

A comprehensive review on the deformation behavior of refractory high entropy alloys at elevated temperatures

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Abstract. Thermo-mechanical processing of refractory high entropy alloys (RHEAs) at high temperatures is very important. It is an effective method of modifying the microstructure, properties, and shaping into final components after casting. Using the Scopus database, 57 articles relating to the hot deformation of refractory high entropy alloys were extracted from 2011 to 2022. Despite the limited number of articles on hot deformation of RHEAs, it is important to find out if the dominant softening mechanisms reported in other metallic alloys are evident. This is the main impetus for this study since the hot deformation behavior has not been comprehensively studied. All the probable mechanisms influencing deformation in metallic alloys, such as work hardening, dynamic recrystallization, and dynamic recovery, have also been observed in RHEAs. The bulging phenomenon, serrated grain boundaries, and necklace-like structures reported in metallic alloys have also been detected in hot deformed RHEAs. Unsafe deformation behavior such as cracks that have been reported in metallic alloys, have also been observed in RHEAs. This review has provided a comprehensive study on the hot working processes of RHEAs and highlighted critical gaps for future research direction with some suggested limitations.

Keywords: Softening mechanism / refractory high entropy alloy / hot deformation / flow stress / microstructural evolution

1 Introduction

The world's technological growth and development drive a persistent demand for new materials having excellent mechanical properties to fulfil high temperature structural applications. In 2004, two different research groups showed that new alloy systems with simple microstructures could be synthesized from multi-principal elements in equimolar or near-equimolar ratios [1–4]. These new alloy systems have come to be known as high entropy alloys (HEA) [1] or multi-component alloys (MCA) [4,5] or compositionally complex alloys [6]. They are different from other known engineering alloys because they comprise at least five main elements or multi-principal elements at equiatomic percentages [1,7]. Some researchers have questioned the

original definition of HEA as alloys comprising at least five elements with equiatomic percentages. These researchers reported that some HEAs can be developed with three or four elements, and the maximum element concentration may be more than 35%. These sets of alloys such as CoCrNi [8]; NiFeCoCr [9]; MoReRuRh [10]; ScGdHo [11] and AlFeTiV [12] have been called medium entropy alloys. In addition, many HEAs may contain intermetallic compounds and complex phases, which have major effects on the microstructures and properties of the materials. HEAs possess combination of properties such as high strength, high hardness, good ductility, outstanding wear resistance, good oxidation and corrosion resistance [13–15].

Refractory high entropy alloys (RHEAs) which have high-melting-point refractory elements, were introduced in 2010 as potential candidates for high temperature functional materials due to their high hardness, wear resistance, and high temperature strength above 1100 °C

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when compared with Inconel 718 and Haynes 230 [16,17]. These characteristics are the result of increased melting temperature and solid solution strengthening [13,18]. The main potential applications of refractory high entropy alloys include tribology and thin film [19,20], hydrogen storage [21–24], nuclear materials [25–27], surface and coatings technology [28–30], and hard particles for wear resistance [31]. To optimise the behaviour of RHEAs at high temperatures for structural and functional applications, knowledge of the flow behaviour, microstructural evolution, and deformation mechanisms is important for determining their optimum thermo-mechanical processing parameters. According to Eleti [32], thermo-mechanical processing (TMP) at high temperatures is an effective method of modifying the microstructure and properties of RHEAs after casting. It is also used for shaping alloys into final components. TMP such as hot deformation, results in microstructural improvement, which often promotes strength and ductility of the material [33]. TMP describes the processes that combine plastic deformation of metallic materials with heat treatments. These processes are typically carried out at temperatures higher than 0.4 of the melting temperature (T_m) [34]. During hot deformation, a metal or alloy deforms under conditions of high temperature (greater than $0.6 T_m$) and different strain rates of 10^{-1} to 10^{-3} s^{-1} [35,36].

The plastic deformation at such high temperatures frequently results in simultaneous changes in microstructures [34]. Deformation mechanisms of work hardening (WH), dynamic recovery (DRV), and dynamic recrystallization (DRX) are examples of mechanisms that can cause microstructural changes in metals and alloys [37,38]. Comprehensive understanding of the flow behaviour of metals and alloys and their microstructures under hot deformation conditions is crucial during metal forming processes such as forging, hot rolling, and extrusion. Currently, studies on refractory high entropy alloys have at least 700 scientific papers based on the Scopus database. Among those papers are the high temperature deformation behaviour of RHEAs [32,39–41]. Recently, some reviews on hot deformation behavior of high entropy alloys were published [42–46]. However, review papers regarding the hot deformation behaviour of RHEAs are still limited and not comprehensive. For example, Savaedi et al. [42] focused on the deformation mechanism of HEAs at high temperatures, but only one RHEA was found, highlighting microstructural evolution. In George et al. [47] high temperature compressive behaviour of HEAs, high temperature compressive behaviour was reported for only one RHEA.

This review paper will highlight some important findings on the hot deformation behaviour of RHEAs. Emphasis on flow stress, flow softening mechanisms, and microstructural evolution in deformed RHEAs. This review is structured as follows: Section 2 discusses the method for selecting the publications related to the scope of this study. It also describes the trend of scientific publication and citation analysis of hot deformed RHEAs. The flow stress behaviour of RHEA during hot deformation is presented in Section 3. The section also examines the occurrence of discontinuous dynamic recrystallization and continuous dynamic recrystallization, sharp stress drop,

and steady-state flow processes in hot deformed RHEAs. Microstructural evolution in hot deformed refractory high entropy alloys is reported in Section 4. Dislocation motion mechanisms in hot deformed RHEAs are reported in Section 5. Section 6 depicts unsafe deformation behavior in hot deformed RHEAs. Section 7 illustrates constitutive modelling and processing maps in hot deformed RHEA. Section 8 contains discussion on the yield strength and compressive strength of reported RHEA during hot deformation. This review closes with a summary of this work (Sect. 9) and future research directions (Sect. 10).

2 Method for selecting publications considered in this review

Published literature was assessed to determine the dominant mechanisms influencing the hot deformation of RHEAs. In the selection phase for the databases for the review, Scopus and Web of Science were considered due to their popularity and efficiency [48]. After a preliminary search in both databases, Scopus was chosen. It has more published articles related to the topic and most articles on Web of Science are also indexed in Scopus. All the journal articles, conferences and monographs were searched and retrieved from the Scopus database using the following keywords: “hot deformation*” AND “refractory high entropy alloy*” OR “high temperature*” AND “refractory high entropy alloy*” OR “hot compression*” AND “refractory high entropy alloy*” OR “high temperature compression*” AND “refractory high entropy alloy*” OR “high temperature compression*” AND “refractory multi-component alloy*” OR “high temperature mechanical properties*” AND “refractory high entropy alloy*”. These keywords were utilised in searching the title, abstract and keyword field which produced 115 papers, but only 57 papers were related to the scope of this study. These 57 documents, comprising of mostly journal articles, were used for the review of the deformation behavior of RHEA.

2.1 Trend of scientific publication on hot deformation of RHEAs from 2011 to 2022

The annual scientific outputs on the hot deformation of RHEAs are shown in Figure 1. According to Scopus, the first article on the hot deformation of RHEAs was published in 2011. There were 2 articles published in that year. However, no publications were found in 2013 relating to hot deformation. This can be attributed to the area of focus within this period, which is mainly on alloy composition search, modelling and simulation of RHEAs. Also, RHEA contains expensive alloying elements that might make it difficult for researchers to explore in much detail considering the cost of making sufficient alloys for hot deformation testing. The publications gradually increased to five from 2014 to 2016 and later dropped to zero in 2017. In 2017, most publications on RHEA are centered on elemental alloying to improve mechanical properties, RHEA coating, oxidation studies of RHEA. Thereafter, in 2018, publications gradually increased to 5

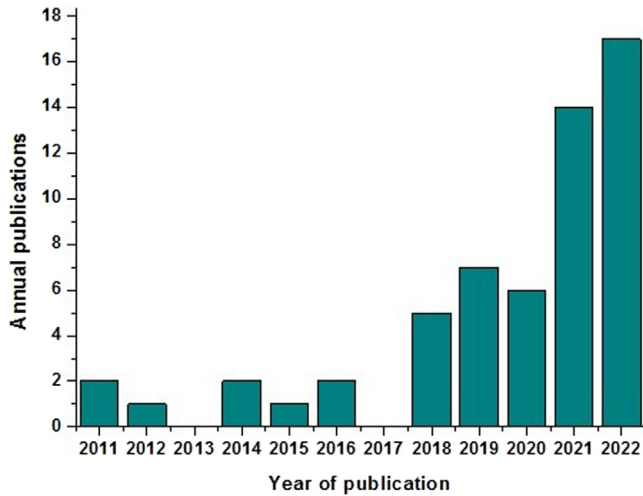


Fig. 1. Annual scientific outputs on hot deformation of RHEAs.

before increasing significantly to 49 from 2018 to 2022. The slight drop in 2020 might be due to the COVID-19 pandemic leading to lockdown and many restrictions which reduced research activities all over the world. In 2022, 17 articles have been published. It is projected that this increase will continue as researchers develop more RHEAs and conduct thermo-mechanical processing.

To determine the publishers with the most publications and citations, consideration was given to publications with at least five citations. Consequently, 39 articles meet these criteria and are shown in Table 1. In Table 1, Mater Sci Eng A had 7 publications with 220 citations to rank first. This is followed by the J Alloys Compd with 6 publications and 108 citations. The next are Mater Lett, Intermetallics, Acta Mater and the Int J Refract Met Hard Mater. Put together, these four journals have a combined total of 15 articles and 1887 citations on the deformation of RHEAs. Table 2 shows the publication analysis per country, with 11 countries contributing to the study on hot deformation of RHEA. Some countries overlap on the same article because the authors collaborated from different countries. From Table 2, China has the highest number of publications (34) and citation counts (616). This is followed by the United States (8 publications and 2313 citations), Japan (4 publications and 108 citations), and the Russian Federation (4 publications and 15 citations). Table 3 shows the citation analysis of the 10 most influential articles on the hot deformation of RHEA. The minimum number of citations for a document is selected to be 50, with 10 of the 50 articles meeting this criterion. The most influential article in terms of citation count is Senkov et al. [49] with 1347 citations. This is not strange because pioneering research articles usually have the highest citation counts as newer articles continue to refer to them.

