Dynamic deformation behavior of 93W-5.6Ni-1.4Fe heavy alloy prepared by spark plasma sintering

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Abstract: In this study, split Hopkinson pressure bar was used to evaluate the dynamic deformation behavior of the 93W–5.6Ni–1.4Fe heavy alloy (93WHA) prepared by spark plasma sintering (SPS) and conventional liquid-phase sintering (CLS). The influence of the microstructural characteristics (such as W grain size, W–W contiguity and volume fraction of the matrix) on the dynamic deformation behavior was investigated. In contrast to the conventional liquid-phase sintered 93WHA, the spark plasma sintered 93WHAs exhibit high yield strength and flow stress during high strain rate compression, due to the decreased mean matrix thickness (the mean matrix thickness is related to the W grain size, W–W contiguity and volume fraction of the matrix). The decreased matrix mean thickness and increased number of grain boundaries in the spark plasma sintered 93WHAs result in an increase of aspect ratio of W grains in the core of the deformed specimen and a decreased width of shear band along the direction of maximum shear stress.

Key words: tungsten heavy alloy; SPS; grain refinement; dynamic compression

1. Introduction

Tungsten heavy alloys (WHAs) are attractive candidate materials for using as kinetic energy penetrators to replace the depleted uranium alloys (DUAs), because they possess unique combination of properties including high density, excellent ductility, good corrosion resistance, good thermal

conductivity, as well as good formability and non-radiation pollution [1]. The conventional fabrication process for WHAs includes mixing the desired amount of elemental powders, followed by cold pressing and liquid-phase sintering at temperature above 1460 °C in hydrogen atmosphere to almost full density [2]. The WHA after liquid-phase sintering manifests a typical microstructure where spherical body-centered cubic (BCC) W grains of 30~50 μm are dispersed in the face-centered cubic (FCC) Ni-Fe-W solid solution matrix [3]. In general, the liquid phase sintered WHA exhibits 10~20% lower penetration performance than DUA at high strain rate [4,5]. Deformation strengthening was done to improve the dynamic mechanical properties of the WHA penetrators. Kim et al. [6] have reported that the swaged WHA exhibits a higher possibility of adiabatic shear band (ASB) formation under dynamic torsion than the sintered one. In the recent work of Liu et al. [7], WHA with fibrous W grains fabricated by hot-hydrostatic extrusion shows an anisotropic dynamic deformation behavior, i.e., the dynamic mechanical behaviors and susceptibility to ASB of the extruded WHA varied with the angles between the loading direction and the fiber orientation. When the loading axis is perpendicular to the fiber orientation, the extruded WHA exhibits highly localized shearing under uniaxial dynamic compression; however, when the loading axis is parallel to the fiber orientation, it only exhibit uniform plastic deformation [7]. In order to improve the susceptibility of the extruded WHA to ASB on the axial direction, hot torsion was adopted to change the fiber orientation of the extruded WHA [8,9]. However, there are two main shortcomings with deformation strengthening: first, the processing operations and costs increase; and second, the deformation processing can only change the aspect ratios of W grains to a limited extent, leading to a restrained increment of the strength, because of the coarse W grains.

Grain refinement has been proven to be another approach to improve the dynamic mechanical properties of the WHAs [10–12]. Wei et al. [10] have reported that ultra-fine grained tungsten fabricated via severe plastic deformation shows a significant flow softening and a reduction in strain rate sensitivity. The recent investigations of Gong et al. [11] and Fan et al. [12] show that fine grained WHAs (average tungsten grain size is about 6 μm) added with 0.03% (weight percent) of yttrium (Y) or yttrium oxide (Y_2O_3) exhibit low strain rate-sensitivity and increased plastic flow localized instability under lower high-strain-rate compression (strain rate is lower than $2 \times 10^3 \text{ s}^{-1}$). In order to refine the W grain size, three effective methods are used. The first method of grain refinement of WHA is to add other refractory metals such as molybdenum (Mo), rhenium (Re), tantalum (Ta), etc. [13-15]. The second route is to add inhibitors (such as oxide dispersoids) to prevent grain growth [12,16–19]. The last one is to adopt

