5.1 Introduction

Ni-Zn ferrites are one of the most versatile soft magnetic materials. Recently, the technological application of these materials has been studied extensively [1-4], primarily due to their applicability in many electronic devices owing to their high permeability at high frequency, remarkably high electrical resistivity, low eddy current loss and reasonable cost [5-9].

The spinel structure of these ferrites possesses the general formula of \((A)[B_2]O_4\), where A represents cations in tetrahedral sites and B represents cations in the octahedral positions in a cubic structure. However, the formula \((A_{1-i}B_i) [A_iB_{2-i}]O_4\) represents many possible intermediary distributions that denote considerable cation disorder, indicating that this structure requires special attention in terms of magnetic characterization [10]. The Ni-Zn ferrite is a well-known mixed inverse spinel [11] whose unit cell is represented by the formula

Results and discussion of \(\text{Ni}_{0.7}\text{Zn}_{0.3}\text{Cr}_x\text{Fe}_{2-x}\text{O}_4\) ferrite system.
(Zn\(_x\)Fe\(_{1-x}\))[Ni\(_{1-x}\)Fe\(_{1+x}\)]O\(_4\) [12,13], whereas the intrinsic magnetization (magnetic moment or total theoretical Bohr magneton of the lattice spinel) known results from the inverse and normal phase, i.e., the distribution of cations in the spinel lattice. Based on these facts, the intrinsic magnetization of Ni–Zn ferrite can be calculated with and without the addition of 0.1 mol of chromium ions. Based on Hund’s Rules [14], the magnetic moments of Fe\(^{3+}\), Cr\(^{3+}\), Ni\(^{2+}\) and Zn\(^{2+}\) ions are 5, 3, 2 and 0 µ\(_B\), respectively. The addition of impurities induces changes in the defect structure and texture of the crystal [15], significant modifications in the magnetic and electrical properties of these materials. Several researchers have studied the effects of the substitution of Cr in the spinel structure of ferrites [15-17]. However, we have not found any reports in the literature on Ni–Zn ferrite powders obtained by combustion reaction. Combustion synthesis, which is employed in the field of propellants and explosives, involves an exothermic and self-sustaining chemical reaction between the desired metal salts and a suitable organic fuel [18]. A key feature of the process is that the heat required to trigger the chemical reaction is provided by the reaction itself and not by an external source, such as urea [19]. Compared with other synthesis methods, the combustion reaction process offers the advantages of being fast and simple, without requiring subsequent intermediary calcination stages, and consumes less energy during synthesis [20]. Like various other
methods that have been proposed and applied in the preparation of ceramic powders, combustion synthesis allows for low-temperature synthesis, yielding products that are usually dry, crystalline, fine and chemically pure and homogeneous [21]. This chapter therefore presents a study of nanosized Ni-Zn ferrite doped with Cr$^{3+}$ produced by sol-gel auto-combustion synthesis and its morphologic, magnetic and electric characterization.

5.2 Experimental

5.2.1 Synthesis method

The powders were synthesized by the sol-gel auto-combustion method. A.R. grade citric acid ($\text{C}_6\text{H}_8\text{O}_7\cdot\text{H}_2\text{O}$), nickel nitrate ($\text{Ni(NO}_3\text{)}_2\cdot\text{6H}_2\text{O}$), zinc nitrate ($\text{Zn(NO}_3\text{)}_2\cdot\text{6H}_2\text{O}$), chromium nitrate ($\text{Cr(NO}_3\text{)}_3\cdot\text{9H}_2\text{O}$) and iron nitrate ($\text{Fe(NO}_3\text{)}_3\cdot\text{9H}_2\text{O}$) were dissolved in distilled water to obtain a mixed solution. The reaction procedure was carried out in an air atmosphere without the protection of inert gases. The molar ratio of metal nitrates to citric acid was 1:3. The metal nitrates were dissolved together in the minimum amount of double-distilled water needed to obtain a clear solution. An aqueous solution of citric acid was mixed with the metal-nitrate solution, and ammonia solution was slowly added to adjust the pH to 7. The mixed solution was placed on a hot plate with continuous stirring at 90 °C. During evaporation, the
solution formed a very viscous brown gel. When all of the water molecules were removed from the mixture, the viscous gel began to froth. After few a minutes, the gel ignited and burnt with glowing flints. The decomposition reaction continued until the entire citrate complex was consumed. The auto-combustion was completed within a minute, yielding brown-colored ashes referred to as the precursor. The as-prepared powders of all the samples were sintered at 600 °C for 4 h to obtain the final product. The powder was granulated using 2 % PVA as a binder and pressed into discs and toroids (internal and outer diameters of 2 and 1 cm, respectively, and a thickness of 2 cm), with an applied pressure of 5000 kg/cm$^2$ to obtain the toroid-shaped samples for permeability measurements.

5.2.2 Characterization techniques

The samples were powdered for X-ray investigation. One portion of the powder was examined using a Phillips X-ray diffractometer (Model 3710) with Cu-$K_\alpha$ radiation ($\lambda=1.5405$ Å). The scanning step was 2 °/min and scanning rate was 0.02°. A 1-D detector was used for the XRD measurement. The X-ray generator was operated at 40 kV and 30 mA. A specially processed Si powder sample was used as the instrumental standard. The (1 1 1) reflection of Si at 28.5° indicates that the instrumental broadening is very small (0.5 Å).
The morphology and structure of the powder samples were studied using a JEOL-JSM-5600 N scanning electron microscope (SEM). The infrared spectrum of each sample was recorded at room temperature in the range of 350 cm\(^{-1}\) to 800 cm\(^{-1}\) using a Perkin Elmer infrared spectrophotometer.

The magnetic measurements were performed at room temperature using a commercial PARC EG&G VSM 4500 vibrating sample magnetometer. The magnetic hysteresis loops were measured at room temperature with maximum applied magnetic fields of 0.8 T. The magnetic field sweep rate was 5 Oe/s for all measurements.

The DC electrical resistivity of each sample was measured using the two-probe technique in the temperature range 300-800 K on a disc-shaped pellet of 10 mm diameter and 3 mm thickness. The dielectric properties, i.e., the dielectric constant (\(\varepsilon'\)), dielectric loss (\(\varepsilon''\)) and dielectric-loss tangent (\(\tan\delta\)), were studied using LCR-Q meter (Hioki) at room temperature and in the frequency range of 100 Hz to 1 MHz. By measuring the capacitance (Cp), dielectric constant (\(\varepsilon'\)), dielectric loss (\(\varepsilon''\)) and dielectric-loss tangent (\(\tan\delta\)) were obtained.