Next are Senkov et al. [50] with 440 citations and Juan et al. [51] with 262 citations. Completing the top five most influential are two articles by Senkov et al. [52–53] with 255

Table 1. Analysis of the top research journals.

Journal and where it was published	Number of articles	Citations
J Alloys Compd	6	108
Int J Refract Met Hard Mater.	3	13
Mater Lett	5	114
J Mater Sci Technol	2	17
Mater Sci Eng A	7	220
Mater Des	2	22
Scr Mater	1	13
Mater Charact	1	9
Intermetallics	4	1652
J Mater Sci	1	440
Int J Mech Sci	1	17
Acta Mater	3	108
Entropy	1	39
Materials	1	22
JOM	1	226

Table 2. Publication analysis per country.

Country	Number of articles	Citations
China	34	616
United States	8	2313
Russian Federation	4	15
Japan	4	108
India	3	37
Germany	3	11
Taiwan	2	299
Ukraine	1	12
Australia	1	2
Iran	1	7
South Korea	1	3

and 226 citations. The top five most cited articles were published in 2011, 2012, and between 2014 and 2016. More details from Table 3 showed that four of the top five articles that made the top five were from Oleg Senkov, with a combined citation of 2268. Studies on the hot deformation of RHEA in the last decade need to be reviewed so that critical gaps can be identified, and this may draw the attention of more researchers into the field. Despite the limited number of articles on the hot deformation of RHEAs, it is interesting to find out if the dominant deformation mechanisms reported in other metallic alloys are evident in deformed RHEAs. This is the main impetus for reviewing these articles. The next section of the review paper will cover the occurrence of DRV and DRX in the flow curves of hot deformed RHEAs.

Table 3. Top 10 most influential studies on hot deformation of RHEAs.

First author and Year	Article title (doi)	Journal	Citations
Senkov et al. [49]	10.1016/j.intermet.2011.01.004	Intermetallics	1347
Senkov et al. [50]	10.1007/s10853-012-6260-2	J Mater Sci	440
Juan et al. [51]	10.1016/j.intermet.2015.03.013	Intermetallics	262
Senkov et al. [52]	10.1016/j.msea.2011.09.033	Mater Sci Eng A	255
Senkov et al. [53]	10.1007/s11837-014-1066-0	JOM	226
Guo et al. [54]	10.1016/j.msea.2015.10.113	Mater Sci Eng A	66
Eleti et al. [32]	10.1016/j.actamat.2019.04.018	Acta Mater	54
Zhang et al. [55]	10.1016/j.matlet.2016.03.092	Mater Lett	54
Guo et al. [56]	10.1016/j.jallcom.2018.10.230	J Alloys Compd	51
Xu et al. [57]	10.1016/j.msea.2019.03.054	Mater Sci Eng A	50

3 Flow stress behavior of RHEA during hot deformation

Dynamic recovery and dynamic recrystallization are desirable softening mechanisms during the deformation of metallic alloys. The DRV is one of the processes caused by the rearrangement of dislocations during deformation, which results in decreased strain hardening [34]. The DRV flow stresses have simple shapes, and there is strain hardening occurring at low strains followed by steady-state flow at high strains. In the case of DRX, it occurs during straining if the temperature is above $0.5T_m$ [33]. When DRX occurs, the flow stress is reduced and the materials' ability to deform is increased. DRX occurrence is established by a hump in the flow stress due to the softening effect of DRX [58]. It can also be specified by broad peaks and/or oscillations prior to the steady-state flow stress. The mechanisms can be classified into two types: continuous dynamic recrystallization CDRX and discontinuous dynamic recrystallization DDRX [57]. CDRX is characterised by a broad peak stress in the stress-strain curve, while DDRX has a single peak or multiple peaks in the stress-strain curve. The flow stress curves signifying softening mechanisms of Dynamic recovery and dynamic recrystallization have been observed in metallic alloys. For example, in the nickel-based superalloy (GH4169), there was a drop in flow stress related to the dynamic recrystallization softening mechanism at 980–1040 °C and strain rates (0.01 – 0.001 s^{-1}) [59]. For the $\beta 21\text{S}$ alloy, the flow stress stays steady once it has reached its peak signifying dynamic recovery [60]. At 1100 °C and strain rates of 10^{-3} s^{-1} , the flow curves of IN718 alloy revealed a dynamic recrystallization mechanism. When the strain rates decrease to 10^{-3} s^{-1} , the dynamic recrystallization occur during the hot deformation of TB8 titanium alloy [61]. The flow stress of hot deformed C276 superalloy increases sharply with strain until a peak flow stress because of WH at the start of deformation, and then the flow stress reduces gradually with the strain because of dynamic recrystallization.

The flow stress behaviour of refractory high entropy alloys, especially the yield-drop like phenomena has been reported in some literatures [50,54,62,63]. In 2011, Senkov

et al. [49] studied the high temperature compression of $\text{Nb}_{25}\text{Mo}_{25}\text{Ta}_{25}\text{W}_{25}$ and $\text{V}_{20}\text{Nb}_{20}\text{Mo}_{20}\text{Ta}_{20}\text{W}_{20}$ RHEAs at 600–1600 °C with a strain rate of 0.001 s^{-1} . At 600 and 800 °C, $\text{Nb}_{25}\text{Mo}_{25}\text{Ta}_{25}\text{W}_{25}$ RHEA undergoes continuous strengthening, whereas at higher temperatures (1000–1400 °C) it reaches a high stress at an intermediate strain [49]. According to Senkov et al. [49], the strain softening at these temperatures is an indicator of plastic uncertainty and spallation of the surface material observed at larger strains. At 1600 °C, nearly steady state flow at a fixed stress of ~ 590 – 600 MPa was observed, and so the yield stress for $\text{Nb}_{25}\text{Mo}_{25}\text{Ta}_{25}\text{W}_{25}$ RHEA steadily decreased as a function of temperature. For the $\text{V}_{20}\text{Nb}_{20}\text{Mo}_{20}\text{Ta}_{20}\text{W}_{20}$ RHEA deformed at temperatures (1400 and 1600 °C), a fast decline in yield stress and obvious softening was observed. The flow stress in most metals and alloys decreases with the increase in deformation temperature and the decrease in strain rate [64–67]. Also, when there is a reduction in the strain rate or an increase in the deformation temperature, the critical strain for dynamic recrystallization is decreased. Eleti et al. [32] observed a distinct sharp stress drop around yielding, followed by continuous flow softening, while deforming a HfNbTaTiZr RHEA at 1000–1200 °C, strain of 0.69, and strain rate of 10^{-2} to 10^{-4} s^{-1} . The sudden drops of the flow stress are visibly detected at initial stages of hot deformation (Fig. 2a). To reveal the mechanism of sharp stress drop, Eleti et al. [32] performed an interrupted deformation test at 1000 °C, held for 600 s, and then deformed at a strain rate of 0.001 s^{-1} (Fig. 2b). The unusual stress drop in the HfNbTaTiZr RHEA is due to either dislocation unlocking by solute atoms or the ending of short-range order by dislocation motion. Short-range order can have a profound effect on the stacking-fault energy and dislocation motion, which have noticeable influence on the plastic deformation of HEAs [68]. Similarly, in 2022, Li et al. [69] observed a sharp stress drop in $\text{Ti}_2\text{ZrHfV}_{0.5}\text{Ta}_{0.2}$ RHEA at 900–1100 °C and a strain rate of 10^{-2} to 10^{-1} s^{-1} (Figs. 2c–e). To reveal the mechanism of sharp stress drop, Li et al. [69] performed an interrupted hot deformation and strain aging experiment (Fig. 2f) at 900 °C/ 0.1 s^{-1} with a strain of 0.1 and held for intervals, 0 s and 600 s. The sharp stress drop was also attributed to the unlocking of dislocations pinned by the Cottrell atmosphere or short-range order. The unusual yield-drop phenomena in RHEAs are commonly observed in alloy compositions

