

Electrical properties and scaling behaviour of Sm^{3+} doped CaF_2 -bismuth borate glasses

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Abstract. The electrical properties for $20\text{Bi}_2\text{O}_3$ – $60\text{B}_2\text{O}_3$ ($20-x$)– CaF_2 – $x\text{Sm}_2\text{O}_3$ glasses ($0 \leq x \leq 2$) were measured in the temperature range 297 K up to 629 K and in the frequency range 0.1–100 kHz. The d.c. and a.c. conductivity values and the dielectric loss ($\tan \delta$) values were found to increase with increasing Sm_2O_3 content, whereas the activation energy of conductivities and the dielectric constant decreased. These results were attributed to the introduction of the rare earth ions; promote the formation of a high number of non-bridging oxygen atoms, which facilitate the mobility of charge carriers. The frequency dependence of the a.c. conductivity follows the power law $\sigma_{ac}(\omega) = A\omega^s$. The frequency exponent (s) values ($0.64 < s < 0.8$) decrease with increasing temperature. This suggested that the a.c. conduction mechanism follows the correlated barrier hopping model (CBH). The dielectric constant (ϵ') and dielectric loss ($\tan \delta$ values) were found to increase with increasing temperature and increasing Sm_2O_3 concentration in the glass. The a.c. conductivities as a function of frequency at different temperatures of a given glass superimposed onto a master curve (Rohling scaling model). Furthermore, we have performed to scale the data as a function of composition. Two master curves were obtained, which suggested that there are differences in dominant charge carriers between glasses having Sm_2O_3 concentration ≥ 1 and glass of Sm_2O_3 concentration < 1 .

Keywords. Electrical properties; rare earth; glass.

1. Introduction

Borate glasses are structurally more intricate as compared to silicate or phosphate glasses due to two types of coordination of boron atoms with oxygens (3 and 4). It is well established that the addition of metal oxides converts the boron coordination and the structural groups from one to another depending on the type and concentration of metal oxides (Zhong and Bray 1989; Prasad *et al* 2006; Shaw *et al* 2006). Bismuth borate glasses are interesting because their properties are exploited in applications such as wave-guides in non-linear optics, scintillation detectors in high-energy physics, reflecting windows and layers for optoelectronic devices (Ardelean *et al* 1996; Hazra and Ghosh 1997; Stehle *et al* 1998; Murugan and Varma 2001). As Bi^{3+} ions have small field strength, they cannot work as a network former; however, in combination with other network formers, like B_2O_3 , glass formation is possible. The variation in physical, optical and electrical properties has been investigated as a function of Bi_2O_3 content in 20CaO – $x\text{Bi}_2\text{O}_3$ – $(80-x)\text{B}_2\text{O}_3$ ($0 \leq x \leq 60$) glasses. The d.c. conductivity is found to decrease with the addition of Bi_2O_3 content up to 30 mol%, and with further addition conductivity increases. The decrease in conductivity could be interpreted in terms of

the decrease in mobility and proportion of Ca^{2+} ions available for conduction because Ca^{2+} may be interacting with $[\text{BiO}_6]^{3-}$ and $[\text{BiO}_5]^{2-}$ units. The conductivity is ionic and there is no contribution of electronic conductivity. Beyond 30 mol%, the electronic contribution to the total conductivity becomes relevant (Sanghi *et al* 2004). The introduction of CaF_2 into the glass matrix lowers its viscosity, decreases its liquidus temperature and further it leads to the formation of a large concentration of colour centres (Rao *et al* 2002). The effect of replacing CaO with CaF_2 on electrical conductivity of borosilicate glass ceramics was studied (Assem 2005). It was found that the existence of fluorine ions increases the electrical conductivity. Rare-earth ion-doped glasses are technologically important materials as they find applications in lasers, white light emitting diodes, light converting, and optoelectronic devices (Canalejo *et al* 1988; Zheng and Clare 2005).

The addition of Sm_2O_3 to PbO – PbF_2 – B_2O_3 – Sm_2O_3 glass induced the formation of the $[\text{BO}_4]$ units, beside the disruption of diborate units which induces the formation of “loose” diborate, chain- and ring-metaborates (Souza Filho *et al* 2000). They concluded that the rare earth ions promote the formation of a high number of non-bridging oxygen atoms. Ni *et al* (2007) reported that Sm^{3+} ions exist as network modifier in B_2O_3 – Al_2O_3 – SiO_2 glasses containing Sm_2O_3 and the higher Sm^{3+} ions concentration was the more $[\text{BO}_3]$ transitioned to $[\text{BO}_4]$.

