

Complete superconducting phase diagram and flux pinning mechanisms in the yttrium-containing high-entropy alloy YNbTiZrHf

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We report on the synthesis and characterization of an yttrium-containing high-entropy alloy (HEA) superconductor obtained from a nominal YNbTiZrHf composition. Structural analysis reveals a body-centered-cubic matrix with minor yttrium-rich hexagonal precipitates. Comprehensive measurements of electrical resistivity, magnetization, ac magnetic susceptibility, and specific heat confirm bulk type-II superconductivity with a transition temperature $T_c \approx 6.1$ K, a critical current density $J_c \approx 45$ kA/cm², and an upper critical field $\mu_0 H_{c2} = 12.5$ T, exceeding the Pauli limit due to strong spin-orbit scattering. The superconducting phase diagram, constructed from $J_c(T)$, $J_c(\mu_0 H)$, and $\mu_0 H(T)$ data, is reported for a HEA superconductor. Flux pinning analysis indicates dominant surface pinning with a secondary point contribution, while collective pinning theory reveals a crossover from δT_c -type to δl -type pinning with increasing field. The estimated depairing current density $J_d \approx 21$ MA/cm², which is more than 2 orders of magnitude higher than the experimental J_c , suggests significant potential for performance enhancement through microstructural optimization.

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I. INTRODUCTION

High-entropy alloys (HEAs), composed of multiple principal elements in near-equimolar ratios, have emerged as a versatile platform for exploring novel physical phenomena due to their high configurational entropy and inherent chemical disorder. Among them, superconducting HEAs have attracted considerable interest for their potential to combine mechanical robustness with superconducting behavior [1–3]. In particular, the discovery of superconductivity in the Ta₃₄Nb₃₃Hf₈Zr₁₄Ti₁₁ [4] HEA initiated extensive research into the interplay between composition, structure, and superconducting properties in chemically complex systems. Subsequent studies demonstrated that the superconducting transition temperature T_c , upper critical field $\mu_0 H_{c2}$, critical current density J_c , and flux pinning characteristics are highly sensitive to both the choice of constituent elements and microstructural features [1–3,5–7].

In 2022, Krnel *et al.* investigated ScHfNbTaTiZr alloys [8]. They reported that, unlike the parent HfNbTaTiZr HEAs, which form a single-phase body-centered-cubic (bcc) structure with a homogeneous microstructure, the addition of Sc induces a two-phase structure comprising a Sc-poor bcc

phase and a Sc-rich hexagonal-close-packed (hcp) phase. Superconductivity in these alloys originates from the bcc phase fraction, which exhibits superconducting parameters comparable to those of the parent alloys. However, the determined $\mu_0 H_{c2}$ for these systems (i.e., for Sc₁₈Nb₃₂Ti₂₄Zr₂₆ $\mu_0 H_{c2} \approx 14$ T) suggests that adding group III elements to a bcc HEA may have a positive influence on this crucial superconducting parameter. Similarly, von Rohr and Cava [9] studied five nonequimolar HEAs containing yttrium but reported on only transition temperatures, leaving other key superconducting parameters unexplored.

In this work, we report on the synthesis and comprehensive characterization of an yttrium-containing high-entropy alloy obtained from a nominal YNbTiZrHf composition. For clarity, throughout the remainder of this paper we retain the notation YNbTiZrHf to refer to the nominal composition. However, structural and microstructural analyses reveal that the material forms a two-phase system consisting of a superconducting yttrium-poor bcc phase and a nonsuperconducting yttrium-rich hcp secondary phase, indicating pronounced compositional partitioning. The superconducting properties are investigated using electrical resistivity, dc magnetization, ac susceptibility, and specific heat measurements. In particular, the superconducting phase diagram was constructed using the determined $J_c(T)$, $J_c(\mu_0 H)$, and $\mu_0 H(T)$ dependencies, and the flux pinning behavior was analyzed in detail within the frameworks of the Dew-Hughes model and collective pinning theory. The obtained results are discussed in the context of previously studied HEA superconductors, allowing for direct comparison of critical parameters such as T_c , $\mu_0 H_{c2}$, J_c , flux pinning mechanisms, and electron-phonon coupling strength.

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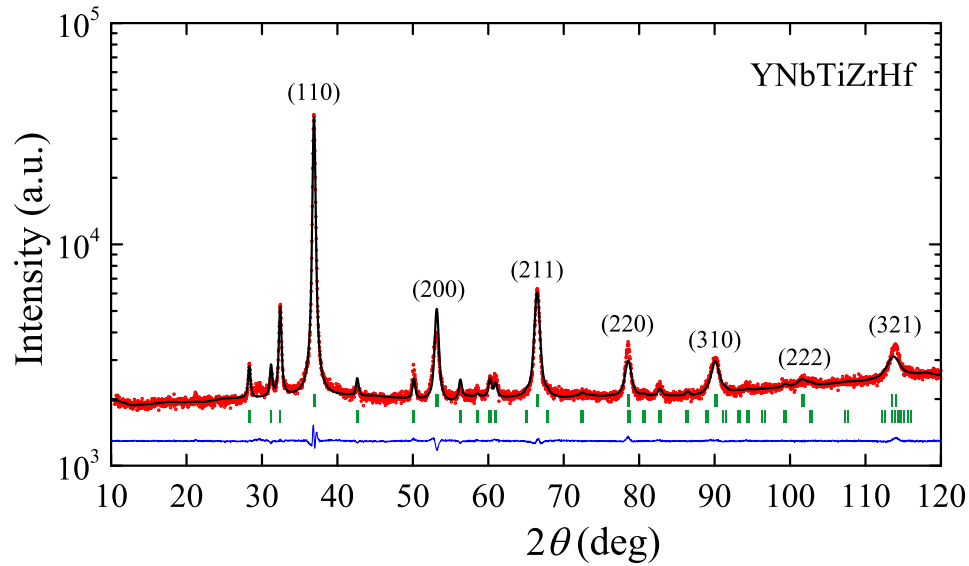


FIG. 1. X-ray powder diffraction pattern of YNbTiZrHf with Rietveld refinement of its crystal structure. Red circles represent the experimental data, the black line corresponds to the calculated pattern, and the blue curve shows the difference between the two. Upper and lower green ticks indicate the positions of Bragg reflections for both the main and impurity phases, respectively (see text for details). The listed Miller indices refer to the main phase.

II. EXPERIMENTAL DETAILS

A polycrystalline sample of YNbTiZrHf was synthesized via arc melting in a high-purity argon atmosphere using a Ti-Zr alloy as a getter. To enhance chemical homogeneity, the ingot was remelted multiple times. The total mass loss during the melting process did not exceed 0.2%. All elemental precursors used in the synthesis had a nominal purity of $\geq 99.9\%$, and the nominal starting composition was Y:Nb:Ti:Zr:Hf = 1:1:1:1:1. The resulting sample was sectioned using a diamond saw.

Structural characterization was performed by x-ray diffraction (XRD) using a PANalytical X'Pert Pro diffractometer equipped with a Cu $K\alpha$ radiation source. The diffraction data were analyzed using the FULLPROF software [10]. The chemical composition was examined via energy-dispersive x-ray spectroscopy using a FEI Nova NanoSEM 230 scanning electron microscope (SEM) equipped with an EDAX Genesis XM4 spectrometer.

Magnetic susceptibility, magnetization, electrical resistivity, and specific heat measurements were carried out using a Quantum Design physical property measurement system. Experiments were conducted over a temperature range of 1.8–300 K and in applied magnetic fields up to 9 T.