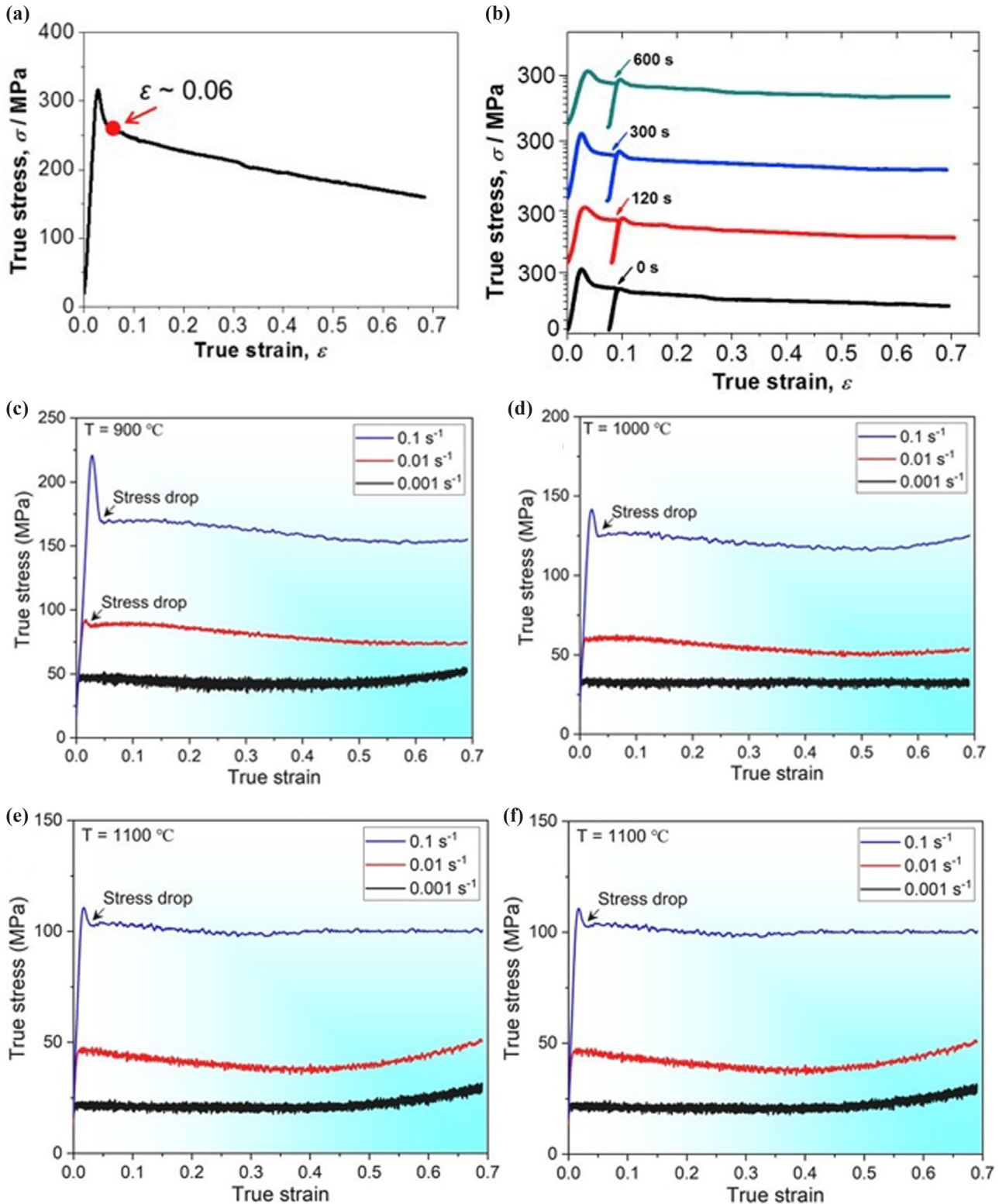


Fig. 2. (a) Sharp stress drop in hot deformed HfNbTaTiZr RHEA at $1000^\circ\text{C}/10^{-3}\text{s}^{-1}$ and strain of 0.06 [32]. (b) Interrupted test on hot deformed HfNbTaTiZr RHEA at $1000^\circ\text{C}/10^{-3}\text{s}^{-1}$ and a strain of 0.06 and then restarted after holding for 0 s, 120 s, 300 s, or 600 s [32]. (c-e) Stress-strain curves of $\text{Ti}_2\text{ZrHfV}_{0.5}\text{Ta}_{0.2}$ RHEA at different strain rates (10^{-1} to 10^{-3}) and temperatures (900–1000 °C) [69]. (f) Stress-strain curve of the strain aging experiment on the $\text{Ti}_2\text{ZrHfV}_{0.5}\text{Ta}_{0.2}$ RHEA at $^\circ\text{C}/10^{-1}\text{s}^{-1}$ through to the strain of ~ 0.1 for three times, with intervals between the first and third hot deformation of 0 s and 600 s [69].

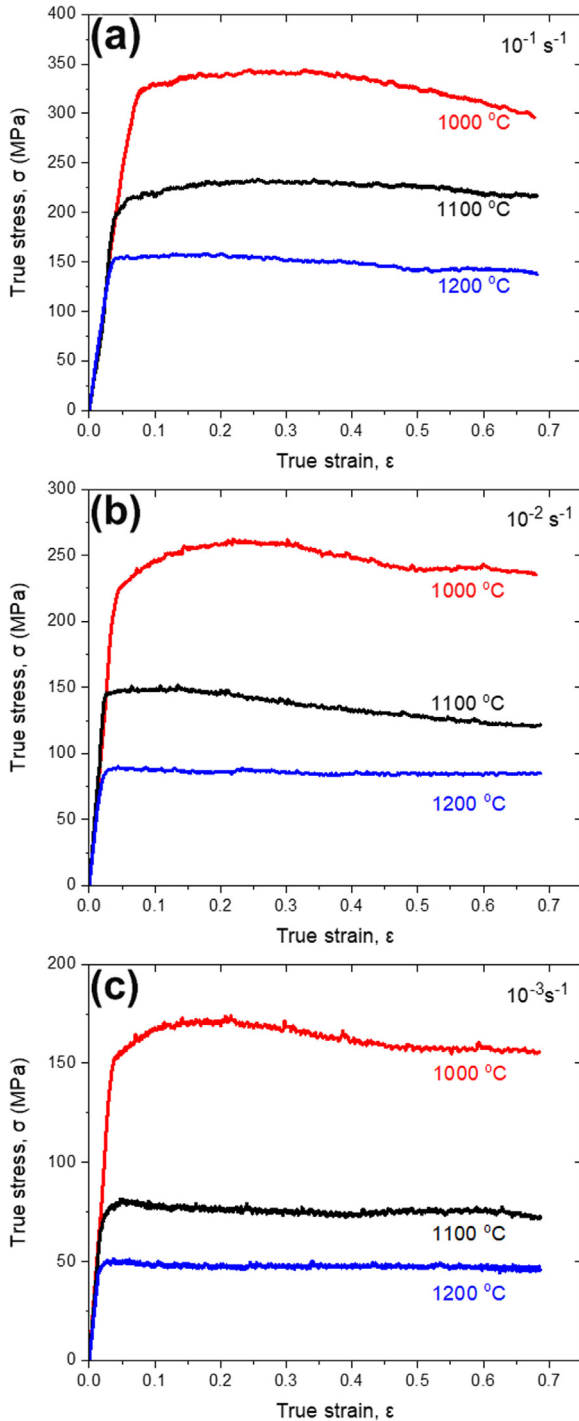


Fig. 3. Flow stress curves of the TiAlVNb₂ RHEA at different deformation temperatures of 1000–1200 °C and strain rates of 10^{-1} to 10^{-3} s⁻¹ [72].

comprising TiZrHfVTa [69], AlNbTaTiZr [70], HfNbTaTiZr [30], AlNbTiVZr [71], TaNbHfZrTi [48], and MoNbHfZrTi [54].

Bai et al. [72] investigated the high temperature study of a TiAlVNb₂ RHEA at strain rates from 10^{-3} to 10^{-1} s⁻¹ and temperatures from 1000 to 1200 °C. The flow stress of the TiAlVNb₂ RHEA at the start of deformation

show an obvious work hardening phenomenon (Fig. 3). This is followed by flow stress drops due to temperature increase or decrease in strain rate, which is commonly linked to more thermal activation [73]. The increased temperature leads to the quicker movement of grain boundary and dislocation, thereby showing softening and stress decreasing. In the event of a decrease in strain rate, the deformation time will be delayed, and the dislocation stacking degree will be reduced, resulting in a flow stress decrease. Bai et al. [72] also found that at strain rates (0.1, 0.01 and 0.001) and temperatures (1000–1100 °C), the flow stresses in some curves display decreasing trends after peak flow stress. The decreasing trend can be attributed to DRX taking place in the TiAlVNb₂ RHEA. Similarly, the flow stress of deformed TiAlVNb₂ RHEA remains stable after peak flow stress at 0.1 s⁻¹/1100 °C, 0.001 s⁻¹/1100 °C, 0.1 s⁻¹/1200 °C, 0.01 s⁻¹/1200 °C, and 0.001 s⁻¹/1200 °C (Fig. 3). After deformation, some flow curves of some alloys could show fluctuations (Fig. 3). For example, in Figure 4, Ni₃Al-based superalloy hot compressed at 1050–1250 °C with a strain rate of 0.1 s⁻¹. The fluctuations, according to Wu et al. [74], can be attributed to an alternation between the effect of WH and the dynamic softening, which is caused by the fast-growing grains at high temperatures and DRX.

The hot deformation characteristics of MoNbHfZrTi RHEA were studied by Guo et al. [54] at a strain rate of 10^{-3} – 10^{-1} s⁻¹ and temperatures of 800–1200 °C. The stress–strain curve revealed WH before the sample fractured at 800 °C when the strain rate is 10^{-1} s⁻¹ and 10^{-3} s⁻¹ (Fig. 5). A peak stress with increasing strain was identified at a strain rate of 10^{-1} s⁻¹, and then the stress decreased as the strain was further increased at 900 °C (Fig. 5a). Whereas at low strain rates (0.01 s⁻¹ and 0.001 s⁻¹), peak stress was identified with increasing strain but then decreased and slowly reached a stable value as the strain increased. Flow curve characteristics, as identified by Guo and co-workers, belong to dynamic recrystallization occurrence during hot compression of alloys with low SFE [75]. The stress decreases with the increasing strain since the effect of WH can be neutralised or partly neutralised by the formation of the dynamic softening mechanism such as DRV and DRX. In general, for the hot deformation mechanism of MoNbHfZrTi RHEA, both the DDRX and CDRX occur concurrently. This is evident in Figure 5 where the serrated flow stress shows the DDRX and the broad peak stress shows the CDRX.

In 2021, Wei et al. [76] observed work hardening after yielding in W_{0.4}MoNb_xTaTi ($x=1.1, 1.3$, and 1.5) RHEA hot deformed at strain rates of 0.001 s⁻¹ and temperatures of 800, 1000, and 1100 °C. At 800 and 1000 °C, a decrease in the yield stress was seen and related to the energy provided by the high temperature. After reaching the yield point, Wei et al. [76] indicated a distinct steady-state flow stages which were seen on the true stress–strain curves. During hot deformation of mechanical alloyed and spark plasma sintered MoNbTaTiV RHEA, Liu et al. [77] observed that flow stresses at temperatures (1100 and 1200 °C) exhibit similar trends. Also, the flow stresses are slowly reduced after attaining the maximum compressive stress when the

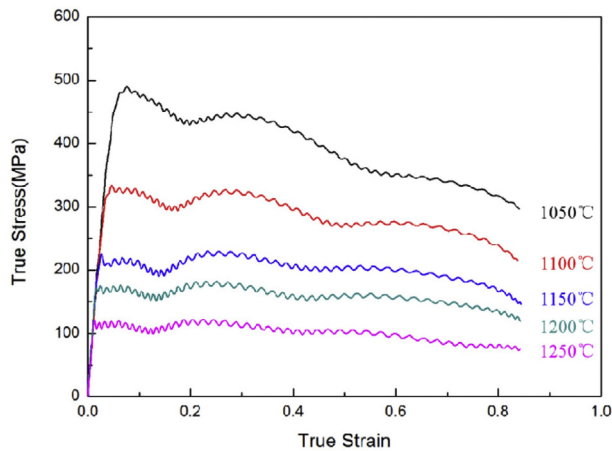


Fig. 4. The flow stress–strain curves in hot deformed Ni_3Al -based superalloy at 1050–1250 °C with strain rate of 0.1 s^{-1} [74].

strain rates are reduced to $5 \times 10^{-2} \text{ s}^{-1}$ and $5 \times 10^{-3} \text{ s}^{-1}$. An obvious peak stress and stress reduction process were observed at 1300 °C and strain rates of $5 \times 10^{-1} \text{ s}^{-1}$ and $5 \times 10^{-2} \text{ s}^{-1}$, followed by a steady state flow stress. Liu et al. [77] further noticed that the flow stresses instantly reached steady state at strain rates of $5 \times 10^{-3} \text{ s}^{-1}$ and $5 \times 10^{-4} \text{ s}^{-1}$, which is comparable to the flow stress evolution at 1100 °C and $5 \times 10^{-4} \text{ s}^{-1}$ deformation conditions and 1200 °C and $5 \times 10^{-4} \text{ s}^{-1}$ deformation conditions. This clearly shows that smaller stresses are obtained during higher deformation temperatures and lower strain rates. This is further evident in the spark plasma sintered MoNbTaTiV RHEA at a strain rate of $5 \times 10^{-2} \text{ s}^{-1}$ and with temperature reduction (1200–1300 °C), the yield stress is rapidly reduced, and the flow stress quickly reaches a steady state. Typical DDRX flow stresses were formed at strain rates of 0.5 s^{-1} and temperatures (1200–1300 °C). Feng et al. [78] discovered serrated flows in deformed CrMoNbV RHEA alloy when the temperature exceeded 500 °C and the strain rate was 0.0002 s^{-1} (Fig. 6). This was attributed to the contact between dislocations and the local stress area due to the rugged local atomic environment [79]. The serrated flows became less obvious at 1000 °C compared to 500–900 °C, signifying a diffusion-mediated softening mechanism could become important.

