Structural, dielectric and low temperature magnetic response of Zn doped cobalt ferrite nanoparticles

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ABSTRACT
The finely controlled nanostructured cubic spinel ferrites pave the way to synthesize nanomaterials with specific properties for particular applications. In this paper, we report sol-gel synthesis of Zn doped spinel Co$_{1-x}$Zn$_x$Fe$_2$O$_4$ (where x= 0.0, 0.1, 0.2, and 0.3) ferrite nanoparticles. X-ray diffraction confirms the single phase cubic structure of nano ferrites with average particle size estimated between 55.38 to 32.87 nm and validated by Transmission electron microscopy (TEM) results (±1). The lattice parameter was found to increase with increasing Zn doping concentration. The samples exhibit normal dielectric behaviour of Maxwell-Wagner type of interfacial polarization that decreases with increasing frequency of the applied field. Temperature-dependent magnetic properties were investigated with the aid of physical property measurement system. The hysteresis measurements of the samples show clearly enhancement in magnetic parameters as the temperature goes down to 20 K. Tuning of magnetic properties has been witnessed as a function of doping and temperature under the influence of externally applied magnetic field, has been discussed in this paper.

I. INTRODUCTION
Magnetic materials are of great interest since their discovery to the scientific community because of their interesting properties and applications in modern technology. Among various magnetic materials, ferrites are one of the important class of materials, which have attracted the interest of scientific community since last 5 decades due to their promising properties. When treated on nano-scale, they show different/tunable electrical and magnetic properties by chemical manipulation. These materials show very important applications in high-frequency devices (MHz to GHz), high-density perpendicular recording media devices and ferrofluids. Among various spinel ferrite materials, the unique and novel size-dependent chemical and physical properties displayed by materials on nanoscale have initiated the current wide range of intensive research on nanomaterials. The nanocrystalline cobalt ferrite has been extensively studied due to its distinctive properties such as cubic magnetocrystalline...
anisotropy, high coercivity, moderate saturation magnetization, high chemical stability, wear resistance, and electrical insulation.\textsuperscript{7-10} The cobalt ferrite has shown promising applications in high-frequency devices, memory cores, recording media, and also in the biomedical field.

In a normal spinel structure of $\text{CoFe}_2\text{O}_4$, Co is a divalent atom that occupies the tetrahedral A-site, whereas Fe is a trivalent atom that occupies the octahedral B-site. With A-sites being populated by Fe\textsuperscript{3+} ions, while B-sites equally being populated by Co\textsuperscript{2+} and Fe\textsuperscript{3+} ions, the structure is known as inverse spinel. The interaction between two sub-lattices A and B, in a spinel system ($\text{AB}_2\text{O}_4$), consist of inter-sublattice (A-B) super-exchange interactions, intra-sublattice (A-A), and (B-B) exchange interactions. Among the three interactions, the super-exchange interactions (A-B) is more dominant than the other two exchange interactions.\textsuperscript{11}

In this paper, we report influence of Zn doping on the structural, morphological and transport electrical and magnetic properties of $\text{Co}_{1-x}\text{Zn}_x\text{Fe}_2\text{O}_4$ ferrite nanoparticles in order to improve physiochemical properties. The substitution of non-magnetic Zn ion is expected to modify crystal as well as other properties by altering the cation distribution between the two crystallographic A and B-sites.

II. EXPERIMENTAL DETAILS

A. Materials and preparation

$\text{Co}_{1-x}\text{Zn}_x\text{Fe}_2\text{O}_4$ ferrite nanoparticles, were prepared by conventional sol-gel method using AR grade cobalt nitrate [Co(NO\textsubscript{3})\textsubscript{2}.6H\textsubscript{2}O] (99%), zinc nitrate [Zn(NO\textsubscript{3})\textsubscript{2}.4H\textsubscript{2}O] (99%), ferrie nitrate [Fe(NO\textsubscript{3})\textsubscript{3}.9H\textsubscript{2}O] (99%), urea (CH\textsubscript{2}N\textsubscript{2}O) (99%), of high purity in stoichiometric proportion. Metal nitrates were dissolved in double distilled water with respect to weight percent as calculated in 20 mL deionized water and 20 mL acetone. The coloured solutions formed were then mixed very well on a magnetic stirrer with constant heating of 70 $^\circ$C followed by grinding using mortar and pestle for 1 hour. The powder samples obtained were further heat treated at 600 $^\circ$C in a furnace for 6 hours while maintaining heating and cooling rate at 5 $^\circ$C/min followed by grinding using mortar and pestle for 1 hour.