III. RESULTS AND DISCUSSION

A. Crystal structure and chemical composition

Figure 1 presents the XRD pattern of the YNbTiZrHf sample measured at room temperature. Rietveld refinement of the experimental data indicates the presence of two crystalline phases. The derived values of the weighted profile R factor $R_{wp} = 18.6$, the expected R factor $R_{exp} = 7.43$, and the chi-square parameter $\chi^2 = 6.25$ indicate good quality of the overall fit.

The primary phase adopts a bcc structure (space group $Im\bar{3}m$, W type) with a refined lattice parameter $a = 3.445(1) \text{ \AA}$. This value is larger than those reported for other bcc HEAs composed of transition metals [4–6,8,11–13]. The relatively large atomic radius of Y compared to Nb, Zr, Hf, and Ti likely contributes to the increased lattice parameter observed in YNbTiZrHf. A similar effect has been reported for scandium-containing bcc HEAs, such as $\text{Sc}_{21}\text{Hf}_{26}\text{Nb}_{28}\text{Zr}_{25}$ and $\text{Sc}_9\text{Hf}_{21}\text{Nb}_{26}\text{Ta}_{26}\text{Zr}_{18}$, in which the inclusion of Sc (also a group III element) leads to lattice expansion [8].

The secondary phase identified in the diffraction pattern accounts for 11.4(2) wt% of the sample and crystallizes in a hcp structure (space group $P6_3/mmc$). The refined lattice parameters, $a = b = 3.641(1) \text{ \AA}$ and $c = 5.738(1) \text{ \AA}$, closely match those of pure yttrium [$a = b = 3.6474(7) \text{ \AA}$, $c = 5.7306(8) \text{ \AA}$] [14]. This similarity suggests that the secondary phase consists predominantly of yttrium, with minor contributions from other constituent elements.

Figure 2 presents SEM micrographs and energy-dispersive x-ray spectroscopy (EDXS) elemental maps of the studied sample. As shown in Fig. 2(a), two distinct regions are observed in the cross section: a near-surface layer and the sample interior. High-magnification images [Figs. 2(b) and 2(c)] reveal a pronounced mosaiclke microstructure composed of yttrium-poor crystallites separated by chemically distinct intergranular regions. This chemical segregation is consistent with the x-ray diffraction results, which indicate the coexistence of two crystalline phases: a bcc phase associated with the Y-poor crystallites and an hcp phase corresponding to the Y-rich intergranular regions. Similar microstructural features have been reported in HEAs containing group III elements such as Sc [8] and Tb [15].

The quantitative EDXS analysis summarized in Table I confirms strong chemical partitioning between two phases. In

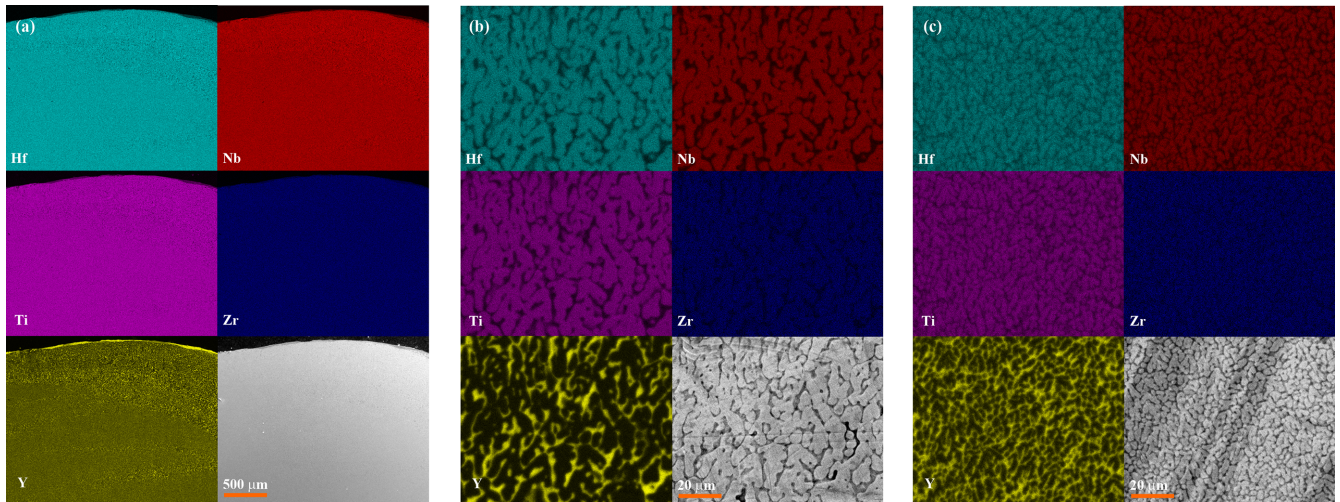


FIG. 2. SEM micrographs and EDXS elemental maps of the YNbTiZrHf sample. (a) Low-magnification cross-sectional image. (b) High-magnification images and corresponding elemental maps of the near-surface region. (c) High-magnification images and elemental maps of the bulk (central) region. The bottom right panel in each set shows the corresponding SEM image.

the near-surface region, the interiors of crystallites (Y-poor phase) contain 5.6(3) at. % Y, while Nb, Ti, Zr, and Hf are present in nearly equiatomic proportions [25.7(1.6), 21.5(0.4), 23.7(0.6), and 23.6(0.7) at. %, respectively]. In contrast, the intergranular regions (Y-rich phase) are strongly enriched in yttrium, reaching 57.4(1.7) at. % Y, accompanied by reduced concentrations of the transition metals. A similar chemical segregation is observed in the bulk region [Fig. 2(c)], although the compositional contrast is reduced. The Y-poor crystallites contain 9.7(1.0) at. % Y, while the Y-rich intergranular regions contain 39.1(2.2) at. % Y. The remaining elements again remain close to equiatomic proportions within each phase.

Compared to that in the near-surface region, the reduced yttrium concentration difference between the two phases in the bulk indicates partial homogenization of the elemental distribution. At the same time, the bcc grains are significantly finer in the sample interior. This suggests that differences in solidification and cooling conditions play a key role in determining the final microstructure. In particular, the slower cooling rate in the central part of the sample likely promotes enhanced atomic diffusion, leading to reduced compositional gradients and finer spatial separation between the phases.

B. Physical properties

Figure 3 presents the temperature dependence of the electrical resistivity ρ for YNbTiZrHf, measured in applied

magnetic fields up to 9 T. All $\rho(T)$ curves exhibit a sharp drop in resistivity to unmeasurable values, indicating the onset of a superconducting state. The critical temperature T_c^ρ was defined as the intersection point of the extrapolation of the steepest section of the transition with zero resistivity for each $\rho(T)$ curve. In zero magnetic field, $T_c^\rho = 6.1(1)$ K and (as expected) decreases gradually with increasing magnetic field. Notably, superconductivity persists up to $\mu_0 H = 9$ T, with $T_c^\rho = 2.5(1)$ K, suggesting that this alloy may exhibit one of the highest upper critical fields among bcc-structured HEAs [1,5,6,8].

As shown in Fig. 3(a), above the superconducting transition, the electrical resistivity of YNbTiZrHf exhibits metallic behavior, with ρ increasing systematically with temperature. The residual resistivity ratio, defined as $\rho(300\text{ K})/\rho(7\text{ K})$, is close to 2. This relatively low value is comparable to those reported for other HEAs and reflects the high degree of chemical and structural disorder characteristic of this class of materials [16].