The as-cast $\text{Al}_{10}\text{Nb}_{15}\text{Ta}_5\text{Ti}_{30}\text{Zr}_{40}$ alloy deformed at temperatures ranging from 250 to 1200 °C and a strain rate of 10^{-3} s^{-1} exhibited peak stress at 600 °C at the start of deformation and then continuously decreases until the end of deformation [70]. At 800 °C, the yield stress dropped considerably at a strain of 0.05 and later displayed a very minor decrease in the flow stress with increasing strain. The reduction in the flow stress indicated DRX. In addition, at a temperature range of 250–500 °C, the deformed alloy has almost the same stress values. At 600 °C, a fast stress drop occurred at the start of deformation, which was followed by a constant decrease in the flow stress. Senkov et al. [70] described the deformation behaviour at temperatures ($T \geq 600 \text{ °C}$) as a softening feature with the occurrence of a sharp stress drop almost immediately after yielding. The

softening feature was likened to dislocation recovery and DRX. Furthermore, the sharp stress drop is followed by a slow flow stress decrease or the formation of a steady-state flow process. The occurrence of a large peak stress followed by a sharp stress drop at the start of deformation can be a result of the short mobility of dislocations present before deformation.

In summary, all the probable mechanisms influencing deformation in metallic alloys, such as WH, DRX, and DRV, have also been observed in refractory high entropy alloys. The sharp drops in the flow stress experienced in some reported RHEAs were attributed to the unlocking of dislocations pinned by the Cottrell atmosphere or short-range order. The DDRX and the CDRX occurred concurrently in some of the hot deformed RHEAs. This is evident in the flow curves of the RHEAs by the serrated and broad peak flow stress. A summary of the flow stress in other hot deformed RHEAs is shown in Table 4 at different strain rates, strains and temperatures. The next section covers the microstructural features observed in hot worked RHEAs.

4 Microstructural evolution during hot deformation of RHEA

In metallic alloys such as GH4169, Nb-1Zr-0.1C and IN718, bulging and serration formation, necklace-like structure, and LAGB movement to HAGB have been attributed to dynamic recrystallization, continuous and discontinuous dynamic recrystallization [59,64,88]. These microstructural features have also been seen in hot worked RHEAs. Senkov et al. [89] reported the microstructural evolution of hot deformed NbTaTi and NbTaZr RHEAs. Prior to the deformation, the microstructure of the NbTaTi RHEA comprised of coarse, polygonal grains of $\sim 0.4 \text{ mm}$ in diameter, whereas the NbTaZr RHEA comprised of finer equiaxed grains of $\sim 40 \mu\text{m}$ in diameter. Furthermore, coarse, needle-like or plate-like, dark second-phase particles were observed at NbTaZr RHEA grain boundaries, some of which spread inside the grains. After deformation at 1000 °C, the NbTaTi RHEA microstructure showed elongation of grains and the creation of distinctive patterns in the directions inclined about 60–70° to the compression direction. This signifies dynamic recrystallization of the grains. The microstructure of the NbTaZr RHEA after deformation at 1000 °C showed elongated equiaxed grains (DRX) and many plate-like Zr-rich second phase particles situated at grain boundaries. Jia et al. [90] study on the NbTiVZr refractory medium entropy alloy showed an equiaxed grain microstructure before hot compression. At 800 °C, the grain boundaries are many and are bounded by an increasing number of precipitates as the temperature is increased. For the inner grain, a relatively loose microstructure can also be observed.

In the silicide containing HfNbTiVSi_{0.5} RHEA by Zhang et al. [55], deformation bands from the longitudinal sections at 800 °C and 1000 °C were observed. This demonstrated that the matrix softened due to the strong thermal activation effect. The silicides are severely deformed by the plastic flow effect and become more

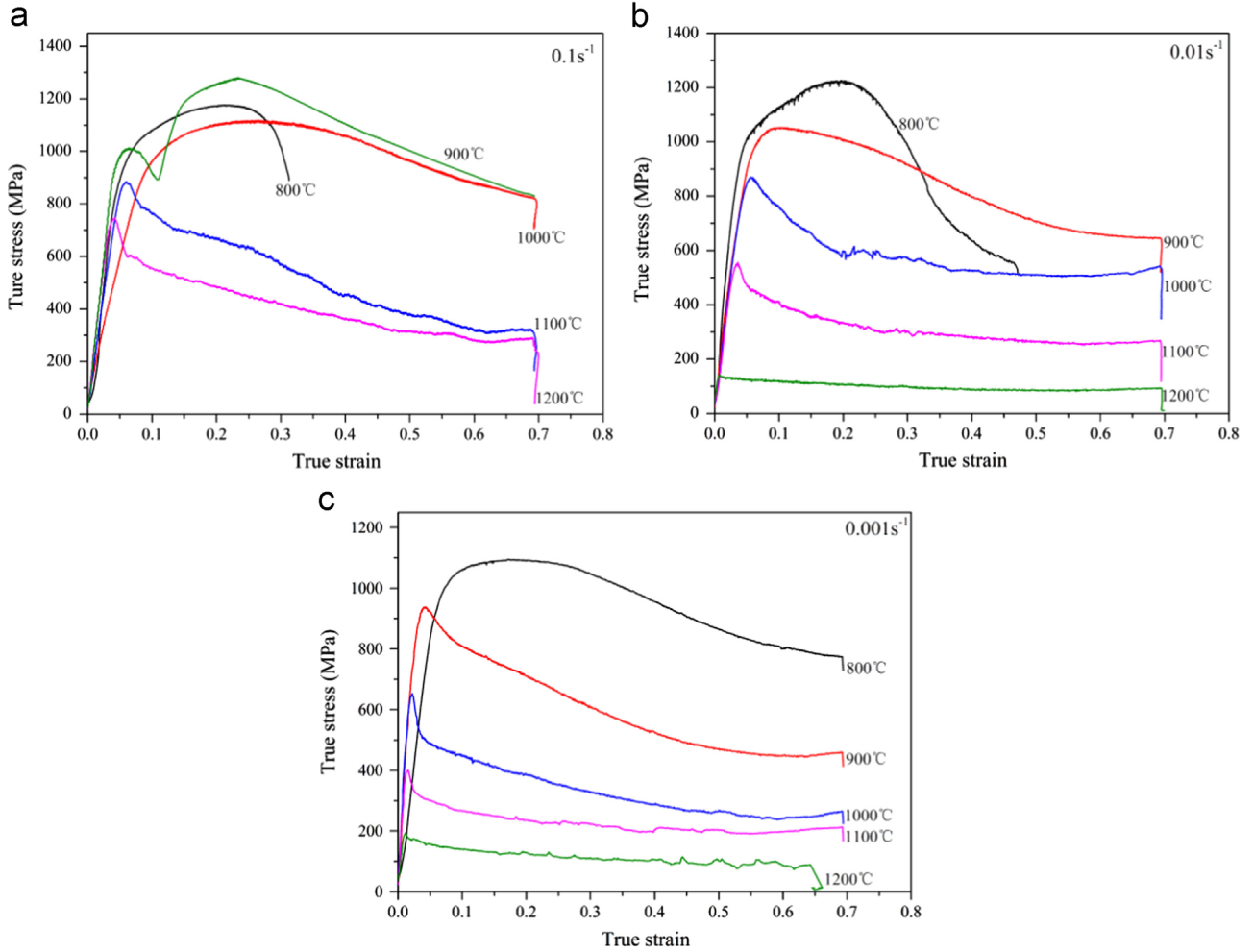


Fig. 5. Flow stress curves of MoNbHfZrTi RHEA at 800 °C, 900 °C, 1000 °C, 1100 °C, 1200 °C with different strain rates: (a) 10^{-1} s^{-1} , (b) 10^{-2} s^{-1} , (c) 10^{-3} s^{-1} [54].

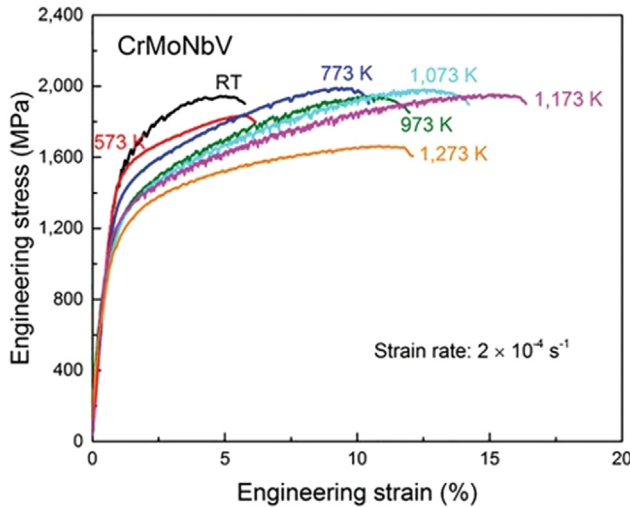


Fig. 6. Serrated flows in deformed CrMoNbV RHEA alloy at wide range of temperatures [78].

