

Enhanced Magnetization in Low-Concentration Yttrium-Doped Cobalt Ferrite: A Structural and Magnetic Study

Mukesh Kumar¹, Amitabh Kumar¹, Rajnish Kumar^{2,*}

¹Department of Physics, Patliputra University, Patna, India, 800020

²Department of Physics, College of Commerce, Arts and Science, Patliputra University, Patna, India, 800020

*Corresponding Author. E-mail address: rajnish@cocaspatna.ac.in

Received: 15.06.2026; Accepted: 06.08.2026; Published online: 31.08.2026

DOI: <https://doi.org/10.54280/jse.260809>

Abstract: Cobalt ferrite (CoFe_2O_4) is a chemically stable inverse spinel ferrite with tuneable magnetic properties, making it useful for memory devices, sensors, and biomedical applications. In this study, the effect of low-level yttrium (Y^{3+}) substitution on the structural, microstructural, and magnetic properties of $\text{CoY}_x\text{Fe}_{2-x}\text{O}_4$ ($x = 0.00, 0.02, 0.06, 0.10$) was investigated. Samples were synthesized by a citric-acid-assisted sol-gel auto-combustion method and heat-treated at 600°C . X-ray diffraction confirmed the formation of a single-phase cubic spinel structure without secondary phases up to $x = 0.10$. The lattice parameter increased from $8.370 \text{ \AA}$ to $8.398 \text{ \AA}$ with Y^{3+} substitution, indicating lattice expansion due to the larger ionic radius of Y^{3+} . Crystallite size decreased while tensile micro strain increased with increasing Y content. Density slightly decreased and porosity increased, consistent with FESEM observations of finer and more porous particles. FTIR analysis confirmed the preservation of the spinel structure, with small shifts in metal-oxygen vibration bands suggesting local bond distortion caused by Y^{3+} incorporation. Magnetic measurements revealed a strong non-monotonic behaviour. The saturation magnetization (M_S) increased from 52.16 emu g^{-1} for the undoped sample to a maximum of 66.82 emu g^{-1} at $x = 0.06$, followed by a sharp decrease to 36.22 emu g^{-1} at $x = 0.10$. This behaviour is attributed to favourable cation redistribution and enhanced A-B superexchange interactions at low Y concentrations, while higher Y content causes exchange dilution and spin disorder. The coercivity decreased continuously from 1450 to 1060 Oe , indicating reduced magnetic anisotropy and particle-size-related domain changes. The results demonstrate that low-level Y^{3+} substitution, particularly at $x = 0.06$, effectively tunes the crystal structure and significantly enhances magnetization while maintaining phase stability and controllable coercivity. These findings highlight yttrium-doped CoFe_2O_4 as a promising material for advanced magnetic and multifunctional nanodevices.

Key words: Sol-gel synthesis, Magnetic enhancement, Superexchange interactions, Spinel structure.

1 Introduction

Magnetic spinel ferrites, denoted by the formula AB_2O_4 , have attracted considerable attention from researchers due to their diverse electronic, magnetic, and catalytic properties. Here, A represent divalent metal ion (CO^{2+}) whereas B represents a trivalent iron ion (Fe^{3+}). Among several spinel ferrite materials, CoFe_2O_4 (CFO) has gained attention due to its applications in various fields, including targeted drug delivery, biomedical imaging, memory devices, magneto-optical devices, magnetic sensors, and gas sensors [1]. This wide range of

applications stems from its physical properties, including moderate saturation magnetization, a tuneable coercive field, and anisotropy constant, as well as chemical and thermal stability, and mechanical hardness [2].

CFO, a magnetic spinel oxide, has gained prominence over the years due to its exceptional magnetic hardness, moderate saturation magnetization, strong magnetocrystalline anisotropy, and excellent chemical stability [3]. These properties make it suitable for applications ranging from magnetic data storage and microwave devices to biomedical technologies, such as tar-

geted drug delivery and magnetic hyperthermia [4]. The fundamental structure of CFO, characterized by the AB_2O_4 spinel configuration, offers notable flexibility for ionic substitution which allow compositional modification, providing fine control over its physical and magnetic properties [5].

One of the most effective ways to tune the performance of CFO is by substituting ions at either the tetrahedral (A) or octahedral (B) sites. Various dopants-including non-magnetic ions like Zn^{2+} and Al^{3+} , as well as magnetic ones such as Mn^{2+} , Cr^{3+} , and rare-earth ions-have been investigated for their ability to tailor coercivity, saturation magnetization, and grain morphology [6], [7], [8], [9], [10]. For instance, the substitution of Zn^{2+} at the Co^{2+} site has been widely studied. Zn^{2+} , preferring the tetrahedral site, disrupts the magnetic exchange pathways and generally leads to reduced coercivity but improved saturation magnetization up to a certain concentration [10], [7]. Similarly, Al^{3+} substitution at Fe^{3+} sites often results in decreased magnetization due to the non-magnetic nature of aluminum and its influence on the cation distribution [6].

However, the selection of dopant elements should be made carefully, as not all substitutions result in favorable structural or magnetic properties. While magnetic ions like Mn^{2+} or Cr^{3+} can enhance coercivity, they sometimes introduce unwanted secondary phases or disrupt the uniformity of the spinel structure if not carefully optimized [7], [8], [9]. In contrast, rare-earth substitutions, particularly with trivalent ions have shown greater promise due to their ability to influence magnetic anisotropy and structural distortion simultaneously. Among rare-earth elements, yttrium (Y^{3+}) presents a particularly interesting case.

Although Y^{3+} is non-magnetic, it has a relatively large ionic radius (0.90 Å compared to 0.645 Å for Fe^{3+}) and exhibits a strong preference for the octahedral site. Its presence in the CFO lattice can lead to significant lattice distortion, which in turn

affects cationic distribution and enhances magnetic interactions under certain conditions. More importantly, Y^{3+} substitution avoids some of the complications seen with other rare-earth ions-such as excessive lattice stress or phase separation-due to its intermediate size and chemical stability [10].

A comprehensive literature review of the structural and magnetic properties of rare-earth doped CFO, with a specific emphasis on yttrium (Y^{3+}) substitution, indicates highly adjustable outcomes influenced by cation distribution, ionic radii mismatches, and composition-dependent superexchange modifications. Haik Dunn et al. examined $CoY_xFe_{1.98-x}O_4$ synthesized via sol-gel combustion, demonstrating that Y^{3+} ions effectively substitute into the octahedral B-sites up to a 10% cationic threshold. This substitution expands the inversion parameter and reduces both the saturation magnetization (M_S) and Curie temperature (TP) due to the weakening of the A-B and B-B super-exchange interactions [11]. The suppression of M_S through magnetic dilution and localized spin canting is a consistent feature, as Kumari et al. synthesized $CoY_xFe_{2-x}O_4$ nanoparticles using the citrate precursor method. They reported a dual single-phase cubic spinel configuration, where Y^{3+} inclusion restricted grain growth, decreased magnetic saturation (M_S), and enhanced coercivity (H_C) via a transition from a multidomain to a single-domain regime [12]. Similarly, Chakrabarty et al. synthesized nanocrystalline CFO doped with up to 15 mol % Y, identifying optimized oxygen vacancies, Yafet-Kittel spin canting angles, and a composition-dependent evolution of magnetocrystalline anisotropy alongside maximum electrical conductivity at the highest doping limits [13]. To further support the magnetic softening behavior, Rahardjo et al. processed $CoY_{0.10}Fe_{1.90}O_4$ through sol-gel auto-combustion and reported a stable cubic $Fd\bar{3}m$ phase where Y^{3+} substitution for Fe^{3+} reduced the magnetic saturation as a function of processing and annealing temperatures [14]. Beyond single-phase