advanced fast sintering techniques such as microwave sintering [20–22] and spark plasma sintering (SPS) [23–26]. Our previous work shows that SPS is an effective method to consolidate blended W-Ni-Fe powders, and 93W-5.6Ni-1.4Fe heavy alloys (93WHAs) with fine spherical W grains of about 6 μm have been fabricated by control of the SPS process [25]. The spark plasma sintered (SP-sintered) 93WHAs exhibit an increased bending strength and yield strength under quasi-static loading due to the fine microstructure, in contrast to the conventional liquid-phase sintered 93WHAs [25]. In this study, the SP-sintered 93WHAs were tested under uniaxial compression using split Hopkinson pressure bar (SHPB) to evaluate the dynamic deformation behavior. The effect of microstructure characteristics (W grain size, W–W contiguity, etc.) on dynamic deformation behavior is investigated in detail.

2. Experimental procedure

Commercial elemental powders of W, Ni, and Fe were selected as the experimental materials. The characteristics of the as-received powders are listed in Table 1. The WHA with the compositions 93.00W–5.60Ni–1.40Fe, by weight percent, was prepared from the starting powders. W, Ni, and Fe powders were weighed accurately to make up the desired compositions. The weighed powders were subsequently blended for 10 h in a V-shaped mixer. Details of the morphology of the blended powders can be found in our previous paper [25].

The blended 93W-5.6Ni-1.4Fe powders were consolidated by SPS (Dr. Sinter825, Sumitomo Coal Mining Co. Ltd., Japan) in a graphite die-punches assembly. Graphite foils were placed between the punches and the powder bed, and between the die and the powder bed for easy removal and reduction of temperature inhomogeneity. Consolidation was performed in vacuum (residual cell pressure of 6 Pa), and a constant pressure of 30 MPa was applied from the beginning of the heating step to the end of the dwell. The pulse sequence was 12:2. During SPS experiments, the heating from room temperature to 600 °C was controlled by a preset heating program and completed within 4 min; from 600 °C to the desired sintering temperature, heating rates ranged from 10 to 380 °C·min⁻¹. After SPS, samples of fine-grained 93W-5.6Ni-1.4Fe heavy alloys with sizes of 20 mm in diameter and 5 mm in thickness were obtained. In addition, 93WHA with coarser W grains (average grain size is about 28 μm) was prepared by conventional liquid-phase sintering (CLS) for comparison. The blended powders were compacted isostatically into a cylinder of 15 mm in diameter and 75 mm in height using a cold isostatic press (CIP32260, AVURE, USA) at a pressure of 250 MPa. The green compact was then liquid-phase sintered

in hydrogen atmosphere. The liquid-phase sintered 93WHA was heated at 1000 °C for 1 h in vacuum to dehydrogenation. The sintering schedules for both SPS and CLS are shown in Table 2. The prepared 93WHAs by both SPS and CLS manifested typical liquid-phase sintered microstructures (the morphologies of the 93WHAs were shown in our previous paper) [25]. The microstructural characteristics after sintering (including relative density, W grain size, W–W contiguity and dihedral angle at the intersection of the boundary between two W grains) of the prepared 93WHAs were shown in Table 3.

Uniaxial dynamic compression tests were performed using a split Hopkinson pressure bar (SHPB) under a strain rate of $5 \times 10^3 \text{ s}^{-1}$ at room temperature. The overall setup of the SHPB system used in this study is shown in Fig. 1. The test pieces of cylindrical shape with dimensions of $\varnothing 3 \text{ mm} \times 3 \text{ mm}$ were machined out from the as-sintered alloys by electrical discharge machining (EDM), and then they were sandwiched between the incident bar and transmitted bar. To reduce end friction on the specimens, the specimen ends were polished and greased with powdered graphite lubricant. In each case (under the same impact pressure), three sets of specimens were utilized to ensure the accuracy and recurrence of the measurements. The differences in obtained flow curves were negligible. After the tests, the impacted specimens were recovered and then diametrically sectioned parallel to the loading direction using EDM. The sectioned surface was polished and chemical etched with a solution of potassium ferricyanate (6 g) and potassium hydroxide (0.5 g) in distilled water (50 mL) to reveal the deformation microstructure by scanning electron microscopy (SEM, Quanta200, FEI Corporation, Hillsboro, OR).