The initial permeability (\(\mu_i\)) measurements on the toroid-shaped samples were conducted as a function of temperature and frequency. The initial permeability measurements of the toroid samples were taken using a HP-4284 A LCR precision meter.
5.3 Results and discussion

5.3.1 Structural Analysis

The X-ray diffraction (XRD) patterns of the Ni$_{0.7}$Zn$_{0.3}$Cr$_x$Fe$_{2-x}$O$_4$ spinel ferrite system with $x = 0.0$-0.5, in the steps of 0.1 are shown in Fig. 5.1. All of the Bragg peaks of the XRD patterns are broad and do not contain any extra peak other than cubic spinel phase. The X-ray diffraction analysis of these samples reveals the formation of single phase cubic spinel structure. Table 5.1 gives the Miller indices and inter-planner spacing of the samples.

The lattice constant (a) of all the samples was determined by using the equation discussed elsewhere [22]. The lattice-constant values were obtained for every sample using XRD data with an accuracy of ±0.002 Å and are listed in Table 5.2. The lattice constant initially increases and then begins to decrease, as shown in Fig. 5.2. The initial increase of the lattice constant from $x = 0.0$ to 0.2 may be due to the fact that the substitution of Cr up to $x = 0.2$ does not affect the lattice. The decrease in the lattice constant above $x > 0.2$ is related to the difference in ionic radii of Fe$^{3+}$ and Cr$^{3+}$. In the present ferrite system Fe$^{3+}$ ions (0.67Å) ions are replaced by the relatively small Cr$^{3+}$ ions (0.64Å). A similar nature was reported for other Cr-substituted spinel ferrite [23].

The X-ray density ‘$d_x$’ was calculated using the expression 4.2. The values of the X-ray density are listed in Table 5.2 and the variation is
shown in Fig. 5.3 It can be seen from Table 5.2 that the X-ray density increases with increasing in Cr content. The variation of the X-ray density with x is opposite to that of the lattice constant with x because the X-ray density is inversely proportional to the lattice constant $a^3$.

The average crystallite size ($t$) was determined using the line broadening of most intense (311) diffraction peak using the Debye Scherrer formula (eq. 4.3) [22]. The values of crystallite size ($t$) are given in Table 5.3. The crystallite size is decreases from 35 nm to 20 nm with increasing Cr content x. The variation of the crystallite size with Cr content is shown in Fig. 5.4.

The specific surface area ($S$) was calculated from the diameter of the particle in nanometers and the measured density in gm/cm$^3$ using the relation 4.4 [24]. The values of surface area are given in Table 5.2. The specific surface area decreases with increasing Cr$^{3+}$ content. The decrease in $S$ is due to the decrease in crystallite size.

The bulk density $d_B$ of the specimens was determined by Archimedes’ method (eq. 4.5). The values of the bulk density are shown in Table 5.2, and the variation of bulk density is shown in Fig. 5.5. The bulk density was found to decrease with increasing Cr content x. In the present series, both the molecular weight of the Ni$_{0.5}$Zn$_{0.5}$Fe$_2$O$_4$ spinel ferrite and the volume of the unit cell decrease with increasing Cr substitution, but the rate of the decrease of molecular weight is more
than that of volume. Therefore, the bulk density decreases with Cr substitution in the present case. This led to increase in porosity. The porosity ‘P’ of the ferrite nano particles was then determined using the relation 4.6 [24]. The values of percentage porosity are given in Table 5.2 and its variation with Cr content x is shown in Fig. 5.5. It is clear from Fig. 5.5 that the sample density decreases and the porosity increases with increasing Cr content. The increase in porosity may be due the decrease in particle size, which increases the grain boundaries of the particle and accordingly the porosity. The high porosity demonstrate the porous structure of the prepared Ni-Zn-Cr-Fe spinel ferrite samples.

The hopping length (L_A and L_B) between magnetic ions (the distance between the ions) in the tetrahedral A-site and octahedral B-site can be calculated using the eqs. 4.7 and 4.8 [25]. The values of hopping lengths are given in Table 5.3. The variation of hopping length (L_A and L_B) as a function of Cr content ‘x’ is shown in Fig. 5.6, which shows that hopping length initially increases up to x = 0.2 and then decreases with further increases in Cr content. The results are explained by the variation in lattice constant with increasing Cr content.

The SEM micrographs are shown in Fig. 5.7 (a and b). It is evident from these figures that the microstructure of these samples is affected by Cr substitution. The SEM images were used to study the surface morphology of the prepared samples and to calculate the grain size.
5.3.3 Cation distribution

The cation distribution in spinel ferrite can be obtained from an analysis of the X-ray diffraction pattern. In the present work, the Bertaut method [26] is used to determine the cation distribution. The best information on cation distribution is achieved by comparing the experimental and calculated intensity ratios for reflections whose intensities (i) are nearly independent of the oxygen parameter, (ii) vary with the cation distribution in opposite ways and (iii) do not significantly differ.

The cation distribution for each concentration and the site preferences of cations distributed among the tetrahedral A-site and octahedral B-sites are listed in Table 5.4. In this table, the fraction of $\text{Fe}^{3+}$ ions in either sites is listed. The results demonstrate that $\text{Ni}^{2+}$ ions occupy B sites whereas $\text{Zn}^{2+}$ ions occupy tetrahedral A-site. $\text{Cr}^{3+}$ preferentially replaces $\text{Fe}^{3+}$ from octahedral sites because of favorable crystal field effects ($\text{Cr}^{3+}6/5\Delta_\omega$, $\text{Cr}^{3+}0\Delta_\omega$) [27]. The data in Table 4 show that $\text{Cr}^{3+}$ ions predominately occupy the octahedral sites, which is consistent with the preference for large octahedral site energy. With increasing $\text{Cr}^{3+}$ content, the fraction $\text{Cr}^{3+}$ ions in octahedral sites increases, whereas the fraction of $\text{Fe}^{3+}$ ions in octahedral sites decreases linearly.

The mean ionic radius of the A and B sites ($r_A$ and $r_B$) can be calculated using the relations discussed elsewhere [28, 29]. The values are
given in Table 5.5 and their variation is shown in Fig. 5.8. It is shown that \( r_A \) remains constant and \( r_B \) decreases with increasing Cr content. The decrease in \( r_B \) is due to the increasingly high occupation of the B site by the smaller ionic radii of Cr\(^{3+} \) (0.64 Å) replacing Fe\(^{3+} \) (0.67 Å).

The theoretical values of the lattice parameter can be calculated with the help of equation 4.13 [30]. The values of theoretical lattice parameter \( a_{th} \) obtained by using relation (5.15) are shown in Fig. 5.9. The variation of theoretical values is similar to that observed for experimentally determined lattice parameter.