Figure 4 presents the real (χ') and imaginary (χ'') parts of the ac susceptibility measured for YNbTiZrHf in the zero-field-cooling (ZFC) regime under nominal applied magnetic fields up to $\mu_0 H = 9$ T. The emergence of a diamagnetic signal in the $\chi'(T)$ curves, accompanied by a peak in $\chi''(T)$, provides clear evidence of a superconducting transition in the sample. The determined T_c^{ac} values are in good agreement with those obtained from the resistivity data.

TABLE I. Quantitative EDXS elemental composition (in at. %) of the investigated HEA in the near-surface and bulk regions. The data are given separately for yttrium-poor (bcc) and yttrium-rich (hcp) phases, along with the average composition for each region.

Region Element Phase	Bulk					Near surface				
	Y	Nb	Ti	Zr	Hf	Y	Nb	Ti	Zr	Hf
Y-rich	39.1(2.2)	17.2(1.7)	12.1(0.4)	18.3(1.0)	13.4(0.3)	57.4(1.7)	12.9(1.3)	7.9(0.8)	12.8(0.2)	9.0(0.8)
Y-poor	9.7(1.0)	25.5(0.6)	19.7(0.6)	24.1(0.7)	21.0(0.7)	5.6(0.3)	25.7(1.6)	21.5(0.4)	23.7(0.6)	23.6(0.7)
Averaged	21.8(0.9)	19.3(0.8)	18.3(0.7)	20.7(0.8)	19.9(0.8)	17.1(0.7)	20.7(0.8)	19.5(0.8)	21.3(0.9)	21.3(0.9)

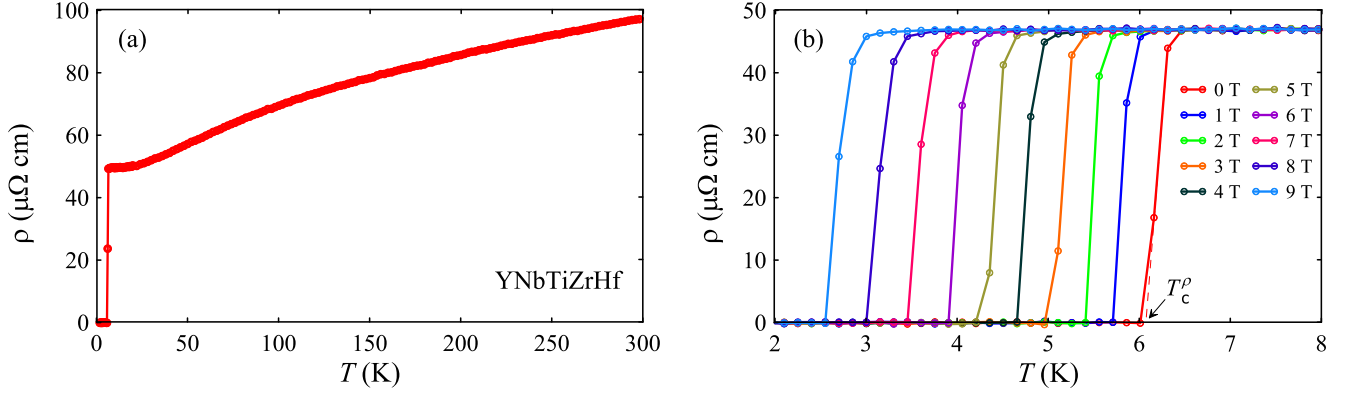


FIG. 3. Temperature dependence of the electrical resistivity measured for YNbTiZrHf (a) in zero magnetic field up to room temperature and (b) at low temperatures in the presence of applied magnetic fields up to 9 T. Solid lines are guides to the eye.

Figure 5(a) presents the dc mass magnetization M measured for YNbTiZrHf in the ZFC regime under applied magnetic fields up to $\mu_0 H = 9$ T. The appearance of a pronounced diamagnetic signal, with negative magnetization values, confirms the onset of superconductivity in the sample. The superconducting transition temperatures T_c^{dc} determined from the magnetization data are in good agreement with those obtained from electrical resistivity and ac susceptibility measurements. Using the estimated demagnetization factor $N = 0.208$, the dimensionless volumetric susceptibility was determined to be $4\pi\chi_v = -0.917(34)$ at $\mu_0 H = 10$ mT, as shown in the upper inset of Fig. 5(b). This corresponds to a superconducting volume fraction of approximately 91.7%. In

combination with the XRD results, this finding indicates that superconductivity originates exclusively from the bcc phase. Finally, it should be noted that at higher magnetic fields a weak paramagnetic contribution becomes noticeable in the $M(T)$ curves. This behavior can be attributed to the normal-state response of the bcc phase, together with the presence of the nonsuperconducting yttrium-rich secondary phase.

Figure 5(b) displays the magnetic hysteresis loops measured at temperatures ranging from 2.5 to 6 K. These loops, along with the $M(\mu_0 H)$ curves obtained in the ZFC regime, shown in the inset of Fig. 5(b), clearly indicate that YNbTiZrHf is a bulk type-II superconductor with a lower critical field $\mu_0 H_{c1} < 0.1$ T at 2.5 K.

As we can see in Fig. 6(a), the temperature dependence of the specific heat C_p of YNbTiZrHf was found to be featureless above T_c and can be well described by the free-electron Sommerfeld model combined with the Debye model for lattice vibrations (corrected for anharmonic effects [17]), represented by the first and second terms, respectively, in the formula

$$C_p(T) = \gamma T + \frac{9Rn}{1 + \alpha T} \left(\frac{T}{\Theta_D^{\text{HT}}} \right)^3 \int_0^{\Theta_D^{\text{HT}}/T} \frac{x^4 e^x}{(e^x - 1)^2} dx, \quad (1)$$

where Θ_D^{HT} denotes the Debye temperature determined from high-temperature data, n is the number of atoms in the formula unit (f.u.), R is the universal gas constant, and α stands for the parameter describing the anharmonicity of the system. Assuming $n = 1$ for YNbTiZrHf and using the Sommerfeld coefficient $\gamma = 8.49(7)$ mJ K⁻² mol⁻¹ (determined later in this section), a least-squares fit of Eq. (1) to the experimental data yields $\Theta_D^{\text{HT}} = 238(3)$ K and $\alpha = 2.6(4) \times 10^{-4}$ K⁻¹; all those values are of the magnitude characteristic of intermetallic alloys and compounds.

At low temperatures, the specific heat data plotted as C_p/T versus T^2 [Fig. 6(b)] reveal a distinct jump, indicating the emergence of bulk superconductivity in YNbTiZrHf. The normal-state specific heat values obtained above the transition were used to extract the Sommerfeld coefficient $\gamma = 8.49(7)$ mJ K⁻² mol⁻¹, along with the lattice heat capacity coefficients $\beta = 0.216(5)$ mJ K⁻⁴ mol⁻¹ and $\delta = 0.0011(1)$ mJ K⁻⁶ mol⁻¹, using the expression

$$\frac{C_p(T)}{T} = \gamma + \beta T^2 + \delta T^4. \quad (2)$$

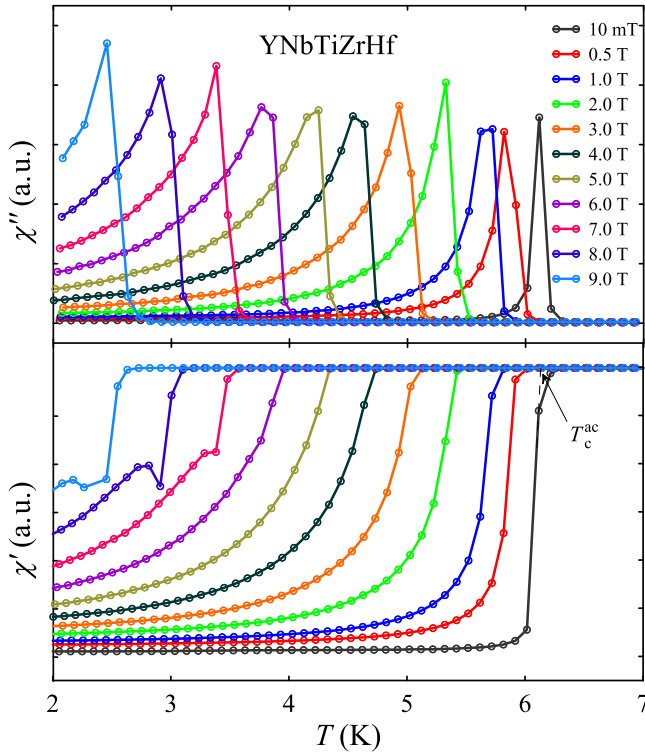


FIG. 4. Real (χ') and imaginary (χ'') parts of the ac susceptibility measured for YNbTiZrHf in nominal applied magnetic fields $\mu_0 H$. Solid lines are guides to the eye.