3. Results

3.1. Flow stress–strain behaviors

Table 4 lists the dynamic compression mechanical properties of the prepared 93WHAs at strain rate of $\sim 5 \times 10^3 \text{ s}^{-1}$. The specimens except SPS0 after dynamic compression are all deformed. The yield strengths of the SP-sintered 93WHAs are about 2000 MPa, which are higher than 1800 MPa of the conventional liquid phase sintered 93WHA. Fig. 2 shows the uniaxial compression true stress–strain curves of the prepared 93WHAs at strain rates of 10^{-3} s^{-1} and $\sim 5 \times 10^3 \text{ s}^{-1}$. The flow stress-strain behaviors of the sintered 93WHAs under dynamic compressive loading are different from that compressed with quasi-static strain rate. It can be seen that the prepared 93WHAs show remarkable strain hardening after yielding at strain rate of 10^{-3} s^{-1} . However, under high strain rate ($\sim 5 \times 10^3 \text{ s}^{-1}$), the

specimens except SPS0 show elastic-nearly perfectly plastic behaviors (namely, the plastic parts of the stress–strain curves are nearly flat), which means that only slight strain softening occurs in the prepared 93WHAs. The flow stress of the SP-sintered 93WHA (except for the SPS0) is higher than that of the conventional liquid-phase sintered 93WHA. Moreover, the SP-sintered 93WHA exhibits an increased flow stress with the decrease of W grain size (W grain sizes of the prepared 93WHAs are shown in Table 3). It means that the microstructure has an important influence on the dynamic mechanical behavior of the prepared 93WHAs. On the other hand, sharp stress drop after yielding is observed in the stress–strain plot of the SPS0 specimen. It should be noted that the SPS0 specimen finally fractures into several fragments and the stress collapse is not a consequence of remarkable strain softening. The dynamic failure of SPS0 is not caused by plastic flow localized instability but it may be by the brittleness due to the high W-W contiguity. The detailed fracture analysis is shown in the next section.

3.2 Microstructure after dynamic compression

The prepared 93WHAs (except the SPS0) after dynamic deformation exhibited slight barreling shapes as shown in Fig. 3(a), though the end friction was minimized. The heights of the 93WHAs specimens after dynamic deformation are almost the same, being about 2.05 mm. The end friction results in an uneven distribution of stress, causing inhomogeneous plastic deformation within the specimen during dynamic compression. Fig. 3 (c)-(e) shows the cross-sectional microstructure of the CLS0 specimen after dynamic deformation, for instance. According to these microstructures, the prepared 93WHAs after dynamic deformation can be roughly divided into three deformation regions. Fig. 3(b) presents the schematic of nonuniform deformation of the cylindrical specimen during dynamic compression. Region I was below the center of the specimen end surfaces, in which the microstructure remains unchanged, i.e., the W grains stay round during dynamic deformation (Fig. 3 (c)). Region II located in the core of the tested specimen. In this region, the W grains were deformed to be subelliptical and arranged in the direction perpendicular to the axial pressure, as shown in Fig. 3 (d). Region III underwent high shear deformation, in which the W grains were fibered and rotated to the direction of shear stress (Fig. 3 (e)). Local shear banding often forms in this region [27]. The microstructural characteristics after sintering have a great effect on the microstructure of the region II and III, which is discussed in more detail in the succeeding chapters.