Using the values of ‘a’, the radius of the oxygen ion \( (R_O = 1.32 \text{ Å}) \) and \( r_A \) in the following expression, the oxygen positional parameter \( (u) \) can be calculated using an equation 4.14 [31]. Fig. 5.9 shows the increasing value of the oxygen positional parameter from 0.3888 - 0.3893 Å. In most oxide spinels the oxygen ions are larger than the metallic ions. In spinel-like structures, the oxygen positional parameter has a value near 0.375 Å, for which the arrangement of \( \text{O}^{2-} \) ions is equals, exactly a cubic closed packing, but in an actual spinel lattice, this ideal pattern is slightly deformed. \( u \) has a value of 0.375 Å when the origin is chosen on the tetrahedral sites. However, the structure is a centric and the structure factor calculation is less direct [32]. Our value of \( u \) is larger than the ideal value \((u = 0.375 \text{ Å})\), which value may probably be due to many reasons, including the history of the samples and, experimental or measurement.
errors, e.g. the precision of the observed X-ray intensity and the theoretical data used for the scattering model of the system. In most spinels $u > 0.375$ is obtained because of a small displacement of the anions due to the expansion of the tetrahedral interstices. In the present work, $u > 0.375$ may be due to an anion displacement from the ideal situation that it to form expanded tetrahedral interstices. The lattice disturbance is confirmed by the data for the lattice constant and the oxygen positional parameter.

Using the experimental values of the lattice parameter and the oxygen positional parameter and substituting them into the equations 4.15a-4.15e [33], the tetrahedral and octahedral bond length $d_{Ax}$ and $d_{Bx}$, tetrahedral edge, shared and unshared octahedral edge ($d_{AXE}$, $d_{BXE}$ and $d_{BXEU}$) were calculated. The values are presented in Table 5.6, which indicates that the tetrahedral and octahedral bond lengths decreases as Cr$^{3+}$ ion substitution increases. The tetrahedral edge and shared and unshared octahedral edges decrease as Cr$^{3+}$ content increases. This trend could be related to the smaller radius of Cr$^{3+}$ ions compared to Fe$^{3+}$ ions.

5.3.3 Infrared spectroscopy

The infrared spectra can give some additional information on valance state and the different vibrational modes of the crystal lattice. The IR spectra of the series $\text{Ni}_{0.7}\text{Zn}_{0.3}\text{Cr}_x\text{Fe}_{2-x}\text{O}_4$ are shown in Fig 5.10. The band position obtained from these IR spectra is given in Table 5.7. The IR
spectra reveal the presence of two absorption bands of frequency \( \nu_1 \) and \( \nu_2 \). The higher frequency band \( \nu_1 \) is at approximately 600 cm\(^{-1}\) and the lower frequency band \( \nu_2 \) is at approximately 400 cm\(^{-1}\). The difference in band position is expected because of the difference in the Fe\(^{3+}\)-O\(^2-\) inter-ionic distance for the octahedral and tetrahedral coordinations. The vibrational frequency depends on the cation mass, cation-to-oxygen distance and bonding force. Similar IR spectra are also observed in other spinel ferrite systems [23].

The force constant is the second derivative of the potential energy with respect to the site radius with the other independent parameters kept constant. The force constant for tetrahedral site (\( K_t \)) and octahedral site (\( K_O \)) were calculated using the method suggested by Waldron [34]. The force constants \( K_t \) and \( K_O \) for the respective sites are given by equation 4.16 and 4.17. The bond lengths \( R_A \) and \( R_B \) have been calculated using the formula given by Gorter [35]. The molecular weights of the tetrahedral \( M_t \) and octahedral \( M_O \) sites have been calculated using the cation distribution data in Table 5.4. The values of \( R_A \), \( R_B \) and the force constants \( K_t \) and \( K_O \) are listed in Table 5.7.

The differences in band position and intensity with increasing Cr content may be explained as follows. The decrease in the Fe\(^{3+}\)-O\(^2-\) intermolecular distance increases the metal-oxygen vibrational energies, which arises from the decrease in the number of Fe\(^{3+}\)-O\(^2-\) complexes
caused by the increase of the number of Cr$^{3+}$-$\text{O}_2^-$ complexes [36] and the formation of Me$^{3+}$-$\text{O}_2^-$ at A and B sites (Me = Ni$^{2+}$)

### 5.3.4 Magnetization

The introduction of Cr$^{3+}$ ions into Ni-Zn ferrite greatly affects its magnetic properties. Fig. 5.11 shows the plots of the hysteresis loops for Ni$_{0.5}$Zn$_{0.5}$Fe$_{2-x}$Cr$_x$O$_4$ specimens. This figure indicates that Ni-Zn ferrite is a soft magnetic material with minimal hysteresis. The magnetic moment in ferrite is mainly due to the uncompensated electron spin of the individual ions and the spin alignments in the two sublattices, which are arranged in an antiparallel fashion. In a spinel ferrite, each ion at the A site has 12 B-site ions as nearest neighbors. According to Neel’s molecular-field model [37], the A-B super exchange interaction predominate the intrasublattice A-A and B-B interactions. Therefore, the net magnetic moment is given by the sum of the magnetic moments of the A and B sublattices.

The magnetic moment per formula unit ($n_B$) was calculated from Neel’s sub-two-lattice model using the relation 4.19. For chromium-substituted Ni-Zn ferrite, the Fe$^{3+}$ ions are replaced by Cr$^{3+}$ ions, leading to a decrease in the B-site sublattice magnetization. Therefore, the magnetization of the B sublattices decreases, which leads to a decrease in the net magnetization (Fig. 5.13). The decrease in magneton number is explained by the A-B interaction. In the present case, Cr$^{3+}$ ions of low magnetic moment values (3$\mu_B$) replace Fe$^{3+}$ ions of high magnetic
moment (5\(\mu_B\)). According to the cation distribution data from Table 5.4, Cr\(^{3+}\) ions occupy octahedral B-sites. This placement leads to a decrease of the magnetic moment of the B-site, and thus the magneton number \(n_B\) decreases with Cr concentration.

The magnetic moments calculated from Neel’s molecular field model are in agreement in the experiment results, which indicate that the saturation magnetization decreases monotonically with increasing Cr-substitution content (Fig. 5.12).

The observed magnetic moment per formula unit in the Bohr magneton (\(\mu_B\)) was calculated using a relation 4.18 [38]. The calculated and observed values of the magneton number are in close agreement, suggesting that the structure is collinear (Fig. 5.12). It has been established that the tetrahedral and octahedral sublattices of ferrite may be subdivided such that the resultant vector of the magnetic moments of the sublattices are aligned in such a direction that will influence the effective magnetization. Thus, the decrease in magnetization may be explained by the formation of a collinear spin arrangement due to the substitution of Cr ions. The observed and calculated magneton numbers along with the magnetization are given in Table 5.8.

