Exploring the structure as well as electrical and photovoltaic mechanism in PrFe$_{0.5}$Ni$_{0.5}$O$_3$/GaAs heterojunction

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Structural and electrical transport study of PrFe$_{0.5}$Ni$_{0.5}$O$_3$ heterojunction prepared by pulsed laser deposition (PLD) on GaAs (001) substrate is presented. The film deposited on this substrate is polycrystalline and has orthorhombic structure with space group Pbnm. The surface roughness was estimated around 3.4 nm and this increase in surface roughness results essentially from the grain formation. This heterojunction shows usual semiconducting behavior. Resistivity data have been fitted with variable range hopping (VRH) model, and small Polaron Hopping (SPH) model, from which different parameters like density of state at Fermi level $N(E_f)$, hopping energy ($E_h$), hopping distance ($R_h$), and Mott’s characteristic temperature ($T_0$) were estimated. The photovoltaic behavior of this heterojunction was investigated via current–voltage (I–V) characteristics. A mechanism based on the well established model for photovoltaics in conventional semiconductors has been utilized to explain the observed trend in the present perovskite-type metal oxides. This observed behavior in the perovskite-type metal oxide may open a new window for its application in the modern optical based electronics.

1. Introduction

The scientific interest during the last few decades has been devoted to study the heteroepitaxial junctions of perovskite-type metal oxides (ABO$_3$), which can be expected to show novel characteristics and display a high sensitivity to external magnetic and electric fields [1–6]. Various investigations have been carried out with regard to the photovoltaic effect on perovskite-based p–n junctions [7,8]. Orthoferrites RFeO$_3$ (R is a rare earth element) have a ABO$_3$ distorted type perovskite crystal structure and show unusual magnetic and electrical properties [9]. Generally, these orthoferrites are antiferromagnetic insulators and show a strong negative indirect exchange interaction in the network of Fe$^{3+}$–O–Fe$^{3+}$, which causes two magnetic sub-lattices usually in anti-parallel direction [10].

PrFe$_2$O$_{4}$ at room temperature is a wide gap, high spin Mott insulator placed in highly correlated regime [11]. The small angle between the magnetic moments of its sub-lattices leads to formation of weak ferromagnetic moment and shows this type of magnetic behavior [12]. PrNiO$_3$ is a Pauli paramagnetic metal at room temperature with orthorhombic structure and experiences first order MIT at 131 K. The carrier density of PrNiO$_3$ is about $4 \times 10^{20}$ e/cm$^3$ (at 300 K) and at low temperature, PrNiO$_3$ is a charge transfer insulator [13].

Recently, we carried out structural, electrical and magnetic studies on PrFe$_1$–xNi$_x$O$_3$ (0 $\leq$ x $\leq$ 0.5) grown on LaAlO$_3$ (LAO) substrate by a pulsed laser deposition technique and it was found that these films exhibit weak ferromagnetic and semiconducting behavior [14,15]. Further, in this direction, the effects of different substrates and shift heavy ion irradiation (SHI) on thin films of PrFe$_{0.7}$Ni$_{0.3}$O$_3$ and PrFe$_{1-x}$Ni$_x$O$_3$ (x = 0 and 0.3) were also studied [16,17]. Pressure developed due to SHI and
different substrates (due to lattice mismatch) is believed to influence their various observed potential properties [18].

As far as this type of junction is concerned, there is no detailed electrical transport and photovoltaic study existing in literature. In this paper, the effect of large lattice mismatch between film and substrate on electrical properties of PLD grown on PrFe0.5Ni0.5O3/GaAs heterojunction was studied. Room temperature photovoltaic study on this heterojunction was also carried out.