boundaries, solubility limits often introduce secondary orthoferrite phases; for instance, Ghone et al. evaluated $\text{CoFe}_{2-x}\text{Y}_x\text{O}_4$ systems and recorded the emergence of a secondary YFeO_3 phase at $x \geq 0.15$, which directly correlated with a progressive reduction in both saturation magnetization and magnetostrictive strain due to nonmagnetic rare earth–transition metal interactions [15]. Multi-element modifications further expand these trends, as seen when Devkar et al. examined $\text{Co}_{0.8}\text{Fe}_{2-x}\text{Zn}_{0.2}\text{Y}_x\text{O}_4$ nano ferrites, determining that 3d-4f coupling dynamics and magnetic dilution lowered M_S while concurrently enhancing the magnetostrictive strain derivative, suggesting strong optimization potential for transducer applications [16]. In related Co-doped frameworks, Ahmed et al. analyzed $\text{Co}_{0.785}\text{Zn}_{0.15}\text{Y}_x\text{Fe}_{2-x}\text{O}_4$ fabricated by double sintering, observing a decrease in grain size and M_S alongside high retentivity and coercivity profiles favorable for high-density magnetic recording [17]. Concurrently, Hossain et al. utilized Rietveld refinement on $\text{Co}_{0.25}\text{Zn}_{0.75}\text{Y}_x\text{Fe}_{2-x}\text{O}_4$ matrices to detail an increasing lattice parameter and porosity, noting that Y^{3+} incorporation decreased initial permeability while introducing temperature-dependent magnetic traits specific to soft ferrites [18]. Extending this to alternate co-doped regimes, Das et al. explored $\text{Co}_{0.5}\text{Zn}_{0.5}\text{Y}_x\text{Fe}_{2-x}\text{O}_4$ and verified that single-phase structures hold until $x < 0.06$ before YFeO_3 precipitates out, showing that the large ionic radius of Y^{3+} hinders domain wall motion, resulting in low-loss, high-resistivity materials optimized for high-frequency electronics [19]. Finally, Das et al. manufactured pure $\text{CoY}_x\text{Fe}_{2-x}\text{O}_4$ nanomaterials through a cost-effective sol–gel route, documenting systematic decreases in crystallite size down to 11 nm, lattice parameters, M_S , and anisotropy constants, which collectively soft-tuned the magnetic response and enhanced their suitability for biomedical magnetic hyperthermia applications [20].

In the present work, we synthesized yttrium-substituted CFO ($\text{CoY}_x\text{Fe}_{2-x}\text{O}_4$;

$x = 0.00, 0.02, 0.06, 0.10$) via citric acid modified sol-gel method, a low-temperature route known for producing highly homogeneous, fine-grained, and phase-pure ferrite powders. The sol-gel approach offers superior control over stoichiometry and microstructure, which is critical when working with rare-earth dopants that can easily destabilize the host lattice under traditional ceramic methods.

This study focuses on evaluating the influence of low concentrations of Y^{3+} substitution at the iron site on the phase formation, crystal structure, and magnetic behavior of CFO. Structural analysis using X-ray diffraction (XRD), followed by confirmation of phase purity with Fourier-transform infrared spectroscopy (FTIR) across all substitution levels. Microstructural study has been carried out using Field Emission Electron Microscopy (FESEM). Magnetic measurements using vibrating sample magnetometer (VSM) revealed that samples with 2% and 3% Y^{3+} substitution exhibited enhanced magnetization compared to undoped CFO, with a peak observed at 3%. This improvement is attributed to favorable cation redistribution and lattice strain-induced modification in superexchange interactions. However, at 5% substitution, a decline in magnetization was observed, likely due to the onset of magnetic dilution and microstructural distortion beyond the optimal threshold.

Overall, our study substantiates the value of yttrium as a dopant of choice in CFO systems, exceptionally when fine control over the synthesis process is maintained. Compared to other dopants, Y^{3+} offers a unique balance of structural stability and magnetic enhancement, making it a compelling choice for future magnetic nanodevice and functional ferrite applications.

2 Experimental Details

$\text{CoY}_x\text{Fe}_{2-x}\text{O}_4$, where $x = 0.00, 0.02, 0.06$, and 0.10 , was synthesized using auto combustion citric acid-modified sol-gel. The re-

quired precursors were used in the stoichiometric ratio necessary. Aqueous solutions of all precursors were prepared separately and poured into the aqueous solution of citric acid with constant stirring using a magnetic stirrer. The solution was then heated at 100°C to obtain a viscous gel. This was then heated at 120°C in an oven. After the auto combustion of this gel, a black ash-like material was obtained. The sample was heated at 600°C for 2 hours. Pellet of the synthesized powder was made and sintered at 600 °C for 3 hours. The samples obtained were characterized using an X-ray diffractometer to study the phase purity of the samples, which was supported by FTIR spectroscopy. FESEM was performed to determine the particle size. A magnetic study was performed using a vibrating sample magnetometer, where a magnetic hysteresis loop was recorded in an applied field of $\pm T$. The magnetic properties of the samples were studied using the law of approach (LA) to saturation magnetization to extract the important parameters including saturation magnetization and anisotropy constants providing insights into their magnetic behavior at high magnetic field. Magnetic characterization, combined with structural and morphological assessments, provided a comprehensive understanding of the prepared nanomaterials and their potential applications across

various fields.

3 Results and Discussion

3.1 Structural Analysis: X-Ray Diffraction (XRD)

The X-ray diffraction patterns of the Y^{3+} -substituted CFO series $CoY_xFe_{2-x}O_4$ ($x = 0.00, 0.02, 0.06, 0.10$) are presented in Figure 1. All four compositions exhibit a consistent set of diffraction peaks that can be unambiguously indexed to the characteristic reflections of the cubic spinel structure with space group $Fd\bar{3}m$. The observed peaks, indexed as (111), (220), (311), (222), (400), (422), (511), (440), and (533), are in excellent agreement with the standard JCPDS card for CFO (JCPDS No. 22-1086), confirming the formation of a single-phase spinel structure for all compositions studied [12], [21]. No extraneous peaks attributable to secondary phases such as $YFeO_3$ or Y_2O_3 were detected within the detection limits of the instrument, indicating that Y^{3+} ions were successfully incorporated into the CFO lattice up to the substitution level of $x = 0.10$. This is consistent with the sol-gel synthesis route employed, which promotes atomic-level mixing and facilitates homogeneous substitution at relatively low concentrations [12].

The XRD data were analyzed using Bragg's law to deduce lattice parameter which is mentioned below [22].

$$2d \sin \theta = \lambda \quad (1)$$

$$d = a \times \sqrt{h^2 + k^2 + l^2} \quad (2)$$

where d , θ , λ and a are the interplanar spacing, Bragg's angle, wavelength of X-ray and lattice parameter, respectively. (hkl) is the Miller index for a given plane. The obtained lattice parameter is tabulated in table 1. The obtained lattice parameters showed an increasing trend with increasing Y concentration. A careful examination

of Figure 2, which presents an enlarged view of the most intense (311) reflection, reveals a systematic and progressive shift of the peak position toward lower 2θ angles with increasing Y^{3+} content. For the unsubstituted sample ($x = 0$), the (311) peak is centered at approximately $2\theta = 35.52^\circ$, and this value decreased monotonically as x increased to 0.10. This left shift of the diffraction peaks is a direct consequence of the lattice expansion induced by the substitution of Fe^{3+} (ionic radius ≈ 0.645 Å in octahedral coordination) by the significantly larger Y^{3+} ion (ionic radius ≈ 0.90 Å in octahedral coordination) [23], [24]. As

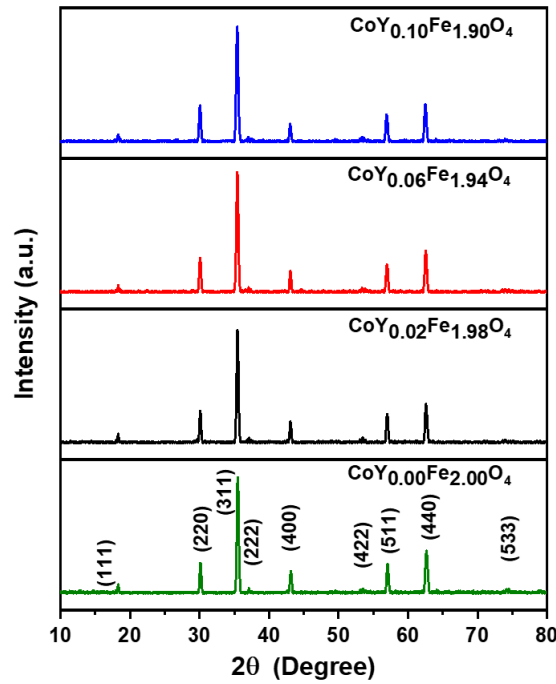


Figure 1: X-ray diffraction (XRD) patterns of $\text{CoY}_x\text{Fe}_{2-x}\text{O}_4$ ($x=0.00, 0.02, 0.06, 0.10$).

confirmed by the lattice parameter data in Table 1, the unit cell parameter “a” increased systematically from 8.370 Å for $x = 0$ to 8.398 Å for $x = 0.10$. From Eq. (1), if “a” increases, θ must decrease to satisfy the equation as λ is constant. This lattice expansion is consistent with Vegard’s law and has been widely reported for rare-earth-substituted spinel ferrites [11], [24]. Thus, the observed increase in lattice

parameter further confirmed the successful incorporation of Y^{3+} ions into the CFO lattice. Because the ionic radius of Y^{3+} is larger than that of Fe^{3+} , the substitution of Y^{3+} at the Fe^{3+} sites leads to lattice expansion, which is reflected in the increase in the lattice parameter and the shift of the diffraction peaks toward lower diffraction angles.