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Influence of rare earth additives on electrical conductivity of different glasses has been previously studied. The results indicated that the effect of rare earth additives on electrical conductivity depend on the glass composition and atomic size of rare earth ions (Reau and Polain 1989). In addition, a decrease in conductivity with increase in atomic weight of rare earth ions has been observed in different rare earth ions doped tellurite glasses and is attributed to the slow mobility of rare earth ions due to their heavy masses (Kumar and Sankarappa 2009). Tellurite glasses in the composition $80\text{TeO}_2-20\text{ZnO}$ doped with Ho^{3+} ions, exhibited a decrease in conductivity with increase in Ho^{3+} ions content. This has been attributed to the formation of quasimolecular complex of rare earth ions with glass network (Shaaban *et al* 2006). In phosphate glasses doped with rare earth ions, the total glass conductivity increased by an order of magnitude and activation energy remained constant or decreased negligibly with increase of La_2O content. They argued that due to an ionic contribution to the total conductivity in addition to electronic (Devidas *et al* 2008). A series of rare earth doped sodium borate glasses $(\text{Na}_2\text{O})_{35.7}(\text{RE}_2\text{O}_3)_{7.2}(\text{B}_2\text{O}_3)_{57.1}$ (RE = Sm, Gd, Dy, Ho, Y, Er and Yb), were prepared and their electrical conductivity were determined (Nakayama *et al* 2010). The electrical conductivity slightly decreased with increasing ionic radius of RE^{3+} and the highest electrical conductivity was achieved for $(\text{Na}_2\text{O})_{35.7}(\text{Yb}_2\text{O}_3)_{7.2}(\text{B}_2\text{O}_3)_{57.1}$ glass. The d.c. and a.c. electrical conductivities of barium tellurite borate glass doped with Nd_2O_3 in the composition, $50\text{B}_2\text{O}_3-(20-X)\text{BaO}-20\text{TeO}_2-10\text{LiF}$ or Li_2O where $x = 0.5, 1, 1.5$ and 2 Nd_2O_3 were measured. The d.c. and a.c. conductivity values increase, whereas the activation energy of conductivities decreases with increasing Nd_2O_3 content in the glasses containing LiF. Whereas the addition of Nd_2O_3 to glasses containing Li_2O led to decreased conductivity and increased activation energy (Ali and Shaaban 2010).

In this work the effect of the addition of Sm_2O_3 on a.c. conductivity and dielectric properties of CaF_2 -bismuth borate glasses, was studied. The frequency, temperature and Sm_2O_3 concentration dependence of the a.c. conductivity has been analysed, in an attempt to investigate the conduction mechanism. A scaling behaviour model of a.c. conductivity of such glass was suggested.

2. Experimental

2.1 Glass preparation

Reagent grade H_2BO_3 , Bi_2O_3 , CaF_2 and Sm_2O_3 were used as starting materials for preparing the glass of compositions $20\text{Bi}_2\text{O}_3-60\text{B}_2\text{O}_3-(20-x)\text{CaF}_2-x\text{Sm}_2\text{O}_3$ mol% ($0 \leq x \leq 2$). For each composition, raw materials were mixed using a mortar. Each mixture was melted in an alumina crucible at 1000 K for 30–40 min in air. The glass melts were stirred occasionally with an alumina rod to achieve good homogeneity melts. The highly viscous melt was cast into a

cylindrically shaped split-mold of mild steel. The glass produced was annealed at 400°C in a second furnace for 1 h. Finally, the furnace was switched off and the glass was allowed to cool gradually *in situ* for 24 h.

2.2 A.C. and d.c. conductivity measurements

The samples were prepared in the form of discs of 1 cm diameter, and polished by usual techniques to a thickness of 0.2 cm. The flat surfaces of the cut samples were highly polished and cleaned. The constructed cell for the electrical measurements consists of a silica tube surrounded by nickel chrome wire as a heater. A chromel–alumel thermocouple (inside the tube) was used for temperature measurements. The a.c. conductivity was measured by applying a complex impedance technique. A constant a.c. voltage ($V_{r.m.s} = 1$ V) was applied to the sample and the current through the sample was determined by measuring the potential difference across an ohmic resistor connected in series with the sample by using a lock-in amplifier (Stanford Research System SR510). The lock-in amplifier simultaneously measures the voltage across the resistor and the phase difference ϕ between this voltage and the applied voltage on the sample. Since the voltage drop on the ohmic resistance is in phase with the current I , therefore, we can say that ϕ is the phase angle between the voltage drops on the sample and the current, I , passing through it. The a.c. conductivity, σ_{ac} , the dielectric constant, ϵ , and the dielectric loss, $\tan \delta = (1/\tan \phi)$, were calculated using a computer program. Furthermore, the d.c. conductivity was measured by applying a constant voltage (10 V) and measuring the current, then application of Ohm's Law. The current was measured using Keithley Electrometer type 617. To overcome the effect of humidity, the electrical conductivity was measured under vacuum. The measurements were made from room temperature, 297 K up to 629 K. Silver electrodes were painted directly on the two flat surfaces. To avoid initial polarizations, if any, the current measurements were actually started after about 5–10 min of keeping constant voltage across the sample.

3. Results and discussion

3.1 D.C. conductivity

The temperature dependence of the measured d.c. conductivity ($\log \sigma_{dc}$) is represented in figure 1. The d.c. conductivity values increase with increasing temperature. The temperature dependence of d.c. conductivity extracted from conductivity spectra at high temperature is found to obey the Arrhenius relation

$$\sigma_{dc} = A \exp(-\Delta E_{dc}/kT), \quad (1)$$

where A is the pre-exponential factor, k the Boltzmann constant and ΔE_{dc} the activation energy. The σ_{dc} and ΔE_{dc}

values at room temperature obtained from figure 1 are given in table 1. In general, it is noticed that the electrical conductivity increased whereas activation energy values decrease with increasing Sm_2O_3 concentration in the glass.