The coercivity (Hc) is an independent parameter, which can be altered by heat treatment or deformation and thus is independent of saturation magnetization. It is clearly observed from Fig. 5.13 and Table
5.8 that the coercivity increases as the Cr$^{3+}$ content increases, similarly to porosity. Porosity affects the magnetization process because the pores work as a generator of a demagnetizing field. As the porosity increases, a higher field is needed to push the domain wall, increasing $H_C$. The saturation magnetization is related to $H_C$ through Brown’s relation 4.20 [39].

According to this relation, $H_C$ is inversely proportional to $M_s$, which is consistent with our experimental results.

5.3.5 Permeability

The initial permeability measurements of the toroid samples were taken using a HP-4284 A LCR precision meter from room temperature to the Curie temperature at 1 kHz from the low field inductance measurements of the coils with the toroidal cores using the relation 4.21 [40].

**Thermal variation of permeability**

The thermal and frequency variations of the initial permeability ($\mu_i$) provide valuable information about the domain nature [41], Curie temperature [42] and factors contributing to the permeability [43]. Many researchers [43-45] have explored the behavior of $\mu_i$ as a function of temperature. Parameters such as Ms, D and $K_1$ are responsible for the diversity of the thermal spectra of $\mu_i$ for the ferrite sample. In present
study, the variation of $\mu_i$ as a function of temperature in the range from room temperature to the Curie temperature are shown in Fig. 5.14.

For the compositions with $x = 0.0, 0.1, 0.2$ and $0.3$

- the initial permeability increases slowly with temperature and exhibits a peak near the Curie temperature, and
- near $T_c$, $\mu_i$ falls rapidly and it reaches zero at $T_c$. The sharp fall suggests the single-phase formation of the ferrite material.
- The anisotropy constant and saturation magnetization usually decrease with the increasing temperature due to thermal agitation, which disturbs the alignment of the magnetic moments [48]. However, the decrease of $K_i$ with temperature is much faster than that of $M_s$ [49]. When $K_i$ goes through zero, $\mu_i$ attains a maximum value and then drops to zero above $T_c$.

For compositions with $x = 0.4$ and $0.5$,

- there is no peaking behavior in the $\mu_i$-$T$ variation,
- the permeability does not change with temperature up to a certain temperature,
- the initial permeability drops with temperature near $T_c$ and becomes zero at $T_c$, and
- all of the compositions are single phase because a double $T_c$ or a gradual variation of $\mu_i$ with $T$ was not observed.
The invariance of the initial permeability of these compositions up to a certain temperature is explained by considering the compensating effects of \( M_s \) and \( K_1 \) with temperature. Kakatkar et al. [50] have reported \( \mu_i - T \) curves for Ni–Zn ferrites and observed a very small decrease in \( \mu_i \) over the temperature range of their experiment. They attributed this behavior to the compensating effects of \( M_s \) and \( K_1 \) with temperature; however, the shape of the curves (except the magnitude of \( \mu_i \)) was not affected by the addition of Cr\(^{3+} \). This behavior may be attributed to fact that the rate of change of \( M_s \) and that of the anisotropy field with temperature is the same. This type of peak may be below the room-temperature region, and some researchers [51,52] have observed such a peak.

Thus, for these compositions, it can be concluded that the peaking behavior is absent due to the values of the anisotropy constant \( K_1 \), which do not change sign within the temperature range of the experiments.

**Compositional variation in permeability**

The compositional variation of the initial permeability (\( \mu_i \)) recorded at 298 K is shown in Fig. 5.15. The initial permeability (\( \mu_i \)) decreases with increasing Cr substitution. The observed variations of the initial permeability can be explained in terms of the Globus model [53]. According to this model, the initial permeability is dependent on the grain size, saturation magnetization and magnetocrystalline anisotropy.
The relationship between grain size and permeability is linear only if the grain growth is normal, i.e., if all grains grow at the same time and at the same rate. This limitation leads to a rather narrow range of final grain sizes. However, when some grains grow very rapidly, they trap pores, which can limit the permeability by pinning down the domain walls. When this exaggerated grain growth occurs, the porosity increases [54,55]. The initial permeability and saturation magnetization are two interdependent terms. The saturation magnetization was observed to decrease with increasing chromium content [56], and thus the observed variation can be correlated.

The Curie temperature (Tc) was obtained using the permeability curves (µ-T). The variation of the Curie temperature with Cr$^{3+}$ content is shown in Fig. 5.16. The value of Curie temperature is tabulated in Table 5.9. The Curie temperature decreases linearly with increasing Cr$^{3+}$ content. This effect can be explained by the number of magnetic ions present in the two sublattices and their mutual interactions. In Cr$^{3+}$ substituted Ni-Zn ferrite, the Fe$^{3+}$ ions are replaced by Cr$^{3+}$ ions at the B site. Because the magnetic moment of the Cr$^{3+}$ ions is 3µ$_B$, whereas the magnetic moment of the Fe$^{3+}$ ions is 5µ$_B$, an increase in the Cr$^{3+}$ concentration results in a reduction of the density of the magnetic ions. This decrease in turn results in a reduction of the net magnetic moment at the B sublattice, weakening the A-B exchange interaction of the type
Fe$^{3+}$(A)-O$^{2-}$Fe$^{3+}$(B). Because the Curie temperature is determined by the overall strength of the AB exchange interaction, the weakening of the Fe$^{3+}$(A)-O$^{2-}$Fe$^{3+}$(B) interaction results in a decrease of the Curie temperature with successive increases in the Cr$^{3+}$ concentration of the Ni-Zn ferrites.

