Enhancement of functional properties of V\textsubscript{0.6}Ti\textsubscript{0.4} alloy superconductor by the addition of yttrium

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We show here that the yttrium is immiscible and precipitates with various sizes in the body centred cubic V\textsubscript{0.6}Ti\textsubscript{0.4} alloy superconductor. The number and size of the precipitates are found to depend on the amount of yttrium added. Precipitates with various sizes up to 30 \(\mu\)m are found in the V\textsubscript{0.6}Ti\textsubscript{0.4} alloy containing 5 at.% yttrium. The large amount of line disorders generated by the addition of yttrium in this alloy are found to be effective in pinning the magnetic flux lines. While the superconducting transition temperature increases with the increasing amount of yttrium in the V\textsubscript{0.6}Ti\textsubscript{0.4} alloy, the critical current density is maximum for the alloy containing 2 at.\% yttrium, where it is more than 7.5 times the parent alloy in fields higher than 1 T. We found that the effectiveness of each type of defect in pinning the flux lines is dependent on the temperature and the applied magnetic field.

I. INTRODUCTION

The V\textsubscript{1-x}Ti\textsubscript{x} alloys are promising materials as an alternate to the Nb based superconductors for high field applications\textsuperscript{[15]} especially in the neutron radiation environment\textsuperscript{[9–10]}. Moreover, the V\textsubscript{1-x}Ti\textsubscript{x} alloys are highly machinable and ductile\textsuperscript{[11,12]}. However, the critical current density (\(J_C\)) of the V\textsubscript{1-x}Ti\textsubscript{x} alloys is about 10\(^7\)A/m\(^2\) at 4 K, which is two orders of magnitude less than the commercially available Nb-based superconductors.\textsuperscript{[2]} Previous attempts to increase the \(J_C\) of these alloys by the introduction of defects through the addition of transition and non-transition elements were ineffective.\textsuperscript{[13]} Recently, we have shown that the polycrystalline V\textsubscript{1-x}Ti\textsubscript{x} alloys form with large grains having sizes ranging from few \(\mu\)m to few millimetres.\textsuperscript{[23]} We have established that the low grain boundary density and the presence of flux flow channels in the V\textsubscript{1-x}Ti\textsubscript{x} alloys are the main reasons for the low \(J_C\) of these alloys.\textsuperscript{[13]}

The rare-earth elements are found to be immiscible in vanadium and titanium.\textsuperscript{[15,21]} Buschow showed by estimating the energy of formation that no binary compound containing rare earth and titanium or vanadium will form.\textsuperscript{[15]} The solubility of rare earths in the liquid vanadium or titanium is limited to very low concentrations (< 1 %).\textsuperscript{[16–19,20]} We have used this property to introduce a large number of pinning centres by adding gadolinium in the V\textsubscript{0.6}Ti\textsubscript{0.4} alloy, which resulted in the enhancement of the \(J_C\) by about 20 times.\textsuperscript{[2]} However, the gadolinium precipitates order ferromagnetically\textsuperscript{[2]}, which seem to hinder the attempts to improve the \(J_C\) further.

In this direction, here we present a detailed study on the yttrium (non-magnetic) containing V\textsubscript{0.6}Ti\textsubscript{0.4} alloys and establish a correlation between the microstructure and the physical properties in the normal and superconducting states. We show that there is an enhancement of \(J_C\) of V\textsubscript{0.6}Ti\textsubscript{0.4} by about 7.5 times in fields higher than 1 T when 2 at.% yttrium is added. Our analysis shows how different defects contribute to the enhancement of \(J_C\) of the (V\textsubscript{0.6}Ti\textsubscript{0.4})-Y alloys at different temperatures and field regimes.

\begin{table}[h]
\centering
\begin{tabular}{|c|c|c|c|}
\hline
Name of the Sample & V & Ti & Y \\
\hline
Y0 & 60 & 40 & - \\
Y1 & 59 & 40 & 1 \\
Y2 & 58 & 40 & 2 \\
Y3 & 57 & 40 & 3 \\
Y5 & 55 & 40 & 5 \\
\hline
\end{tabular}
\caption{Sample name and the at.% of elements in the samples}
\end{table}

