~850°C. After this process, the average pillar height increased by ~1 nm. It is worth noticing that the nature of the nitride species and its amount is unaffected by this etching process, as the N 1s as well as Si 2p signals recorded before and after the etching process were identical. Therefore, the nitride islands formed through the dissociative adsorption of NO are very stable and function effectively as a protective capping layer. Figure 13b is the image of the same sample (same as (a)), which was etched *ex situ* in very dilute HF. The silicon oxide formed due to the exposure to air is expected to be removed in this process. The average pillar height increased by around 1–2 nm due to the removal of Si from the region between the pillars. Image (c) is the same after sample was subjected to three cycles of oxidation and etching in very dilute

HNO₃ and HF, respectively. The average pillar height increased quite significantly by 3–4 nm after this process. Moreover, the small pillars are etched off completely, indicating that size selection is a possibility.

4. Conclusions

We have demonstrated chemical bond manipulation, which is a combination of surface phenomenon and growth, to fabricate novel nanostructures on a full wafer scale emphasizing thereby the role of traditional surface science in nanometre science and technology. For example, by making use of a bonding partner change reaction of oxygen, occurring on the surface, we could fabricate nanometre-sized multiperiod layered structure consisting of Si/SiO₂/Ge on a Si(111) substrate. We have also shown that it is possible to incorporate metallic features into Ge quantum dot network through selective

diffusion. The main challenge in this approach is to control the size and spatial distribution of these nano-structures. We are currently working on Si wafers with various patterns on it to examine the reactivity difference between different regions.

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References

- Ashburn S P, Ozturk M C, Wortman J J, Harris G, Honeycutt J and Maher D M 1992 *J. Electronic Mater.* **21** 81
- Ashoori R C 1996 *Nature* **379** 413
- Avouris Ph, Lyo I-W, Hasegawa Y 1993 *J. Vac. Sci. Technol.* **A11** 1725
- CRC Hand Book of Chemistry and Physics 1989 (Boca Raton, FL: CRC Press) 1989 p. F-177
- Fathy D, Holland O W and White C W 1987 *Appl. Phys. Lett.* **51** 1337
- Fischer P B and Chou S Y 1993 *J. Vac. Sci. Technol.* **B11** 2524
- Grossman H-J and Schubert E F 1993 *Crit. Rev. Solid State Mater. Sci.* **18** 1
- Hegde R I, Tobin P J, Reid K J, Maiti B and Ajuria S A 1995 *Appl. Phys. Lett.* **66** 2882
- Hibino H and Ogino T 1995 *Appl. Phys. Lett.* **67** 915
- Ishizaka A and Shiraki Y 1986 *J. Electrochem. Soc.* **133** 666
- Iyer S S, Patton G L, Stork J M C, Meyerson B S and Hareme D L 1989 *IEEE Trans. Electron. Devices* **36** 2043
- Kamath A, Kwong D L, Sun Y M, Blass P M, Whaley S and White J M 1997 *Appl. Phys. Lett.* **70** 63
- LeGoues F K, Rosenberg R and Meyerson B S 1989 *Appl. Phys. Lett.* **54** 644
- Liou H K, Mei P, Gennser U and Yang E S 1991 *Appl. Phys. Lett.* **59** 1200
- Lu H C, Gusev E P, Gustafsson T, Garfunkel E, Green M L, Brasen D and Feldman L C 1996 *Appl. Phys. Lett.* **69** 2713
- Nayak D, Kamjoo D, Woo J C S, Park J S and Wang K L 1990 *Appl. Phys. Lett.* **56** 66
- Ogino T, Hibino H and Prabhakaran K 1996 *J. Vac. Sci. Technol.* **B14** 4234
- Ogino T, Kobayashi Y, Prabhakaran K, Sumitomo K 1997 *Mater. Res. Soc. Symp. Proc.* (Pittsburgh: MRS) Vol. **438** 211
- Park J S, Karunasiri R P G, Wang J L, Rhee S S and Chern C H 1989 *Appl. Phys. Lett.* **54** 1564
- Patton G L, Iyer S S, Delage S L, Ganin E and McIntosh R C 1988 *Mater. Res. Soc. Symp. Proc.* (Pittsburgh: MRS) Vol. **102** 295
- People R 1986 *IEEE J. Quantum Electron* **22** 1696
- Prabhakaran K and Ogino T 1997 *Surface Sci.* **387** L1068
- Prabhakaran K, Ogino T, Hull R C, Bean J C and Peticolas L C 1994a *Surface Sci.* **316** L1031
- Prabhakaran K, Nishioka T, Kobayashi Y and Ogino T 1994b *Appl. Surface Sci.* **75** 341
- Prabhakaran K, Nishioka T, Kobayashi Y and Ogino T 1993 *Appl. Phys. Lett.* **62** 864
- Prabhakaran K, Ogino T, Scimeca T, Watanabe Y and Oshima M 1994c *Appl. Phys. Lett.* **64** 1839
- Prabhakaran K, Sumitomo K and Ogino T 1996 *Appl. Phys. Lett.* **68** 1241
- Prabhakaran K, Sumitomo K and Ogino T 1997a *Appl. Phys. Lett.* **70** 607
- Prabhakaran K, Sumitomo K and Ogino T 1997b *Appl. Surface Sci.* **117/118** 280
- Prabhakaran K, Matsumoto T, Ogino T and Masumoto Y 1998 *Appl. Phys. Lett.* **72** 3169
- Shen T-C, Wang C, Abeln G C, Tucker J R, Lyding J W, Avouris Ph and Walkup R E 1995 *Science* **268** 1590
- Skuja L 1994 *J. Non-cryst. Solids* **167** 229
- Taber M and Yamamoto T 1996 *Appl. Phys. Lett.* **69** 2222
- von Kanel H *Materials Sci. Rep.* 1992 **8** 193
- Yablonovitch E, Gmitter T J, Meade R D, Rappe A M, Brommer K D and Joannopoulos J D 1991 *Phys. Rev. Lett.* **67** 3380