**Frequency variation of permeability**

The permeability of ferrite is mainly caused by spin rotation and domain wall displacement at microwave frequencies. When excited by an applied alternating magnetic field, the magnetization vector processes around the anisotropy field. According to previous works [57,58], in the presence of an external RF field, the real part $\mu'_i$ can be expressed as given in an equation discussed elsewhere [59]. As per this equation, when $\omega_p > \omega$, $\mu'_i$ increases gradually as $\omega$ increases. When $\omega_p = \omega$, a maximum in $\mu'_i$ is obtained, and when $\omega_p > \omega$, the magnetization vector and the RF magnetic field become out of phase and $\mu'_i$ suddenly falls and becomes negative. This phenomenon is known as resonance. Resonance occurs when the frequency of the applied field coincides with the natural precessional frequency [60]. In Fig. 5.17, the real part of the permeability decreases gradually with increasing frequency. This result is in accordance with Snoek’s law, which can be described by $f = (\gamma M_s/3\pi(\mu_0))$[Hz], where $f$ is the resonant frequency, $M_s$ is the saturation magnetization, and $\gamma$ is the gyromagnetic ratio. The above-mentioned
formula indicates that, because the resonant frequency is low, $\mu_i$ will be high. The value of $\mu_i$ depends significantly on the saturation magnetization. According to above-mentioned magnetization result, the net magnetization decreases gradually with increasing $\text{Cr}^{3+}$ content in chromium-substituted Ni-Zn ferrite. Therefore, the real part of the permeability of Ni-Zn ferrite is higher than that of chromium-substituted Ni-Zn ferrite.

5.3.6 DC resistivity

Fig. 5.18 shows the variation of DC resistivity with the Cr concentration for all of the investigated samples. The DC resistivity shows a linear decrease with temperature. This variation is explained by the location of the cations in the spinel ferrite. It is seen that the variation in D.C. resistivity is almost linear up to a certain temperature, at which a break occurs. This break corresponds to the Curie temperature, indicating a change of magnetic ordering from ferrimagnetism to paramagnetism. An increase in resistivity is observed with an increase in Cr concentration (Fig. 5.19). In ferrites, the resistivity $\rho$ at an absolute temperature $T$ is given by the relation

$$\rho = \rho_0 e^{-\frac{E_g}{kT}} \quad 5.1$$

where $E_g$ is the activation energy and $k$ is Boltzmann constant.
The values of activation energy are listed in Table 5.9. In the table $E_f$ is the activation energy corresponding to the ferrimagnetic region and $E_p$ is that corresponding to the paramagnetic region. It is apparent that the activation energies in the paramagnetic region are higher than those in the ferrimagnetic region. This finding can be attributed to the disordered states of the paramagnetic region and the ordered states of the ferrimagnetic region. This result also suggests that the conduction process is affected by the change in magnetic ordering. In polycrystalline ferrite, cations are surrounded by close-packed oxygen anions and can well be treated as isolated from each other to a first approximation. The electrical conduction in ferrites can be explained by the Verwey model of electron hopping, which involves the exchange of electrons between ions of the same element present in different valence states and distributed randomly over crystallographically equivalent lattice sites. In the present system, the conduction is attributed to the exchange of 3d electrons between Fe$^{2+}$ and Fe$^{3+}$ ions in the octahedral B sites. A small amount of Fe$^{3+}$ is converted into Fe$^{2+}$ during sintering at high temperature. The observed increase in resistivity can be understood by considering the hopping mechanism

$$\text{Fe}^{2+} \leftrightarrow \text{Fe}^{3+}.$$ 

The increase in Cr$^{3+}$ ions at the B sites replaces Fe$^{3+}$ ions, decreasing the number of ferrous ions formed. Although the Cr$^{3+}$ ions do not participate
in the conduction mechanism, they limit the degree of $\text{Fe}^{2+} \leftrightarrow \text{Fe}^{3+}$
transfer, thereby obstructing electron hopping [61] and increasing the resistivity.

5.3.7 Dielectric properties

Frequency Dependence of dielectric properties:

Figures 5.20 and 5.21 show the variation of the dielectric constant and dielectric loss as a function of frequency. The frequency dependence of dielectric constant for $\text{Ni}_{0.7}\text{Zn}_{0.3}\text{Cr}_x\text{Fe}_{2-x}\text{O}_4$ is in very good agreement with spinel ferrite [62, 63], for which the dielectric constant decreases continuously with increasing frequency. It can be seen that both the dielectric constant and the dielectric loss decrease very rapidly with increasing frequency in the lower frequency region. As frequency increases, the dielectric constant and dielectric loss remain almost constant. All of the samples displayed dispersion due to Maxwell-Wagner interfacial polarization [64, 65], in agreement with Koop’s phenomenological theory [66]. The dielectric behavior may be explained qualitatively by supposing that the mechanism of the polarization process in ferrite is similar to that of the conduction process.

Based on the electronic exchange $\text{Fe}^{3+} \leftrightarrow \text{Fe}^{2+} + e^-$, one obtains the local displacement of the electron in the direction of the applied electric field. It is well known that the effect of the polarization is to reduce the
field inside the medium. The decrease in the polarization of the dielectric constant with increasing frequency is due to fact that at a certain electrical-field frequency, the electron exchange between ferrous and ferric ions cannot keep pace with the alternating field. Therefore, the dielectric constant of the material may decrease substantially as the frequency increases.

The variation of the dielectric-loss tangent (\(\tan\delta\)) with the logarithm of frequency (\(\log f\)) for various compositions is shown in Fig. 5.22. All of the compositions show normal dielectric behavior and a exponential decrease in the dielectric-loss tangent with increasing frequency. A maximum in the dielectric-loss tangent versus frequency appears when the frequency of the hopping charge carriers coincides with the frequency of the applied alternating field. The maxima in the dielectric-loss tangent occur at relatively low frequencies and decreases with increasing frequency. The broad peak disappears at higher frequencies. A broad peak of the dielectric-loss tangent indicates the existence of a distribution of relaxation times rather than a single relaxation time [67]. The condition for observing maxima in dielectric-loss tangent of a material is \(\omega \tau = 1\), where \(\omega = 2\pi f_{\text{max}}\) and \(\tau\) is the relaxation time. The relaxation time is related to the jumping probability per unit time \(P\) by the relation

\[
\tau = \frac{1}{2P} \quad \text{or} \quad f_{\text{max}} \propto P .
\]
Temperature dependence of dielectric properties:

The real part i.e. dielectric constant (ε’), the imaginary part i.e. the dielectric loss (ε“) and the dielectric loss tangent (tanδ) of Ni$_{0.7}$Zn$_{0.3}$Fe$_{2-x}$Cr$_x$O$_4$ were computed according to Smit and Wijn [68] as a function of temperature. The measurements were carried out from 403-773K, Figs. 5.23 and 5.24 show the variation of the dielectric constant and dielectric loss with temperature at a fixed frequency of 1 KHz. It is clear from these figures that the dielectric constant and dielectric loss increase with increasing temperature for all samples. This temperature dependence of the dielectric constant and dielectric loss for Ni$_{0.7}$Zn$_{0.3}$Fe$_{2-x}$Cr$_x$O$_4$ is in very good agreement with spinel ferrite [69, 70], for which the dielectric constant increases with increasing temperature.