Table 1: D_S : Particle size calculated by Scherrer formula; D_{MS} : particle size calculated by modified Scherrer formula; D_{WH} : particle size calculated by Williamson–Hall plot; a : lattice parameter; ε : lattice strain calculated from Williamson–Hall plot.

Sample	a (Å)	D_S (nm)	D_{MS} (nm)	D_{WH} (nm)	Strain (ε)
$\text{CoY}_{0.00}\text{Fe}_{2.00}\text{O}_4$	8.370	30	36	46	3.5×10^{-3}
$\text{CoY}_{0.02}\text{Fe}_{1.98}\text{O}_4$	8.387	24	31	42	4.6×10^{-3}
$\text{CoY}_{0.06}\text{Fe}_{1.94}\text{O}_4$	8.393	20	26	38	5.3×10^{-3}
$\text{CoY}_{0.10}\text{Fe}_{1.90}\text{O}_4$	8.398	18	22	34	6.1×10^{-3}

Further, the crystallite size of the synthesized material has been studied using Scherrer equation defined below [22].

$$D_s = \frac{k\lambda}{\beta \cos \theta} \quad (3)$$

where D_s , k and β are crystallite size from

eq (3), shape factor (usually 0.9) and full width at half maximum (FWHM) of the diffraction peak in radians, respectively. Other symbols have their usual meaning. Here β , is taken from the highest-intensity peak. The obtained crystallite size is listed in Table 1. The crystallite size is obtained

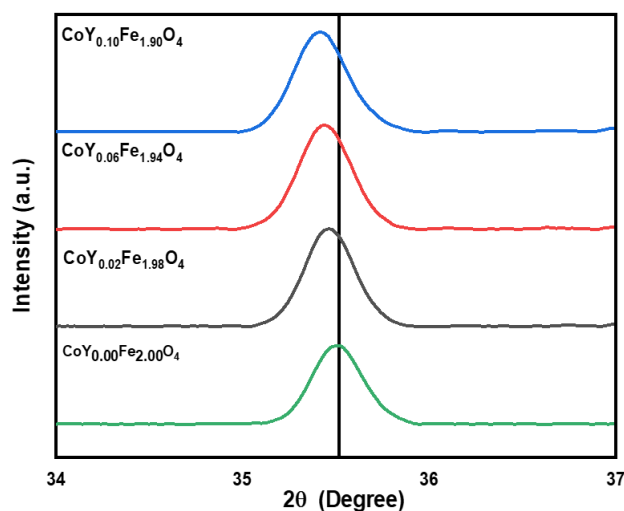


Figure 2: Enlarged view of the (311) diffraction peak of $\text{CoY}_x\text{Fe}_{2-x}\text{O}_4$ ($x=0, 0.02, 0.06, 0.10$) and showed a systematic shift toward lower 2θ value with increased Y^{3+} substitution.

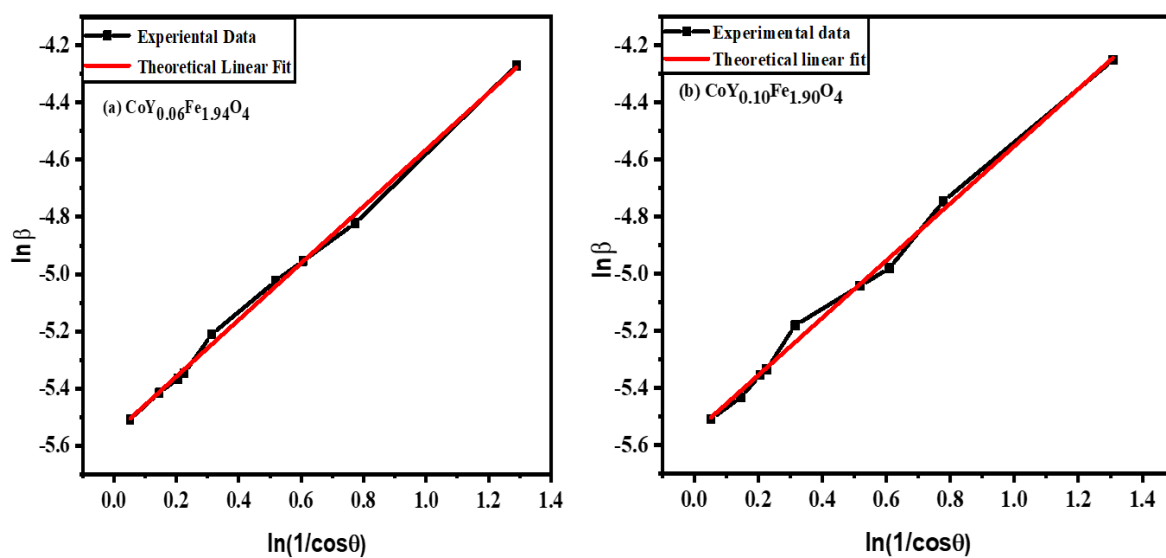


Figure 3: Plot of Modified Scherrer equation for (a) $\text{CoY}_{0.06}\text{Fe}_{1.94}\text{O}_4$ and (b) $\text{CoY}_{0.10}\text{Fe}_{1.90}\text{O}_4$.

from the Eq. (3) decreases from 30 nm for CoFe_2O_4 to 18 nm for $\text{CoY}_{0.10}\text{Fe}_{1.90}\text{O}_4$. However, multiple peaks were recorded for the CFO phase. Therefore, to include the contributions from all recorded peaks, Eq. (3) has been modified by taking the log of both sides, and the obtained equation is known as the modified Scherrer equation, defined below [22].

$$\ln(\beta) = \ln \frac{k\lambda}{D_{MS}} + \ln \frac{1}{\cos \theta} \quad (4)$$

Where D_{MS} is the crystallite size obtained from the above equation. Rest symbols have their usual meaning. β is taken for each peak individually, and $\ln(\beta)$ vs $\ln(1/\cos\theta)$ is plotted to obtain a linear graph, and DMS from the intercept of the graph is obtained. The modified Scherrer equation plot of $\text{CoY}_{0.06}\text{Fe}_{1.94}\text{O}_4$ and $\text{CoY}_{0.10}\text{Fe}_{1.90}\text{O}_4$ is shown in Figure 3. The obtained crystallite sizes are listed in Table 1. The size decreases from 36 nm (for CoFe_2O_4) to 22 nm (for $\text{CoY}_{0.10}\text{Fe}_{1.90}\text{O}_4$). The crystallite size obtained from Eq. (4) is slightly larger than the crystallite size calculated from Eq. (3). This is because, in Eq. (4), the FWHM contribution from

all peaks is considered. Further, in both methods, the contribution from strain is ignored, which also plays an important role in peak broadening. Therefore, to include the contribution from strain, a Williamson-Hall (W-H) plot has been done. The WH equation is defined as follows: [22].

$$\beta \cos \theta = \frac{k\lambda}{D_{WH}} + 4\epsilon \sin \theta \quad (5)$$

where, D_{WH} and ϵ are crystallite sizes obtained from Eq. (5) and strain, respectively. β is taken for all peaks separately, and a plot of $\beta \cos \theta$ vs $4\sin \theta$ is done to obtain a linear graph. The W-H plot of $x=0.06$ and $x=0.10$ are shown in Figure 4. The crystallite size and strain calculated from the graph for all samples are listed in Table 1. The crystallite size obtained from the W-H plot follows a similar trend as obtained from the previous two methods, i.e. crystallite size decreased from 46 nm for $\text{CoY}_{0.06}\text{Fe}_{1.94}\text{O}_4$ to 34 nm for $\text{CoY}_{0.10}\text{Fe}_{1.90}\text{O}_4$. However, the particle size is larger compared to the Scherrer formula and the modified Scherrer formula because the contribution of ϵ in FWHM is considered.