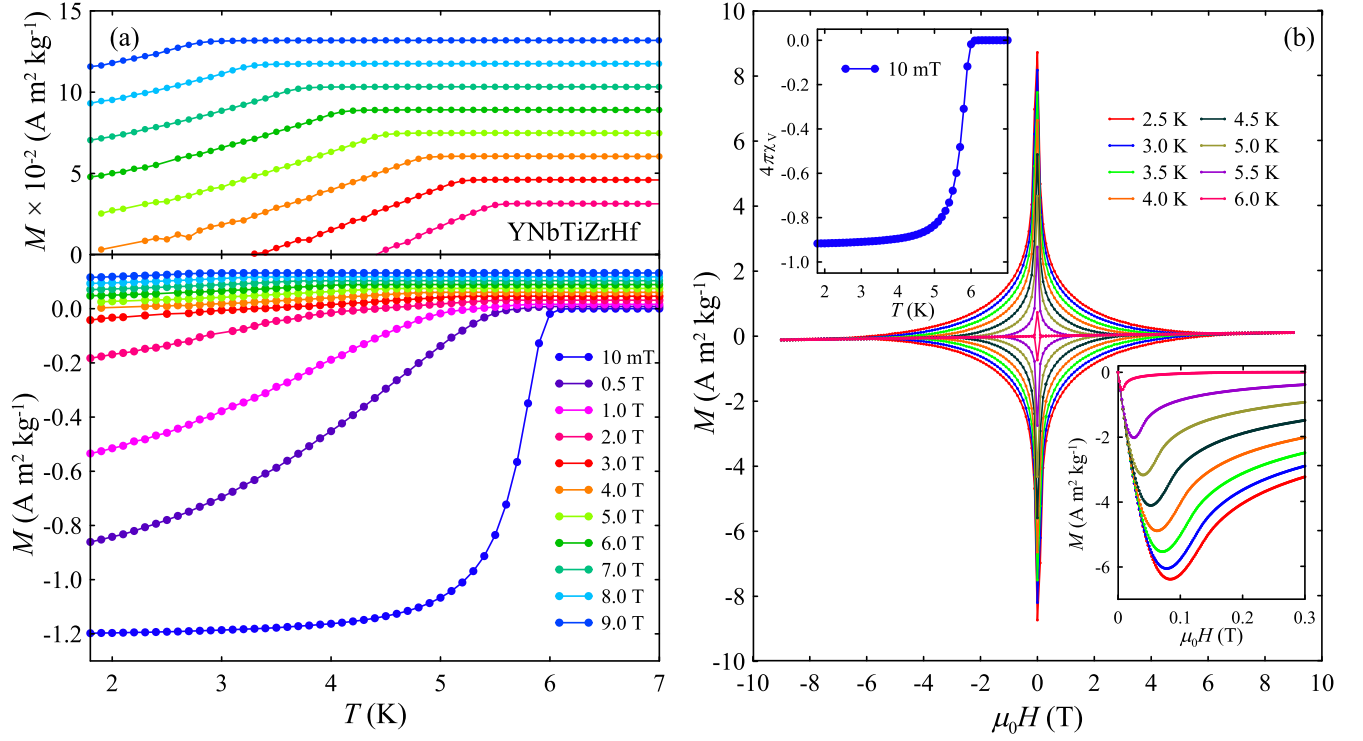


FIG. 5. (a) dc mass magnetization M measured for YNbTiZrHf in nominal applied magnetic fields $\mu_0 H$. (b) Isothermal magnetic hysteresis loops of YNbTiZrHf measured at various temperatures. The upper inset shows the dimensionless volumetric susceptibility $4\pi\chi_v(T)$ measured at $\mu_0 H = 10$ mT. The lower inset shows mass magnetization as a function of magnetic field measured in the zero-field-cooling (ZFC) regime. Solid lines are guides to the eye.

As can be seen in Fig. 6(b), including the δT^4 term enables a more accurate fit to the experimental data, reflecting the subtle influence of anharmonicity on the vibrational properties of YNbTiZrHf. The coefficient β yields the low-temperature Debye temperature $\Theta_D^{\text{LT}} = 208(2)$ K via

$$\Theta_D^{\text{LT}} = \left(\frac{12nR\pi^4}{5\beta} \right)^{\frac{1}{3}}, \quad (3)$$

while γ can be used to determine the density of states of the conduction electrons for both spin orientations at the Fermi level $N(E_F) = 3.60(3)$ states $\text{eV}^{-1} \text{ f.u.}^{-1}$ based on the relation

$$\gamma = \frac{1}{3}\pi^2 k_B^2 N_A N(E_F), \quad (4)$$

where k_B is the Boltzmann constant and N_A is Avogadro's number. The values of $\gamma = 8.49(7)$ $\text{mJ K}^{-2} \text{ mol}^{-1}$ and $\Theta_D^{\text{LT}} = 208(2)$ K obtained for YNbTiZrHf are comparable to those reported for $(\text{NbTi})_{0.75}(\text{ZrHf})_{0.25}$ [6]. In contrast, γ is higher, and Θ_D^{LT} is lower than in most Sc-based HEAs investigated in Ref. [8].

To determine critical temperatures from specific heat data $T_c^{C_p}$, the electron contribution to the specific heat C_{el} was calculated by subtracting the phonon contribution from the total specific heat $C_{\text{el}} = C_p - (\beta T^3 + \delta T^5)$. In the next step, due to the relatively broad, λ -shaped superconducting anomaly, the transition temperature $T_c^{C_p}$ was determined using an equal-entropy construction, as illustrated in Fig. 6(c), which shows $C_{\text{el}}/\gamma T$ as a function of temperature. The critical temperatures determined from electrical resistivity (T_c^{ρ}), ac magnetic susceptibility (T_c^{ac}), dc magnetization (T_c^{dc}), and specific heat

($T_c^{C_p}$) measurements were subsequently used to construct the superconducting phase diagram, which is discussed later in this paper. For the calculations presented in this section, the superconducting transition temperature was taken as $T_c \equiv T_c^{C_p} = 6.17(5)$ K.