The optical absorption, transmission and emission were measured for the glasses used in this work by Ali (2009). The results revealed that the optical bandgap, polaron radius and inter-nuclear distance decrease with increasing concentration of Sm_2O_3 . However, the density and Urbach energy were found to increase. He attributed this behaviour to changes induced in glass structure and as a result the formation of greater number of non-bridging oxygen. Thus the introduction of RE changes the oxygen bonding in glass forming network (the formation of non-bridging oxygen), which in turn changes the optical bandgap. As indicated above (Ni *et al* 2007) the increase of rare earth ions in the glass, induce disruption of diborate units which promote the formation of a high number of non-bridging oxygen atoms. The increase in d.c. conductivity with replacement of CaF_2 by Sm_2O_3 can be discussed as follows: (i) the ionic radius of Sm^{3+} (0.096 nm) is smaller than ionic radius of Ca^{2+} (0.099 nm) and (ii) the substitution of CaF_2 by Sm_2O_3 will decrease F^- anions and increase the number of non-bridging oxygen atoms.

Kalužný *et al* (2009) studied the electrical conductivity of MnF_2 - ZnF_2 - NaPO_3 glasses. They deduced that electroneg-

ative fluorine anions F^- freeze electrons and increase the potential barrier between sodium sites, which increases activation energy and decreases anionic mobility. Thus the substitution of CaF_2 by Sm_2O_3 will decrease F^- anions and increase the number of non-bridging oxygen atoms. This facilitates the mobility of charge carriers Ca^{2+} ions and electrons, which in turn explains the increase of σ_{dc} conductivity and decrease of ΔE_{dc} , with increasing Sm_2O_3 concentration.

3.2 A.C. conductivity

Figure 2 represents the temperature dependence of the a.c. conductivity (σ_{ac}), at chosen frequencies of 1 kHz (as example). The frequency dependence of a.c. conductivity for two of the studied glasses is presented in figure 3(a, b). The results obtained from a.c. conductivity measurements indicate that (i) the a.c. conductivity values in studied frequency range increases with increase in temperature, which reveals that the present samples are semiconducting in nature (Devadas *et al* 2008), (ii) the conductivity at all studied frequencies increased with increase in the rare earth ion concentration giving an indication that the addition of rare earth facilitates the mobility of charge carriers Ca^{2+} ions and electrons, in a similar trend observed in the d.c. conductivity, (iii) the measured a.c. conductivity of all the samples, as shown in table 1, in the temperature range studied is found to be greater than that of their d.c. conductivity, and (iv) in the frequency range studied, the conductivity increased with increase in frequency indicating dispersive behaviour of conductivity.

The frequency dependency of the electrical conductivity analysed using the general relation is (Mansingh and Dhawan 1983)

$$\sigma_{\text{ac}} = A\omega^s \quad \text{or} \quad \ln \sigma_{\text{ac}} = \ln A + s \ln(\omega), \quad (2)$$

where A is a temperature dependent constant and s the frequency exponent. Equation (2) has been found to be applicable for all low mobility amorphous and crystalline materials (Mansingh and Dhawan 1983). The frequency dependence of a.c. conductivity at different temperatures for samples 1 and 5 are shown in figure 3(a, b). Similar nature of variation of $\ln(\sigma_{\text{ac}})$ with $\ln(f)$ has been observed in remaining

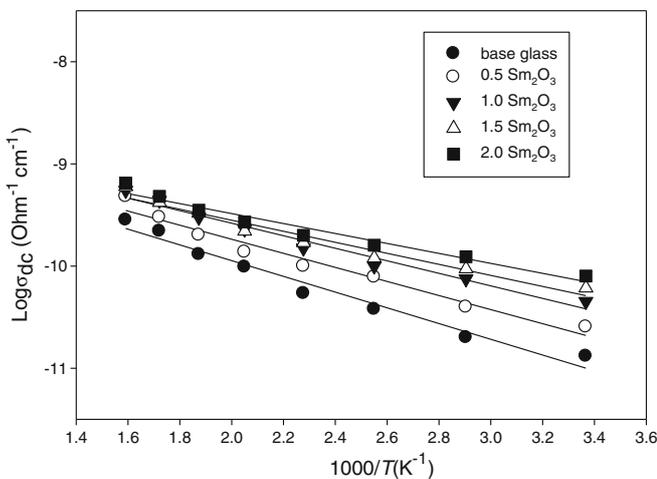


Figure 1. Variation of $\log \sigma_{\text{dc}}$ vs $1000/T$ for all glass samples.

Table 1. Calculated values of the conductivity, activation energy calculated at room temperature and a frequency of 1 kHz, frequency exponent s , the exponent β , W_m and density of states, N , for CaF_2 bismuth borate glasses doped with Sm_2O_3 .

Sample no.	Sm_2O_3 mol%	$\log \sigma_{\text{dc}} (\Omega^{-1} \text{cm}^{-1})$ at room temperature	$\log \sigma_{\text{ac}} (\Omega^{-1} \text{cm}^{-1})$ at 1 kHz and room temperature	ΔE_{dc} (eV)	s	β	W_m	N
1	0.0	-10.699	-9.57	0.6017	0.752	0.248	0.703	3.483×10^{17}
2	0.5	-10.40	-9.5	0.5788	0.735	0.265	0.657	2.91×10^{17}
3	1.0	-10.12	-9.28	0.5254	0.749	0.251	0.613	3.518×10^{17}
4	1.5	-10.02	-9.23	0.4652	0.763	0.240	0.641	4.624×10^{17}
5	2.0	-9.9	-8.89	0.4203	0.795	0.205	0.752	11.79×10^{17}