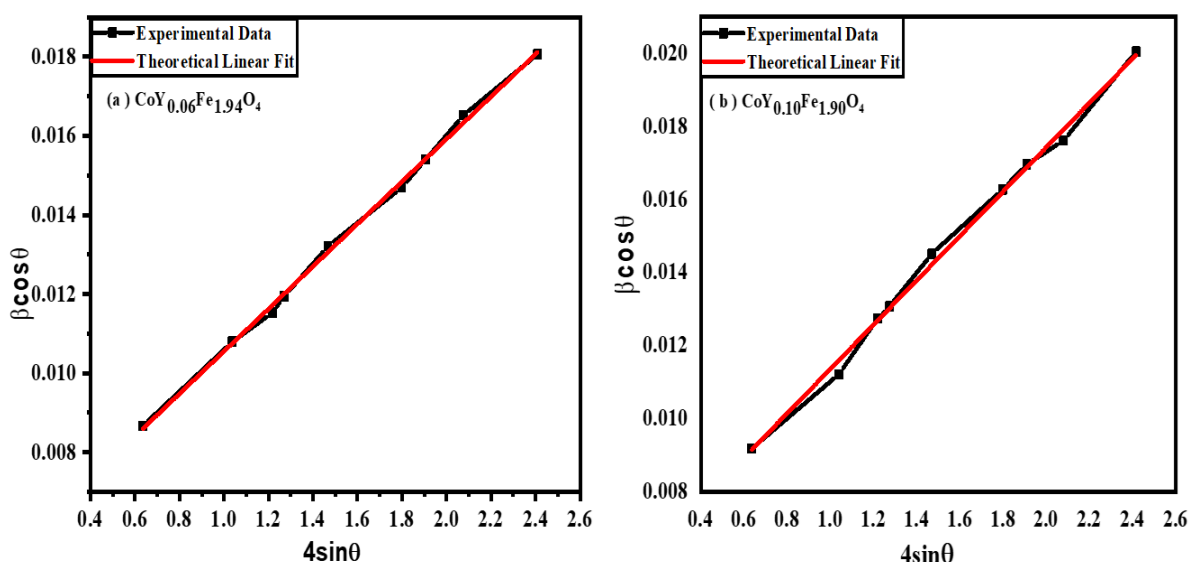


Figure 4: Williamson-Hall (W-H) plots of (a) $\text{CoY}_{0.06}\text{Fe}_{1.94}\text{O}_4$ and (b) $\text{CoY}_{0.10}\text{Fe}_{1.90}\text{O}_4$ used to determine crystallite size of lattice strain..

From Table 1, we clearly note that for the unsubstituted sample, $D_S = 30$ nm, $D_{MS} = 36$ nm, and $D_{WH} = 46$ nm. These values decrease progressively as Y^{3+} content increases, reaching $D_S = 18$ nm, $D_{MS} = 22$ nm, and $D_{WH} = 34$ nm for $CoY_{0.10}Fe_{1.90}O_4$. The consistent reduction in crystallite size with Y^{3+} substitution is attributed to the lattice strain introduced by the size mismatch between Y^{3+} and Fe^{3+} , which impedes grain growth during annealing at $600^\circ C$ [23], [25].

The Williamson–Hall (W-H) plots (Figures 4) provide additional insight into the microstructural state of the samples. In the (W-H) formalism, the intercept of the linear fit to $\beta \cos \theta$ versus $4 \sin \theta$ yields the crystallite size, while the slope is proportional to the lattice microstrain (ϵ). The microstrain values extracted from the (W-H) plots increased monotonically from 3.5×10^{-3} ($CoFe_2O_4$) to 6.1×10^{-3} ($CoY_{0.10}Fe_{1.90}O_4$), as listed in Table 1. The positive slope of the (W-H) plots indicates a tensile-type microstrain regime, which is physically consistent with the lattice expansion caused by the incorporation of the larger Y^{3+} ion [11], [23]. The accumulation of micro strain with increasing substitution level is a well-established consequence of local lattice distortions arising from the size mismatch between the host Fe^{3+} and the dopant Y^{3+} ions, and it contributes to peak broadening beyond the size-broadening effect alone.

The X-ray density (d_x) of the synthesized materials are calculated using the equation mentioned below [22].

$$d_x = \frac{8M}{Na^3} \quad (6)$$

where M , N , and a are the molecular weight of the sample, Avogadro's number (6.022×10^{23}), and the lattice parameter of the sample, respectively. The bulk density (d_B) was experimentally determined from the mass and volume of the sintered pellet made from synthesized materials. The porosity (P) was calculated from the formula mentioned below [26].

$$P = 1 - \frac{d_B}{d_x} \quad (7)$$

The values of d_x , d_B and P are listed in Table 2. The X-ray density decreased slightly from 5.30 g/cm³ ($x = 0$) to 5.26 g/cm³ ($x = 0.10$), which is consistent with the lattice expansion: as the unit cell volume increases, the theoretical mass density decreases if the molecular weight change is proportionally smaller. The bulk density also decreases from 4.34 to 4.20 g/cm³ across the series, while the porosity increases from 18.11% to 20.15% . The increasing porosity with Y^{3+} substitution is a direct consequence of the reduced crystallite size and increased lattice strain, which inhibit densification during sintering and promote the formation of inter-granular voids [19], [21]. Taken together, the XRD analysis confirms the successful synthesis of phase-pure, nanocrystalline Y^{3+} substituted CFO with a cubic spinel structure, in which increasing substitution level introduces progressive lattice expansion, micro strain accumulation, crystallite size reduction, and porosity enhancement.

Table 2: Density and porosity parameters of $CoY_xFe_{2-x}O_4$. Here, d_x is the X-ray density, d_B is the bulk density, and P is the porosity.

Sample	d_x (g/cm ³)	d_B (g/cm ³)	P (%)
$CoY_{0.00}Fe_{2.00}O_4$	5.30	4.34	18.11
$CoY_{0.02}Fe_{1.98}O_4$	5.28	4.28	18.94
$CoY_{0.06}Fe_{1.94}O_4$	5.27	4.22	19.92
$CoY_{0.10}Fe_{1.90}O_4$	5.26	4.20	20.15

3.2 FTIR Spectroscopic Analysis

The FTIR spectra of (a) $\text{CoY}_{0.06}\text{Fe}_{1.94}\text{O}_4$ and (b) $\text{CoY}_{0.10}\text{Fe}_{1.90}\text{O}_4$ samples (Figure 5) display the characteristic absorption features of the spinel ferrite structure. A principal absorption band is observed in the mid-infrared region, i.e., at a higher-frequency band in the range $530\text{--}534\text{ cm}^{-1}$, and another absorption band is expected to occur in the range $350\text{--}400\text{ cm}^{-1}$, which is not observed due to limitation of instrument, as the starting frequency for the instrument is 400 cm^{-1} . However, from the bending at 440 cm^{-1} , it is evident that this absorption band would also have appeared if the starting frequency had been 300 cm^{-1} . These two bands are the hallmark vibrational signatures of spinel oxides and arise from the intrinsic stretching vibrations of the metal-oxygen bonds at the two distinct crystallographic sites of the spinel lattice [27], [28].

The higher-frequency band observed at $\sim 530\text{--}534\text{ cm}^{-1}$ is attributed to the intrinsic stretching vibrations of the metal-oxygen (M-O) bonds at the tetrahedral (A) sites, where Fe^{3+} ions predominantly reside in the inverse spinel structure of CFO [27], [28]. The shorter metal-oxygen bond lengths at the tetrahedral site, compared to the octahedral site, are responsible for the higher vibrational frequency of this band. The lower-frequency band (which doesn't appear due to the limitation of the instrument) is assigned to the M-O stretching vibrations at the octahedral (B) sites, which are occupied by both Co^{2+} and Fe^{3+} ions in CFO [14], [28]. The larger bond lengths at the octahedral site result in a lower restoring force and, thus, a lower vibrational frequency,

consistent with the observed band position.