Figure 6(d) shows $\ln(C_{\text{es}}/\gamma T_c)$ vs $1/T$, which originates from the BCS description of the electronic specific heat [18]:

$$C_{\text{BCS}}(T) = A\gamma T_c \exp\left(-\frac{\Delta_0}{k_B T}\right), \quad (5)$$

where A is a constant and Δ_0 denotes the superconducting energy gap. From this analysis, the normalized energy gap is found to be $2\Delta_0/k_B T_c = 4.38(2)$, exceeding the BCS weak-coupling limit of 3.52 for a single isotropic gap [18], suggesting moderate or even strong coupling in YNbTiZrHf. In the inset to Fig. 6(d) one can see that the normalized superconducting energy gap extracted from the heat-capacity data exhibits a gradual decrease with increasing magnetic field. This behavior is expected for type-II superconductors and reflects the spatial averaging of the superconducting order parameter in the mixed state. With increasing field, the density of vortices and associated normal cores increases, leading to a progressive suppression of the superconducting condensate and enhanced quasiparticle excitations. As a result, the gap value obtained from bulk thermodynamic measurements represents an effective, field-dependent average rather than the maximum gap of the superconductor. The smooth evolution of $2\Delta_0/k_B T_c$ with field is therefore consistent with orbital pair breaking in a single-gap superconductor.

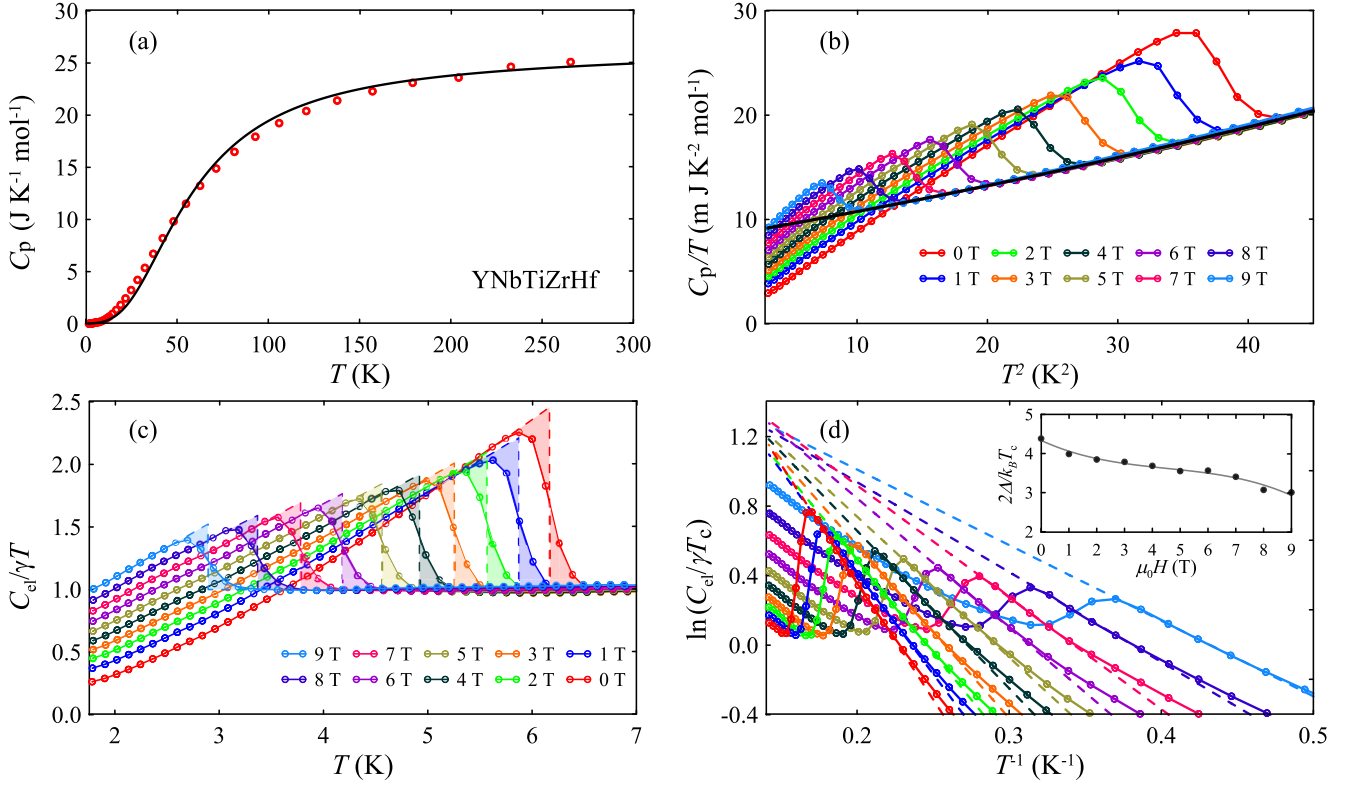


FIG. 6. Specific heat data collected for YNbTiZrHf. (a) Temperature dependence of specific heat C_p ; the solid curve is a fit to Eq. (1) to the experimental data. (b) C_p/T vs T^2 measured in various magnetic fields; the solid black line is a fit to Eq. (2). (c) Temperature dependence of the conduction electron contribution to the total specific heat C_{el} . Equal entropy constructions that allowed for the determination of the critical temperature T_c of the system studied are presented. (d) $\ln(C_{el}/\gamma T_c)$ as a function of inverse temperature, measured in various magnetic fields; the straight dashed lines are fits to Eq. (5) to experimental data. The inset shows the normalized superconducting energy gap $2\Delta/k_B T_c$ as a function of the applied magnetic field.

The values $T_c = 6.1(1)$ K and $\Theta_D^{LT} = 208(2)$ K were used to estimate the electron-phonon coupling constant λ_{el-ph} via McMillan's formula [19]:

$$\lambda_{el-ph} = \frac{1.04 + \mu^* \ln\left(\frac{\Theta_D^{LT}}{1.45T_c}\right)}{(1 - 0.62\mu^*) \ln\left(\frac{\Theta_D^{LT}}{1.45T_c}\right) - 1.04}, \quad (6)$$

where μ^* denotes the Coulomb pseudopotential, typically ranging from 0.10 to 0.15 for d -electron metals. Assuming $\mu^* = 0.125$, the resulting value $\lambda_{el-ph} = 0.77(2)$ suggests intermediate electron-phonon coupling strength in YNbTiZrHf, consistent with the value reported for the equimolar NbTiZrHfTa alloy [20]. However, it is notably lower than the coupling constants obtained for (NbHf)_{0.75}(TiZr)_{0.25}, (NbZr)_{0.75}(TiHf)_{0.25}, and (NbTi)_{0.75}(ZrHf)_{0.25} [6].

C. Upper critical field

The temperature dependencies of the upper critical field $\mu_0 H_{c2}(T)$ derived from the ac magnetic susceptibility, electrical resistivity, specific heat, and dc magnetization measurements are presented in Figs. 7(a)–7(d), respectively. To describe the experimental $\mu_0 H_{c2}(T)$ data, the full Werthamer-Helfand-Hohenberg (WHH) formalism for isotropic-gap BCS superconductors in the dirty limit was employed, incorporating both spin-paramagnetic effects via the Maki parameter α_M

and spin-orbit scattering through the constant λ_{SO} [21–23]:

$$\ln \frac{1}{t} = \left(\frac{1}{2} + \frac{i\lambda_{SO}}{4\gamma} \right) \psi \left(\frac{1}{2} + \frac{\bar{h} + \frac{1}{2}\lambda_{SO} + i\gamma}{2t} \right) + \left(\frac{1}{2} - \frac{i\lambda_{SO}}{4\gamma} \right) \psi \left(\frac{1}{2} + \frac{\bar{h} + \frac{1}{2}\lambda_{SO} - i\gamma}{2t} \right) - \psi \left(\frac{1}{2} \right), \quad (7)$$

where $t = T/T_c$, $\bar{h} = \frac{4}{\pi^2} \frac{H_{c2}}{-dH_{c2}/dT}$, $\gamma = \sqrt{(\alpha_M \bar{h})^2 - (\frac{1}{2}\lambda_{SO})^2}$, i is the imaginary unit, and ψ denotes the digamma function. As shown in Fig. 7, the results obtained from various experimental techniques are in good agreement. However, it should be noted here that among the techniques employed in this study, specific heat measurements are generally considered the most reliable for determining the superconducting transition temperature and related parameters. Unlike electrical resistivity and magnetic susceptibility, which are sensitive to percolative paths and magnetic shielding effects, respectively, specific heat directly probes the bulk thermodynamic response of the material. Therefore, the values $\alpha_M = 1.83(5)$, $\mu_0 H_{c2}(0) = 12.5(3)$ T, and orbital limiting field $\mu_0 H_{c2}^{orb}(0) = 14.7(1)$ T derived from specific heat data are considered the most representative of the intrinsic superconducting properties of YNbTiZrHf.