B. Characterization

The structural and phase formation of the as-grown NPs were studied through X-ray diffraction using Siemens D-5000 powder X-ray diffractometer with CuK\textsubscript{a} source radiation of wavelength 1.54056 Å. The diffractograms were recorded in terms of 2$\theta$ in the range 20$^\circ$ - 80$^\circ$ with a scanning rate of 2$^\circ$/min. The crystalline phases were identified and analyzed by using Powder X-software. The mean crystallite size ($D_{XRD}$) was calculated by using Sherrer equation:\textsuperscript{12}

$$<D_{XRD}>=\frac{K\lambda}{\beta\cos\theta}$$

Where the constant $K$ is related to the crystalline shape and to both $\beta$ full width at half maximum (FWHM) intensity measured in radians and $<D_{XRD}>$, and $\lambda$ is the wavelength of the X-ray (CuK\textsubscript{a} =0.154x10$^{-10}$ nm) and $\theta$ is the Bragg’s angle in degrees. For spinel ferrites, value of $K$ is assumed to be 0.98, while $\beta$ is defined by the warrens correction, $\beta = \sqrt{\beta_{\text{exp}}^2 - \beta_{\text{std}}^2}$, where $\beta_{\text{exp}}$ and $\beta_{\text{std}}$ represent the experimental width of the peaks and instrumental width calculated by means of the pattern of a standard obtained under the same experimental conditions.

The lattice parameter $a$ for all the samples was calculated using the following relation:\textsuperscript{13}

$$a = \sqrt{\frac{2}{d^2}}$$

Where the symbol $d$ stands for the experimentally observed lattice inter-planar spacing and $h k l$ represents Miller indices. The theoretical density ($\rho_t$) of the nanoparticles was estimated using the following equation:\textsuperscript{13,14}

$$\rho_t = \frac{ZM}{NA^3}$$

Where the symbol Z represents the number of molecules per unit cell in the spinel ferrite and is equal to 8, M stands for Molar Mass of the sample estimated from the general formula $\text{Co}_{1-x}\text{Zn}_x\text{Fe}_2\text{O}_4$ for each value of $x$.

The bulk density ($D_b$) was estimated using the Archimedes method of immersion in an inert liquid from the equation:\textsuperscript{15}

$$\rho = \frac{W_0\rho_1}{W_a - W_1}$$

Where the symbol $W_a$ and $W_1$ represent the weight of the sample taken in the air and the weight of the sample taken when fully immersed in liquid and $\rho_1$ represents the density of the liquid (Xylene) in which sample was immersed.

The porosity of the sample Zn doped cobalt ferrite was estimated by the equation:\textsuperscript{13}

$$P = \left(1 - \frac{\rho_a}{\rho}\right) \times 100$$

Where $\rho_a$ and $\rho$ represent the theoretical density and bulk density of the sample.

The strain induced by doping of Zn in $\text{Co}_{1-x}\text{Zn}_x\text{Fe}_2\text{O}_4$ ferrite NPs was calculated using the following relation:\textsuperscript{16}

$$\epsilon = \frac{\beta\cos\theta}{4}$$

While as the dislocation density was calculated using the formula:\textsuperscript{16}

$$\delta = \frac{15 - \epsilon}{a\lambda}$$

The morphology and size estimation of the NPs was carried out using Field Emission Transmission Electron Microscopy (FETEM) of JEOL operated at 200 kV. The Hexane colloidal dispersion was sonicated for five minutes in an ultrasonic bath and then the diluted dispersion was then dropped on a carbon–coated copper grid, which was left to dry at room temperature.

In order to measure the dielectric and ac conductivity response of the synthesized NPs, the composite material was pressed into a circular disc-shaped pellet (12mm x 2mm) with silver pasting done on opposite faces to ensure good electrical contacts and to make parallel palate capacitor geometry. Dielectric measurements were studied at room temperature using LCR HI-Tester (HIOKI 3532-50) from Japan programmable computer interfaced in the selected frequency range of 50Hz to 5MHz.
The magnetic response of the samples was studied using a vibrating sample magnetometer (VSM). The saturation magnetization ($M_s$) was estimated by using the equation:

$$M = M_s \left(1 - \frac{a}{H} - \frac{b}{H^2}\right)$$

For $H$ tending to infinity.