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II. EXPERIMENTAL

A series of samples were synthesized by arc melting\(^3\) the constituent elements (purity better than \(99.9\%\)) in \(99.999\%\) pure Ar atmosphere. The elemental compositions (in at.\%) of the samples are presented in Table I. The samples were cut using diamond saw. The details of metallography experiments can be found elsewhere.\(^2\) Images of the etched samples were taken using a high power optical microscope (Leica DMI 5000M). The elemental analysis was done using the energy dispersive analysis of x-rays (EDAX) setup attached to the scanning electron microscope (SEM, Carl Zeiss, Germany). X-Ray diffraction (XRD) measurements were performed using \(\lambda = 0.817\ \text{Å}\) radiation from the BL12 beamline of the Indus-2 synchrotron facility.\(^2\) The resistivity and heat capacity of the samples were measured using the 9 T Physical Property Measurement System (PPMS, Quantum Design, USA). For resistivity measurements, the sample sizes were about \(8\ \text{mm} \times 1\ \text{mm} \times 0.5\ \text{mm}\). The magnetization was measured using Superconducting Quantum Interference Device based Vibrating Sample Magnetometer (MPMS-3 SQUID-VSM, Quantum Design, USA). The samples used for heat capacity and magnetization measurements have sizes about \(3\ \text{mm} \times 1\ \text{mm} \times 1\ \text{mm}\).

III. RESULTS AND DISCUSSION

A. Microstructural characterization

Figure 1 shows the SEM images of the \((V_{0.6}Tl_{0.4})-Y\) alloys up to 5 at.% of yttrium. These images were taken before etching the samples. Precipitation of secondary phase (white patches) were observed in all the yttrium containing samples. Studies by Love\(^1\) as well as by Komjathy and coworkers\(^1\) on the vanadium-rare earth and titanium-rare earth binaries reveal that the rare earth elements precipitate in the vanadium or titanium matrix. The elemental analysis of a portion of Y3 is shown in Fig. 2. The white precipitates in Fig. 2 are rich in yttrium. The titanium and vanadium are uniformly distributed as a matrix. From Fig. 2 we see that the precipitation follows certain pattern in the Y1 and Y2 samples. This pattern is partially lost in the Y3 sample, whereas the precipitates are more uniformly distributed in the Y5 sample. The size of the precipitates are presented in Table II. While, the precipitate size in the Y1 and Y2 samples are more or less uniform, there is a distribution of precipitate sizes in the Y3 and Y5 samples. Few precipitates in Y5 are found to be bigger than 30 \(\mu\text{m}\) in size.

The optical metallography images in Fig. 3 of the \((V_{0.6}Tl_{0.4})-Y\) alloys reveal the microstructure. These images were taken after etching the polished sample surfaces by a solution containing water, HF and HNO\(_3\) in the 98:1:1 volume ratio. Dendritic growth is observed in all the alloys containing yttrium indicating a spatial compositional variation. The grain size of the \(V_{0.6}Tl_{0.4}\) alloy is about 200-300 \(\mu\text{m}\). The dendritic cell size in Y1, Y2, Y3 and Y5 are 31±9, 24±7, 32±10 and 43±13 microns respectively. Further studies are required to establish whether these cells are different grains. Nevertheless, thin layers of an alloy which has the lowest melting point, solidifies in between these cells. The cell size
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FIG. 3. (a–e) Optical metallography images of the polished (V₀.₆Ti₀.₄)-Y alloys after etching. Dendritic growth in alloys containing yttrium indicates the presence of large amount of disorders. The average dendritic cell size reduces initially with increasing yttrium content, but increases for 3 at.% yttrium or higher.

reduces initially with increasing yttrium content. This indicates that these yttrium-rich precipitate hinders the cell growth due to the presence of strain field. Thus, there is an enhancement of cell boundary density in the (V₀.₆Ti₀.₄)-Y alloys up to 2 at.% yttrium addition and yttrium is precipitated only along the cell boundaries. The cell size is found to increase with further addition of yttrium. The precipitates in these alloys are of larger size and this reduces the number density of precipitates and the strain field in these alloy compositions.