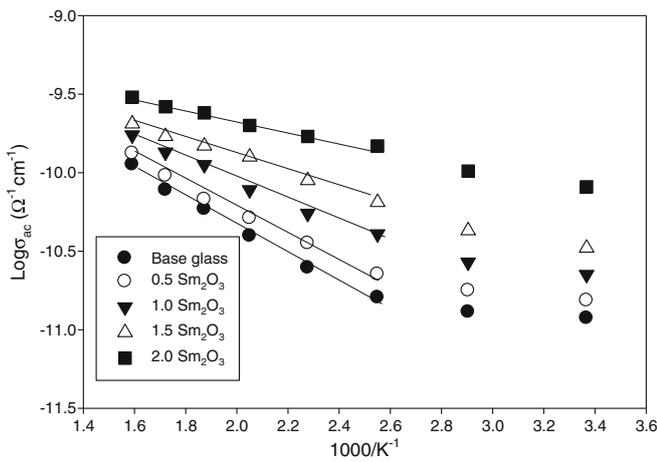


Figure 2. Variation of $\log \sigma_{ac}$ vs $1000/T$ for all glass samples at 1 kHz.

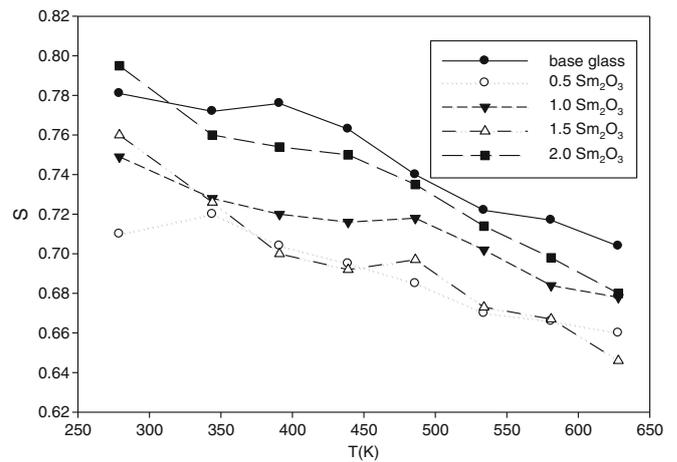


Figure 4. Variation of frequency exponent value (s) vs temperature for all glass samples.

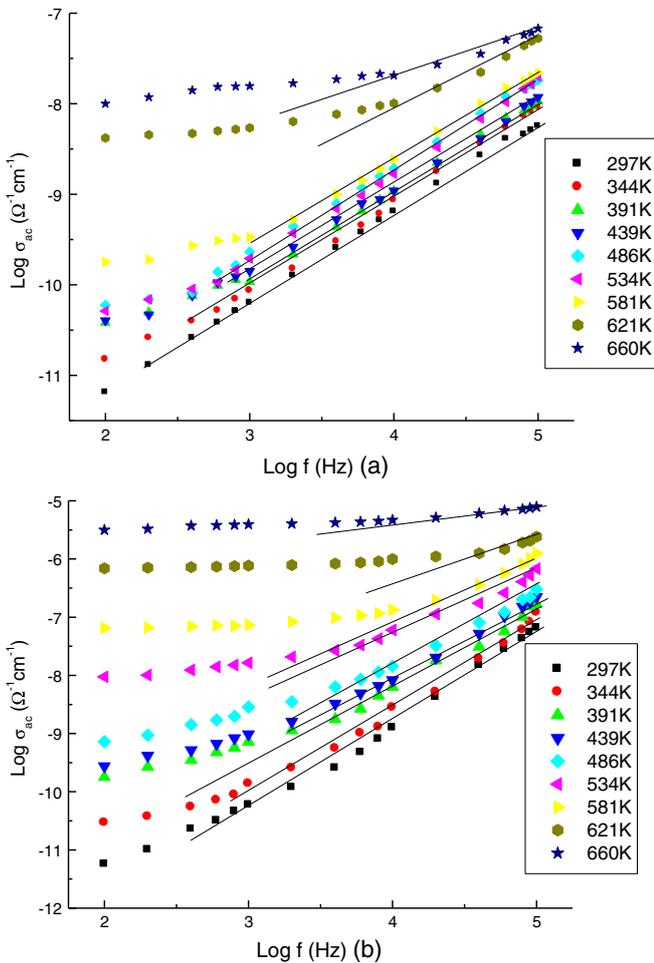


Figure 3. Variation of $\log \sigma_{ac}$ vs \log frequency for base glass (a) $x = 0 \text{ Sm}_2\text{O}_3\%$ and (b) $x = 2 \text{ Sm}_2\text{O}_3\%$.

samples. In figure 3(a, b), curves appear to be linear over a wide range of frequency at lower temperatures and short range of frequency at high temperatures. This observation is

in agreement with the results reported earlier (Hogarth and Basha 1983; Mogus Milankovic *et al* 2005; Bih *et al* 2008). The least square linear lines were fit to the data over the range of frequency in which the linear behaviour is observed. The slopes of the least square fit lines gave directly the frequency exponent s values. The s values are almost less than unity and decrease with temperature. The numerical values of s at room temperature are in the range $0.64 < s < 0.8$. These ranges of s values indicate that the carrier transport is predominantly due to hopping of electrons. Owen (1963) reported that the electrical properties of glasses in an alternating field depend not only on the mobile ions but also on the relatively immobile ions, which also take part in network forming. The a.c. conductivity has been evaluated for single electron motion undergoing quantum mechanical tunneling (QMT). In this model (Ghosh 1993), prediction was made to the effect that the conductivity is linearly dependent on temperature and the frequency exponent; s is independent of temperature but frequency dependent. However, it can be clearly seen from figure 4 that the frequency exponent s decreases with increase in temperature, thereby conflicting with the prediction of QMT model. The limiting value, $s \approx 1$, is observed in some ionic glasses at very low temperatures. Many reports evolving hopping models give the values of s in the range 0.6–1 (Nowick *et al* 1994; Maass *et al* 1995; Lanfredi *et al* 2002). A great number of amorphous materials showed that $s \rightarrow 1$ for $T \rightarrow 0$.