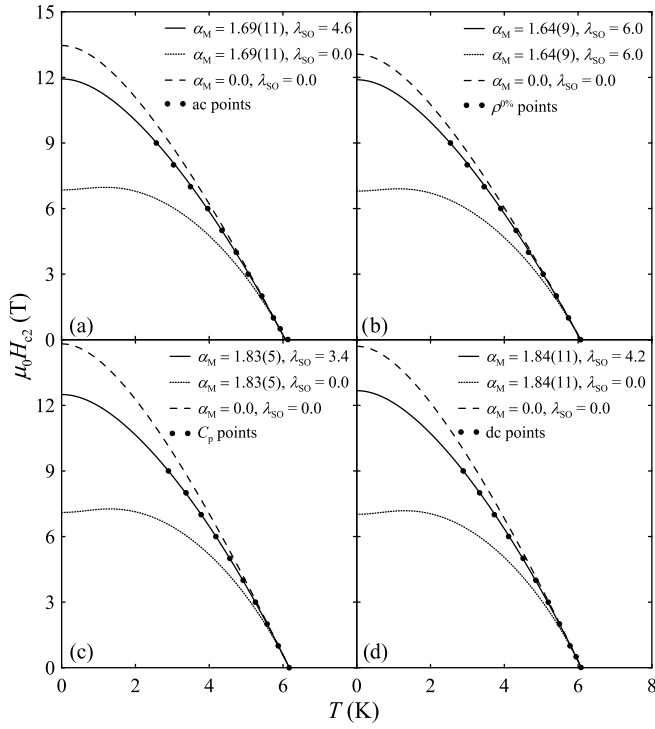


FIG. 7. Temperature dependence of the upper critical fields $\mu_0 H_{c2}$ for YNbTiZrHf determined from measurements of (a) ac magnetic susceptibility, (b) electrical resistivity, (c) specific heat, and (d) dc magnetization. Solid and dashed lines are fits of the WHH model [Eq. (7)] to the experimental points.

and are used for further calculations. Consequently, the obtained $\mu_0 H_{c2}(0) = 12.5(3)$ T places YNbTiZrHf among the bcc-structured HEAs with the highest upper critical fields, comparable to those reported for $(\text{NbHf})_{0.75}(\text{TiZr})_{0.25}$, $(\text{NbZr})_{0.75}(\text{TiHf})_{0.25}$, $(\text{NbTi})_{0.75}(\text{ZrHf})_{0.25}$, $\text{Ti}_{0.5}(\text{ZrNbHfTa})_{0.5}$, $\text{Ti}_{0.5}(\text{VNbHfTa})_{0.5}$, and Sc-based HEAs [5,6,8].

The Pauli limiting field $\mu_0 H_P = 11.3(1)$ T was calculated using the formula [24,25]

$$\mu_0 H_P = 1.84 T_c. \quad (8)$$

As we can see, the upper critical field of $\mu_0 H_{c2}(0) = 12.5(3)$ T exceeds the Pauli paramagnetic limit. At the same time, the relatively large spin-orbit scattering parameter obtained from the WHH analysis, $\lambda_{SO} = 3.4$, points to a significant spin-orbit interaction. This strong spin-orbit coupling is known to suppress the Pauli pair-breaking effect, thereby enabling the superconductor to sustain higher critical fields than those predicted by the Pauli limit alone [22].

The thermodynamic critical field at 0 K, $\mu_0 H_c(0) = 0.155(1)$ T, was derived from the superconducting energy gap Δ_0 and the Sommerfeld coefficient per volume γ_V using the expression [26]

$$\mu_0 H_c(0) = \sqrt{\frac{3\gamma_V \mu_0}{2\pi^2}} \frac{\Delta_0}{k_B}, \quad (9)$$

where γ_V is expressed in $\text{JK}^{-2} \text{m}^{-3}$.

Using the estimated $\mu_0 H_c(0)$ and $\mu_0 H_{c2}^{\text{orb}}(0)$ parameters, we can calculate two fundamental superconducting length scales: the magnetic penetration depth $\lambda_{GL} = 317(8)$ nm and

the Ginzburg-Landau coherence length $\xi_{GL} = 4.74(4)$ nm. These quantities were obtained using the following relations [27]:

$$\xi_{GL} = \sqrt{\frac{\phi_0}{2\pi \mu_0 H_{c2}^{\text{orb}}(0)}}, \quad \lambda_{GL} = \sqrt{\frac{\phi_0}{4\pi} \frac{\mu_0 H_{c2}^{\text{orb}}(0)}{\mu_0 H_c^2(0)}}, \quad (10)$$

where $\phi_0 = 2.067 \times 10^{-15}$ Wb is the magnetic flux quantum. From these values, the Ginzburg-Landau parameter was determined to be $\kappa_{GL} = \lambda_{GL}/\xi_{GL} = 67(1)$, which significantly exceeds the $1/\sqrt{2}$ limit, thereby confirming that YNbTiZrHf is a type-II superconductor. Finally, the lower critical field at 0 K $\mu_0 H_{c1}(0) = 7.7(4)$ mT is estimated from the expression [27]

$$\mu_0 H_{c1}(0) = \frac{H_c(0)}{\sqrt{2}\kappa_{GL}} \ln \kappa_{GL}. \quad (11)$$

The obtained value is consistent with the magnetization data presented in Fig. 4(b).

D. Critical current density

The isothermal magnetic hysteresis loops measured for YNbTiZrHf, shown in Fig. 4(b), were used to estimate the superconducting critical current density J_c . According to the Bean critical state model, the width of the magnetic hysteresis loop reflects the irreversible magnetization arising from the material's response to changes in the external magnetic field under isothermal conditions. This response is governed by the critical current density, which determines the extent of magnetic flux penetration in the superconducting state [28]. As a result, J_c can be expressed as follows [29]:

$$J_c = \frac{3\Delta M d}{R}, \quad (12)$$

where ΔM is the vertical width of the magnetic hysteresis loop, d is the density of the material, and R is the radius of the spherical sample. The J_c values as a function of $\mu_0 H$ calculated for various temperatures are presented in Fig. 8(a). As we can see, the values of J_c are approximately 1 to 2 orders of magnitude lower than those reported for commercial superconducting wires and tapes based on Nb-Ti, Nb₃Sn, or MgB₂ [30]. However, when compared to those of other HEAs, the obtained J_c are significantly higher than those observed in single-phase HEAs such as Ta_{1/6}Nb_{2/6}Hf_{1/6}Zr_{1/6}Ti_{1/6} and Nb_{2/5}Hf_{1/5}Zr_{1/5}Ti_{1/5} and are lower than those reported for NbScTiZr and (NbZrPt)_{0.67}Ti_{0.33} alloys with eutectic microstructures [7].