A systematic and physically meaningful shift was observed in the absorption bands with an increase in the Y^{3+} content. The tetrahedral band shifted from $\sim 530\text{ cm}^{-1}$ ($\text{CoY}_{0.06}\text{Fe}_{1.94}\text{O}_4$) to $\sim 534\text{ cm}^{-1}$ ($\text{CoY}_{0.10}\text{Fe}_{1.90}\text{O}_4$). This shift, although modest, is consistent with the structural changes identified by XRD. The incorporation of Y^{3+} ions, which have a larger ionic radius than Fe^{3+} , into the octahedral sites of the spinel lattice modifies the local bond environment, leading to a redistribution of the metal-oxygen bond lengths and force constants [14], [24]. The slight shift to higher wavenumbers in the band with increasing Y^{3+} content may reflect an increase in the effective force constant associated with the M-O bond, possibly due to the increased lattice strain and bond distortion induced by the larger Y^{3+} substituting ions [28]. Alternatively, the preferential occupation of octahedral B-sites by Y^{3+} can displace Co^{2+} ions, subtly altering the site-specific bond characteristics and vibrational frequencies [14].

The presence of both characteristic spinel bands in all samples, without the appearance of additional absorption features attributable to secondary phases such as YFeO_3 (which typically exhibits absorption bands at different wavenumber positions), corroborates the XRD finding of single-phase spinel formation up to $x = 0.10$ [15], [28]. The FTIR data thus provide complementary molecular-level evidence for the successful substitution of Y^{3+} into the CFO spinel lattice, confirming the structural integrity of the synthesized materials across the entire composition range studied.

3.3 Morphological Analysis: Field Emission Scanning Electron Microscopy (FESEM)

The FESEM micrographs of the representative $\text{CoY}_x\text{Fe}_{2-x}\text{O}_4$ samples ($x = 0.03$ and $x = 0.10$) are presented in Figure 6. Both micrographs, recorded at magnifica-

tions of $50,000\times$ and $60,000\times$ respectively, with a scale bar of 200 nm , reveal a granular, porous microstructure composed of densely packed, irregularly shaped nanoparticle clusters. The surface texture was rough and heterogeneous, with regions of varying brightness and shadows indicating significant topographic relief and depth,

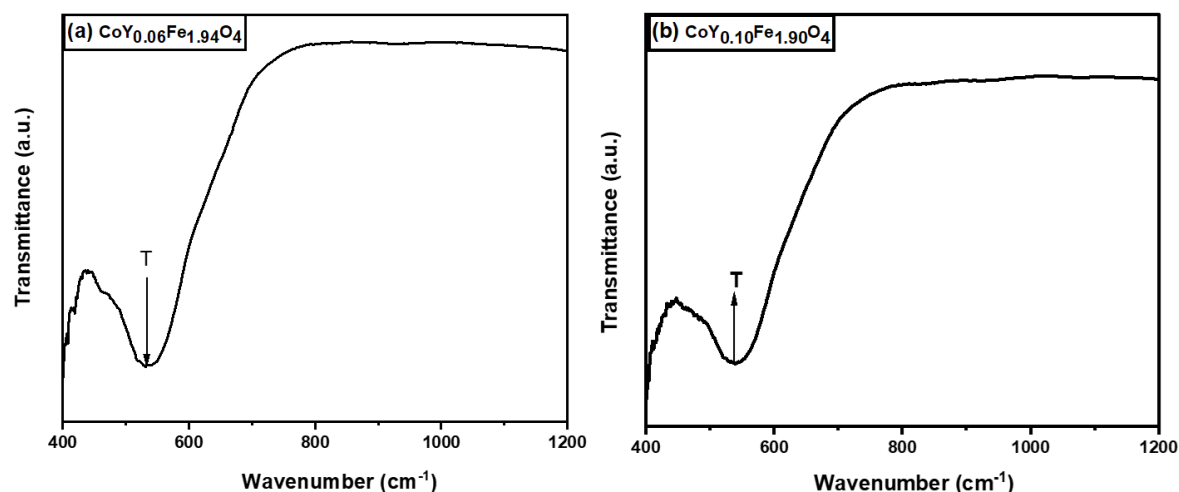


Figure 5: Fourier-transform infrared (FTIR) spectra of (a) $\text{CoY}_{0.06}\text{Fe}_{1.94}\text{O}_4$ and (b) $\text{CoY}_{0.10}\text{Fe}_{1.90}\text{O}_4$.

which is characteristic of nanostructured spinel ferrite materials synthesized by the sol-gel method [21], [25].

For the $x = 0.02$ sample, the particles appeared as fine, rounded clusters with a relatively uniform distribution. The microstructure is consistent with the nanocrystalline nature inferred from XRD, with the apparent particle size estimated to be in the range of 30–50 nm from the micrograph, which is broadly consistent with the crystallite sizes determined by the Williamson-

Hall method ($D_{WH} = 42$ nm for $x = 0.02$, as listed in Table 1). For the $x = 0.10$ sample, the microstructure displayed a slightly finer and more porous texture, with smaller particles and more pronounced inter-particle voids. This observation is in qualitative agreement with the XRD-derived crystallite size reduction from 42 nm ($x = 0.02$) to 34 nm ($x = 0.10$) and the increase in porosity from 18.11% to 20.15% (Table 2) [19], [23].

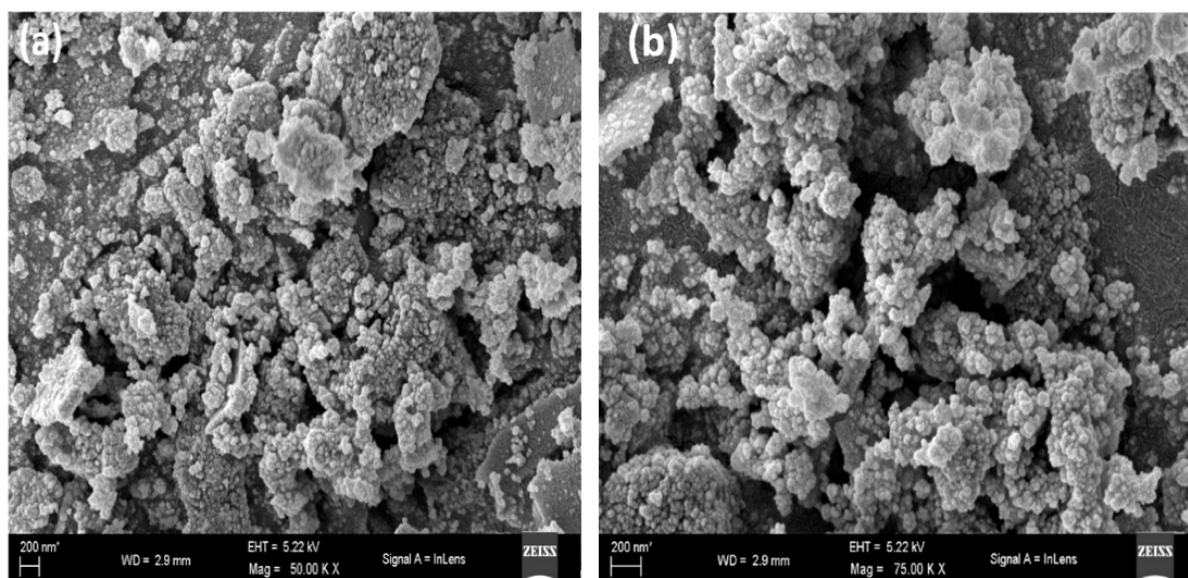


Figure 6: FESEM micrographs of (a) $\text{CoY}_{0.06}\text{Fe}_{1.94}\text{O}_4$ and (b) $\text{CoY}_{0.10}\text{Fe}_{1.90}\text{O}_4$ show particle morphology.

The observed increase in porosity with Y^{3+} substitution, as quantified by the XRD density measurements (Table 2), is consistent with the FESEM observation of more pronounced inter-particle voids in the $x = 0.10$ sample. This increased porosity is a consequence of the reduced crystallite size and the lattice strain introduced by the Y^{3+} substitution, both of which inhibit the densification of the material during annealing [19]. The porous microstructure is not merely a structural curiosity; it has important implications for the surface area, reactivity, and potential functional applications of these materials. Higher porosity increases the accessible surface area, which can be beneficial for catalytic, adsorption, and sensing applications [24].

The FESEM results thus provide direct morphological evidence that is fully consistent with and complementary to the structural information obtained from XRD. The progressive reduction in particle/crystallite size and the increase in porosity with increasing Y^{3+} content, as observed in both the XRD data and the FESEM micrographs, establish a coherent picture of

the microstructural evolution across the $CoY_xFe_{2-x}O_4$ series.