III. RESULTS AND DISCUSSION

A. Structural and morphological studies

Fig. 1 presents the X-ray diffraction patterns of as-synthesized NPs sintered at 600°C. The X-ray patterns were analyzed through powder X-software, which confirmed the nanocrystalline nature of the samples having cubic spinel structure with $Fd\bar{3}m$ space group (ICDD: 01-072-6799). The intensity of the XRD peaks are found to decrease while the peak broadening increases with increasing Zn doping concentration.

The crystallite sizes for all the compositions were calculated from Lorentzian peak (311) by assuming that NPs are spherical in shape. The experimental results show the mean crystallite size of the NPs lies between 55.38 to 32.87 nm.

The lattice constant $a$, x-ray density $\rho_x$, crystallite size $t$, apparent density $\rho_m$, dislocation density $\varepsilon$, was calculated from the XRD data and is presented in Table I. The lattice parameter is found to increase from 8.4564 to 8.5578 Å with increasing Zn doping, which means expansion of the unit cell which may be due to replacement of smaller ionic radii Fe$^{3+}$ ion ($0.064$ nm) by larger ionic radii Zn$^{2+}$ ($0.74$ nm) ion.

From Table I, it is seen that both X-ray ($\rho_x$) and theoretical density ($\rho$) decreases with increasing $x$, which may be due to the fact that the density of dopant Zn ($2.70$ gm/cm$^3$) is less than the density of host Fe ($7.874$ gm/cm$^3$). Further, it is noted from Table I, the porosity of the samples increases $7$ % to almost $16$ % with increasing value of $x$ confirming decrease in densification as a result of more pores generated by doping.

Generally, any doping element changes the crystal symmetry in a material, as a result, many changes are witnessed in a crystal geometry, such as strain and dislocation density. The strain ($\varepsilon$), as well as the dislocation density ($\delta$) values of grown NPs are seen to decrease with increasing Zn substitution, which indicates that the crystal symmetry of grown ferrite NPs also change, however, the changes noticed as shown in Table I.

B. Morphology

The morphology of the samples was investigated by Field Emission Transmission Electron Microscopy (FETEM). Fig. 2(a) presents the FETEM micrographs for the composition $x=0.0$ and 0.1, respectively. The micrographs show NPs are agglomerated and almost spherical in shape. The agglomeration of NPs can be due to two reasons; first the choice of synthesis technique used which is sol-gel in our case and second the NPs are magnetic in nature.

<p>| TABLE I. Presents variation of crystallite size, lattice parameter, X-ray density, bulk density, porosity, dislocation density and strain with composition. |</p>
<table>
<thead>
<tr>
<th>X</th>
<th>Crystallite size (nm)</th>
<th>Lattice Parameter (Å)</th>
<th>X-ray density ($\rho_x$) (g/cm$^3$)</th>
<th>Th. Density ($\rho$) (g/cm$^3$)</th>
<th>Porosity (%)</th>
<th>Strain ($\varepsilon$)</th>
<th>Dislocation ($\delta$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.0</td>
<td>55.38</td>
<td>8.5578</td>
<td>5.46</td>
<td>4.95</td>
<td>7.99</td>
<td>0.5965</td>
<td>0.0590</td>
</tr>
<tr>
<td>0.1</td>
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<td>8.5462</td>
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<td>0.0663</td>
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<td>0.0794</td>
</tr>
<tr>
<td>0.3</td>
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<td>8.5379</td>
<td>5.18</td>
<td>4.37</td>
<td>15.63</td>
<td>0.1125</td>
<td>0.0863</td>
</tr>
</tbody>
</table>

FIG. 1. Shows the XRD pattern of the Co$_{1-x}$Zn$_x$Fe$_2$O$_4$ ferrite nanoparticles.
FIG. 2. (a) Presents the FETEM micrographs of the composition $x=0.0$ and $0.1$. (b) Presents the crystallite size distribution of the composition $x=0.0$ and $0.1$. (c) Presents the SAED patterns of the composition $x=0.0$ and $0.1$. (d) Shows the presents the lattice plane formation of the composition $x=0.0$ and $0.1$. 
crystallite size of the NPs was estimated using Image J software as shown in Fig. 2(b) and was found between 40.52 to 16.52 nm, which agrees very well with the crystallite sizes determined from the XRD data ($\pm 1$). Fig. 2(c) presents the selective area electron diffraction pattern (SAED) for the samples, where we see presence of diffuse rings, which is a signature confirming the high crystalline order of the prepared ferrite samples. Fig. 2(d) shows the lattice plane formation with a lattice spacing value of $d = 2.5875$ Å in the composition $x = 0.0$ and $d = 2.5874$ Å for the composition $x = 0.1$.