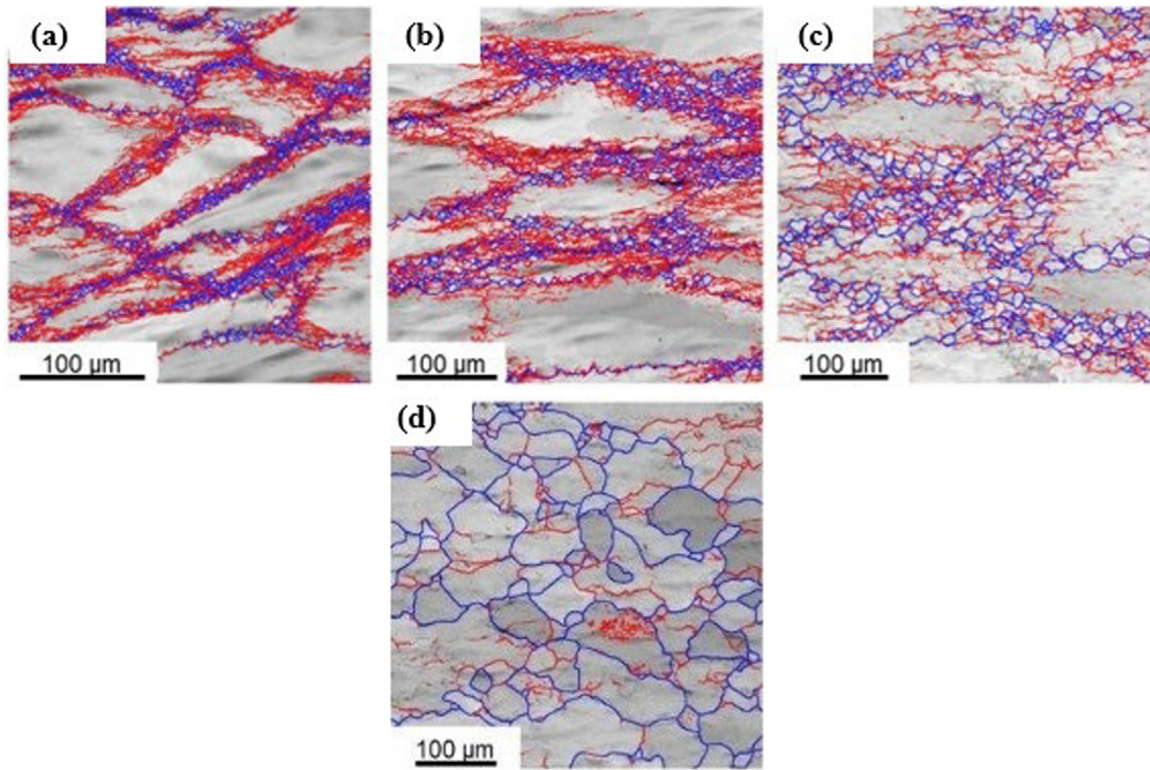
refined and homogeneous than in the as-cast conditions. Guo et al. [91] discovered a dendritic (BCC_A2#1) and inter-dendritic (BCC_A2#2) microstructure in a $\text{TaMo}_{0.5}\text{ZrTi}_{1.5}\text{Al}_{0.1}\text{Si}_{0.2}$ RHEA after casting and 48 hours of annealing at 1300 °C. Also, an HP16 silicide in the main region of the interdendrite later forms a butterfly-like eutectic morphology with the BCC_A2#2 phase. At 1000 °C, DRX start within the inter-dendritic region, leading in the creation of sub-micron grains. A heterogeneous necklace microstructure also formed through dynamic recrystallization (DRX) in the $\text{TaMo}_{0.5}\text{ZrTi}_{1.5}\text{Al}_{0.1}\text{Si}_{0.2}$ RHEA. Also, a necklace-like structure was formed in MoNbHfZrTi RHEA, which composed fine recrystallized grains along the initial grain boundaries, suggesting active DDRX [40]. To be specific, at 1100 °C and 0.5 s^{-1} , necklace-like structures with fine undistorted grains ($1.8 \mu\text{m}$) were observed along the original grain boundaries. Necklace-like structures were also observed with fine undistorted grains ($2.2 \mu\text{m}$) along the shear-deformation region at 1100 °C and 0.1 s^{-1} . These observations confirm the presence of DDRX.

Table 4. Summary of flow stress behaviour of hot deformed refractory high entropy alloys.

Alloy composition	Strain rate (s ⁻¹)	Strain	Temp. (°C)	Findings	References
AlNb ₂ TiV	10 ⁻³	0.33	600, 800 and 1000	An increase in temperature leads to a decrease in yield stress, initially slowly at 600 and 800 °C, and then rapidly at 1000 °C. Softening with steady flow was observed indicating the occurrence of dynamic recovery.	[80]
AlNbTi ₃ VZr _{1.5}	10 ⁻³ to 1	–	1100,1150, 1200,1250	All the flow stress softens, signifying the occurrence of DRX.	[71]
ZrTiHfNb _{0.5} Ta _{0.5}	10 ⁻³	0.5	700, 800 and 900	The stress increased rapidly and then decreased to a stable value with an increase in deformation.	[81]
WReTaMo	10 ⁻³	0.25	1600	At 1600 °C, the WReTaMo alloy shows continuous strengthening without peak stress until the final testing point at strain of 25%.	[82]
Nb ₄₀ Ti ₂₅ Al ₁₅ V ₁₀ Ta ₅ Hf ₃ W ₂	10 ⁻³	0.69	800, 900 and 1000	Samples deformed at 800 °C show strain hardening. Samples deformed at 900 °C and 1000 °C show persistent softening after yielding which signifies dynamic recovery and dynamic recrystallization.	[41]
Nb ₃₀ Mo ₃₀ Hf ₂₀ Co ₂₀	10 ⁻⁴	0.6	600, 800 and 1000	Steep and delayed strain hardening stage was observed in the flow stress curves at 22 °C and 600 °C. Softening with steady-flow stages was observed at 1000 °C, signifying dynamic recovery.	[83]
C _x Hf _{0.25} NbTaW _{0.5}	1 ⁻³	–	1000, 1200 and 1400	A continuous hardening after the yield was observed in all the specimens. As the temperature increased, the steady-state stress appeared at lower strain values.	[84]
TaNbHfZrTi	10 ⁻⁵ to 10 ⁻¹	–	750	Deformation strengthening occurred and the flow stress increased continuously with an increase in the plastic strain at high-strain rates	[50]
Ti _x ZrVNb	10 ⁻³	–	600 and 800	At 800 °C, the strengths of the alloys significantly decreased. Furthermore, a serrated curve was also observed at 800 °C due to the alternating softening deformation and recrystallization	[85]
HfMoNbTaTi	10 ⁻²	0.8	800, 1000 and 1200	A continuous drop in the flow stress started at the peak stress (strain = 0.027), proceeded by a steady-state flow stage from a strain of 0.8. Dynamic recrystallization behaviour was observed during compression at 1200 °C.	[86]

Table 4. (continued).

Alloy composition	Strain rate (s^{-1})	Strain	Temp. ($^{\circ}\text{C}$)	Findings	References
NbCrMo _{0.5} Ta _{0.5} TiZr	10^{-3}	–	800, 1000 and 1200	At 1000 $^{\circ}\text{C}$, dynamic recrystallization of the alloy occurs after a short stage of strain hardening. At 1200 $^{\circ}\text{C}$, further decrease in the flow stress, followed by weak softening and a steady state flow (dynamic recovery).	[52]
HfMoNbTaTiZr	10^{-3}	–	800, 1000 and 1200	At 800 $^{\circ}\text{C}$, work hardening occurred. At 1000 $^{\circ}\text{C}$ and 1200 $^{\circ}\text{C}$, dynamic recrystallization occurred.	[87]

**Fig. 7.** EBSD maps of HfNbTaTiZr RHEA hot compressed to $\varepsilon = 0.69$ at various temperatures and strain rates. Deformed at (a, b, c) 1000 $^{\circ}\text{C}$, (d) 1200 $^{\circ}\text{C}$ [32].

Eleti et al. [32] studied the microstructure evolution in hot deformed HfNbTaTiZr RHEA and discovered fine, equiaxed grains in the necklace structures. The equiaxed grains in the necklace structures are bounded by blue HAGBs, which are recrystallized grains (Fig. 7). In connection to the continuous flow softening observed after the sharp stress-drop in HfNbTaTiZr RHEA, typical necklace microstructures composed of coarse, uncrystallized grains surrounded by fine dynamically recrystallized grains were detected (Figs. 7a, b, c). At 1200 $^{\circ}\text{C}$ and strain rate of 10^{-3} , a more homogeneous microstructure comprised

of coarse grains in the specimen was observed and possibly corresponds to dynamic recrystallization (DRX) grains after substantial grain growth (Fig. 7d).

A low-density TiAlVNb₂ RHEA showed a coarse grain with a dendrite structure in the as-cast condition [72]. Fine grains occur along the original grain boundaries in the samples deformed at lower temperatures or higher strain rates, such as at $0.1 \text{ s}^{-1}/1000^{\circ}\text{C}$, $0.1 \text{ s}^{-1}/1100^{\circ}\text{C}$, $0.01 \text{ s}^{-1}/1000^{\circ}\text{C}$ and $0.001 \text{ s}^{-1}/1000^{\circ}\text{C}$. The fraction and grain size of the fine grains increase with the increasing deformation temperature or decreasing strain rate. Part of the fine

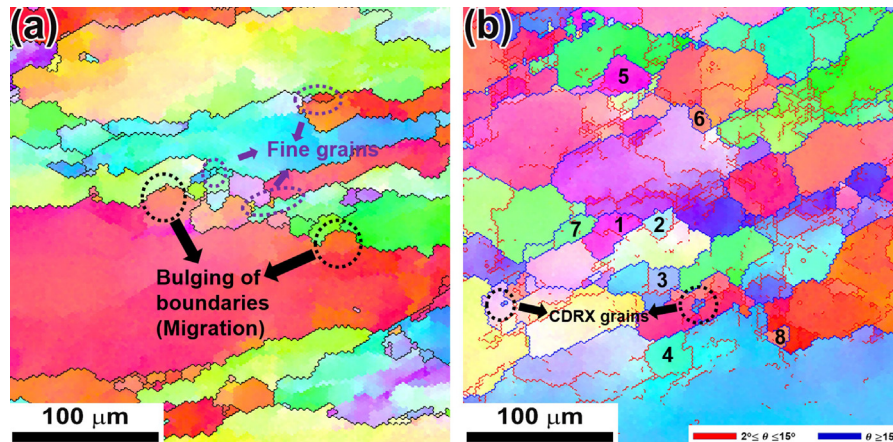


Fig. 8. The inverse pole figure map of TiAlVNb₂ RHEA at 10⁻² s⁻¹/1200 °C (a) and 10⁻³ s⁻¹/1100 °C (b) [72].