The vanadium alloys containing small amount of yttrium (< 0.5 at.%) undergo monotectic transition at about Tₘₜ = 2173 K (1900 °C) from homogeneous liquid above Tₘₜ to solid β-vanadium phase and a yttrium-rich liquid below Tₘₜ. The enrichment of yttrium in this liquid leads to phase separation of yttrium-rich and vanadium-rich liquids. If the enrichment of yttrium in the yttrium-rich liquid reaches 96 at.%, an eutectic transition at about 1730 K (≈ 1460 °C) occurs within this phase. On the other hand, the titanium alloys containing small amount of yttrium (< 20 at.%) undergoes eutectic transition at about Tₑₜ = 1673 K (1400 °C). The melting point (Tₘ) of yttrium is about 1793 K (1520 °C). Our metallography results match closely with those of the V-Y alloys indicating that the phase diagram of the (V₀.₆Ti₀.₄)-Y alloys must be similar to the V-Y phase diagram. On comparing the literature available on the vanadium/titanium-rare earth binary phase diagrams and our experimental observations, we present a schematic phase diagram for the dilute (V₀.₆Ti₀.₄)-Y alloys in Fig. 4.

The small size of the precipitates in the alloys containing 3 at.% or less yttrium indicates that these alloys, while cooling from the melt undergo monotectic transition from the homogeneous V-Ti-Y liquid to solid β-V-Ti alloy and a yttrium-rich liquid. Figure 5 shows the typical compositions of the precipitates and matrix.

FIG. 4. Schematics of the phase diagram of the pseudo-binary dilute (V₀.₆Ti₀.₄)-Y alloys. At low yttrium content, the homogeneous liquid formed at high temperatures phase-separates into a yttrium-rich liquid and solid V₀.₆Ti₀.₄ when cooled below 1935 K. This results in the formation of fine yttrium-rich precipitates below 1430 K.

TABLE II. Average sizes of the precipitates

| Sample | Precipitate size (µm) |
|--------|-----------------------|
| Y₁     | 1.3 ± 0.6             |
| Y₂     | 2.3 ± 1               |
| Y₃     | 2.8 ± 1.8             |
| Y₅     | 4.7 ± 3.3             |
of the Y2 alloy. The composition of the matrix is about V_{0.626}Ti_{0.374} with a minor local variation of the composition. The oxygen content within a precipitate varies substantially among the precipitates. We found that higher the oxygen content, higher is the amount of titanium and vanadium present in the precipitate. In the cases where oxygen is absent in the precipitates, the composition is close to 95 at. % of yttrium. This indicates that eutectic microstructure may be present within the yttrium-rich precipitates. In order to get a clear picture of the different phases present in the samples, we show in Fig. 6, the x-ray diffraction (XRD) pattern of the Y2 alloy. The symbols ‘*’, ’#’ and ‘+’ represent reflections from the β-V_{0.60}Ti_{0.40}, yttrium precipitates and Y_{2}O_{3} phases respectively. The lattice parameter of β-V_{0.60}Ti_{0.40} is about 3.1412 Å while that of Y_{2}O_{3} is about 10.4 Å. The lattice parameters of the yttrium precipitates are about a = 3.647 Å and c = 5.728 Å. The lattice parameters of β-V_{0.60}Ti_{0.40} and yttrium precipitates are in agreement with literature. On the other hand, the lattice parameter of Y_{2}O_{3} is slightly less than that of bulk which may be due to oxygen off-stochiometry. Thus, the reduction of the cell size in these compositions is related to the formation of fine yttrium-rich precipitates during the phase separation of the homogeneous liquid into solid β-V_{0.60}Ti_{0.40} and L-Y phases (Fig. 4). On the other hand, a two phase microstructure seen in between the large V-Ti cells in the Y5 sample (Fig. 3(e)) is caused by the liquid immiscibility.

Earlier, we have shown that the V_{1-x}Ti_{x} alloys have large grain sizes of the the order of few microns to few millimetres. Our microstructural studies reveal that the yttrium can introduce large amount of defects in the V_{1-x}Ti_{x} alloys which is helpful in improving the critical current density in the mixed state of these superconducting alloys. Therefore, we characterize the present alloys in the superconducting as well as normal states.