Fig. 4 shows the microstructure of the region II for the prepared 93WHAs under high strain rate

compression. It is clear that the W grain size influences the plastic deformation of the W grain when the strain is the same of ~ 0.38 . In order to determine the effect of W grain size on the deformation of W grains in the region II, a quantitative analysis is adopted by the aspect ratio of W grains base on assuming that the W grains in the region II after dynamic deformation are standardly elliptical. Fig. 5 shows the aspect ratio of W grains in the region II and the inset is the schematic of determination of the aspect ratio of elliptical W grains. The aspect ratio of W grains increases with the decrease of W grains. It means that W grain refinement promotes the plastic deformation of W grains in the region II during dynamic compression. Combining the experiments and FEM simulation, Muszka et al. [28] also found that in the core of the testing specimen plastic strain increased with the decrease of grain size when microalloyed steels were dynamic deformed.

Fig. 6 shows the microstructure of the region III for the prepared 93WHAs under high strain rate compression. Local shear bands were observed along orientation at about 45° to the compression direction or along the direction of maximum shear stress in the region III. The W grains are severely fibred within the shear band. Outside the band, nevertheless, W grains are not deformed as heavily as those within the band. Moreover, the deformation of W grains along the propagation direction of the shear band (i.e., the aspect ratios of W grains) increased with the decrease of W grain size. It has been demonstrated that local shear band width affects the penetration performance of WHA [3,4,7,8]. Local shear band with narrower width and more amounts of shear strain leads to an easier microcrack initiation, which is beneficial to the self-sharpening effect [12]. The boundary between the shear band and its outside materials is identifiable, which is indicated by two black straight lines in Fig. 6. We thus consider the distance between the two straight lines is the width of the shear band. Fig. 7 shows the shear band width of the prepared 93WHAs after dynamic compression. The shear band width in the CLS0 specimen after dynamic loading was about $110\text{ }\mu\text{m}$, and it then decreased with the W grain size in the SP-sintered 93WHAs. Yang et al. [29] reported that the propagation of local shear band could absorb more energy and restrict the growth of shear bands in the surrounding area. With a narrower shear band and more amount of shear strain, the energy exhausted in the surrounding area is less than that with wider shear bands. Therefore, the local shear band propagates more deeply into the interior with decreasing the W grain size, as shown in Fig. 6.

Fig. 8 presents the macrograph and fractographs of the SPS0 specimen after dynamic compression. The SPS0 specimen after dynamic compression was broken into pieces (Fig. 8 (a)). Observation on the

fracture surface shows that the SPS0 specimen was failed by W–W intergranular separation during high strain rate compression. The fractured W grains below the ends of the SPS0 are undeformed (Fig. 8 (b)); whereas the ruptured W grains in the core of the specimen are fibrous, as shown in Fig. 8 (c). Further, on the fracture surface at higher magnification, local W grains experienced shear deformation as indicated by the black arrows in Fig. 8 (d). It suggests that the SPS0 is first deformed and then fractured with increasing strain.

4. Discussion

According to Xu's work on the thermomechanical behavior and constitutive modeling of 93W–4.9Ni–2.1Fe heavy alloy over wide temperature and strain rate ranges [30], the KHL viscoplastic model (which is introduced by Khan, Huang and Liang [31,32]) shows a good description of room temperature flow stress of WHA at high strain rate. This model is also demonstrated to have the ability to describe dynamic deformation process of metallic materials with a wide range of grain sizes (from tens of nanometers up to hundreds of micrometers) [28,32]. The KHL model is expressed as

$$\sigma = \left(a + \frac{k}{\sqrt{d}} \right) \left[1 + B^* \left(1 - \frac{\ln \dot{\epsilon}}{\ln D_0^p} \right)^{n_1} (\epsilon^p)^{n_2} \right] \left(\frac{\dot{\epsilon}}{\dot{\epsilon}^*} \right)^C \left(\frac{T_m - T}{T_m - T_r} \right)^m \quad (1)$$