In high temperature we may suppose that the a.c. conduction mechanism follows the correlated barrier hopping model (CBH) proposed by Elliot (1987).

In this model the electrons in the charged defect state hop over the columbic barrier whose height is given as W according to:

$$W = W_m - \frac{4ne^2}{\epsilon R}, \tag{3}$$

where W_m is the maximum height of energy of the energy band, ε an effective dielectric constant, e the electronic charge, n the number of hopping electrons. In the case of glass, $n = 2$ and R is the distance between the hopping sites. The relaxation time τ for the electron hopping over a barrier of height, W , is given by

$$\tau = \tau_0 \exp(-W/k_B T), \quad (4)$$

where τ_0 is the order of atomic vibration period $=10^{-13}$ and k_B the Boltzmann constant. The final expression for the a.c. conductivity can be expressed by

$$\sigma_{ac} = \left[\frac{\pi^2 N^2}{24} \varepsilon \left(\frac{8e^2}{\varepsilon W_m} \right) \frac{s}{\tau_0^\beta} \right], \quad (5)$$

where N is the concentration of localized states (the density of states) and β is given by

$$\beta = \frac{6}{W_m} kT, \quad (6)$$

and

$$s = 1 - \beta. \quad (7)$$

Using (5)–(7) the density of state (N) can be calculated. The N , s , β and W_m values for the studied glasses at room temperature and at frequency equal to 1 kHz are listed in table 1. N values increase with increasing concentration of Sm_2O_3 (except for sample 2) due to the increase in non-bridging oxygen ion content with increasing Sm_2O_3 ion concentration. As shown in figure 4, the measured frequency exponent values are in agreement with the CBH model of single electron hopping. This confirms the fact that CBH model is adequate to explain the frequency dependence of conductivity of the present glasses

3.3 Dielectric properties

Figures 5a and 6a represent the variation of dielectric constant (ε') and dielectric loss ($\tan \delta$) values with frequency at different temperatures for glass containing 1.5% Sm_2O_3 (as example). It can be noticed that both the dielectric constant and dielectric loss values increase with increasing temperature. Such increase in both values is more pronounced at lower frequencies. This behaviour can be attributed to the applied electric field, which assists electron hopping between two different sites in glasses. At high temperatures, the jump frequency of the charge carrier becomes large and comparable with the frequency of the applied field. Accordingly at low frequency the charge carriers hop easily out of the sites with low free energy and tend to accumulate at sites with high free energy barriers. This leads to a net polarization and gives an increase in the dielectric constant. However at high frequency, the charge carriers will no longer be able

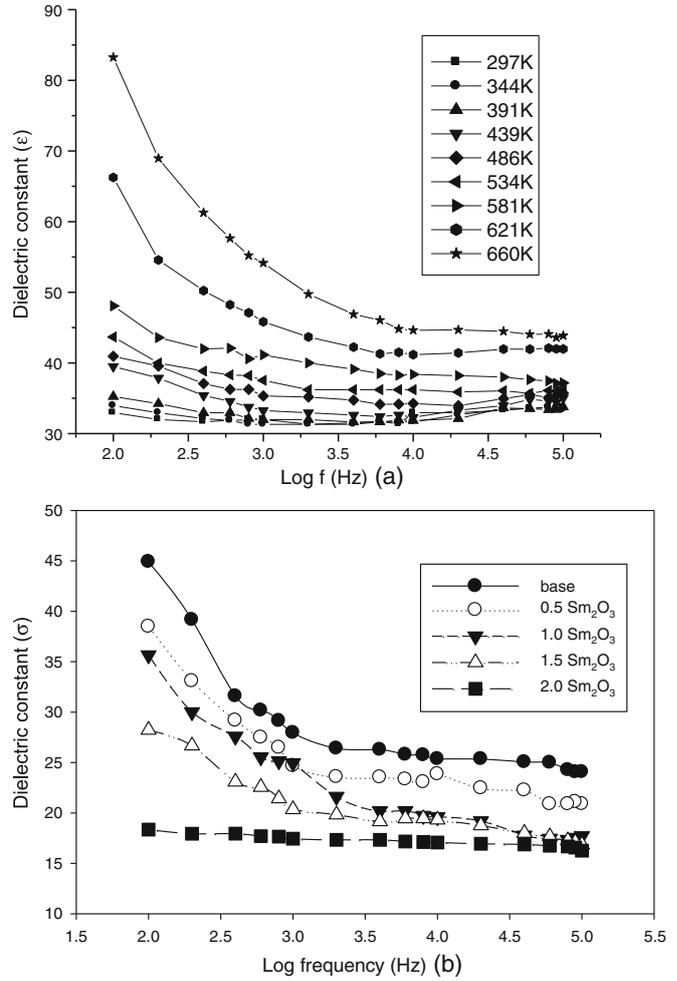


Figure 5. (a) Variation of dielectric constant (ε) with frequency for glass samples ($x = 1.5 \text{ Sm}_2\text{O}_3\%$) at different temperature and (b) frequency response of the dielectric constant (ε) for all glass samples at 629 K.