2. Experimental

Polycrystalline bulk target of PrFe0.5Ni0.5O3 was synthesized using a standard solid state reaction technique [19]. Using this well characterized polycrystalline bulk target of PrFe0.5Ni0.5O3, pulsed laser deposition (PLD) technique (KrF excimer laser λ=248 nm, τp=20 ns) was used to grow thin film. Single crystal GaAs (001) wafers were used as substrate and were cleaned sequentially in acetone, methanol and deionized water prior to deposition. Prior to deposition, the chamber was evacuated to the base pressure of 10-5 Torr. During deposition, the partial pressure of oxygen was maintained at 300 mTorr and substrate temperature was maintained at 750 °C. Target to substrate distance was kept at 5 cm and laser energy density was 1.8 mJ/cm2. After deposition, the deposited films were cooled down to room temperature in the ambient oxygen pressure at 20 °C/min. The films thickness was estimated around 200 ± 5 nm using a stylus profilometer. These films were characterized by X-ray diffraction (XRD) with a Rigaku powder diffractometer, a Nanoscope-E atomic force Microscope (AFM) from Digital Instruments, USA, in contact mode and four probe electrical resistivity measurement in the temperature range 130–300 K. To fabricate metal contact (as Schottky diode: Au/PrFe0.5Ni0.5O3/GaAs heterojunction diode with planner geometry) high purity Gold (Au) metal (99.999%) with a thickness of 100 nm was thermally evaporated onto the whole front surface of the wafers at the pressure of 4.3 x 10 -5 Torr. Then, low resistivity Ohmic contacts were followed by a temperature treatment at 590 °C for 3 min in N2 atmosphere. A current source meter, Keithley 2400, was employed to measure the current–voltage (I–V) characteristics of the PrFe0.5Ni0.5O3/ GaAs heterostructure. To investigate the photo-response of the heterojunction, an ordinary incandescent bulb (220 V, 100 W) with wavelengths of about 400–780 nm was used as the light source. During the measurement all the electrodes were kept in dark to avoid the photovoltaic effect due to contact barrier. The inset of Fig. 6 is a schematic representation of the Au/PrFe0.5Ni0.5O3/GaAs device.

3. Results and discussion

3.1. Structure and particle size

The XRD diffractogram of the as-deposited thin films on GaAs(001) is shown in Fig. 1. The sharp peaks of the sample occur at specific Bragg’s angles. The observed peaks are randomly oriented. These peaks are consistent those of with its bulk sample used as thin film target for the present film [18]. Therefore, a polycrystalline growth is observed on this substrate. The obtained diffraction pattern of the present sample was indexed using Powder X software. The data is fitted with orthorhombic structure having Pbnm space group and the fitted lattice parameters are a=5.46734 Å, b=5.5442 Å, and c=7.7655 Å, with interfacial angles α=β=γ=90.00°. These structural results matched JCPDS File Card 782424. However, a slight reduction in observed parameters is seen in comparison with the present system (as this JCPDS File Card is for PrFeO3). This could be due to the Ni3+ ion as its ionic radius rNi3+(rNi3+=0.56 nm) is smaller than that of Fe3+ ion (rFe3+=0.64 nm) (visibly present system is half-doped system). The data is also in close agreement with similar systems reported in existing literature [14,19].

Also, the average particle size was estimated with the help of Scherrer’s equation [20]

\[ D = \frac{0.89\lambda}{\beta \cos \theta} \]  

(1)

where \( \beta = \left(\rho^2_M - \rho^2_i\right)^{1/2} \). Here \( \lambda \) is the X-ray wavelength (1.543 Å for Cu Kα), \( \rho_M \) and \( \rho_i \) are the measured and instrumental broadening in radians respectively and \( \theta \) is Bragg’s angle in degrees. The calculated average grain size for the as-deposited thin film is about 22.3 nm.

The lattice stress (\( \varepsilon \)) is given by

\[ \varepsilon = \left( a_s - a_l \right) / a_l \]  

(2)

where \( a_s = 3.912 \) Å and \( a_l = 3.8115 \) Å are pseudo-cubic lattice parameters of the substrate and film, respectively.

It should also be noted that there is a large lattice mismatch \( \sim 2.63\% \) between the substrate and deposited film. Hence a significant stress is seen in the present film. Therefore, due to the great difference between the structures of the substrate, a polycrystalline growth is observed. This indicates that the existing large strain between film and substrate affects the crystalline quality as well as the different physical (electrical) properties of the film [19].

The dislocation density (\( \delta \)) in a crystal is defined as the total length of dislocations per unit volume and an alternative definition for \( \delta \) is the number of dislocation intersecting a unit area. Here \( \delta \) was calculated by the following relation:

\[ \delta = 1 / D^2 \]  

(3)
As observed here, the calculated dislocation density is $2.01 \times 10^{15}$ line/m$^2$. At such a density, the surface states can play the role of trapping or recombination centers and can significantly affect the electrical properties of heterostructures [21–23]. In some compound semiconductors, dislocations act mainly as centers of non-radiative recombination [22]. Here in our case, the dislocation network may influence the leakage current and its related parameters.