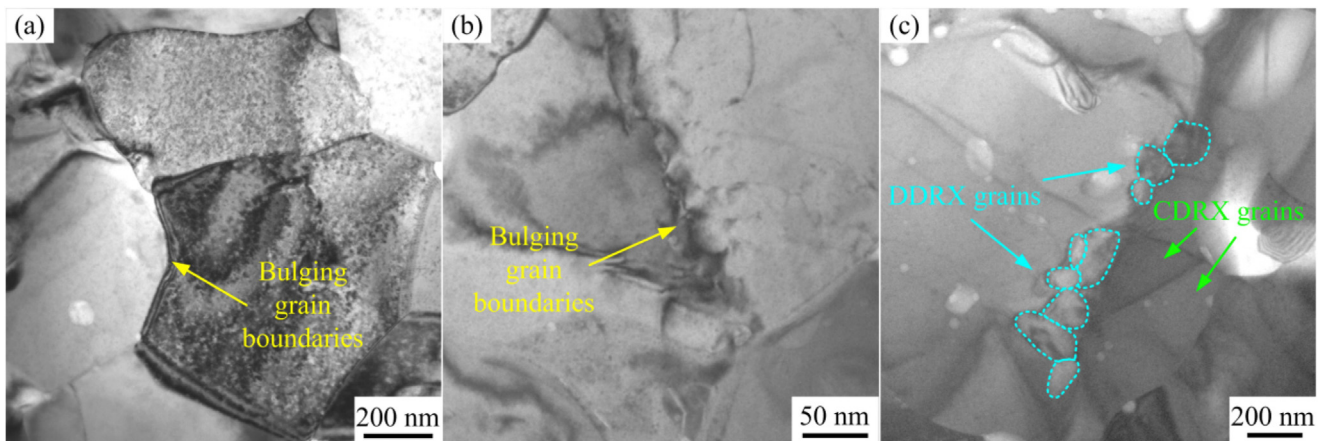


Fig. 9. Transmission electron microscopy images of the hot compressed MoNbTaTiV RHEA showing (a) and (b) bulging grain boundaries and (c) discontinuous dynamic recrystallized grains and continuous dynamic recrystallized grains [63].

grains are equiaxed, while the others appear elongated due to on-going compression. The TiAlVNb₂ RHEA is composed of relatively uniform coarse grains at the highest temperature (1200 °C) and lowest strain rate (0.001 s⁻¹). Some fine dynamic recrystallized grains with clear misorientation from the initial grains can be found in grain boundaries and triple junctions of grains (Fig. 8a). Serrated grain boundaries of coarse grains bulge towards the adjacent grains by migration of high angle grain boundaries, which form some sub-grains (labelled as 1–4) surrounded by sub-boundaries (Fig. 8b). This is classified as a DRX grain formation mechanism and is strongly dependent on increased energy storage caused by the propagation and pile-up of dislocations near to the boundaries during the high-temperature compression [33,66].

In the hot deformation processing of a MoNbTaTiV RHEA fabricated by powder metallurgy, the microstructural characteristics revealed nearly-equiaxed grains, apart from some deformation microstructures seen at 1100 °C and 0.05 s⁻¹ [63]. With increasing deformation temperature

and decreasing strain rate, the number of these ultrafine dynamically recrystallized grains gradually decreases. As shown from TEM images in Figure 9, the grain boundaries of the ultrafine grains act as a nucleation point for the deformed MoNbTaTiV RHEA. Furthermore, smaller bulging shapes were observed to be present (Figs. 9a and b). As a result, several ultrafine discontinuous dynamic recrystallized grains are produced, as shown in Fig. 9c. The grain boundaries bulging, which is caused by the nucleation of new grains and subsequent grain expansion, distinguishes discontinuous dynamic recrystallized grains from other types of recrystallized grains [75]. The bulging phenomenon was also observed in some studies on the hot deformation of arc-melted RHEA [54,72,92].

In summary, the bulging phenomenon, serrated grain boundaries, and necklace-like structures reported in metallic alloys have also been detected in hot deformed RHEAs. The necklace-like structures have been typically observed in FCC metals and alloys with medium or low stacking-fault energies but not often in BCC metals and alloys. The following elemental compositions:

TiZrHfNbTa, TiZrHfNbMo, TaNbHfZrTi, TiZrHfVTa, HfNbTaTiZr, MoNbHfZrTi, TaMoZrTiAlSi, and AlNb-TiVZr RHEAs commonly have the necklace-like structures. The next section discusses the dislocation motion mechanisms observed in hot deformed RHEAs.

5 Dislocation motion mechanisms in hot deformed refractory high entropy alloy

Dynamic recrystallization and dynamic recovery are governed by dislocation geometry and movement during deformation. In the plastic deformation of metallic materials, the important mechanism that governs mechanical properties is dislocation slip [93]. The dislocation multiplication rate rises with the increase in strain rate at the same deformation temperature and only happens in metallic alloys with positive strain rate sensitivity [94]. According to Liu et al. [95], the softening mechanism in coarse grain alloys is dominated by dislocation movement, while that for ultra-fine-grained alloys is mostly boundary gliding and rotation. Due to the lattice distortion effect in high entropy alloys, dislocations are likely to possess a rare dislocation core structure and the Peierls potential is fluctuated, which has a significant effect on mechanical properties [93]. Although there is still a lack of availability of widespread research on dislocation motion mechanisms in high temperature compressed refractory high entropy alloys. This section briefly discusses the types of dislocation motion mechanism in high temperature compressed RHEA.

A study of HfNbTaTiZr RHEA single crystals by Yasuda et al. [96] showed screw dislocations ($1/2\langle 111 \rangle$) which control the deformation behaviour and it is similar to conventional BCC metals. The dislocation slip traces are a little wavy and are inclined to cross-slip and less sensitive to deformation temperature. Eleti et al. [32] observed plastic deformation by dislocation slip in a RHEA deformed at a strain of 0.69, similar to conventional BCC metals and alloys. In another study by Hu et al. [97] on the dynamic behaviour of TaNbHfZrTi RHEA at higher temperatures, it was found that the dominant softening mechanism is linked to the thermally activated screw dislocation at quasi-static conditions and the phonon drag controlled screw dislocation motion at high strain rates and high temperatures. Li et al. [98] studied ZrNbMoTaW RHEA with in-situ forming heterogeneous structure at 1000 °C. It was found that screw dislocation cross slip was the dominant plastic deformation mechanism. Pang et al. [41] studied the high temperature strength of the Nb₄₀Ti₂₅Al₁₅V₁₀Ta₅Hf₃W₂ RHEA deformed after 50% compression at 800 °C. A massive dislocation motion was found in the deformed region, which is multiple dislocation interactions, such as cross-slip and tangles. The massive dislocation motion pile up at the grain boundaries, and the resultant high stress concentrations in these regions facilitate the concentrated deformation along grain boundaries. This occurred because the deformation temperature at 800 °C was not high enough to facilitate dislocation movement and

diffusion to relieve stress concentration. For a Re_{0.5}MoNbW (TaC)_{0.5} RHEA composite deformed at 1200 °C, a high density of dislocations was observed in the BCC and FCC-MC phases upon severe compression compared to the as-cast microstructure [99]. The phase interface after deformation revealed screw dislocation motion in BCC close to the interface. Furthermore, severe dislocation entanglement and dipolar dislocation walls existed in the Re_{0.5}MoNbW(TaC)_{0.5} high-entropy alloy. Dislocation dipoles created by dislocation cross-slip and distinguishable pinning effects were evidently observed in the compressed BCC phase. In general, the substructures of dipolar dislocation walls originate from the BCC phase.

In the hot deformation study of MoNbTaTiV RHEA produced by powder metallurgy, increase in temperatures and decrease in strain rates promote the involvement of the grain boundary gliding mechanism [95]. This resulted in a decrease in intragranular dislocation generation. Grain deformation is dominant at low deformation temperature and high strain rate, which results in an increase in the dislocation density and many dislocation entanglements. Feng et al. [100] investigated the dislocation features of CrMoNbV RHEA at 700, 800 and 900 °C. It was found that non-screw dislocation that are less mobile due to pinning are the dominant dislocation motion during high-temperature deformation in the CrMoNbV RHEA. This effect contributed to the observed superior high-temperature strength in the deformed CrMoNbV RHEA. Prior to the mechanical deformation, no clear pre-existing dislocations were observed. However, at 900 °C and plastic compression of 3.3%, considerable pinned and curved dislocation motion are seen. This means that the strong pinning effect applied on the dislocation motion specifies strong connections between dislocation motion and severe distorted local lattice/local compositional fluctuations. The solute atoms along the pathways of a gliding dislocation are independent point obstacles that pin the dislocation, which bows out in the regions between the solutes.

6 Unsafe deformation behavior in hot deformed RHEA

During the deformation process, stress concentration occurs, which causes a sharp increase in local energy. When the stress concentration reaches a value, crack propagation begins to appear, releasing energy. According to Prasad et al. [101], the unsafe deformation behavior include: flow localisation, void formation, inter-crystalline cracking, wedge cracking, kink bands and adiabatic shear bands. Unsafe deformation behavior have been reported in metallic alloys: IN738LC (deformation bands and grain boundary cracking) [102]; Ni3Al (flow localization, adiabatic shear bands and cracks) [74]; Ti₁₅Mo₃Al₃Nb_{0.2}Si (cracks) [60] and TB8 titanium alloy (flow localization and shear bands) [61]. The unsafe deformation behavior have also been reported in RHEAs.