The variation of the dielectric-loss tangent with temperature is given in Fig. 5.25, in which it is observed that the dielectric-loss tangent increases with increasing temperature. The increase in the dielectric-loss tangent is fast up to 700 K, but the dielectric-loss tangent decreases beyond this temperature. According to Rabkin and Novikova [71], the process of dielectric polarization in ferrite occurs through a mechanism similar to the conduction process. From the electronic exchange Fe$^{3+}$↔ Fe$^{2+}$ and Ni$^{2+}$↔ Ni$^{3+}$, one obtains the local displacement of the electron in the direction of the applied electrical field. This displacement determines the polarization of both types of charge carrier, n and p, which
contributes to the polarization and depends on temperature. The temperature dependence of the dielectric constant, dielectric loss and dielectric-loss tangent can be explained by the polarization effect. As temperature increases, the electrical conductivity increases due to the thermal activity and mobility of the electrical charge carriers according to the hopping mechanism. Thus, the dielectric polarization increases, increasing the dielectric constant and dielectric loss and decreasing the dielectric-loss tangent.

5.4 Conclusions

The substitution of Cr$^{3+}$ has induced significant changes in the structural, electrical and dielectric properties of Ni-Zn ferrite. Experimental results revealed that the lattice constant and cell volume decrease with increasing Cr$^{3+}$ content in Ni-Zn ferrite. The crystallite size is decreases from 35 nm to 20 nm with increasing the Cr$^{3+}$ content x. The cation distribution suggests that Cr$^{3+}$ and Zn$^{2+}$ both have a strong preference towards the octahedral B site and that Ni$^{2+}$ also occupies the B site, whereas Fe$^{3+}$ occupies both the A and B sites. The calculated and observed values of the magneton number are in good agreement, suggesting that the structure is collinear. The IR spectra confirm the formation of the spinel structure and give the distribution of ions between A- and B- sites.
The saturation magnetization decreases linearly from 58.31 to 42.9 emu/g with increasing Cr\(^{3+}\) content. This behavior is ascribed to the fact that the increasing concentration of nonmagnetic ions weakens the inter-site exchange interaction, decreasing the value of saturation magnetization. The crystallite size decreases from 35 nm to 20 nm with increasing Cr\(^{3+}\) content. The initial permeability (\(\mu_i\)) decreases with increasing Cr\(^{3+}\) substitution. The decrease in initial permeability (\(\mu_i\)) is attributed to decrease in magnetization on addition of Cr\(^{3+}\). The real part of the permeability decreases gradually with increasing frequency in accordance with Snoek’s law. Weakening of the Fe\(^{3+}\)(A)-O\(^{2-}\)Fe\(^{3+}\)(B) interaction results in a decrease of the Curie temperature with successive increases in the Cr\(^{3+}\) ions.

The Cr\(^{3+}\) ions do not participate in the conduction mechanism but limit the degree of Fe\(^{2+}\) \(\leftrightarrow\) Fe\(^{3+}\) transfer, thereby obstructing electron hopping and resulting in an increase in resistivity. Although the resistivity increases with increasing Cr\(^{3+}\) content, the dielectric constant decreases. The broad peak of the dielectric-loss tangent indicates the relaxation-time distribution rather than a single relaxation time. Dielectric constant and dielectric loss tangent increases as temperature increases.
References


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Table 5.1
Inter planer spacing (d) of \( \text{Ni}_{0.7}\text{Zn}_{0.3}\text{Fe}_{2-x}\text{Cr}_x\text{O}_4 \)

<table>
<thead>
<tr>
<th>Plane</th>
<th>x = 0.0</th>
<th>x = 0.1</th>
<th>x = 0.2</th>
<th>x = 0.3</th>
<th>x = 0.4</th>
<th>x = 0.5</th>
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</thead>
<tbody>
<tr>
<td>220</td>
<td>2.9563</td>
<td>2.9582</td>
<td>2.9563</td>
<td>2.9621</td>
<td>2.9525</td>
<td>2.9430</td>
</tr>
<tr>
<td>311</td>
<td>2.5585</td>
<td>2.5223</td>
<td>2.5182</td>
<td>2.5236</td>
<td>2.5154</td>
<td>2.5578</td>
</tr>
<tr>
<td>222</td>
<td>2.4111</td>
<td>2.4123</td>
<td>2.4073</td>
<td>2.4136</td>
<td>2.4198</td>
<td>2.3987</td>
</tr>
<tr>
<td>400</td>
<td>2.0906</td>
<td>2.0915</td>
<td>2.0897</td>
<td>2.0906</td>
<td>2.0851</td>
<td>2.0814</td>
</tr>
<tr>
<td>422</td>
<td>1.7080</td>
<td>1.7080</td>
<td>1.7062</td>
<td>1.7091</td>
<td>1.7009</td>
<td>1.6986</td>
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<tr>
<td>511</td>
<td>1.6103</td>
<td>1.6103</td>
<td>1.6088</td>
<td>1.6098</td>
<td>1.6037</td>
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<tr>
<td>440</td>
<td>1.4786</td>
<td>1.4791</td>
<td>1.4778</td>
<td>1.4791</td>
<td>1.4765</td>
<td>1.4748</td>
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</table>
Table 5.2

Lattice constant (a), X-ray density (d_x), crystallite size (t), specific surface area (S), bulk density (d_B) and porosity (P) of Ni_{0.7}Zn_{0.3}Fe_{2-x}Cr_xO_4

<table>
<thead>
<tr>
<th>Comp. x</th>
<th>a (Å)</th>
<th>d_x (Å)</th>
<th>t (nm)</th>
<th>S (m²/gm)</th>
<th>d_B (gm/cm³)</th>
<th>P (%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.0</td>
<td>8.3631</td>
<td>5.369</td>
<td>35</td>
<td>35.54</td>
<td>4.823</td>
<td>10.16</td>
</tr>
<tr>
<td>0.1</td>
<td>8.3677</td>
<td>5.351</td>
<td>32</td>
<td>39.05</td>
<td>4.802</td>
<td>10.26</td>
</tr>
<tr>
<td>0.2</td>
<td>8.3724</td>
<td>5.333</td>
<td>29</td>
<td>43.55</td>
<td>4.751</td>
<td>10.92</td>
</tr>
<tr>
<td>0.3</td>
<td>8.3607</td>
<td>5.347</td>
<td>25</td>
<td>51.94</td>
<td>4.621</td>
<td>13.58</td>
</tr>
<tr>
<td>0.4</td>
<td>8.3482</td>
<td>5.362</td>
<td>22</td>
<td>59.89</td>
<td>4.554</td>
<td>15.07</td>
</tr>
<tr>
<td>0.5</td>
<td>8.3335</td>
<td>5.382</td>
<td>20</td>
<td>67.67</td>
<td>4.433</td>
<td>17.63</td>
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</tbody>
</table>
Table 5.3
Hopping length $L_A$ and $L_B$ of $\text{Ni}_{0.7}\text{Zn}_{0.3}\text{Fe}_{2-x}\text{Cr}_x\text{O}_4$