In addition, besides the role of substrate, the partial pressure of oxygen can affect the oxygen stoichiometry, texture and orientation of this kind of films [24]. The peaks of the film on GaAs are quite wider than those of the target [18], pointing out a smaller particle size (as calculated above) or smaller grain boundaries. This may be due to the existing stress in the film or due to the prevailing structural difference between the substrate and the film.

### 3.2. Surface morphology

The effect of the substrate on the surface morphology of PFN thin films was examined by AFM. Fig. 2 shows the AFM results for the film deposited on GaAs. The root mean square of the roughness (RMS roughness) was found to be around 3.4 nm. Also the film prepared on GaAs predominantly consists of platform grains with a diameter of up to 200 nm (see Fig. 2). Furthermore, from the AFM image it is also revealed that the film deposited on GaAs is dense (with higher specific surface area), which could greatly affect its conductive properties, as discussed later [19].

### 3.3. DC conductivity

Fig. 3 shows the electrical resistivity as a function of temperature for this film. Semiconducting behavior is observed in the measured temperature range. It is clear from the graph that the resistivity at room temperature is 22.893 Ω cm. In order to examine the electrical conduction mechanism of PrFe$_{0.5}$Ni$_{0.5}$O$_3$ thin film, $\rho(T)$ is compared to

$$\rho = \rho_0 \exp(W_a/k_B T)$$

which represents simple thermally activated electrical conductivity [25], where an electron or hole (as charge carrier) moves from one localized state to another due to an exchange of energy between the charge carrier and phonon. In the above equation, $\rho_0$ is the electrical resistivity at extremely high temperatures, $W_a$ is the thermal activation energy (or simply activation energy) of the electron or hole, and $k_B$ is the Boltzmann constant. The electrical resistivity of PrFe$_{0.5}$Ni$_{0.5}$O$_3$ is plotted as ln $\rho$ versus $1/T$ (Fig. 4), since the data does not exhibit linear fit throughout the studied temperature range. Therefore, Eq. (4) is not the perfect mechanism to explain the transport for this system. However, this equation can provide a basic idea of thermal activation energies in narrow temperature regions where the data can be fitted. Also it is interesting to see that the film shows (see inset of Fig. 4) different activation energy ($W_a$) regions, the calculated values of $W_a$ show irregular variation with temperature. One possible reason for this may be the effect of temperature on grain boundaries of this material on this substrate. Alternatively, we can say that there may be variation of grain sizes with respect to temperature variation. Another possible reason may be the difference in thermo-electric coefficient (TEC) of substrate and material.
Hence, for this system, we have fitted our data with phenomenon known as variable-range hopping (VRH) doping sites. The carriers in the localized states move by a system, which may tend to localize the carriers at the by chemical substitution generally brings disorder in the system, which may tend to localize the carriers at the doping sites. The carriers in the localized states move by a phenomenon known as variable-range hopping (VRH) [25]. Hence, for this system, we have fitted our data with Mott’s variable-range hopping (VRH) model [25]. The expression for VRH model is

\[
\rho = \rho_0 \exp\left(\frac{T_0}{T}\right)^{1/4}
\]

The parameter \(T_0\) is related to localization length \(\alpha\) by the expression

\[
T_0 = 18\alpha^3/N(E)k_b
\]

where \(N(E)\) is electronic density of states.

Furthermore, mean hopping distance \(R_h(T)\) and hopping energy \(E_h(T)\) as a function of temperature are given below [14]

\[
R_h(T) = \frac{3}{8}\alpha(T_0/T)^{1/4}
\]

\[
E_h(T) = \frac{1}{4}k_b\left(\frac{T_0}{T}\right)^{3/4}T_0^{1/4}
\]

Inset of Fig. 3 shows the \(\ln\rho\) versus \(T^{-1/4}\) plot for the sample. It has been observed that the data fit by VRH model to a certain range of temperature window. This is consistent with other similar systems and suggests that the conduction is governed by the disorder induced localization of charge carriers [14,18].

We have obtained different parameters for DC conductivity by fitting data with above equations and these are summarized in Table 1. The values of \(T_0\) and \(\alpha = 2.5\) Å are used for similar type of systems [14]. The calculated value of \(N(E)\) was found to be larger compared to low Ni doped PrFeO\(_3\) [14,18]. This could be the possible reason for the enhanced conductivity in the present system. It is clear from Table 1 that the values of \(R_h, E_h, W_d\) are quite low.