where σ is the flow stress; d is the grain size; $\dot{\epsilon}$ is the current strain rate; $D_0^p (= 10^6 \text{ s}^{-1})$ is the arbitrarily chosen upper bound strain rate; ϵ^p is plastic strain; $\dot{\epsilon}^* (= 1 \text{ s}^{-1})$ is the reference strain rate; T_m , T , and T_r is the melting temperature, current, and reference temperature in Kelvin, respectively; a , k , B^* , n_0 , n_1 , C , and m are material constants. The part of $(a + k d^{-0.5})$ in Eq. (1) is the classical Hall–Petch relation for conventional polycrystals with grain sizes ranging from several to hundreds of micrometers [32]. However, WHA is a typical double-phase material, in which the W phase is harder than the matrix phase. Ashby [33] has reported that under quasi-static loading, the yield strength of double-phase alloy is related to the inverse square root of average thickness of soft phase between hard phases because yielding of double-phase alloy begins by deformation of the soft phase. An adapted Hall–Petch relation for WHA was developed by Lee et al. [17,18] by assuming deformation of WHA began with deformation of the soft matrix phase, as expressed by the following formula:

$$\sigma_y = \sigma_0 + K \left[\frac{C_{WW} (1 - V_M)}{G_W V_M} \right]^{1/2} \quad (2)$$

where σ_y is the yield strength of WHA, σ_0 and K are constants, C_{WW} is the W–W contiguity, G_W is the W

grain size, V_M is volume fraction of the matrix and $G_W V_M / [C_{WW}(1-V_M)]$ is mean thickness of the matrix. Our previous study showed that the room temperature yield strength of the prepared 93WHAs under quasi-static compression agreed well with the Eq. (2) [25]. During high strain rate compression, the deformation of WHAs also starts with deformation of the matrix [11]. Hence, the KHL model for WHAs can be modified as follows:

$$\sigma = \left[\sigma_0 + K \left(\frac{C_{WW}(1-V_M)}{G_W V_M} \right)^{1/2} \right] \left[1 + B^* \left(1 - \frac{\ln \dot{\epsilon}}{\ln D_0^p} \right)^{n_1} (\epsilon^p)^{n_2} \right] \left(\frac{\dot{\epsilon}}{\dot{\epsilon}^*} \right)^c \left(\frac{T_m - T}{T_m - T_r} \right)^m \quad (3)$$

From the above formula, it is known that for the dynamic flow stress of WHA at a certain plastic strain is in proportion to the square root of the microstructural characteristics (or the inverse square root of mean thickness of the matrix) when the dynamic compression testing temperature and strain rate are constant. Fig. 9 shows the dynamic compression flow stress as a function of the square root of microstructural characteristics for the prepared 93WHAs at strain rate of $\sim 5 \times 10^3 \text{ s}^{-1}$. It is clear that the straight lines obtained by linear fitting are approximatively parallel, suggesting that the microstructural characteristics play an important role in dynamic deformation behavior of the prepared 93WHA. Furthermore, the amount of grain boundaries increases significantly with the decrease of W grain size. According to the work on dynamic behaviors of BCC metals with ultrafine grained and nanocrystalline microstructures by Wei et al. [10,34], the flow stress is dominated by the long-range stress that arises from the interaction between dislocations and abundant grain boundaries.

The microstructural characteristics or the mean matrix thickness affect not only the flow stress of the prepared 93WHAs but also the microstructure after dynamic compression. The mean matrix thickness δ of the prepared 93WHAs (except for the SPS0 specimen) is shown in Fig. 10, according to the formula $\delta = G_W V_M / [C_{WW}(1-V_M)]$. The mean matrix thickness decreases with the W grain size. The plastic deformation of WHA starts with deformation of the matrix, as previously stated. Along the direction of maximum shear stress in the region III, shear localization occurs first within the matrix phase and it then creates a path by deforming nearby tungsten grains to form a shear band with the plastic strain increasing. Therefore, with finer matrix thickness, it is easier to establish a path through the tungsten grains to form a narrow shear band. Furthermore, during dynamic compression plastic work is converted mostly into heat, causing temperature increase in material. The temperature increase within the deformed matrix diffuses to the neighboring W grains. A simple but close estimate of the characteristic time Δt for thermal diffusion through the plastic deformed matrix is $\delta^2 / (2\alpha)$ [10], where α is the thermal