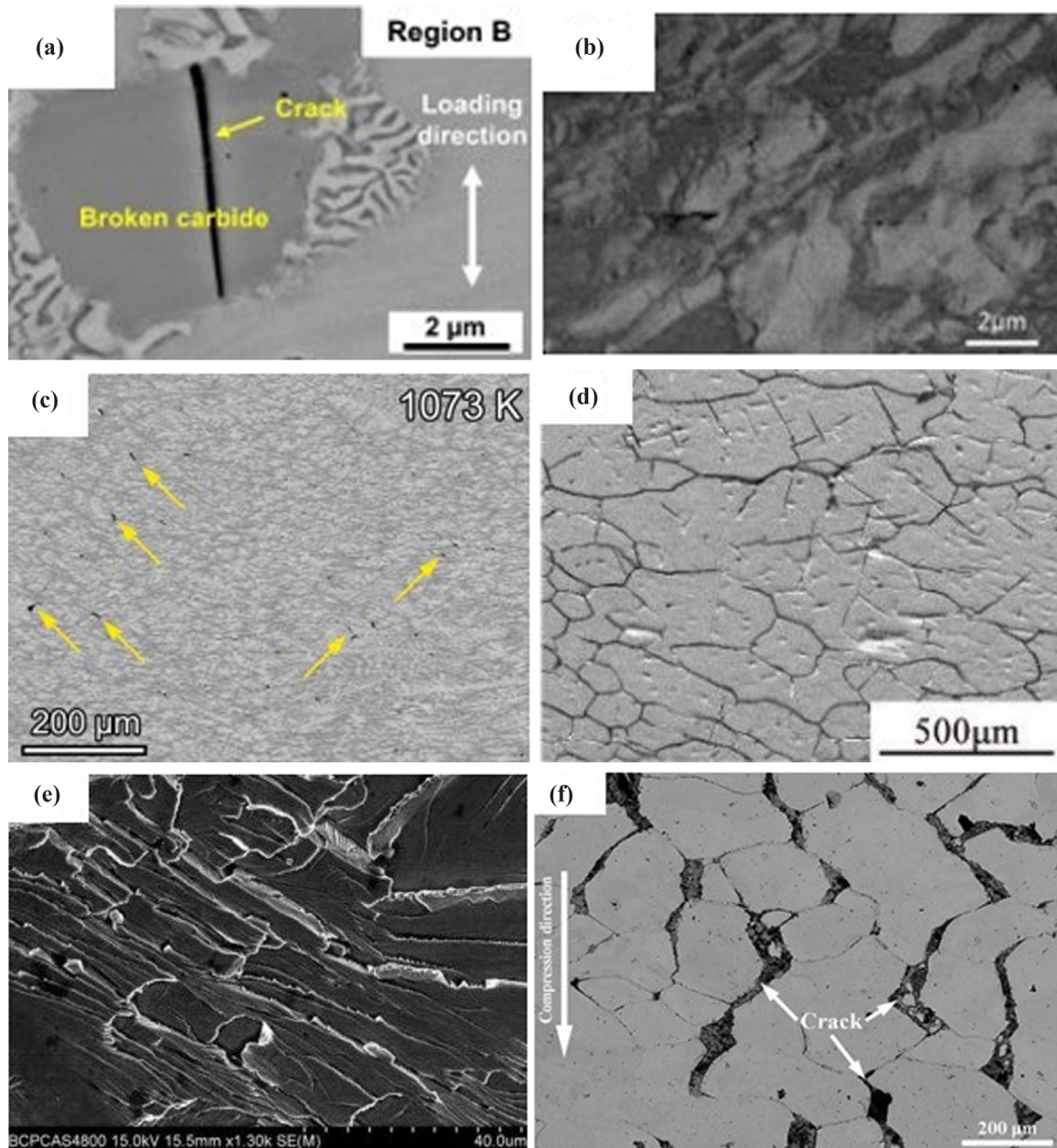


Fig. 10. Crack propagation in hot compressed refractory high entropy alloys at 800 °C (a–e) and 1600 °C (f): (a) Broken carbide and crack in TiNbTaZrHf RHEA based composite [104]. (b) Microcracks observed inside the silicides containing HfNbTiVSi_{0.5} RHEA [55]. (c) Microcracks observed inside the deformed Nb₄₀Ti₂₅Al₁₅V₁₀Ta₅Hf₃W₂ RHEA [41]. (d) Severe cracks at grain boundaries of W_{0.4}MoNb_xTaTi RHEA [76]. (e) Fracture feature of river-like appearance for ZrTiHfNbMo₂ RHEA [105]. (f) Propagation of cracks in WReTaMo RHEA [82].

The V₂₀Nb₂₀Mo₂₀Ta₂₀W₂₀ RHEA exhibited much lower plastic strain, and its fracture started via crack propagation along a direction inclined at 10° relative to the loading axis, with significant involvement of grain boundaries during crack growth [49]. In the deformed microstructures of (VNbTiTa)_{100-x}Si_x ($x = 2.5, 5$) RHEAs at 1000 °C, silicide was discovered in eutectics which became broken after compression and the broken fragments are circulated within the original inter-dendrite areas [103]. At 1200 °C, silicide elongated and broke in Si_{2.5}

and Si₅ alloys. Similarly, some of the carbides are cracked along the loading direction at the side regions of the TiNbTaZrHf-based composites by Li et al. [104] compressed at 800 °C (Fig. 10a). The appearance of microcracks at the sides of carbides confirms their load-bearing effect during the high temperature deformation process. Furthermore, at 1000 and 1200 °C, a small amount of carbides crack along the loading direction but drops significantly, indicating that carbides continue to play a role in sharing and transferring applied stress from the

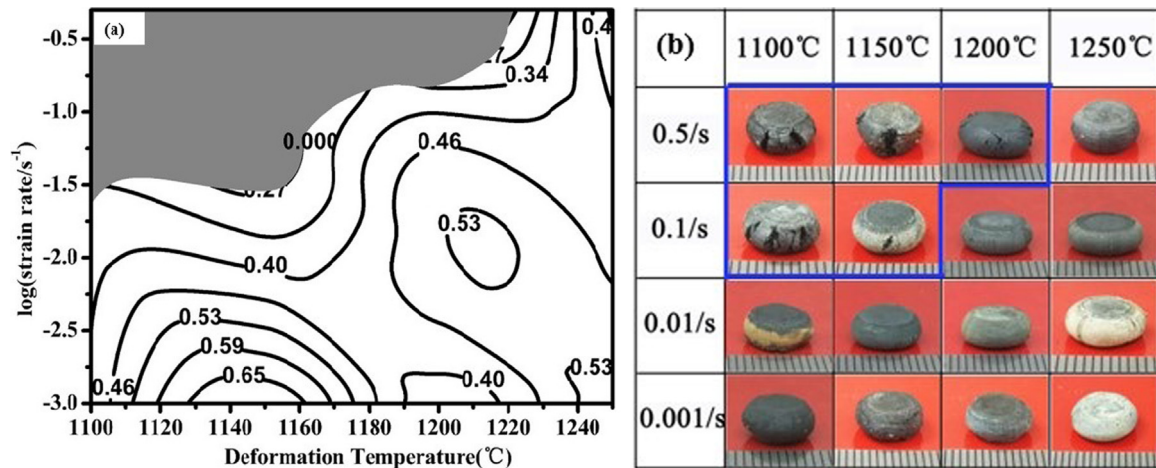


Fig. 11. The processing maps for the MoNbHfZrTi RHEA (grey part is the instability zone) (a); Distribution of cracks of the deformed specimens at a series of processing parameters (blue region is the instability zone) (b) [40].

matrix during the deformation process. In Figure 10b, the silicide containing HfNbTiVSi_{0.5} RHEA deformed at 800 and 1000 °C showed severe deformation due to plastic flow, and a few microcracks were observed inside the silicides [55]. A few microcracks observed in the silicide containing HfNbTiVSi_{0.5} RHEA correspond to high interfacial adhesion between the matrix and silicides.

In the deformation of Nb₄₀Ti₂₅Al₁₅V₁₀Ta₅Hf₃W₂ RHEA, Pang et al. [41] observed some microcracks at 800 and 900 °C. The microcracks in Figure 10c is intergranular, and the widespread microcracks originated from triple grain boundaries at 800 °C after compression. Microcracks were also observed at 900 °C but decreased with increasing temperatures until no microcracks could be seen at 1000 °C. According to Pang et al. [41], the pile up of massive dislocations at grain boundaries and its resultant stress concentration speeds up the formation and extension of microcracks (Fig. 10c), which are detrimental to deformation stability. Crack propagation occurred in hot deformed W_{0.4}MoNb_xTaTi RHEA at temperatures of 800, 1000, and 1100 °C [76]. At 800 °C, the Nb_{1.3} alloy in Figure 10d showed severe cracks at grain boundaries. The cracks originated at the grain boundaries and later circulated along the grain boundaries. Meanwhile, few cracks extended into the grains. However, at an increased temperature of 1000–1100 °C with an increased niobium concentration (Nb_{1.5}), the propagation of cracks was suppressed. As the deformation temperature increases, the atomic activity and dislocation migration rate increase significantly. Therefore, the local stress concentration could be efficiently improved, which would prevent crack initiation and propagation. When the molybdenum content of the ZrTiHfNbMo_{2.0} alloy was increased, poor plasticity and a river-like appearance for the fracture feature (Fig. 10e) were observed at 800 °C [105]. Rhenium containing WReTaMo RHEA deformed at 1600 °C by Wan et al. [82] experienced grain boundary sliding that contributes to the deformation as only intergranular cracks are observed. As shown in Figure 10f, the cracks originate at the grain boundaries initially and then spread along the grain boundaries.

In summary, microcracks and fracture which is an unsafe deformation behavior, were observed in some hot deformed RHEAs but decreased with increasing temperatures. The microcracks in the RHEAs are mostly intergranular. The next section briefly describes the constitutive modelling and processing maps in hot deformed RHEAs.

7 Optimising deformation conditions in hot deformed RHEAs

Hot working processes can be optimised using a combination of experimental stress–strain curves, constitutive equations, and processing maps. However, the selection of an appropriate constitutive model to accurately describe the stress–strain behaviour and the dominant mechanisms requires a detailed understanding of the various constitutive models. The activation energy for hot working is an important parameter which gives an indication of the material's resistance to deformation and is also used for the prediction of flow stress [65,106,107]. In addition, it could also provide insights into the microstructural mechanisms influencing the deformation process [65,108]. Constitutive models are largely classified into phenomenological models, physical models, and artificial neural network models [109,110]. The most common phenomenological model is the Johnson-Cook model [111] while the Arrhenius model [112] is the simplest form of phenomenological model. Understanding hot deformation behaviour is critical for controlling the microstructures and properties of RHEAs. Hot workability of RHEAs has rarely been reported. Especially in establishing a constitutive connection between the flow stress (σ), deformation temperature (T), strain rate ($\dot{\epsilon}$) and strain (ϵ). As in other conventional alloys, constitutive modelling has been used on hot worked RHEA to predict flow stress. Processing maps have also been used in RHEA to determine optimum deformation parameters. According to [101], processing maps not only identify the ideal combination of processing parameters for

shaping and microstructural control, but they can also distinguish the “safe” and “unsafe” combinations of processing parameters.