Table 1

<table>
<thead>
<tr>
<th>(T_0) (K)</th>
<th>(\rho) (mΩ cm)</th>
<th>(W_d) (eV cm(^{-1}))</th>
<th>(N(E)) (\times 10^{10})</th>
<th>(R_h) at 300 K</th>
<th>(E_h) at 300 K</th>
</tr>
</thead>
<tbody>
<tr>
<td>3.561 (\times 10^5)</td>
<td>1528.401</td>
<td>120.12</td>
<td>9.15</td>
<td>5.502</td>
<td>37.912</td>
</tr>
</tbody>
</table>

Also with the addition of Ni, the \(e_g\) electrons of Fe\(^{3+}\) ions become delocalized. It may be possible that the electronic property in the system is due to induced distortion effect caused by two different homovalent transition metals (Fe and Ni). When Fe\(^{3+}\) (high spin with \(r_{p\alpha} = 0.64\) Å) is replaced with Ni\(^{3+}\) (low spin with \(r_{p\alpha} = 0.56\) Å) [14], it will change Fe–O–Fe or Ni–O–Fe bond angle, which affects the tolerance factor \(t = d_{\text{Fe}–\text{O}}/\sqrt{2(d_{\text{Fe}–\text{O}})}\) (Re=rare earth ion and M=transition metal ion) causing change in single particle band width and will have direct effect on the transport properties. It is now well established that transport properties are closely related to the structural distortion and in particular to the arrangement of FeO\(_6\) octahedra [14,18,26].

Further, to understand the hopping mechanism and nature of the strength of electron–phonon (e–ph) interaction, the resistivity versus temperature data was fitted with small polaron hopping model \(\ln(\rho/T)\) versus \(1/T\) [25]. For rare earth transition metal oxides the high temperature part of conduction is believed to be thermally activated by small polarons, similar to the case of manganites [27]. Polaronic transport is observed also in La\(_{1-x}\)Sr\(_x\)FeO\(_3\) (0 \(< x \leq 0.4\)), PrNiO\(_3\), LaNi\(_{1-x}\)Fe\(_x\)O\(_3\) and Bi\(_{1-x}\)Nd\(_x\)FeO\(_3\) at high temperature [13,28,29]. According to this model the expression for electrical resistivity is given by

\[
\rho/T = \rho_0 \exp\left(\frac{E_p}{k_bT}\right)
\]

with \(\rho_0\) being a pre-exponential factor given by \(\tau_{ph}N\epsilon^2R^2C(1-C_e)\exp(-2\alpha R)/k_b\), where \(N\) is the number of transition metal oxides (TMI) per unit volume, \(C_e\) is the ratio of lower valence state to the total TMI concentration, \(\tau_{ph}\) is an optical-phonon frequency (\(\sim 10^{12}\) Hz), \(\epsilon\) is the electronic charge, \(R\) is the average Fe–Fe spacing (\(\sim N^{-1/3}\)), \(\alpha\) is the wave function decay constant, and \(E_p\) is the (polaron) activation energy. The exponential factor \(\exp(-2\alpha R)\) reduces to 1.0 in adiabatic case. The activation energy \(E_p\) is given by [25]

\[
E_p = W_h + W_d/2 \quad \text{for} \quad T > \theta_D/2
\]

\[
E_p = W_d \quad \text{for} \quad T < \theta_D/4,
\]

where \(\theta_D\) is the Debye temperature and is defined by \(\hbar\nu_{ph} = k_b\theta_D\), \(\hbar\) is Planck’s constant, \(W_h\) is the polaron-hopping energy and is equal to \(W_d/2\), \(W_p\) is the polaron binding energy and \(W_d\) is the disorder energy arising due to the energy difference of the neighboring sites at low temperature.

In order to calculate the hopping energy, the resistivity versus temperature is plotted as \(\ln(\rho/T)\) versus \(1/T\) (Fig. 5). From the slope above \(\theta_D/2\), activation energy is estimated. The values of \(\theta_D/2\) are estimated from the
temperature where deviation from linearity occurs in the high temperature zone (Fig. 5). Recently in structurally similar type of thin film deposited on LAO(001), small polaron hopping model was used to explain the conduction mechanism at high temperature [30]. The important parameters obtained while fitting are shown in Table 2.