to rotate sufficiently rapidly, so their oscillation will begin to lag behind this field resulting in a decrease of dielectric constant (Szu and Lin 2003; Sindhu *et al* 2005). At low temperatures, jump frequency of the charge carriers becomes smaller than the frequency of the applied field. The periodic reversal of the applied field takes place so rapidly that there are no excess charge carrier jumping in the field direction, and the polarization due to charge piling up at high free energy barrier sites disappears, which lead to a decrease in the values of ε and $\tan \delta$ (Sindhu *et al* 2005). The dielectric loss values increase whereas the dielectric constant values decrease with increasing Sm_2O_3 concentration in the glass (figures 5b and 6b). As mentioned before, the RE ions exist in glass network as a modifier and increase in the non-bridging oxygen. Thus the introduction of Sm^{3+} ions weakens the glass network and creates pathways suitable for migration of free ions that build up space charge polarization leading to an increase in the dielectric parameters as observed earlier (Balaji Rao

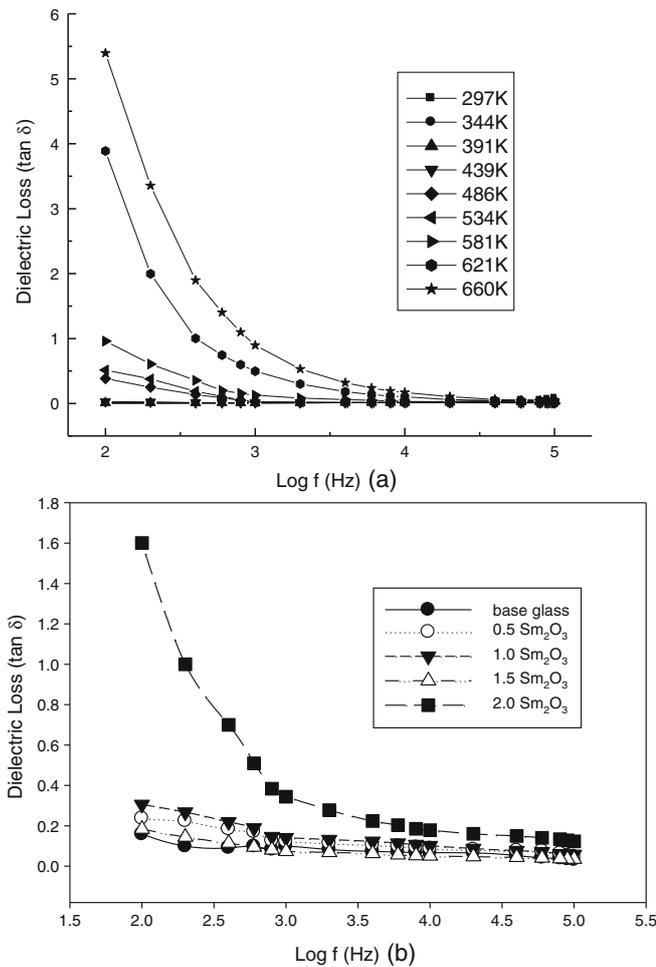


Figure 6. (a) Variation of dielectric loss ($\tan \delta$) with frequency for glass samples (1.5 $\text{Sm}_2\text{O}_3\%$) at different temperatures and (b) variation of dielectric loss ($\tan \delta$) with frequency for glass samples at 548 K.

et al 2004; Shyam Prasad *et al* 2004). The decrease in dielectric constant values with the increase in Sm_2O_3 concentration in the glass may be attributed to cross-linkage between bismuth and samarium ions.

The dielectric loss is equivalent to the a.c. conductivity as can be seen by the relation:

$$\tan \delta = \frac{4\pi\sigma_{ac}}{\omega\varepsilon'}. \quad (8)$$

Thus the same comments made on the a.c. conductivities will also apply here. As mentioned above the change in glass structure accompanied by the introduction of rare earth ions into glasses, which in turn increases the electrical conductivity and dielectric loss with increasing Sm_2O_3 concentration.

3.4 Scaling model

The d.c. conductivity is a parameter that is frequently used in scaling methods. Therefore, we have performed a scaling