B. Electrical and thermal properties of the (V_{0.6}Ti_{0.4})-Y alloys and the superconducting transition temperature (T_{C})

Figure 7 shows the temperature dependence of electrical resistivity (ρ(T)) of the (V_{0.6}Ti_{0.4})-Y alloys in the range 2-300 K. The yttrium containing alloys have higher ρ(T) in comparison with that of the parent V_{0.6}Ti_{0.4} alloy. Residual resistivity (ρ_{0}) increases up to 3 at. % yttrium due to the increased static defects (precipitates, grain/cell boundaries, dislocations and point defects). The reduced number density of defects in the Y5 alloy due to the larger size of the precipitates results in the lower ρ_{0} as compared to the other yttrium containing alloys. The inset to Fig. 7 shows the expanded view of resistivity around the T_{C}. The T_{C} is obtained as that temperature at which the temperature derivative of resistivity shows a peak. The T_{C} increases from 7.68 K for the V_{0.6}Ti_{0.4} alloy to 7.85 K for the Y5 sample.

Normally, the T_{C} is expected to decrease with increasing disorder. However, the increase in the T_{C} in the present alloys can be due to (i) suppression of spin fluctuations by the disorder (ii) increase in the electron-phonon coupling due to phonon softening by the defects and/or (iii) removal of the trace oxygen by yttrium from the V-Ti matrix.

The T_{C} of the V_{1-y}Ti_{y} alloys is limited by the spin fluctuations and increases from 5.4 K for y = 0 to about 7.68 K for the y = 0.4 alloy. We have argued that the suppression of spin fluctuation by the disorder introduced when vanadium is alloyed with titanium increases the T_{C} of the V_{1-y}Ti_{y} alloys. Palladium is also found to be

FIG. 5. Compositional analysis of yttrium-rich precipitates and V_{0.60}Ti_{0.40} matrix using EDAX. Presence of oxygen is found in many of the yttrium-rich precipitates. No trace of yttrium is found in the matrix.

FIG. 6. X-ray diffraction pattern of the Y2 alloy (a) The major peaks are indexed to β- V_{0.60}Ti_{0.40} phase. (b) Weak reflections corresponding to α-Y and Y_{2}O_{3} phases are also seen.
superconducting when certain type of disorder is introduced to suppress the spin fluctuations. Thus, the increase in the $T_C$ of the yttrium containing alloys can be attributed to the suppression of spin fluctuations with increase in disorder. However, this should result in the reduction of the normal state $\rho(T)$ which is contrary to the observed result that the yttrium containing alloys have higher normal state $\rho(T)$ as compared to the parent V$_{0.6}$Ti$_{0.4}$ alloy. The normal state $\rho(T)$ in the range 20-120 K is fitted using the equation:

$$\rho(T) = \rho_0 + \rho_{sf} \left( \frac{T}{T_{sf}} \right)^2 F_2 \left( \frac{T_{sf}}{T} \right) - \rho_{sf} \left( \frac{T}{T_{sf}} \right)^5 F_5 \left( \frac{T_{sf}}{T} \right) + \rho_{sd} \left( \frac{T}{\theta_D} \right)^3 F_3 \left( \frac{\theta_D}{T} \right)$$

(1)

where $\rho_0$ is residual resistivity, $\rho_{sf}$ is the coefficient of resistivity for spin-fluctuation, $\rho_{sd}$ is the coefficient of resistivity corresponding to interband scattering, $T_{sf}$ is the spin fluctuation temperature, $\theta_D$ is the Debye temperature and $F_k(M)$ is the Fermi-integral given by

$$F_k(M) = \int_0^M \frac{dz}{z^k \exp(z) - 1}$$

(2)

The fitting is shown in Fig. 7 and coefficients obtained by the fitting are presented in the Table III. The errors in the estimation of parameters are within 10%. We found that $\rho_{sf}$ increases with increasing yttrium content. The large $\theta_D$ and small $\rho_{sd}$ for Y3 and Y5 alloys indicate a negligible contribution from the interband scattering to resistivity in these alloys. The $T_{sf}$ also increases with yttrium content up to 3 at. %. Therefore, the enhancement of $T_C$ in the yttrium containing alloys is not due to the suppression of spin fluctuations.