<table>
<thead>
<tr>
<th>Comp.</th>
<th>Hopping length</th>
<th></th>
<th></th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>$L_A$ (Å)</td>
<td>$L_B$ (Å)</td>
<td></td>
</tr>
<tr>
<td>0.0</td>
<td>3.621</td>
<td>2.957</td>
<td></td>
</tr>
<tr>
<td>0.1</td>
<td>3.623</td>
<td>2.958</td>
<td></td>
</tr>
<tr>
<td>0.2</td>
<td>3.625</td>
<td>2.960</td>
<td></td>
</tr>
<tr>
<td>0.3</td>
<td>3.620</td>
<td>2.956</td>
<td></td>
</tr>
<tr>
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<td>3.615</td>
<td>2.952</td>
<td></td>
</tr>
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<td>3.609</td>
<td>2.946</td>
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</table>
Table 5.4

Cation distribution and intensity ratios of $\text{Ni}_{0.7}\text{Zn}_{0.3}\text{Fe}_{2-x}\text{Cr}_x\text{O}_4$

<table>
<thead>
<tr>
<th>Comp. x</th>
<th>Cation distribution</th>
<th>$I_{(220)}/I_{(400)}$</th>
<th>$I_{(422)}/I_{(440)}$</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.0</td>
<td>$(\text{Zn}<em>{0.3}\text{Fe}</em>{0.7})^A [\text{Ni}<em>{0.7}\text{Fe}</em>{1.3}]^B \text{O}_4$</td>
<td>2.551</td>
<td>2.973</td>
</tr>
<tr>
<td>0.1</td>
<td>$(\text{Zn}<em>{0.3}\text{Fe}</em>{0.7})^A [\text{Cr}<em>{0.1}\text{Ni}</em>{0.7}\text{Fe}_{1.2}]^B \text{O}_4$</td>
<td>1.809</td>
<td>1.825</td>
</tr>
<tr>
<td>0.2</td>
<td>$(\text{Zn}<em>{0.3}\text{Fe}</em>{0.7})^A [\text{Cr}<em>{0.2}\text{Ni}</em>{0.7}\text{Fe}_{1.1}]^B \text{O}_4$</td>
<td>1.569</td>
<td>1.582</td>
</tr>
<tr>
<td>0.3</td>
<td>$(\text{Zn}<em>{0.3}\text{Fe}</em>{0.7})^A [\text{Cr}<em>{0.3}\text{Ni}</em>{0.7}\text{Fe}_{1.0}]^B \text{O}_4$</td>
<td>2.680</td>
<td>2.571</td>
</tr>
<tr>
<td>0.4</td>
<td>$(\text{Zn}<em>{0.3}\text{Fe}</em>{0.7})^A [\text{Cr}<em>{0.4}\text{Ni}</em>{0.7}\text{Fe}_{0.9}]^B \text{O}_4$</td>
<td>2.136</td>
<td>2.136</td>
</tr>
<tr>
<td>0.5</td>
<td>$(\text{Zn}<em>{0.3}\text{Fe}</em>{0.7})^A [\text{Cr}<em>{0.5}\text{Ni}</em>{0.7}\text{Fe}_{0.8}]^B \text{O}_4$</td>
<td>1.675</td>
<td>1.504</td>
</tr>
</tbody>
</table>
Table 5.5

Ionic radii of tetrahedral A-site ($r_A$), octahedral B-site ($r_B$), theoretically lattice constant ($a_{th}$) and oxygen positional parameter ($u$) of Ni$_{0.7}$Zn$_{0.3}$Fe$_{2-x}$Cr$_x$O$_4$

<table>
<thead>
<tr>
<th>Comp. ‘x’</th>
<th>$r_A$ (Å)</th>
<th>$r_B$ (Å)</th>
<th>$a_{th}$ (Å)</th>
<th>$u$ (Å)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.0</td>
<td>0.691</td>
<td>0.677</td>
<td>8.422</td>
<td>0.3888</td>
</tr>
<tr>
<td>0.1</td>
<td>0.691</td>
<td>0.6755</td>
<td>8.418</td>
<td>0.3888</td>
</tr>
<tr>
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<td>0.691</td>
<td>0.674</td>
<td>8.414</td>
<td>0.3887</td>
</tr>
<tr>
<td>0.3</td>
<td>0.691</td>
<td>0.6725</td>
<td>8.410</td>
<td>0.3889</td>
</tr>
<tr>
<td>0.4</td>
<td>0.691</td>
<td>0.671</td>
<td>8.406</td>
<td>0.3891</td>
</tr>
<tr>
<td>0.5</td>
<td>0.691</td>
<td>0.6695</td>
<td>8.402</td>
<td>0.3893</td>
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</table>
Table 5.6

Tetrahedral bond ($d_{AX}$), octahedral bond ($d_{BX}$), tetra edge ($d_{AXE}$) and octahedral edge ($d_{BXE}$) (shared and unshared) of Ni$_{0.7}$Zn$_{0.3}$Fe$_{2-x}$Cr$_x$O$_4$

<table>
<thead>
<tr>
<th>Comp. ‘x’</th>
<th>$d_{AX}$ (Å)</th>
<th>$d_{BX}$ (Å)</th>
<th>Tetra edge (Å)</th>
<th>Octa edge $d_{BXE}$ (Å)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td></td>
<td>$d_{AXE}$</td>
<td>Shared</td>
</tr>
<tr>
<td>0.0</td>
<td>1.898</td>
<td>2.044</td>
<td>3.098</td>
<td>2.814</td>
</tr>
<tr>
<td>0.1</td>
<td>1.899</td>
<td>2.045</td>
<td>3.100</td>
<td>2.816</td>
</tr>
<tr>
<td>0.2</td>
<td>1.900</td>
<td>2.046</td>
<td>3.102</td>
<td>2.818</td>
</tr>
<tr>
<td>0.3</td>
<td>1.897</td>
<td>2.043</td>
<td>3.097</td>
<td>2.814</td>
</tr>
<tr>
<td>0.4</td>
<td>1.894</td>
<td>2.040</td>
<td>3.093</td>
<td>2.809</td>
</tr>
<tr>
<td>0.5</td>
<td>1.891</td>
<td>2.037</td>
<td>3.087</td>
<td>2.804</td>
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Table 5.7
Band position ($\nu_1$ and $\nu_2$), Force constant ($K_0$ and $K_t$) and Bond length ($R_A$ and $R_B$) of $\text{Ni}_{0.7}\text{Zn}_{0.3}\text{Fe}_{2-x}\text{Cr}_x\text{O}_4$