To elucidate the flux pinning mechanism, the first step involves calculating the magnetic field dependence of the flux pinning force density, defined as

$$F_p = |\mu_0 H \times J_c|. \quad (13)$$

Next, the normalized flux pinning force $f_p = F_p/\max(F_p)$ is evaluated as a function of the reduced field $h = H/H_{irr}$ at various temperatures, where $\max(F_p)$ is the maximum pinning force and H_{irr} denotes the irreversibility field determined from Kramer plots shown in Fig. 8(b) [31]. Within the Kramer scaling approach, which applies to superconductors with dominant surface (grain-boundary) pinning, the

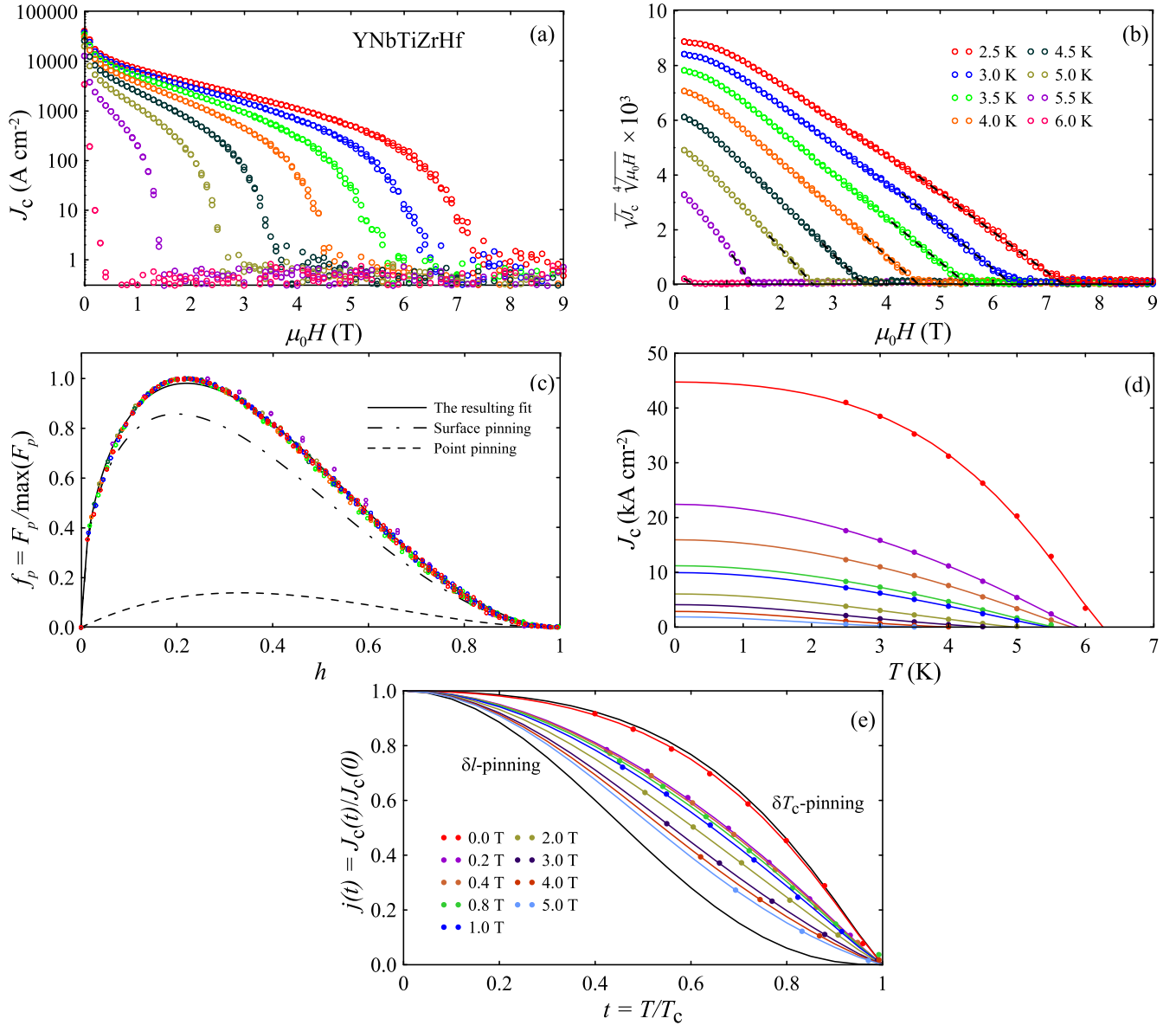


FIG. 8. Critical current analysis for YNbTiZrHf. (a) J_c as a function of $\mu_0 H$ determined at various temperatures. (b) the Kramer plots (dashed lines) to determine the irreversible fields H_{irr} . (c) $f_p(h)$ dependence constructed using all measured points in the temperature range from 2.5 to 6 K; the solid line is a fit of the two-component Dew-Hughes model with dominant surface pinning and a secondary point pinning contribution to the experimental data. (d) Temperature dependence of J_c measured at various magnetic fields; solid lines are fits of the two-component Griesen model to the experimental data. (e) Normalized temperature dependence of the critical current density; solid lines are fits of the two-component Griesen model, and the “ δI -pinning” and “ δT_c -pinning” lines correspond to Eq. (15).

quantity $J_c^{1/2}(\mu_0 H)^{1/4}$ is expected to exhibit a linear dependence on $\mu_0 H$ in the high-field regime. Extrapolation of this linear dependence to zero provides an estimate of the irreversibility field $\mu_0 H_{irr}$. The shape of $f_p(h)$ is governed by the underlying flux pinning mechanism. A widely used framework for interpreting this behavior is the Dew-Hughes model [32], which expresses the field dependence of the pinning force as

$$f_p(h) \propto h^p (1 - h)^q, \quad (14)$$

where the exponents p and q are characteristic of the type of defects or microstructural features responsible for flux

pinning. This empirical relation enables differentiation between various pinning regimes, such as point defects and grain boundaries, based on the shape and peak position of the $f_p(h)$ curve. Figure 8(c) presents the $f_p(h)$ dependence determined for YNbTiZrHf, which is constructed using all measured points in the temperature range from 2.5 to 6 K. As we can see, the experimental data can be well described by a two-component model. In particular, the best fit is obtained using the exponents $p = 0.5$ and $q = 2$ for normal surface (grain-boundary) pinning and $p = 1$ and $q = 2$ for normal point pinning [32]. The relative contributions of these components suggest that surface pinning dominates, with an approximate intensity ratio of 3:1 compared to point pinning. Interestingly,

the normalized pinning force analysis reveals no evidence of $\Delta\kappa_{\text{GL}}$ -type pinning, which is typically expected in chemically disordered systems such as HEAs. $\Delta\kappa_{\text{GL}}$ pinning arises from spatial fluctuations in the Ginzburg-Landau parameter, often linked to compositional inhomogeneities or local variations in superconducting properties. Given the multicomponent nature and high configurational entropy of YNbTiZrHf, such fluctuations are generally presumed to be present. However, the experimental data are well described by the two-component Dew-Hughes model involving surface and point pinning, with no indication of the characteristic maximum located at $h > 0.5$, which is typical of $\Delta\kappa_{\text{GL}}$ pinning [32].