TABLE III. Parameters obtained from the fitting of resistivity

| Sample | $\rho_0$ (\(\mu\Omega\ cm\)) | $\rho_{sf}$ (\(\mu\Omega\ cm\)) | $\rho_{sd}$ (\(\mu\Omega\ cm\)) | $T_{sf}$ (K) | $T_D$ (K) |
|--------|----------------|----------------|----------------|-------------|----------|
| Y0     | 54.6           | 9.1            | 25.8           | 196         | 300      |
| Y1     | 56.9           | 18.0           | 14.0           | 198         | 366      |
| Y2     | 56.9           | 19.6           | 12.3           | 204         | 382      |
| Y3     | 57.7           | 30.4           | 3.4            | 308         | 561      |
| Y5     | 55.7           | 26.1           | 7.9            | 217         | 433      |

The higher $\rho(T)$ with a different curvature for the yttrium containing alloys as compared to that of the parent V$_{0.6}$Ti$_{0.4}$ alloy hints at the enhancement of electron-phonon coupling due to the softening of phonon modes by disorder. To verify the correlation between the $T_C$ and phonon softening, we have measured the heat capacity of the yttrium containing alloys in zero and 8 T fields (Fig. 5(a)). The normal state $C/T$ vs $T^2$ data in 8 T in the range 5-10 K is used to fit a straight line to obtain the Sommerfeld coefficient of electronic heat capacity ($\gamma$) and the Debye temperature ($\theta_D$) using the relation $C = \gamma T + \beta T^3$ where $\theta_D^3 = 1943.66/\beta$. The maximum change in the $\theta_D$, $\gamma$ (Fig. 8(b)) and $T_C$ with the yttrium content are about 6.5%, 4%, and 2.3% respectively. Since, change in $T_C$ is quite smaller in comparison with the other two variables, it is unlikely that the enhancement of the $T_C$ is due to changes in the electron-phonon coupling by the addition of yttrium.

The $T_C$ of vanadium is about 5.4 K. The presence of oxygen suppresses the $T_C$ of vanadium. It is well known that the addition of 0.5-2 at.% yttrium in titanium or vanadium improves the ductility. This improvement is caused by the scavenger effect of yttrium in removing oxygen from the grain boundaries of vanadium and titanium. Therefore, we infer that the addition of yttrium to the V$_{1-x}$Ti$_x$ alloys improves the $T_C$ by removing oxygen from the matrix. Our studies on the (V$_{0.6}$Ti$_{0.4}$)$_{50}$Y$_{50}$ alloy showed that the superconductivity is induced in yttrium-rich phase by the proximity effect. Therefore, we infer that the region of yttrium-rich precipitates where the oxygen is absent become su-

FIG. 7. Temperature dependence of electrical resistivity of the (V$_{0.6}$Ti$_{0.4}$)-Y alloys. The residual resistivity and critical temperature increases with yttrium addition. Open symbols are the experimental data points and the solid lines are the fits using eq.(1). The parameters of fitting are presented in table III. The inset shows expanded view of the resistivity around the superconducting transition.
Critical current density in V-Ti-Y alloys

perconducting below $T_C$ due to proximity effect and the boundaries between the precipitates and the matrix can act as effective pinning centres.

C. Magnetic properties of the (V$_{0.6}$Ti$_{0.4}$)-Y alloys: Role of microstructure on the enhancement of critical current density

The temperature dependence of magnetization ($M(T)$) of the Y2 and Y5 alloys are shown in Fig. 9. The $M(T)$ is measured in the presence of 10 mT field while warming up after cooling down the sample from $T > T_C$ to 2 K in zero field (ZFC), while cooling down in the same field after warming above $T_C$ (FCC), and then again while warming up in the same field (FCW). The $T_C$ is estimated as that temperature at which the $M(T)$ starts to decrease towards negative values when the temperature is decreased from 10 K. The $T_C$ estimated from $M(T)$ is in agreement with that estimated from the resistivity measurements. The insets to the Fig. 9 show the expanded view of $M(T)$ measured during FCC and FCW cycles. The Meissner fraction ($M_f$) is estimated as the ratio $M_{FCC}/M_{ZFC}$ at 2 K. The $M_f$ is about 0.095% for the V$_{0.6}$Ti$_{0.4}$ alloy which decreases with increasing amount of yttrium in this alloy. The $M_{FCC}/M_{ZFC}$ is about 0.032% for the Y3 alloy indicating that the magnetic flux line pinning improves when yttrium is added to the V$_{0.6}$Ti$_{0.4}$ alloy. This indicates that the yttrium containing alloys have higher $J_C$ than the parent V$_{0.6}$Ti$_{0.4}$ alloy.