3.4 Magnetic Analysis: Vibrating Sample Magnetometry (VSM)

The room-temperature magnetic hysteresis loops (M-H curves) of the $CoY_xFe_{2-x}O_4$ series, measured in an applied field range of $\pm 20,000$ Oe, are shown in Figure 7. All four samples exhibited well-defined closed hysteresis loops with non-zero coercivity and remanence, confirming the ferrimagnetic nature of the synthesized materials at room temperature [23], [24]. The loops are characteristic of a magnetically hard ferrimagnetic material, consistent with the known high magnetocrystalline anisotropy of CFO arising from the spin-orbit coupling of Co^{2+} ions at the octahedral sites [11], [23]. The quantitative magnetic parameters, saturation magnetization (M_S), remanent magnetization (M_r), coercive field (H_C), and magnetocrystalline anisotropy constant (K), extracted from the VSM measurements and the Law of Approach to Saturation (LAS) analysis are summarized in Table 3.

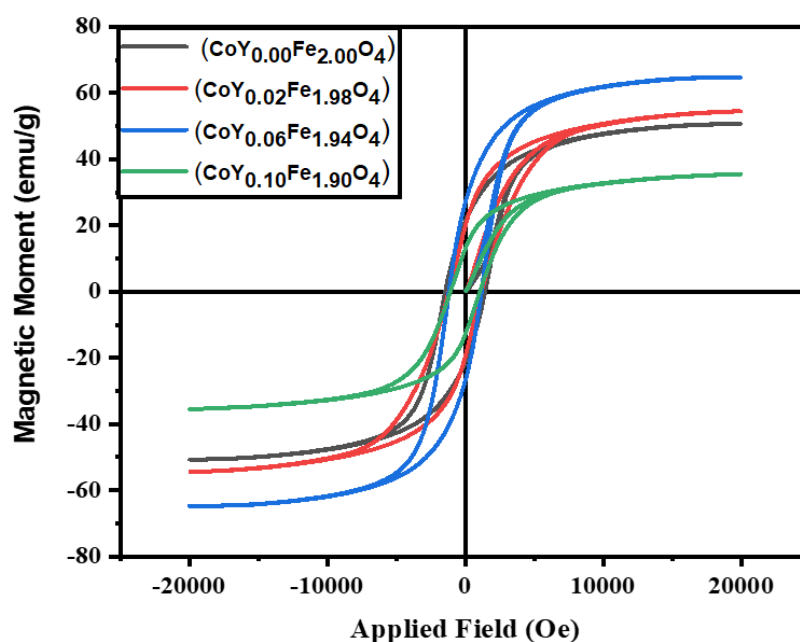


Figure 7: M-H hysteresis loop of $CoY_xFe_{2-x}O_4$ ($x=0,0.02,0.06,0.10$).

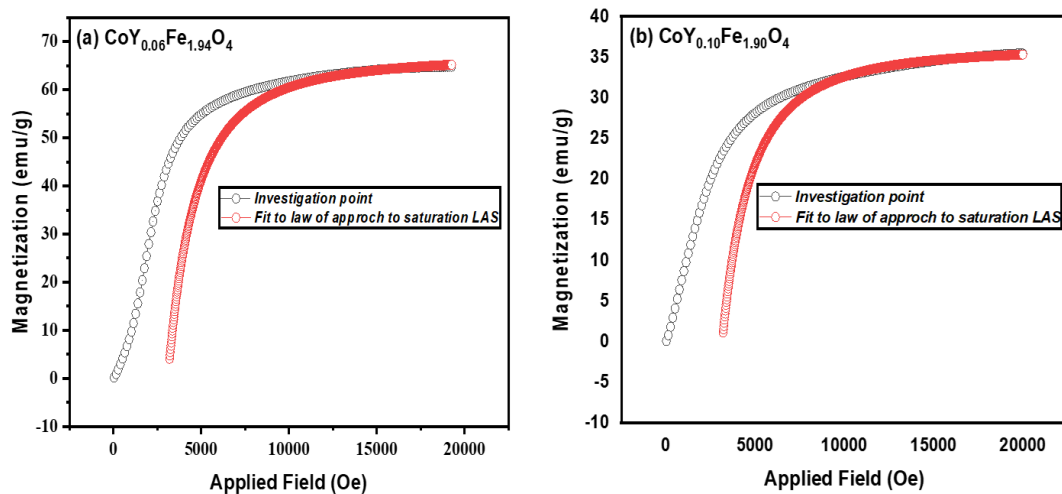


Figure 8: LA fit to Saturation for (a) (a) $\text{CoY}_{0.06}\text{Fe}_{1.94}\text{O}_4$ and (b) $\text{CoY}_{0.10}\text{Fe}_{1.90}\text{O}_4$ & Solid line represents the fitted LAS model.

The M_S exhibited a non-monotonic variation with Y^{3+} content. Starting from $M_S = 52.16$ emu/g for the unsubstituted sample ($x = 0$), the value increases to 55.82 emu/g for $x = 0.02$ and reaches a maximum of 66.82 emu/g for $x = 0.06$, before declining sharply to 36.22 emu/g for $x = 0.10$. The initial increase in M_S up to $x = 0.06$ may be attributed to favorable cation redistribution and enhanced A-B superexchange interactions. At higher Y^{3+} concentration ($x = 0.10$), the replacement of magnetic Fe^{3+} ions by non-magnetic Y^{3+} ions weakens the magnetic exchange interactions and promotes spin disorder, leading to a reduction in saturation magnetization. This non-monotonic behavior is physically significant and can be understood in terms of the

competing effects of cation redistribution and magnetic exchange dilution. At low substitution levels ($x = 0.02$ and 0.06), the incorporation of Y^{3+} ions into the octahedral (B) sublattice preferentially displaced Fe^{3+} ions from the B-sites to the tetrahedral (A) sites. This redistribution strengthens the net A-B superexchange interaction, as the magnetic moment difference between the A and B sublattices is effectively increased, leading to an enhancement in M_S [11], [14]. The LAS plots (Figures 8) confirm that the samples approach saturation at high applied fields, and the fitted curves are in excellent agreement with the experimental data, thus validating the extracted M_S values.

Table 3: Magnetic parameters of $\text{CoY}_x\text{Fe}_{2-x}\text{O}_4$ obtained from VSM measurements. M_S : saturation magnetization, M_r : remanent magnetization, H_C : coercive field, and K : magnetocrystalline anisotropy constant.

Sample	M_S (emu/g)	M_r (emu/g)	H_C (Oe)	K (10^6 erg/cm ³)
$\text{CoY}_{0.00}\text{Fe}_{2.00}\text{O}_4$	52.16	21	1450	2.33
$\text{CoY}_{0.02}\text{Fe}_{1.98}\text{O}_4$	55.82	19	1200	2.94
$\text{CoY}_{0.06}\text{Fe}_{1.94}\text{O}_4$	66.82	27	1174	3.88
$\text{CoY}_{0.10}\text{Fe}_{1.90}\text{O}_4$	36.22	13	1060	1.76

At a higher substitution ($x = 0.10$), the non-magnetic Y^{3+} ions (with a $4d^0$ electronic configuration and zero magnetic moment)

began to significantly dilute the magnetic exchange network. The substitution of magnetic Fe^{3+} ($5 \mu\text{B}$) by non-magnetic Y^{3+} (0

μB) at the B-sites reduces the total magnetic moment of the B sublattice, weakening the A-B superexchange interaction and thereby reducing M_S [23], [24]. Furthermore, at the highest substitution level, the increased lattice strain and the associated surface spin disorder on the smaller nanoparticles contributed to a magnetically disordered surface layer, which effectively reduced the net magnetization [25]. The sharp decrease in M_S from 66.82 emu/g ($x = 0.06$) to 36.22 emu/g ($x = 0.10$) reflects the dominance of exchange dilution and surface spin disorder effects over the cation redistribution mechanism at higher Y^{3+} concentrations.

The H_C decreased monotonically from 1450 Oe ($x = 0$) to 1060 Oe ($x = 0.10$), as shown in Table 3. This systematic reduction in coercivity with increasing Y^{3+} content is primarily attributed to the decrease in crystallite size observed in the XRD analysis. As the crystallite size decreased from 46 nm ($x = 0$, D_{WH}) to 34 nm ($x = 0.10$, D_{WH}), the domain structure of the nanoparticles evolved, and the coercivity was expected to decrease as the single-domain critical size was approached or crossed. In addition, the reduction in the magnetocrystalline anisotropy constant K from $2.33 \times 10^6 \text{ erg cm}^{-3}$ ($x = 0$) to $1.76 \times 10^6 \text{ erg cm}^{-3}$ ($x = 0.10$) contributes directly to the observed decrease in coercivity, since H_C is proportional to K/M_S in the Stoner–Wohlfarth model [11], [23]. The dilution of the Co^{2+} content at the B-sites by the non-magnetic Y^{3+} ions reduces the spin-orbit contribution to the magnetocrystalline anisotropy, further lowering H_C [24].