The temperature dependence of J_c measured at various magnetic fields is shown in Fig. 8(d). These data were analyzed within the framework of collective pinning theory [33], which predicts two primary pinning mechanisms: δl pinning, arising from spatial variations in the charge-carrier mean free path, and δT_c pinning, associated with randomly distributed fluctuations in the superconducting transition temperature. According to the approach proposed by Griessen *et al.* [34], these mechanisms can be described by the following equations:

$$J_c(t) = J_c(0)(1 - t^2)^{5/2}(1 + t^2)^{-1/2},$$

$$J_c(t) = J_c(0)(1 - t^2)^{7/6}(1 + t^2)^{5/6} \quad (15)$$

for δl and δT_c pinning, respectively. As shown in Fig. 8(e), both models were fitted to the experimental $J_c(T)$ curves. At zero magnetic field, the data are well described by the δT_c pinning model, indicating that compositional fluctuations dominate the pinning landscape. With increasing magnetic field, a gradual crossover to a mixed pinning regime is observed, characterized by a decreasing contribution from δT_c pinning and an increasing contribution from δl pinning. This behavior suggests that, while compositional inhomogeneities govern vortex pinning at low fields, structural defects affecting the mean free path become increasingly relevant as the vortex density increases. Such a crossover is consistent with expectations for HEAs, where chemical disorder and lattice imperfections may coexist as pinning sources.

At first glance, the Dew-Hughes and collective pinning analyses may appear mutually inconsistent. However, this apparent discrepancy can be resolved by considering the different aspects of vortex pinning probed by the two approaches. The Dew-Hughes model classifies pinning in terms of the nature of the pinning centers (normal or $\Delta\kappa$ type) and their effective dimensionality. Importantly, this effective dimensionality depends on the magnetic field since pinning centers are characterized by the number of their spatial dimensions that exceed the inter-flux-line spacing $d \propto B^{-1/2}$ [32]. The Dew-Hughes analysis indicates that the pinning mechanism remains essentially unchanged over the investigated temperature range (2.5–6 K), with the dominance of normal surface pinning. Such pinning can be naturally associated with grain boundaries, which coincide with fine regions of the yttrium-rich hcp secondary phase. In contrast, the collective pinning analysis reveals a clear field dependence of the pinning mechanism, with an increasing contribution from δl pinning at higher magnetic fields. This behavior can be understood by noting that different pinning mechanisms may dominate in

different field regimes. In particular, contributions associated with $\Delta\kappa$ -type pinning may be present at low magnetic fields but remain unresolved in the Dew-Hughes analysis, as the normalized pinning force is primarily governed by mechanisms that are most effective at intermediate and high fields. As the magnetic field increases and the inter-flux-line spacing decreases, extended defects such as grain boundaries become increasingly efficient pinning centers within the Dew-Hughes framework, leading to the apparent dominance of surface normal pinning.

Furthermore, the depairing current density J_d , which represents the maximum supercurrent that can flow before Cooper pairs break, was estimated using Ginzburg-Landau theory:

$$J_d = \frac{4H_c}{3\sqrt{6}\lambda_{\text{GL}}}. \quad (16)$$

Using the calculated values of λ_{GL} and $\mu_0 H_c$ values, one can obtain $J_d \approx 21$ MA/cm². Although approximate, this estimate demonstrates that there is significant potential for enhancing J_c in YNbTiZrHf, as J_d exceeds the experimental J_c by more than 2 orders of magnitude. The collective pinning length L_c , estimated using the equation [33]

$$L_c \simeq \xi_{\text{GL}} \left(\frac{J_d}{J_c} \right)^{1/2}, \quad (17)$$

is approximately 100 nm, which is significantly larger than the coherence length $\xi_{\text{GL}} = 4.74(4)$ nm. This indicates that the system is in the weak collective pinning regime, which is characterized by strong suppression of the critical current density. To improve J_c , microstructural optimization strategies should be considered. Mechanical or thermal treatments, such as directional rolling, could improve current percolation paths and reduce weak links, while controlled heat treatments might introduce nanoscale inclusions or generate eutectic microstructures that could act as effective pinning centers [7,30].

The superconducting phase diagram of YNbTiZrHf is shown in Fig. 9, summarizing the experimentally determined dependencies $J_c(T)$, $J_c(\mu_0 H)$, $\mu_0 H_{\text{irr}}(T)$, and $\mu_0 H_{c2}(T)$. The irreversibility field $\mu_0 H_{\text{irr}}(T)$ curve was obtained using the empirically established relation [35]

$$H_{\text{irr}}(T) = H_{\text{irr}}(0) \left[1 - \left(\frac{T}{T_c} \right)^2 \right]^n, \quad (18)$$

with $\mu_0 H_{\text{irr}}(0) = 8.87(24)$ T and an exponent $n = 1.11(10)$.

IV. SUMMARY AND CONCLUSIONS

The new high-entropy alloy YNbTiZrHf was synthesized using the arc-melting method. Structural analysis confirmed a bcc matrix with minor precipitation of a hexagonal yttrium-rich secondary phase. Magnetization, electrical resistivity, ac magnetic susceptibility, and specific heat measurements confirmed bulk type-II superconductivity with a transition temperature close to 6.1 K and $\mu_0 H_{c2} = 12.5$ T, placing YNbTiZrHf among the HEAs with the highest known upper critical fields. The WHH analysis revealed that the Pauli limit is exceeded due to significant spin-orbit scattering.

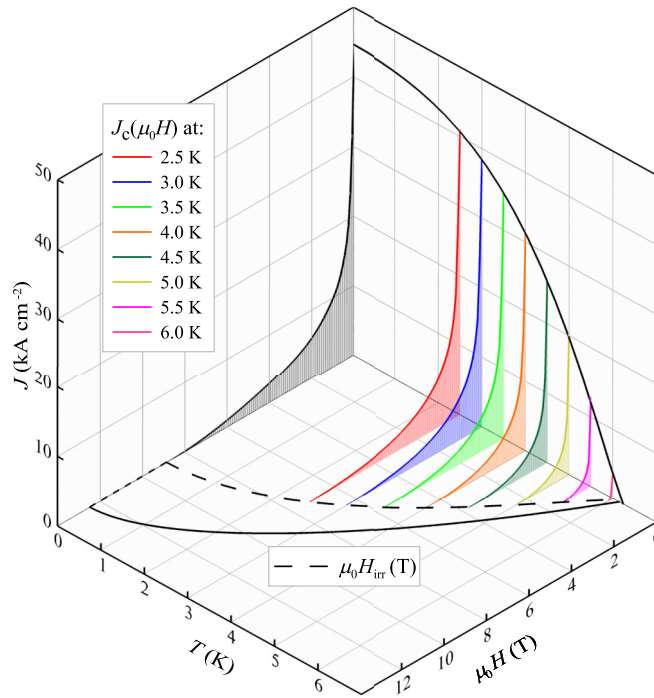


FIG. 9. Superconducting phase diagram of YNbTiZrHf determined from the collected experimental data.

Detailed analysis of the flux pinning behavior indicates a dominant contribution from surface pinning, with a secondary point pinning component, while collective pinning theory

reveals a crossover from δT_c -type to δl -type pinning with increasing field. These results highlight the interplay between chemical disorder, microstructure, and vortex dynamics in multicomponent systems.

Importantly, the combined $J_c(T)$, $J_c(\mu_0 H)$, $\mu_0 H_{irr}(T)$, and $\mu_0 H_{c2}(T)$ data enabled the construction of a complete superconducting phase diagram for YNbTiZrHf. This diagram not only delineates the stability region of the superconducting state but also provides a framework for comparing HEA superconductors.

The estimated $J_d \approx 21 \text{ MA/cm}^2$, which is more than 2 orders of magnitude higher than the experimental J_c , suggests that proper optimization strategies could dramatically improve the performance of HEA superconductors.

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DATA AVAILABILITY

The data presented in this study are openly available in the OSF repository [36] (subsection: Complete superconducting phase diagram and flux pinning mechanisms in a yttrium-containing high-entropy alloy YNbTiZrHf).

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