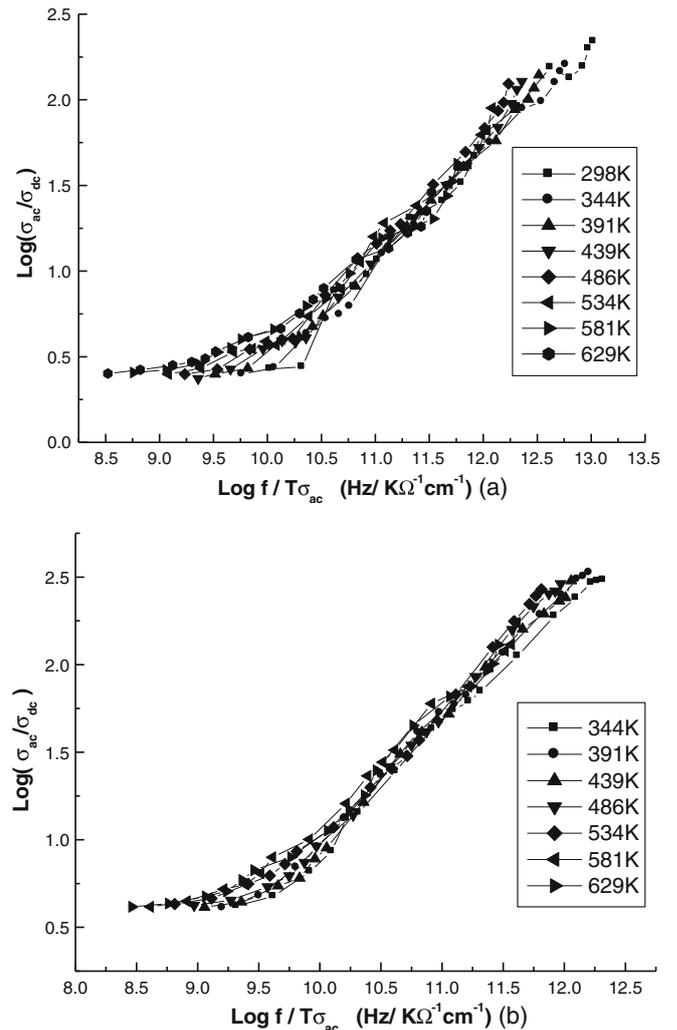


Figure 7. Relation between $\log(\sigma_{ac}/\sigma_{dc})$ and $\log(f/T\sigma_{dc})$ at different temperatures for (a) sample 2 and (b) sample 5.

process of the a.c. conductivity as a function of frequency (figures 7a, b), which show the conductivity master curves of the samples 2 and 5, where we have used $\log(\sigma_{ac}/\sigma_{dc})$ as the y-axis scaling parameter and $\log(f/T\sigma_{dc})$ as the x-axis scaling parameter. This scaling was proposed by Roling *et al* (1997) who had found that the product $T\sigma_{dc}$ obeys an Arrhenius relation. He had obtained conductivity master curves for some ionic conducting glasses by using this scaling parameter. From figure 7(a, b), it is obviously seen that a quite satisfying overlap of the data at different temperatures on a single master curve illustrates well the dynamic processes occurring at different frequencies need almost the same thermal activation energy. Another indication of those scaled master curves is that all ‘‘Arrhenius’’ temperature dependence of conductivity is embedded in the d.c. conductivity term (Saafan 2008). Furthermore we have performed to scale the data as a function of compositions, where we have used $\log(\sigma_{ac}/\sigma_{dc})$ as the y-axis scaling

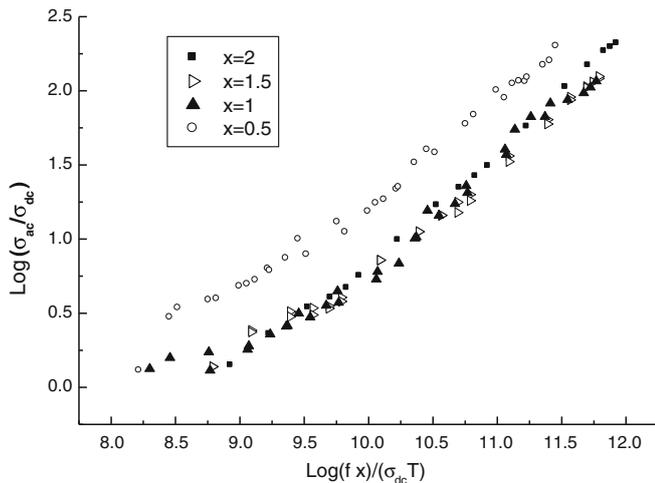


Figure 8. Relation between $\log(\sigma_{ac}/\sigma_{dc})$ and $\log(fx/T\sigma_{dc})$ for all glass samples.

parameter and $\log(fx/T\sigma_{dc})$ as the x -axis scaling parameter (figure 8). We use x = the concentration of Sm_2O_3 as an additional scaling factor for the frequency axis. Figure 8 indicates that glass composition of Sm_2O_3 concentration of 1, 1.5 and 2 overlap in a single master curve. However, glass of Sm_2O_3 concentration of 0.5 fail to overlap with other glasses. A similar trend was also observed in d.c., a.c. and N values. This indicate that there are differences in dominant charge carriers between glasses having Sm_2O_3 concentration ≥ 1 and glass of Sm_2O_3 concentration < 1 (Maass *et al* 1995).