The value of polaron coupling constant $\gamma_p$ which is a measure of electron–phonon (e–ph) interaction can be estimated from the relation $\gamma_p = \frac{2W_n}{h\nu_p}$ [31]. The estimated values of $\gamma_p$, using $\nu_p = 10^{14}$ Hz for these materials are around 5.93 (Table 2). It is suggested that if $\gamma_p > 4$ then strong e–ph interaction exists in the solids and less than this is considered weak [28]. Therefore, an almost strong e–ph interaction is observed in this system. Also it is observed that the change in bond angle and bond length of the network Fe$^{3+}$–O–Fe$^{3+}$ or Fe$^{3+}$–O–Ni$^{3+}$ has due effect on the electronic transport properties (due to impact on band width) of the system [14,17]. From the value of $\gamma_p$, we have evaluated the ratio of polaron mass $m_p$ to the rigid-lattice effective mass $m^*$ using the relation [31]

$$m_p = \left(\frac{h^2}{8\pi^2}R^2\right)\exp(\gamma_p) = m^* \exp(\gamma_p). \quad (11)$$

The calculated value of $m_p/m^*$ is found to be 55.51 (Table 2), indicating a nearly strong e–ph interaction in these RETMOs at this concentration. Hence, the increase of Ni doping decreases the rigid lattice effective mass, which is also coherent with larger polaron size. According to Eq. (11) for most couplings, the number of phonons is closely related to the exponent of the effective mass. For small electron–phonon coupling, the polaron cloud is virtually empty.

### 3.4. Current–voltage characteristics

The $I$–$V$ characteristics (in dark and presence of light) of the Au/PrFe$_{0.5}$Ni$_{0.5}$O$_3$/GaAs heterojunction diode are shown in (Fig. 6). The $I$–$V$ characteristics of the diode can be analyzed by the following relation [32]:

$$I = I_s \left(\exp\left(\frac{qV}{k_BT}\right) - 1\right) \quad (12)$$

in which $q$ is unit electronic charge, $V$ is the applied bias voltage, $I_s$ is the saturation current and $\zeta$ is the ideality factor. In this equation nearly all kinds of mechanisms, such as thermionic emission, image-force lowering, generation–recombination, interface states, tunneling, and inhomogeneous Schottky barrier, can be taken into account by the change of ideality factor. The ideality factor $\zeta$ is a measure of conformity of the diode to pure thermionic emission and it is determined from the slope of linear region of forward bias $\ln I$–$V$ characteristics through the following relation:

$$\zeta = \frac{q}{k_BT}\left(\frac{dV}{d\ln I}\right) \quad (13)$$

In this system the value of $\zeta$ is around 4.7 at 300 K. This large ideality factor is quite larger than the ideal value of 1 in semiconductor p–n junctions and has been observed also in perovskite-based oxide p–n junctions, where charge trapping at defects in the bulk seems important for transport properties [33].

In order to explain the current transport in present oxide heterojunction, there are several mechanisms which can control metal–oxide–semiconductor structure, Schottky, Poole–Frenkel and Fowler–Nordheim emission mechanisms. Since in an ideal Schottky junction, the main transport mechanism is thermionic emission, in the present case, two possible mechanisms are explored in the metal–oxide interface i.e. Richardson–Schottky (RS) and Poole–Frenkel (PF) emissions [34]. The RS emission generated by the thermionic effect is caused by electron

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**Table 2**

Parameters for small polaron hopping conduction of PrFe$_{0.5}$Ni$_{0.5}$O$_3$/GaAs heterojunction.

<table>
<thead>
<tr>
<th>$\theta_0$(K)</th>
<th>$\nu_p$(Hz)</th>
<th>$W_p$(meV)</th>
<th>$W_a$(meV)</th>
<th>$W_h$(meV)</th>
<th>$\gamma_p$</th>
<th>$m_p/m^*$</th>
</tr>
</thead>
<tbody>
<tr>
<td>201.78</td>
<td>4.14 × 10$^{12}$</td>
<td>45.25</td>
<td>10.46</td>
<td>49.82</td>
<td>5.93</td>
<td>316.15</td>
</tr>
</tbody>
</table>

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**Fig. 5.** Fitting of PrFe$_{0.5}$Ni$_{0.5}$O$_3$/GaAs(001) heterojunction by SPH model.

**Fig. 6.** $I$–$V$ characteristic curves for PrFe$_{0.5}$Ni$_{0.5}$O$_3$/GaAs(001) heterojunction (in dark and in light). Inset shows device diagram.
transport across the potential energy barrier at a metal–oxide interface [34].