In order to quantify the the enhancement of $J_C$ in the yttrium containing alloys, we have measured the field dependence of magnetisation ($M(H)$) for all the alloys at different temperatures. The Fig. 10(a) shows the $M(H)$ at 4 K for all the alloys. The size of the hysteresis increases with increasing yttrium content in the V$_{0.6}$Ti$_{0.4}$ alloy up to 2 at.%, and then it starts shrinking. Figure 10(b) shows the $M(H)$ of the Y2 alloy at different temperatures. The hysteresis in the $M(H)$ is symmetric along the $H$ axis indicating that the Bean-Livingston surface barrier effect is negligible in Y2 alloy. The upper critical field ($H_{C2}$) and magnetic irreversibility field ($H_{irr}$) at various temperatures below $T_C$ are estimated from the isothermal magnetisation curves. The magnetic field at which $M(H)$ deviates from its behaviour in the normal state is taken as the $H_{C2}$. The magnetic field at which the $M(H)$ for increasing $H$ bifurcates from that during the $H$ decreasing cycle is taken as the $H_{irr}$.

The $H_{C2}$ and $H_{irr}$ as a function of temperature for all the alloys are plotted in Fig. 11. The $H_{C2}(T)$ is almost the same for all the yttrium containing alloys and $H_{irr} < H_{C2}$. Addition of yttrium to V$_{0.6}$Ti$_{0.4}$ alloy increases the $H_{irr}$. The solid line in Fig. 11 represents the fit to the $H_{C2}(T)$ using the Werthamer-
Helfand-Hohenberg (WHH) formalism for dirty limit superconductors. \[ \lambda_{SO} \] From the fitting, the parameters \( \alpha \) (corresponding to the Pauli paramagnetic effect) and \( \lambda_{SO} \) (which is the measure of strength of spin-orbit interaction) are found to be 1.49 and 2.5 respectively. The \( H_{C2} \) in the limit of absolute zero (\( H_{C2}(0) \)) is found to be about 13.2 T which is comparable to the Nb-Ti alloys.

The critical current density \( J_C \) of the \( (V_{0.6}Ti_{0.4}) \)-Y alloys estimated from the \( M(H) \) is shown in Fig. 10. The \( J_C \) is estimated using the Bean’s critical state model.

\[
J_C = 2\Delta M \left[ a \left( 1 - \frac{a}{3b} \right) \right]^{-1}.
\]  

Here, the \( \Delta M \) at every \( H \) is the difference between the \( M \) measured during increasing and decreasing \( H \) cycles. The parameters \( a \) and \( b \) \( (b > a) \) are the dimensions of the rectangular cross section of the sample in the direction normal to the applied magnetic field.

The zero field \( J_C \) value at 2 K for the \( V_{0.6}Ti_{0.4} \) alloy is estimated to be \( 2 \times 10^{6} \) A/m², which is in close agreement with the literature. The \( J_C \) increases significantly with the addition of yttrium up to 2 at.\%, and then it decreases with further addition. The \( Y2 \) alloy has a \( J_C \) of \( 7 \times 10^{6} \) A/m² at 4 K and 1 T field, which is an order of magnitude higher than that of the \( V_{0.6}Ti_{0.4} \) alloy. The drop in \( J_C \) with the application of low magnetic fields is found to be less steep for all the yttrium containing alloys in comparison with the parent \( V_{0.6}Ti_{0.4} \) alloy. The \( J_C \) of the parent \( V_{0.6}Ti_{0.4} \) alloy exists only up to 5 T at 4 K. On the other hand, the yttrium containing alloys have significant \( J_C \) above 5 T at 4 K. The value of \( J_C \) as well as the range over which a significant \( J_C \) is present increase with the increasing yttrium content up to 2 at.\%. The \( J_C \) at 2 K (Fig. 12) exceeds \( 2 \times 10^{7} \) Am⁻² at 7 T for the \( Y2 \) alloy. The increase in the defects when yttrium is added to the \( V_{0.6}Ti_{0.4} \) alloy (section 3.1) results in the enhancement of flux line pinning, which in turn increases the \( J_C \). Though the enhancement of \( J_C \) by adding yttrium in \( V_{0.6}Ti_{0.4} \) alloy is significant, the \( J_C \) is still considerably lower than that of commercial Nb-Ti alloys.