C. Dielectric response

The dielectric constant ($\varepsilon'$) in ferrite materials is contributed by several structural and microstructural factors. In the present study, we have investigated the frequency dependent response of the dielectric constant at room temperature. The effect of frequency on the real ($\varepsilon'$) and imaginary ($\varepsilon''$) part of the dielectric constant for different $Co_{1-x}Zn_xFe_2O_4$ ferrite compositions is illustrated in Fig. 3(a).

**FIG. 3.** (a) and (b) shows the variation of real and imaginary part of dielectric constant with respect to the frequency.

**FIG. 4.** Shows the variation of dielectric loss with frequency.

**FIG. 5.** Shows the variation of ac conductivity with frequency.
and (b). The dispersion in dielectric constant shows normal behaviour representing the Maxwell-Wagner\textsuperscript{18,19} type of interfacial polarization in accordance with the Koop’s phenomena logical theory.\textsuperscript{20} The dielectric constant decreases rapidly in the low-frequency region and remains almost unaffected in the high-frequency region, which can be understood on the basis that beyond a certain frequency limit of the applied field, the charge exchange cannot follow the changes in the applied field. The imaginary part of the dielectric constant also shows a similar behaviour where it decreases with increasing frequency of the applied field.

The imaginary dielectric constant ($\varepsilon''$) is found to decrease faster than the real dielectric constant ($\varepsilon'$) in the low-frequency region and the variation is almost same as both approaches to the higher frequency region. It is seen that the dielectric constant value decreases with increasing concentration of the Zn doping. The maximum value of the dielectric constant is obtained for the pure sample ($x = 0.0$).

### D. Dielectric loss

Fig. 4 shows the variation of the dielectric loss with the frequency of the applied field. It is seen that the dielectric loss decreases rapidly with increasing frequency of the applied field in low-frequency region while the decrease is slowed down in high-frequency region. A very low dielectric loss value has been observed in the investigated samples ($\approx 1.05$ for pure sample), which shows the synthesized materials are a promising candidate for the higher frequency applications.\textsuperscript{21}

### E. ac conductivity

Fig. 5 shows the variation of ac conductivity with frequency at room temperature. The ac conductivity is found to increase gradually with increasing frequency of the applied field. A ferrite structure is assumed to be composed of highly conducting grains separated from low conductivity grain boundaries. At

![Graphs of M-H loops at different temperatures](image)

**FIG. 6.** (a) Shows the M-H loops at 300 K. (b) Shows the M-H loops at 200 K. (c) Shows the M-H loops at 100 K. (d) Shows the M-H loops at 20 K.
lower frequencies, the grain boundaries are more active, therefore, the hopping of electrons is minimized and conductivity is almost constant, while at higher frequencies, grains are more active in promoting the electrons through grain boundaries, which causes increase in electron hopping and hence ac conductivity increases.22–24

The compositional variation of ac conductivity shows that ac conductivity decreases with increasing concentration of Zn doping which may be due decrease in the ratio of Fe$^{2+}$/Fe$^{3+}$ ions and hence decrease in electron hopping.