The $I$–$V$ behavior is given by [34]

$$I = AT^2 \exp\left(-\frac{\phi_s}{k_B T} + \beta_{RS} V^{1/2}\right)$$

(14)

where $A$ is the junction area, $A$ is the effective Richardson constant, $\phi_s$ is the Schottky barrier height, $\beta_{RS} = 1/(k_B T) (e^2/4\pi\varepsilon_0 d)$ is Schottky-field lowering constant, $\varepsilon_0$ is the permittivity of free space, $\varepsilon$ is the high-frequency dielectric constant of the material, $d$ is the length of the sample, and $V$ is the applied voltage.

Eq. (14) predicts a linear relationship between $\ln(I)$ and $V^{1/2}$ with a slope $\beta_{RS}$ at a constant temperature. The PF bulk-limited process also predicts a linear relationship between $\ln(I)$ and $V^{1/2}$ similar to Eq. (14) with $\beta_{RS}$ replaced by $\beta_{PF}$ and $\phi_s$ replaced by $\phi_{PF}$, which is trap depth.

That is, in order to analyze the leakage current conduction (in dark and light) behaviors of Au/PrFe$_{0.5}$Ni$_{0.5}$O$_3$/GaAs heterojunction film, the experimental values of $\beta_{RS}$ and $\beta_{PF}$ are estimated from the slope of $\ln(I)$ versus $V^{1/2}$ plots given in Fig. 7 using $I$–$V$ characteristic of Au/PrFe$_{0.5}$Ni$_{0.5}$O$_3$/GaAs. Typically two possible mechanisms have been explored in the metal–insulator interface to be the most suitable mechanisms that can explain the obtained behavior: RS emission and PF emission [34]. RS emission is caused by thermionic effect in the lower field region. PF emission is due to field enhanced thermal excitation of trapped electrons in the higher field. The figure shows that the behavior of $I$ versus $V^{1/2}$ is linear in both low and high field regions [34]. The slopes were obtained and used for estimating the $\beta$ values i.e. $k_B T \times$ slope [34]. The experimental values of $\beta_{RS}$ and $\beta_{PF}$ were calculated and compared (for dark and light) and are presented in Table 3.

A significant variation in $\beta_{RS}$ and $\beta_{PF}$ is observed for dark and light conditions. This establishes that the present system is photo-sensitive and can be utilized for technology under proper fabrication conditions.

**Table 3**

<table>
<thead>
<tr>
<th>Sample</th>
<th>$\beta_{RS}$ (meV m$^{1/2}$ V$^{1/2}$)</th>
<th>$\beta_{PF}$ (meV m$^{1/2}$ V$^{1/2}$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Light</td>
<td>0.905</td>
<td>38.97</td>
</tr>
<tr>
<td>Dark</td>
<td>0.727</td>
<td>30.877</td>
</tr>
</tbody>
</table>

It is a well established fact that the grain boundary may influence both the leakage current and breakdown voltage [35]. Leakage current features are determined by grain boundary intersection and breakdown field is determined by grain boundary itself. Also the grain boundaries in dielectrics have higher resistance than that of the bulk of grains; hence, high electric field will be developed across these grain boundaries under applied DC field, and therefore can cause local breakdown or field–assisted emission of trapped charge carriers [36,37]. In addition, space charges like oxygen vacancies in the oxide film are created at the grain boundaries during the deposition process [38] and as a result, the potential barrier height of grain boundary reduces to decrease the density of states at the grain boundary [39]. Also the height of the grain boundary potential barrier and activation energy for electronic conduction can tune the current level [40]; therefore, resultant leakage current rises with the domain area of grain boundaries. It should be noted that large values of $\beta$ and $\phi_s$ are caused by a high density of surface states (misfit dislocations) at the heterojunction boundary.

4. Conclusion

Thin films of PrFe$_{0.5}$Ni$_{0.5}$O$_3$ were deposited by PLD on GaAs (001) substrate. The film deposited on GaAs(001) is polycrystalline, of single phase and has orthorhombic structure with Pbnm space group. It shows semiconducting behavior. AFM of film deposited on GaAs also supported the polycrystalline growth. Resistivity data have been fitted with VRH model and SPH model, from which different parameters were estimated. The photovoltaic behavior of this heterojunction was investigated. The Schottky barrier height and Schottky-field lowering constant were also calculated. All the observed properties are explained by the substrate induced strain effect in the system. The heterojunction of the perovskite oxides and conventional semiconductors might open up scope for advanced generation microelectronic and optoelectronic devices.

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