<table>
<thead>
<tr>
<th>Comp. 'x'</th>
<th>$\nu_1$ (cm$^{-1}$)</th>
<th>$\nu_2$ (cm$^{-1}$)</th>
<th>$K_0 \times 10^5$ (dyne/cm)</th>
<th>$K_t \times 10^5$ (dyne/cm)</th>
<th>$R_A$(Å)</th>
<th>$R_B$(Å)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.0</td>
<td>575</td>
<td>400</td>
<td>1.478</td>
<td>1.509</td>
<td>0.691</td>
<td>1.207</td>
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<tr>
<td>0.1</td>
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<td>406</td>
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<td>1.551</td>
<td>0.691</td>
<td>1.209</td>
</tr>
<tr>
<td>0.2</td>
<td>583</td>
<td>406</td>
<td>1.520</td>
<td>1.548</td>
<td>0.691</td>
<td>1.211</td>
</tr>
<tr>
<td>0.3</td>
<td>587</td>
<td>417</td>
<td>1.541</td>
<td>1.629</td>
<td>0.691</td>
<td>1.206</td>
</tr>
<tr>
<td>0.4</td>
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<td>418</td>
<td>1.546</td>
<td>1.633</td>
<td>0.691</td>
<td>1.201</td>
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<tr>
<td>0.5</td>
<td>589</td>
<td>420</td>
<td>1.551</td>
<td>1.645</td>
<td>0.691</td>
<td>1.194</td>
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</table>
Table 5.8

Saturation magnetization ($M_s$), Magneton number ($\eta_B$) and Coercivity ($H_C$) of Ni$_{0.7}$Zn$_{0.3}$Fe$_{2-x}$Cr$_x$O$_4$

<table>
<thead>
<tr>
<th>Comp. $x$</th>
<th>$M_s$ (emu/g)</th>
<th>$\eta_B$ ($\mu_B$)</th>
<th>$H_C$ (Oe)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>Obs.</td>
<td>Cal.</td>
</tr>
<tr>
<td>0.0</td>
<td>58.31</td>
<td>2.47</td>
<td>4.4</td>
</tr>
<tr>
<td>0.1</td>
<td>56.65</td>
<td>2.39</td>
<td>4.2</td>
</tr>
<tr>
<td>0.2</td>
<td>53.23</td>
<td>2.25</td>
<td>4</td>
</tr>
<tr>
<td>0.3</td>
<td>49.09</td>
<td>2.07</td>
<td>3.8</td>
</tr>
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<td>0.4</td>
<td>44.95</td>
<td>1.89</td>
<td>3.6</td>
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<td>0.5</td>
<td>42.90</td>
<td>1.80</td>
<td>3.4</td>
</tr>
</tbody>
</table>
Table 5.9

Curie temperature ($T_c$) and activation energy ($\Delta E$) from DC resistivity plots of Ni$_{0.7}$Zn$_{0.3}$Fe$_{2-x}$Cr$_x$O$_4$

<table>
<thead>
<tr>
<th>Comp. $x$</th>
<th>$T_c$ (°C)</th>
<th>$E_p$ (eV)</th>
<th>$E_f$ (eV)</th>
<th>$\Delta E$ (eV)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.0</td>
<td>518</td>
<td>0.21</td>
<td>0.11</td>
<td>0.10</td>
</tr>
<tr>
<td>0.1</td>
<td>510</td>
<td>0.26</td>
<td>0.14</td>
<td>0.12</td>
</tr>
<tr>
<td>0.2</td>
<td>504</td>
<td>0.32</td>
<td>0.16</td>
<td>0.16</td>
</tr>
<tr>
<td>0.3</td>
<td>495</td>
<td>0.37</td>
<td>0.19</td>
<td>0.18</td>
</tr>
<tr>
<td>0.4</td>
<td>478</td>
<td>0.43</td>
<td>0.21</td>
<td>0.22</td>
</tr>
<tr>
<td>0.5</td>
<td>466</td>
<td>0.51</td>
<td>0.24</td>
<td>0.27</td>
</tr>
</tbody>
</table>
Fig. 5.1: X-Ray diffraction patterns of Ni$_{0.7}$Zn$_{0.3}$Cr$_x$Fe$_{2-x}$O$_4$
Fig. 5.2: Variation of lattice constant with Cr content x
**Fig. 5.3:** Variation of X-ray density with Cr content x.
**Fig. 5.4**: Variation of Crystallite size with Cr content $x$. 
Fig. 5.5: Variation of bulk density ($d_B$), and porosity (P) with Cr content $x$. 
Fig. 5.6: Variation of hopping lengths ($L_A$ and $L_B$) with Cr content $x$. 
Fig. 5.7(A): SEM images for \( x = 0.1 \) and \( x = 0.2 \)

Fig. 5.7(B): SEM images for \( x = 0.3 \), and \( x = 0.4 \)
**Fig. 5.8:** Variation of ionic radii ($r_A$ and $r_B$) with Cr content $x$. 
**Fig. 5.9:** Variation of theoretical lattice constant and oxygen parameter with Cr content x.
**Fig. 5.10:** Infrared spectra of Ni$_{0.7}$Zn$_{0.3}$Cr$_x$Fe$_{2-x}$O$_4$
Fig. 5.11: Variation of magnetization with applied field.
Fig. 5.12: Variation of magneton number with Cr content x.
Fig. 5.13: Variation of magnetization and coercivity with Cr content x.
Fig. 5.14: Variation of initial permeability with temperature.
**Fig. 5.15:** Variation of initial permeability with Cr content $x$. 
**Fig. 5.16:** Variation of Curie temperature with Cr content x.
Fig. 5.17: Variation of initial permeability with logarithm of frequency.
Fig. 5.18: Variation of logarithm of resistivity with temperature.
Fig. 5.19: Variation of logarithm of resistivity with Cr content $x$. 

\[ \log \rho (\Omega \text{-cm}) \] at 400 K
Fig. 5.20: Variation of dielectric constant with logarithm of frequency.
Fig. 5.21: Variation of dielectric loss with logarithm of frequency.
Fig. 5.22: Variation of dielectric loss tangent with logarithm of frequency.
**Fig. 5.23**: Variation of dielectric constant with temperature.
Fig. 5.24: Variation of dielectric loss with temperature.
Fig. 5.25: Variation of dielectric loss tangent with temperature.