F. Magnetic measurements

Magnetic hysteresis loops determine the ordering of spins in magnetic materials. Magnetic hysteresis measurements were performed for the prepared samples at different temperatures, 20 K, 100 K, 200 K, and 300 K, respectively, under an applied field of 1 kOe as shown in Fig. 6(a), b, c and (d). The inset of Figure 6(a) shows the detailed hysteresis loops near the origin, to make coercivity visible. Magnetic parameters such as saturation magnetization ($M_s$), remanence magnetization ($M_r$) and coercivity ($H_c$), remanent ratio ($R$), magnetic moment ($\eta B$), were calculated for all compositions at different temperatures and are listed in Table II. The magnetic measurements confirm that the Co$_{1-x}$Zn$_x$Fe$_2$O$_4$ ferrites are ferromagnetic in nature as they all exhibit ‘S’ shaped magnetic hysteresis loops. The samples exhibited narrow hysteresis loops at 300 K confirming weak ferromagnetism but broadened loops at 20 K confirming strong ferromagnetism. $M_r$ values were found to decrease from 36.8 emu/g for $x = 0.0$ to 21.9 emu/g for $x = 0.3$ at 300 K, with increasing concentration of Zn except for the composition $x = 0.1$ ($M_r = 43.12$ emu/g). However, the $M_r$ values increase as the temperature goes down to 20 K ($M_r = 40.3$ emu/g for $x = 0.0$ and $M_r = 36$ emu/g for $x = 0.3$, respectively). The increase in saturation magnetization with increasing Zn doping can be explained on the basis of magnetic moment, which initially increases, then decreases as the doping level is increased by more than 10%. The non-magnetic Zn$^{2+}$ ions have an occupational preference for A–site, the Fe$^{3+}$ ions at A-site migrate to B-site, which in turn decreases the magnetic moment at A–site, but increases the Fe$^{3+}$ number at B-site. However, the net magnetization of spinel ferrite crystal is the difference of magnetization between A- and B-site and hence we see an increase in magnetization up to 10 % doping. The further increase in Zn doping results in the decrease of overall magnetization which may be attributed to spin canting, which leads to canted spins with non-collinear magnetic order in which orientation of magnetic ions aligns by making angles from the easy or preferred direction and hence magnetization decreases. Similar results have been reported by Nandapure et al. 

and Yadav et al.26

<table>
<thead>
<tr>
<th>TABLE II. Presents the variation of magnetic parameters with composition at 20 K, 100 K, 200 K, and at 300 K.</th>
</tr>
</thead>
<tbody>
<tr>
<td>x</td>
</tr>
<tr>
<td>------------------------------------------</td>
</tr>
<tr>
<td>T=20K</td>
</tr>
<tr>
<td>0.0</td>
</tr>
<tr>
<td>0.1</td>
</tr>
<tr>
<td>0.2</td>
</tr>
<tr>
<td>0.3</td>
</tr>
<tr>
<td>------------------------------------------</td>
</tr>
<tr>
<td>T=100K</td>
</tr>
<tr>
<td>0.0</td>
</tr>
<tr>
<td>0.1</td>
</tr>
<tr>
<td>0.2</td>
</tr>
<tr>
<td>0.3</td>
</tr>
<tr>
<td>------------------------------------------</td>
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<tr>
<td>T=200K</td>
</tr>
<tr>
<td>0.0</td>
</tr>
<tr>
<td>0.1</td>
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<tr>
<td>0.2</td>
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<tr>
<td>0.3</td>
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<tr>
<td>------------------------------------------</td>
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<tr>
<td>T=300K</td>
</tr>
<tr>
<td>0.0</td>
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<tr>
<td>0.1</td>
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<td>0.2</td>
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<td>0.3</td>
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</table>
At room temperature, the magnetization of nanomaterials is supposed to be mainly dependent on size effects, whereas, at low temperature, size confinement, quantum effects, spin glass transition, thermal dependence adds to the magnetization. Also, at nanoscale, a nanoparticle is expected to be quantized of spin wave excitation, while at high temperature, they have broader energy levels with a continuous excitation spectrum. In such a case, the temperature dependence of magnetization can be ascribed by Bloch’s law, while at low temperatures, it is considered that at nanoscale ‘long spin wave’ excitations cannot propagate and hence the spectrum becomes discrete, which is supposedly due to the increase in magnetization. The behaviour can be expressed using the equation:

\[ M_s(T) = M_s(0) - C \exp\left(\frac{-E_1}{kT}\right) + \exp\left(\frac{-E_2}{kT}\right) \]

Where the symbols, \( M_s(T) \) represent temperature dependent magnetization, \( M_s(0) \) is magnetization at 0 K, \( C \) is constant, \( E_1 \) and \( E_2 \) represent energy levels, and \( K_b \) stands for Boltzmann constant. This shift in the spin wave spectrum with temperature increases the magnetic energy levels which increases the magnetic response.

The response of remanent magnetization \( (M_r) \) with composition shows it decreases with increasing substitution of Zn doping. In order to further understand the remanence magnetization behaviour, we calculated the reduced remanance \( (R) \), i.e., ratio of \( M_r \) and \( M_s \). It is worth to note that as expected, the value of the reduced remanence magnetization decreases from 0.31 to 0.2, which may be attributed to the decreasing magnetic moment due to the incorporation of a non-magnetic Zn²⁺ ion. Also, the decrease in reduced remanence can be attributed to the decreasing size of the NPs, where we observe formation of single domains.