The M_r shows a non-monotonic trend like M_S , increasing from 21 emu/g ($x = 0$) to 27 emu/g ($x = 0.06$) and then decreasing to 13 emu/g ($x = 0.10$). The ratio M_r/M_S provides information about the domain structure and switching mechanism of nanoparticles. For the present series, M_r/M_S ranges from approximately 0.40 ($x = 0$) to 0.36 ($x = 0.10$), values that are consistent with a uniaxial anisotropy system and indicate that the particles are in

or near the single-domain regime [23]. The magnetocrystalline anisotropy constant K , derived from the LAS analysis, peaked at $x = 0.06$ ($K = 3.88 \times 10^6 \text{ erg/cm}^3$) and decreased to its minimum at $x = 0.10$ ($K = 1.76 \times 10^6 \text{ erg/cm}^3$), consistent with the trend in H_C and reflecting the progressive dilution of the Co^{2+} driven anisotropy with increasing Y^{3+} substitution.

4 Integrated Discussion: Structure-Morphology-Magnetic Property Correlations

The comprehensive characterization of the $\text{CoY}_x\text{Fe}_{2-x}\text{O}_4$ series by XRD, FTIR, FE-SEM, and VSM revealed a coherent and physically consistent picture of how Y^{3+} substitution modifies the crystal structure, microstructure, and magnetic properties of CFO. The structural modifications identified by XRD including lattice expansion, crystallite size reduction, and microstrain accumulation are the primary drivers of the observed changes in morphology and magnetic behavior.

The lattice expansion confirmed by XRD (Table 2) and the associated peak shift to lower 2θ angles (Figure 2) are direct consequences of the larger ionic radius of Y^{3+} relative to Fe^{3+} . This lattice distortion introduces microstrain (Table 1), which impedes grain growth during annealing and results in the progressive reduction in the crystallite size from 46 nm ($x = 0$) to 34 nm ($x = 0.10$) as determined by the WH method. The FESEM micrographs (Figure 6) provide direct visual evidence of this size reduction and the associated increase in porosity, which was further quantified by the XRD density measurements (Table 2). The FTIR spectra (Figure 5) confirm that the spinel structure is preserved across the entire substitution range and provide molecular-level evidence for the modification of the metal–oxygen bond environment by Y^{3+} incorporation, consistent with the lattice expansion observed by XRD [14],

[28].

The magnetic properties are intimately linked to the structural and microstructural changes. The non-monotonic variation of M_S with Y^{3+} content — increasing from $x = 0$ to $x = 0.06$ and then decreasing sharply at $x = 0.10$ - reflects the interplay between two competing mechanisms: (i) cation redistribution, in which Y^{3+} displaces Fe^{3+} from octahedral to tetrahedral sites, strengthening the A-B superexchange and increasing M_S at low substitution levels; and (ii) exchange dilution and surface spin disorder, which dominate at higher substitution levels and reduce M_S [11], [23]. The decreasing crystallite size, as confirmed by both XRD and FESEM, plays a critical role in this context: smaller nanoparticles have a higher surface-to-volume ratio, and the magnetically disordered surface spins constitute a larger fraction of the total magnetic moment, leading to a net reduction in M_S at $x = 0.10$ [25].

The monotonic decrease in H_C and K with increasing Y^{3+} content is directly linked to the structural changes identified by XRD and FESEM. The reduction in crystallite size brings the nanoparticles toward the superparamagnetic limit, reducing H_C , while the dilution of the Co^{2+} spin-orbit coupling contribution by the non-magnetic Y^{3+} ions directly reduces the magnetocrystalline anisotropy K [23], [24]. The increasing lattice strain and porosity, which are structural consequences of Y^{3+} substitution, further modulate the inter-particle exchange interactions and the effective anisotropy of the system [19].

The sample with $x = 0.06$ emerges as the composition with the most favorable combination of magnetic properties in this series, exhibiting the highest M_S (66.82 emu/g) and K (3.88×10^6 erg/cm³) while maintaining a moderately reduced H_C (1174 Oe). This composition represents an optimal balance between the beneficial effects of cation redistribution (which enhances M_S) and the detrimental effects of exchange dilution and surface spin disorder (which reduce M_S at higher substitutions). These find-

ings are of practical significance for the design of Y^{3+} substituted CFO materials for high-density magnetic recording, permanent magnet, and biomedical applications, where a high M_S combined with a tunable coercivity is desirable [11], [19], [23].

In summary, the integrated analysis of the XRD, FTIR, FESEM, and VSM data demonstrated that Y^{3+} substitution in CFO is an effective strategy for tailoring the structural and magnetic properties of this technologically important material. The sol-gel synthesis route employed in this work enables homogeneous substitution and precise control over the composition, microstructure, and magnetic behavior of the $CoY_xFe_{2-x}O_4$ series. The systematic correlations established between crystal structure, morphology, and magnetic properties provide a fundamental understanding that can guide the rational design of rare-earth-substituted CFO materials for advanced applications.

5 Conclusion

This study investigated the effect of low-level Y^{3+} substitution on the structural, microstructural, and magnetic properties of $CoY_xFe_{2-x}O_4$ ($x = 0.00, 0.02, 0.06, 0.10$) synthesized by a citric-acid-assisted sol-gel auto-combustion method and heat-treated at 600°C. XRD analysis confirmed the formation of a single-phase cubic spinel structure without any detectable secondary phases up to $x = 0.10$. The gradual shift of diffraction peaks toward lower angles and the increase in lattice parameter from 8.370 Å to 8.398 Å indicated lattice expansion due to the larger ionic radius of Y^{3+} . Substitution of Y^{3+} reduced crystallite size, increased micro strain, slightly lowered density, and increased porosity, which agreed well with FESEM observations of finer and more porous particle aggregates. FTIR results confirmed the preservation of the spinel structure, while small shifts in metal-oxygen vibration bands suggested local bond distortion caused by Y^{3+} incor-

poration. All samples exhibited ferrimagnetic behaviour at room temperature. The Saturation magnetization showed a non-monotonic trend, increasing from 52.16 emu g⁻¹ for the undoped sample to a maximum of 66.82 emu g⁻¹ at $x = 0.06$, followed by a decrease at higher Y contents. This behaviour is attributed to favourable cation redistribution and enhanced A-B superexchange interactions at low Y concentrations, whereas higher substitution levels caused exchange dilution and increase surface spin disorder. Coercivity and magnetic anisotropy decreased continuously with increasing Y content, indicating reduced anisotropy and particle-size-related domain evolution. Overall, Y³⁺ substitution effectively modified the lattice structure, particle characteristics, and magnetic properties of the CFO. The composition $x=0.06$ provided the best balance of enhanced magnetization and controllable coercivity. These results demonstrate that yttrium-substituted CFO is a structurally stable and tenable material with strong potential for advanced magnetic and multifunctional ferrite-based nanodevices.