In order to establish a correlation between the nature of the defects and the enhancement of \( J_C \), we have estimated the pinning force density \( (F_P) \) at 4 K as \( F_P = J_C \times H \) and the same is plotted in Fig. 13 (a). In the \( V_{1-y}Ti_y \) alloys, the grain boundaries are the major pinning centres in the low-field regime, whereas the point defects and dislocations are the effective pinning centres in the high-field regime. The maximum \( F_P \) of the \( V_{0.6}Ti_{0.4} \) alloy at 4 K is in the range of \( 10^{7} \) Nm⁻³. Addition of yttrium increases this value up to 2 at.\%. The maximum \( F_P \) of the \( Y2 \) alloy at 4 K is about \( 7.6 \times 10^{7} \) Nm⁻³ which is about 7.6 times that of the parent \( V_{0.6}Ti_{0.4} \) alloy. Significant pinning strength above 5 T is found in the yttrium containing alloys. In all the samples at 4 K, the \( F_P \) increases sharply in very low fields and falls of gradually in high magnetic fields.

FIG. 10. (a) Magnetic field dependence of magnetisation for the \( (V_{0.6}Ti_{0.4}) \)-Y alloys at 4 K. The hysteresis is observed to increase with yttrium addition and is maximum for the \( Y2 \) alloy. (b) Magnetic field dependence of magnetisation for the \( Y2 \) alloy at 2 K, 4 K and 6 K.

FIG. 11. The temperature dependence of upper critical fields (closed symbols) and irreversibility fields (open symbols) for the \( (V_{0.6}Ti_{0.4}) \)-Y alloys. The solid line is the fit to the \( H_{C2} \) using WHH formalism.
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The Fig. 12(b) presents the plot of reduced pinning force density \( f = F_p/F_{p_{\text{max}}} \) of all the alloys as a function of reduced field \( h = H/H_{irr} \). According to Dew-Hughes, the \( f \) depends on the spacing, size and nature of the pinning centres and is proportional to \( h^p(1-h)^q \). Each type of pinning centre yields an unique set of \( p \) and \( q \) values and a maximum of \( f \) at \( h_m = p/(p+q) \). The values of \( p \), \( q \) and \( h_m \) provides information on the nature of the pinning centres responsible for the \( J_C \). All the curves in Fig. 13(b) tend to scale on the falling edge, while significant variation of \( f \) among the alloys is observed for \( h < 0.1 \). This indicates that the pinning mechanism at high magnetic fields is same for all the alloys. The experimental value of \( h \) at which \( F_p \) becomes maximum \( (h_m) \) for the \( V_{0.6}Ti_{0.4} \) alloy is about 0.06. The \( h_m \) increases with the addition of yttrium and is about 0.14 for the Y2, Y3 and Y5 alloys at 4 K. The comparison of the curvature of the different pinning mechanisms for \( h \) close to unity with the curvature of \( f \) of the present alloys indicates that the pinning at high fields is due to the regions with large change in the superconducting properties (large change in the Ginsburg-Landau parameter \( \kappa \) or large \( \Delta \kappa \)). The known mechanisms of flux pinning have \( h_m \geq 0.2 \). The \( h_m \) < 0.2 for all the alloys indicates that the pinning that exists at low magnetic fields may not be effective at high fields.