4. Conclusions

The a.c. and d.c. electrical conductivities and dielectric loss values for $20\text{Bi}_2\text{O}_3$ 60BO_3 $(20-x)\text{CaF}_2$ $x\text{Sm}_2\text{O}_3$ glasses ($0 \leq x \leq 2$), were increased with increasing Sm_2O_3 concentration. Both were explained as increasing of the Sm_2O_3 concentration promotes the formation of a high number of non-bridging oxygen atoms, which enhance the mobility of charge carrier ions and electrons. The frequency exponent (s) values, measured at different temperatures, suppose that the a.c. conduction mechanism follows the correlated barrier hopping model (CBH). The observed decrease in dielectric constant with increasing Sm_2O_3 concentration in the glass could be explained as to cross-linkage between bismuth and samarium ions, i.e. the formation of Bi–O–Sm bond. The d.c. conductivity values were used as a parameter in scaling the a.c. conductivities as a function of frequency (Rohling scaling model). The data at different temperatures of a given glass is superimposed in a master curve. This suggested that the scaled master curves illustrate (i) the dynamic processes occurring at different frequencies need almost the same thermal activation energy, (ii) all ‘‘Arrhenius’’ temperature dependence of conductivity is embedded in the d.c. conducti-

vity term. Furthermore, we have performed to scale the data as a function of composition. Glass of Sm_2O_3 concentration ≥ 1 overlap in a single master curve, whereas glass of Sm_2O_3 concentration of < 1 fails to overlap with other glasses. This suggested that there are differences in dominant charge carriers between glasses having Sm_2O_3 concentration ≥ 1 and < 1 . On attempting to scale the data as a function of Sm_2O_3 concentration, two master curves were obtained. These results suggested also differences in dominant charge carriers between glasses having Sm_2O_3 concentrations ≥ 1 and < 1 .

References

- Ali A A 2009 *J. Lumin.* **129** 1314
 Ali A A and Shaaban M H 2010 *Solid State Sci.* **12** 2148
 Ardelean I, Ilonca Gh, Simon V, Cozar O, Ioncu V and Filip S 1996 *Solid State Commun.* **98** 651
 Assem E E 2005 *J. Phys.* **D38** 942
 Balaji Rao R, Gopal N O and Veeraiah N 2004 *J. Alloys Compd.* **368** 25
 Bih L, Abbas L, Nadiri A, Khemakhem H and Elouadi B 2008 *J. Mol. Struct.* **872** 1
 Canalejo M, Cases R and Alcalá R 1988 *Phys. Chem. Glasses* **29** 187
 Devidas G B, Sankarappa T, Prashant Kumar M and Santosh Kumar 2008 *J. Mater. Sci.* **43** 4856
 Elliot S 1987 *Adv. Phys.* **36** 53
 Ghosh A 1993 *Phys. Rev.* **B47** 23
 Hazra S and Ghosh A 1997 *Phys. Rev.* **B56** 8021
 Hogarth C A and Basha M J 1983 *J. Phys D: Appl. Phys.* **16** 869
 Kalužný J, Kubliha M, Vladimír L, Torkia D and Marcel P 2009 *J. Non-Cryst.* **355** 2003
 Kumar M P and Sankarappa T 2009 *J. Non-Cryst. Solids* **355** 295
 Lanfredi S, Saia P S, Lebullenger R and Hernandez A C 2002 *Solid State Ionics* **146** 329
 Maass P, Meyer M and Bunde A 1995 *Phys. Rev.* **B51** 816
 Mansingh A and Dhawan V K 1983 *J. Phys. C: Solid State Phys.* **16** 675
 Mogus Milankovic A, Santic A, Reis S T, Furic K and Day D E 2005 *J. Non-Cryst. Solids* **351** 3246
 Murugan G and Varma K B R 2001 *J. Non-Cryst. Solids* **279** 1
 Nakayama S, Watanabe T, Asahi T, Kiyono H, Aung Y L and Sakamoto M 2010 *Ceram. Int.* **36** 2323
 Ni Y, Lu C, Zhang Y, Zhang Q and Xu Z 2007 *J. Rare Earths* **25**(Suppl. 1) 94–98
 Nowick A S, Lim B S and Vaysleyb A V 1994 *J. Non-Cryst. Solids* **172** 1243
 Owen A 1963 *Prog. Ceram. Soc.* **77** 256
 Prasad S, Clark T M, Sefzik T H, Kwak H, Gan Z and Grandinetti P J 2006 *J. Non-Cryst. Solids* **352** 2834
 Rao G V, Yadagiri Reddy P and Veeraiah N 2002 *Mater. Letts.* **57** 403
 Reau J M and Polain M 1989 *Mater. Chem. Phys.* **23** 2189
 Rohling B, Happe A, Funke K and Ingram M D 1997 *Phys. Rev. Lett.* **78** 2160
 Saafan S A 2008 *Physica B: Condens. Matter* **403** 2049
 Sanghi S, Sindhu S, Agarwal A and Seth V P 2004 *Radiat. Eff. Solids* **159** 369

- Shaaban M H, Ali A A and El-Nimr L K 2006 *Mater. Chem. Phys.* **96** 423
- Shaw J L, Zwanziger U W and Zwanziger J W 2006 *Eur. J. Glass Sci. Technol.* **47** 513
- Shyam Prasad P, Raghavaiah B V and Veeraiah N 2004 *Solid State Commun.* **132** 235
- Sindhu S, Sanghi S, Agarwal A, Seth V P and Kishore N 2005 *Mater. Chem. Phys.* **90** 83
- Souza Filhoa A G, Mendes Filhoa J, Melo F E A, Custodiob M C C, Lebullengerb R and Hernandesb A C 2000 *J. Phys. Chem. Solids* **61** 1535
- Stehle C, Vira C, Hogan D, Feller S and Affatigato M 1998 *Phys. Chem. Glasses* **39** 836
- Szu S-P and Lin C-Y 2003 *Mater. Chem. Phys.* **82** 295
- Zheng Y and Clare A G 2005 *Phys. Chem. Glasses* **46** 467
- Zhong J and Bray P J 1989 *J. Non-Cryst. Solids* **111** 67