diffusivity of the matrix ($\alpha = \lambda/(\rho \cdot c) = 0.285 \text{ cm}^2 \cdot \text{s}^{-1}$, where the density ρ , thermal conductivity λ and specific heat c for the matrix are $9200 \text{ kg} \cdot \text{m}^{-3}$, $100 \text{ W} \cdot (\text{m} \cdot \text{K})^{-1}$ and $382 \text{ J} \cdot (\text{kg} \cdot \text{K})^{-1}$, respectively [35]). Fig. 11 shows that the characteristic time Δt for thermal diffusion through the deformed matrix declines steeply from the CLS0 to the SPS specimens. The estimated characteristic time for thermal diffusion is much smaller than the elapsed time during dynamic loading (typically $100 \text{ } \mu\text{s}$). It is therefore reasonable to believe that adequate time for the generated heat transfers from the deformed matrix to the neighboring W grain is allowed with the plastic strain increasing. In the SP-sintered 93WHAs, an increased number of grain boundaries make the heat tend to concentrate in a narrow region due to the increased phonon scattering effect and the fine W grains do not require excessively high heat for softening due to their smaller volume [11]. As a result, the deformation of W grains in region II increases with the decrease of the W grain size and the shear band width in region III decreases with the W grain size, depending on the thermal conduction length scale. On the other hand, the microstructure of the prepared 93WHAs after dynamic compression in this study is different from that of fine-grained 93W–4.9Ni–2.1Fe–0.03Y alloy under high strain rate compression reported by Fan and Gong, et al. [11,12], in which the plastic strain is highly localized at the direction of 45° to the impact direction to form narrow ASBs, resulting in plastic flow localized instability. The main reason for this discrepancy is that the in situ formed Y_2O_3 particles in 93W–4.9Ni–2.1Fe–0.03Y alloy act as the nucleation sites for ASBs during high strain rate compression [11].

5. Conclusion

In this paper, dynamic deformation behaviors of the 93W–5.6Ni–1.4Fe heavy alloys prepared by SPS and CLS were evaluated using split Hopkinson pressure bar. The relationship between the microstructural characteristics (W grain size, W–W contiguity, etc.) and the dynamic deformation behavior was studied. The results are summarized as follows:

(1) Compared with conventional liquid-phase sintered 93WHA (CLS0), the SP-sintered 93WHA exhibit high yield strength and flow stress during high strain rate compression, due to the fine-grained microstructure. The dynamic mechanical behavior of the 93WHAs is dependent on the microstructural characteristics (W grain size, matrix volume fraction and W–W contiguity) or the mean matrix thickness.

(2) After dynamic deformation, the W grains in the core of the tested specimen were deformed to be subelliptical and arranged in the direction perpendicular to the loading direction (region II). Local shear

band initiated and propagated along orientation at about 45° to the compression direction. Within the shear band, the W grains were severely fibered along the propagation direction of the shear band (region III).

(3) The microstructural characteristics affect the dynamic deformation microstructure of the 93WHAs. In contrast to the conventional liquid-phase sintered 93WHA (CLS0), a decreased mean thickness of the matrix and an increased number of grain boundaries in the SP-sintered 93WHA lead to an increase of aspect ratio of W grains in region II and a decrease of shear band width.

Acknowledgments

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Fig. 1. Overall setup of the SHPB system used in this study.

Fig. 2. Compression true stress–strain curves of the prepared 93WHAs at strain rates of (a) 10^{-3} s^{-1} and (b) $\sim 5 \times 10^3 \text{ s}^{-1}$.

Fig. 3. (a) Overalls and heights of the specimens after dynamic compression; (b) schematic of nonuniform deformation of the cylindrical specimen during dynamic compression; SEM images of the CLS0 specimen after dynamic compression at a strain rate of $\sim 5000 \text{ s}^{-1}$: (c) region I, (d) region II and (e) region III. The inset in (e) is an enlarged image showing the shear band structure in region III.