References

- [1] M. Nidhin, S. S. Nazeer, R. S. Jayasree, M. S. Kiran, B. U. Nair, and K. J. Sreeram, "Flower shaped assembly of cobalt ferrite nanoparticles: application as T2 contrast agent in MRI," *RSC Adv.*, vol. 3, no. 19, p. 6906, Jan. 2013. doi:10.1039/c3ra23232h.
- [2] T. Shiroka, "Fundamentals and applications of magnetic materials," *Contemporary Physics*, vol. 64, no. 1, pp. 83–84, Jan. 2023. doi:10.1080/00107514.2023.2207871.
- [3] K. Freindl, G. Soria, J. Prieto, N. Spiridis, A. Quesada, and J. Figuera, "Ultrathin CoFe₂O₄ films grown by molecular beam epitaxy on Pt(111)," in *Underline Science Inc.*, Mar. 2021. doi:10.48448/jtzq-q231.
- [4] A. Tiwari, "Magnetic materials and devices: Research and applications," *JOM*, vol. 63, no. 10, p. 24, Oct. 2011. doi:10.1007/s11837-011-0168-1.
- [5] H. P. Rooksby and B. T. M. Willis, "Crystal Structure and Magnetic Properties of Cobalt Ferrite at Low Temperatures," *Nature*, vol. 172, no. 4388, pp. 1054–1055, Dec. 1953. doi:10.1038/1721054c0.
- [6] A. Arrott and H. Sato, "Transitions from Ferromagnetism to Antiferromagnetism in Iron-Aluminum Alloys. Experimental Results," *Phys. Rev.*, vol. 114, no. 6, pp. 1420–1426, June 1959. doi:10.1103/physrev.114.1420.
- [7] N. Abinaya, M. C. Robert, N. Srinivasan, and S. Saravanakumar, "Electron density mapping and bonding in Mn doped CoFe₂O₄ using XRD, and its correlation with room temperature optical and magnetic properties," *Journal of Magnetism and Magnetic Materials*, vol. 580, p. 170938, June 2023. doi:10.1016/j.jmmm.2023.170938.
- [8] G. Raju et al., "Effect of chromium substitution on the structural and magnetic properties of cobalt ferrite," *Materials Science for Energy Technologies*, vol. 2, no. 1, pp. 78–82, Nov. 2018. doi:10.1016/j.mset.2018.11.001.
- [9] H. Mamiya, N. Terada, H. Kitazawa, A. Hoshikawa, and T. Ishigaki, "Structural and Magnetic Properties of Dilute Spinel Ferrites: Neutron Diffractometry and Magnetometry Investigations," *Journal of Magnetism*, vol. 16, no. 2, pp. 134–139, June 2011. doi:10.4283/jmag.2011.16.2.134.
- [10] J. Kapoor et al., "Magnetism, Electronic Structure and Lattice Sites of Fe Ions Implanted into Yttrium," *Europhys. Lett.*, vol. 24, no. 4, pp. 299–304, Nov. 1993. doi:10.1209/0295-5075/24/4/011.
- [11] I. Haïk Dunn, S. E. Jacobo, and P. G. Bercoff, "Structural and magnetic influence of yttrium-for-iron substitution in cobalt ferrite," *Journal of Alloys and Compounds*, vol. 691, pp. 130–137, Aug. 2016. doi:10.1016/j.jallcom.2016.08.223.
- [12] S. Kumari, V. Kumar, P. Kumar, M. Kar, and L. Kumar, "Structural and magnetic properties of nanocrystalline yttrium substituted cobalt ferrite synthesized by the citrate precursor technique," *Advanced Powder Technology*, vol. 26, no. 1, pp. 213–223, Oct. 2014. doi:10.1016/j.appt.2014.10.002.
- [13] S. Chakrabarty, A. Dutta, and M. Pal, "Effect of yttrium doping on structure, magnetic and electrical properties of nanocrystalline cobalt ferrite," *Journal of Magnetism and Magnetic Materials*, vol. 461, pp. 69–75, Apr. 2018. doi:10.1016/j.jmmm.2018.04.051.

- [14] D. T. Rahardjo, S. Budiawanti, S. Suharno, R. Suryana, A. Supriyanto, and B. Purnama, "Decreasing of Magnetic Saturation of Yttrium Doped Cobalt Ferrite Prepared by the Sol-Gel Auto-Combustion," *Trends Sci*, vol. 21, no. 4, p. 7320, Feb. 2024. doi:10.48048/tis.2024.7320.
- [15] D. M. Ghone, K. K. Patankar, V. L. Mathe, and S. D. Kaushik, "Influence of substitution of yttrium in cobalt ferrite on the structural, magnetic and magnetostrictive properties," in *AIP conference proceedings*, Cornell University, Jan. 2018, p. 130026. doi:10.1063/1.5029096.
- [16] J. V. Devkar, K. K. Patankar, D. M. Ghone, and V. L. Mathe, "Investigations of yttrium-doped cobalt-zinc ferrite as potential material for transducer application," *emergent mater.*, vol. 4, no. 6, pp. 1725–1733, Apr. 2021. doi:10.1007/s42247-021-00209-2.
- [17] M. Ahmed et al., "Structural and magnetic properties of $\text{Co}_{0.85}\text{Zn}_{0.15}\text{Y}_x\text{Fe}_{2-x}\text{O}_4$ ferrites," *Physica B: Condensed Matter*, vol. 645, p. 414267, Aug. 2022. doi:10.1016/j.physb.2022.414267.
- [18] M. A. Hossain, M. D. Hossain, M. N. I. Khan, S. S. Sikder, and M. A. Hakim, "Frequency and temperature dependent magnetic properties with structural Rietveld refinement of $\text{Co}_{0.25}\text{Zn}_{0.75}\text{Y}_x\text{Fe}_{2-x}\text{O}_4$ ferrites," *Journal of Magnetism and Magnetic Materials*, vol. 493, p. 165696, Aug. 2019. doi:10.1016/j.jmmm.2019.165696.
- [19] M. Das, M. N. I. Khan, M. A. Matin, and M. M. Uddin, "Structural, Morphological, Electrical and Magnetic Properties of Yttrium-Substituted Co-Zn Ferrites Synthesized by Double Sintering Technique," *J Supercond Nov Magn*, vol. 32, no. 11, pp. 3569–3577, May 2019. doi:10.1007/s10948-019-5104-6.
- [20] S. B. Das, R. K. Singh, V. Kumar, N. Kumar, P. Singh, and N. Kumar Naik, "Structural, magnetic, optical and ferroelectric properties of Y^{3+} substituted cobalt ferrite nanomaterials prepared by a cost-effective sol-gel route," *Materials Science in Semiconductor Processing*, vol. 145, p. 106632, Mar. 2022. doi:10.1016/j.mssp.2022.106632.
- [21] S. S. Satpute, S. R. Wadgane, K. Desai, D. R. Mane, and R. H. Kadam, "Substitution effect of Y^{3+} ions on the structural, magnetic and electrical properties of cobalt ferrite nanoparticles," *Cerâmica*, vol. 66, no. 377, pp. 43–49, Mar. 2020. doi:10.1590/0366-69132020663772734.
- [22] A. Kumar et al., "Structural and gas sensing properties of sol-gel synthesized Zn-doped nickel ferrite nanoparticles," *J Mater Sci: Mater Electron*, vol. 37, no. 4, Feb. 2026. doi:10.1007/s10854-026-16702-5.
- [23] T. E. P. Alves, H. V. S. Pessoni, and A. Franco Jr, "The effect of Y^{3+} substitution on the structural, optical band-gap, and magnetic properties of cobalt ferrite nanoparticles," *Phys. Chem. Chem. Phys.*, vol. 19, no. 25, pp. 16395–16405, Jan. 2017. doi:10.1039/c7cp02167d.
- [24] X. Wu, Z. Ding, N. Song, L. Li, and W. Wang, "Effect of the rare-earth substitution on the structural, magnetic and adsorption properties in cobalt ferrite nanoparticles," *Ceramics International*, vol. 42, no. 3, pp. 4246–4255, Dec. 2015. doi:10.1016/j.ceramint.2015.11.100.
- [25] T. K. C. Nguyen and A. T. Nguyen, "Structural, optical and magnetic properties of Y-doped CoFe_2O_4 nanoparticles prepared by a simple co-precipitation method," *J Mater Sci: Mater Electron*, vol. 34, no. 5, Feb. 2023, doi:10.1007/s10854-023-09914-6.
- [26] E. C. M. Pennings and W. Grellner, "Precise Nondestructive Determination of the Density of Porous Ceramics," *Journal of the American Ceramic Society*, vol. 72, no. 7, pp. 1268–1270, July 1989. doi:10.1111/j.1151-2916.1989.tb09724.x.
- [27] S. M. Asgarian, S. Pourmasoud, Z. Kargar, A. Sobhani-Nasab, and M. Eghbali-Arani, "Investigation of positron annihilation lifetime and magnetic properties of $\text{Co}_{1-x}\text{Cu}_x\text{Fe}_2\text{O}_4$ nanoparticles," *Mater. Res. Express*, vol. 6, no. 1, p. 015023, Oct. 2018. doi:10.1088/2053-1591/aae55d.
- [28] N. Khatun et al., "Influence of Y^{3+} and La^{3+} ions on the structural, magnetic, electrical, and optical properties of cobalt ferrite nanoparticles," *Heliyon*, vol. 9, no. 2, p. e13019, Jan. 2023. doi:10.1016/j.heliyon.2023.e13019.