In such cases where multiple types of pinning centres become available for pinning at different magnetic fields, \( F_p \) can be expressed as:

\[
F_p = F_{p1} + F_{p2} + F_{p3} + \ldots
\]  

(4)

where, \( F_{p1} \) is the pinning force density of individual pinning centres. The solid blue line in Fig. 13(b) shows the fit to the \( f \) by considering three types of pinning centres viz.,

(i) normal surface pinning \( (h^{0.5}(1-h)^2) \) by grain/cell boundaries (violet dot-dashed line)
(ii) \( \Delta \kappa \) surface pinning \( (h^{1.5}(1-h)) \) by dislocations (purple dashed line) and
(iii) normal surface pinning with \( H_{irr}' = 0.22 \times H_{irr} \) (orange dotted line).

One can see that pinning centres of type (i) and (ii) combined (red dash-dot-dotted line in Fig. 13(b)) can account for pinning in the high field range \( h > 0.35 \). The variation of \( F_{p_{\text{max}}} \) with the composition follows the cell boundary and dislocation density in these alloys. The \( F_p \) is maximum for the Y2 alloy which has the smallest...
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We believe that the pinning mechanism of type (iii) with a very low irreversibility field \( H_{irr} \) is about 0.22 times \( H_{irr} \). We have observed a positive magnetization signal just below \( T_C \) for all the \((V_{0.6}Ti_{0.4})-Y\) alloys in low magnetic fields. This is an effect what is known as 'High Field Paramagnetic Effect (HFPME)' where the positive magnetization arises from flux compression as well as creeping of flux lines from weak pinning-centres to strong pinning-centres leading to anisotropic distribution in flux line pinning. The flux compression is found to be due to the yttrium-rich precipitates become superconducting by the proximity effect. We believe that the pinning mechanism of type (iii) is related to the pinning centres that contribute to HFPME.

The shape of the \( F_p \) curve and its peak position change with temperature (Fig. 14(a)). The position of the maximum \( F_p \) changes from \( h_m = 0.08 \) at 6 K to about 0.23 at 2 K (Fig. 14(b)). The \( F_p \) curves are extrapolated linearly to get \( h_m \) \( (H_{irr}) \) at temperature 3 K and 2 K. The sharpness of the peak reduces and the curvature of the falling edge changes with decreasing temperature. This behaviour is an indication of different pinning centres becoming effective at different temperatures. The Fig. 14(b) shows that the pinning mechanism present in high fields and low temperatures is absent at temperatures close to \( T_C \). The thermal conductivity studies on \((V_{0.6}Ti_{0.4})-Gd\) alloys indicated that the dislocations become effective in scattering the phonons at low temperatures and high magnetic fields which in turn renormalizes the electron-phonon coupling. We infer that the dislocation become effective in pinning flux lines in high magnetic fields at low temperatures which is inline with the analysis related to Fig. 13(b).

IV. CONCLUSION

We have shown that the yttrium is immiscible and precipitates with various sizes in the \((V_{0.6}Ti_{0.4})-Y\) alloy. Dendritic microstructure is observed in all yttrium containing alloys. For \( \leq 2 \) at.% yttrium in the \((V_{0.6}Ti_{0.4})-Y\) alloy, fine yttrium-rich precipitates are generated because of phase separation of the homogeneous V-Ti-Y liquid into a solid \( \beta \)-\((V_{0.6}Ti_{0.4})-Y\) alloy and a yttrium-rich liquid. The size of the yttrium-rich precipitates increases for higher yttrium content due to liquid immiscibility. The dendritic cell size reduces with increasing yttrium content up to 2 at.% in the \((V_{0.6}Ti_{0.4})-Y\) alloy which results in the generation of a large number of line defects. Yttrium removes oxygen from the \((V_{0.6}Ti_{0.4})-Y\) alloy matrix due to which the \( T_C \) of the yttrium containing alloys is enhanced. The defects generated by the addition of yttrium are found to be effective in pinning the flux lines and also increase the \( H_{irr} \). The critical current density is increased by more than 7.5 times in fields higher than 1 T for the \((V_{0.58}Ti_{0.40})-Y_{0.02}\) alloy at 4 K.

V. DATA AVAILABILITY STATEMENT

The data that support the findings of this study are available from the corresponding author upon reasonable request.

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