Fig. 4. Microstructure in the region II of the prepared 93WHAs after dynamic compression at a strain rate of $\sim 5000 \text{ s}^{-1}$:

(a₁) and (a₂) CLS0; (b₁) and (b₂) SPS1; (c₁) and (c₂) SPS3; (d₁) and (d₂) SPS5.

Fig. 5. Aspect ratio of W grains in the region II. Inset is the schematic of determination of the aspect ratio of W grains in the region II.

Fig. 6. Microstructures in the region III of the prepared 93WHAs after dynamic compression:

(a₁) and (a₂) CLS0; (b₁) and (b₂) SPS1; (c₁) and (c₂) SPS5. The strain rate is constant at $\sim 5000 \text{ s}^{-1}$.

Fig. 7. Shear band width of the prepared 93WHAs after dynamic compression. The strain rate is constant at $\sim 5000 \text{ s}^{-1}$.

Fig. 8. (a) Macrograph and (b), (c), (d) fractographs of the SPS0 specimen after dynamic compression at strain rate of $\sim 5000 \text{ s}^{-1}$. Black arrows indicate the W grains underwent shear deformation.

Fig. 9. Flow stress (σ) vs. square root of microstructural characteristics ($[C_{ww}(1-V_M)/(G_W V_M)]^{0.5}$) for the prepared 93WHAs dynamic compressed at a strain rate of $\sim 5 \times 10^3 \text{ s}^{-1}$.

Fig. 10. The mean matrix thickness of the prepared 93WHAs.

Fig. 11. The characteristic time Δt for thermal diffusion through the deformed matrix in the prepared 93WHAs.

Table 1 Characteristics of the powders used to prepare 93WHAs.

Elemental powder	Fabrication method	Purity	Particle size (μm)	Main impurity (wt %)			
				C	O	P	N
W	Oxide reduction	0.9992	2.4-2.6	0.002	0.07	0.0008	–
Ni	Carbonyl process	0.9950	2.2-2.5	≤ 0.25	≤ 0.11	0.0003	–
Fe	Carbonyl process	0.9950	3.0-5.0	≤ 0.1	≤ 0.3	–	≤ 0.1

Table 2 Sintering schedules to consolidate the blended powders [25].

Sintering methods	Specimens	Sintering conditions		
		Heating rate(°C·min ⁻¹)	Reference temperature (°C)	Dwell duration (min)
SPS	SPS0	10	1430	0
	SPS1	20	1420	0
	SPS2	50	1410	0
	SPS3	100	1400	0
	SPS4	200	1380	0
	SPS5	380	1360	0
CLS	CLS0	10	1480	90

Table 3 Microstructural characteristics of the prepared 93WHAs [25].

Specimen	Relative density	Tungsten grain size (μm)	W-W contiguity	Dihedral angle ($^{\circ}$)
CLS0	0.987 ± 0.003	28.4 ± 8.8	0.496 ± 0.049	44.3 ± 24.5
SPS0	0.956 ± 0.005	14.0 ± 4.5	0.632 ± 0.069	59.4 ± 37.7
SPS1	0.950 ± 0.004	10.9 ± 3.5	0.571 ± 0.102	57.2 ± 32.7
SPS2	0.950 ± 0.005	8.1 ± 2.5	0.541 ± 0.101	56.5 ± 26.6
SPS3	0.954 ± 0.002	7.1 ± 2.3	0.521 ± 0.078	55.0 ± 25.6
SPS4	0.951 ± 0.004	6.1 ± 1.9	0.530 ± 0.090	55.2 ± 30.0
SPS5	0.948 ± 0.005	6.0 ± 2.2	0.532 ± 0.066	57.7 ± 24.7

Table 4 Dynamic compression mechanical properties of the prepared 93WHAs at strain rate of $\sim 5 \times 10^3 \text{ s}^{-1}$.

Specimen	Strain rate (s^{-1})	Yield strength(MPa)	Tested
CLS0	5000	1800	deformed
SPS0	5200	2000	collapsed
SPS1	5000	2000	deformed
SPS2	5100	1950	deformed
SPS3	5200	2000	deformed
SPS4	5200	2100	deformed
SPS5	5000	2000	deformed