The M-H loops show an appreciable increase in coercivity as the temperature goes down from 300 K to 20 K shown in Fig. 6(d). The higher values of coercivity are linked with the magneto-crystalline anisotropy, which is measured by the strength of spin-orbital coupling. At low temperature, the magnetocrystalline anisotropy is pronounced and therefore, particles are scattered in the direction of the anisotropic field due to which coercivity increases. Further, it is to be noted that at low temperature, the effect of thermal fluctuations of blocked moments across the anisotropy barrier is responsible for the increase in coercivity. Therefore, with decreasing temperature, the decrease in thermal fluctuations tend to make magnetic moments isotropic, which causes increase in the coercivity.

The magnetic moment, shape anisotropy constants, and the magnetocrystalline constant were calculated with the help of Mathematica code. The calculations were carried out by using the hysteresis loop data by measuring experimental parameters such as saturation magnetization \( (M_s) \), coercivity \( (H_c) \), and threshold magnetization \( (M_{th}) \) at zero fields. The benefits of non-zero threshold magnetization enable us to calculate easily the shape anisotropy constants by using the equation:

\[ H = \frac{2K_1}{M_s} \left( \frac{M_s}{M_r} \right) + \frac{4K_2}{M_s} \left( \frac{M_s}{M_r} \right)^3 \]

Where the symbols \( H \), \( M_s \), \( M_r \), \( K_1 \), and \( K_2 \) stand for the applied magnetic field, saturation magnetization, magnetization, and magnetocrystalline anisotropy constants. The equation was solved to determine the constant \( K_1 \) and \( K_2 \) using the two conditions:

\[ M = \begin{cases} M_T & \text{at } H = 0 \\ M_S & \text{at } H = 5000 \end{cases} \]

Where \( M_T \) represents threshold magnetization. When applying above two conditions, we get the value of \( K_1 \) and \( K_2 \) turns out equal to:

\[ K_1 = -2K_2 \frac{M_T^2}{M_S^2} \]

And

\[ K_2 = \frac{1250M_S^4}{M_T^4} \]

The estimated values of anisotropy constants, \( K_1 \) and \( K_2 \) are presented in Table II. The difference between \( K_1 \) and \( K_2 \), determined under different magnetic field unveils a fundamental issue in magnetism: the shape anisotropy not only is determined by the shape itself but also depends on the applied magnetic field. Moreover, both shape anisotropy constants shows a composition independent response while the magnetocrystalline anisotropy constant shows composition dependent response at all considered temperatures and increases with decreasing temperature.

Magnetic moment per formula unit (\( \eta_a \)) for the grown nanoparticles was calculated using the formula:

\[ \eta_a = \frac{M_w \times M_s}{585} \]

Where the symbol \( M_w \) is the molecular weight of the sample, \( M_s \) is the saturation magnetization measured at different temperatures. It is seen that the net magnetic moment decreases with increasing substitution of non-magnetic Zn ion which can be due to fractional effects that are caused by lattice defects or decrease in the magnetic interaction between two sublattices. According to Neel’s theory of magnetism, exchange interaction between electrons or ions in the sub-lattices leads to individual magnetic moment. The net magnetic moment mainly depends on the distribution of magnetic ions occupying the B-sites and A-sites. Magnetic ions in sub-lattices have the main contribution to the A-B interaction which is strongest in nature, followed by A-A interaction and B-B interaction. The contribution of magnetic moment (MB) in B-sites is greater than A-sites (MA) and thus the net magnetic moment is defined as:

\[ M_n = \frac{M_B}{M_A} \]

The magnetic moment of Fe³⁺ is 5BM while Zn²⁺ has a zero magnetic moment. Therefore, overall magnetization is decreased with increasing concentration of Zn doping.

IV. CONCLUSION

We have successfully synthesized single phase cubic spinel Zn doped Co₁₋ₓZnₓFe₂O₄ ferrite nanoparticles by sol-gel method with crystallite sizes between 55.38 to 32.87 nm. The frequency dependent dielectric response shows normal behaviour and has been explained in the light of Maxwell-Wagner two-layer model. The dielectric constant, loss tangent and ac conductivity decreases with non-magnetic Zn doping. The saturation magnetization, remanence and coercivity increases with lowering of temperature but they decrease with increasing the Zn concentration.
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