Dong et al. [40] performed high temperature compression tests on the MoNbHfZrTi RHEA at temperatures of 1100–1250 °C and strain rates: 10^{-3} – 0.5 s^{-1} to determine the hot workability and processing maps. In their study, the apparent activation energy (Q) was determined to be 326.1 kJ/mol. Constitutive equation predicted the flow stress of the hot deformed TaNbHfZrTi RHEA. The hot-processing map of the MoNbHfZrTi RHEA shown in Figure 11 indicates a contour area with a power dissipation factor (η) and the shaded portion, which is the instability zone. Dong et al. [40] reported the instability region to be at 1100–1220 °C and $10^{-1.5}$ – 0.5 s^{-1} , with values fluctuating from 0.15 to 0.27 in the instability area (Fig. 11a). This coincides with the occurrence of cracks (Fig. 11b) in the deformed alloy at a strain rate of 0.5 s^{-1} at temperatures (1100–1200 °C) and a strain rate of 10^{-1} s^{-1} at temperatures (1100–1150 °C). In the stable region, the η occur at 1110–1170 °C and 10^{-3} – $10^{-2.5} \text{ s}^{-1}$. In Wang et al. [113] study on the effects of V concentration on the mechanical properties of V_xNbMoTa RHEAs at temperatures 900–1100 °C, it was shown that at a strain range of 0.1–0.5, the activation energy decreases continuously when increasing the vanadium concentration. This indicates that the alloys of higher V content are easier to deform at high temperatures and, therefore, have lower strain hardening rates. Using an improved Johnson-Cook (J-C) constitutive model, Hu et al. [97] describe the deformation behaviour of TaNbHfZrTi RHEA at a strain rate of 400 s^{-1} to 2600 s^{-1} . It was found that fitting values of the J-C constitutive model corresponded well with the experimental findings, indicating that the modified J-C model can be used to accurately predict the hot deformation behaviours of the TaNbHfZrTi RHEA.

At $\varepsilon = 0.1$, Bai et al. [72] discovered that the Q value for TiAlVNb₂ RHEA is approximately 401 kJ mol^{-1} . Furthermore, the Q value at strain of 0.3, 0.5, and 0.69 were calculated to be 387, 379, and 375 kJ mol^{-1} , respectively. The Q value vary in a slight range between 401 and 375 kJ mol^{-1} and are nearly constant at strain of 0.5 (379 – 375 kJ mol^{-1}), though with increasing strain they show a slight decreasing trend. According to Bai et al. [72], the results showed the thermal deformation process in TiAlVNb₂ RHEA is quite stable. Similarly, Yul et al. [71] investigated the flow stress analysis and hot processing region of AlNbTi₃VZr_{1.5} RHEA using Arrhenius constitutive equation and processing maps. The Q value was calculated to be 228.1 kJ/mol. The Q value was lower than TiAlVNb₂ RHEA [72] when Zr was added to the composition. Arrhenius constitutive equation was found suitable to predict the flow stress of AlNbTi₃VZr_{1.5} RHEA. Using the processing map, the appropriate processing region is determined at 1200–1250 °C and $10^{-0.75}$ – 1 s^{-1} .

Eleti et al. [34] studied the deformation behaviour in high temperature compression of HfNbTaTiZr RHEA. It was discovered that the Q value was approximately 258 kJ mol^{-1} at strain of 0.1. The value was lesser than the Q for self-diffusion of the elements comprised in the

alloy, such as Nb and Ta, but greater than the activation energies for self-diffusion in Ti, Hf and Zr. Also, the Q value gradually decreased when strain was increased, but all values were in a slight range of 258 – 232 kJ mol^{-1} . The attained slopes gave the Q value of $\sim 258 \text{ kJ mol}^{-1}$ at the definite strain of 0.1. In particular, the Q value after $\varepsilon = 0.3$ was nearly constant at 232 – 229 kJ mol^{-1} . Eleti and co-workers attribute the large activation energy (258 kJ mol^{-1}) at strain of 0.1 to being likely affected by the drop in stress noticed in the stress–strain curves. A higher value of strain-rate sensitivity ($m = 0.33$) was reported for HfNbTaTiZr RHEA, which signifies the occurrence of grain boundary sliding. Similarly, an earlier study on the deformation behaviour of HfNbTaTiZr RHEA by Senkov et al. [50] reported activation energy of $\sim 226 \text{ kJ mol}^{-1}$ which is in the range (258 – 232 kJ mol^{-1}) obtained from the Eleti et al. [34] study. Li et al. [69] found a low Q value of 144 – 119 kJ mol^{-1} over a range of strains (0.1–0.6) for Ti₂ZrHfV_{0.5}Ta_{0.2} RHEA. Since a high Q value correlates with high resistance to softening at elevated temperature, Ti₂ZrHfV_{0.5}Ta_{0.2} RHEA became less competitive at higher temperatures. The low Q value was attributed to excessive concentration of Ti, Zr, and Hf elements, which tend to have low self-diffusion activation energies. The relationship among flow stress, temperature, and strain rate of Ti₂ZrHfV_{0.5}Ta_{0.2} RHEA was predicted by the Arrhenius-type constitutive equations.

This section has shown that the apparent activation energy values in RHEAs decrease with an increasing strain. RHEAs with a high Q value are more competitive at higher temperatures than RHEAs with a low Q value. Based on the available literatures, constitutive model predicted the flow stress of hot worked RHEAs. Also, the hot-processing map can be used to determine the safe and unsafe region of hot deformed RHEAs. Section 8 reports the yield strength and resistance to high temperature softening of hot deformed RHEAs.

8 Yield strength of RHEA during hot deformation

A fast decrease in the strength of RHEA is known to mostly occur due to the intensification of diffusion-related processes above 0.5 – $0.6 T_m$ [50,114]. The high solid solution hardening effect and strong atomic bonding contribute to the good mechanical properties of RHEAs at high temperatures [115]. According to Senkov et al. [116], the RHEAs have very high strength at temperatures of 800 – 1000 °C, single-phase RHEAs are usually stronger at $T \geq 1000$ – 1200 °C than multi-phase RHEAs. The rapid loss of strength of multi-phase RHEAs at 800 – 1000 °C was discovered to be caused by secondary phase dissolution at solvus temperatures.

Senkov and Woodward [52] deformed Nb₂₀Cr₂₀Mo₁₀Ta₁₀Ti₂₀Zr₂₀ RHEA at 800 , 1000 , and 1200 °C and a strain rate of 10^{-3} s^{-1} in 2011. At 800 °C, the yield stress and peak strength of the alloy decreased to 983 and 1100 MPa . When the temperature increased to 1000 °C, at strain rate of 0.001 s^{-1} , the yield stress and peak strength of the alloy decreased to 546 and 1100 MPa . Finally, at a temperature of

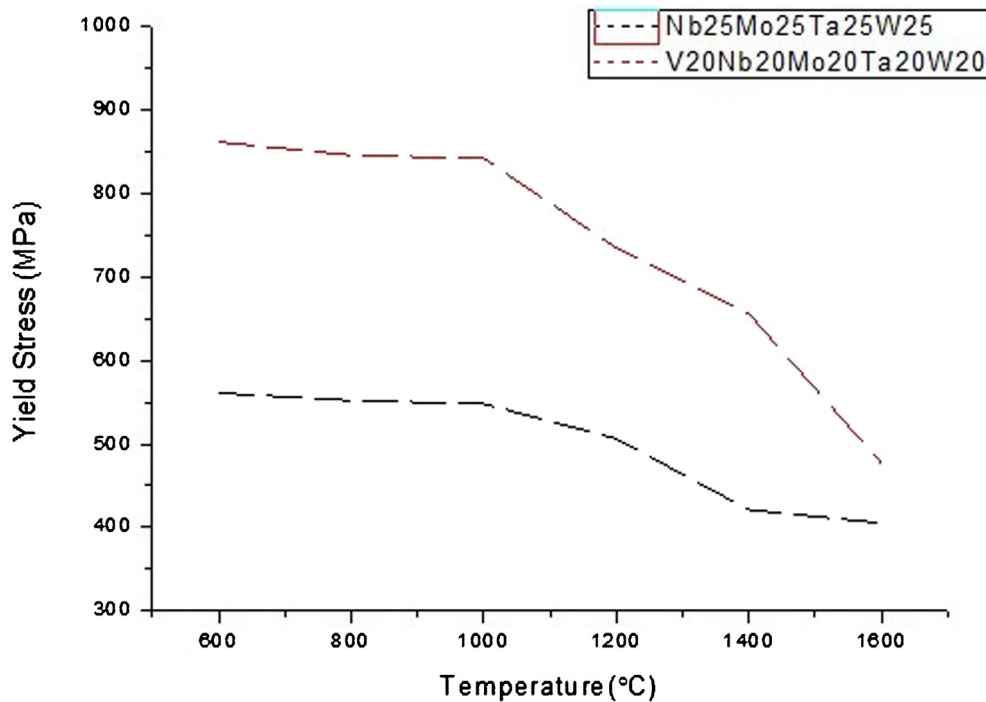


Fig. 12. The effect of temperature on the yield stress of Nb₂₅Mo₂₅Ta₂₅W₂₅ and V₂₀Nb₂₀Mo₂₀Ta₂₀W₂₀ RHEAs.

1200 °C, the yield stress was 170 MPa while the peak stress of 190 MPa was attained shortly after yielding and followed by weak softening. Senkov et al. [49] studied the high temperature properties of Nb₂₅Mo₂₅Ta₂₅W₂₅ and V₂₀Nb₂₀Mo₂₀Ta₂₀W₂₀ RHEAs at 600–1600 °C and a fixed strain rate of 0.001 s⁻¹. Figure 12 shows the temperature effect on the yield stress of Nb₂₅Mo₂₅Ta₂₅W₂₅ and V₂₀Nb₂₀Mo₂₀Ta₂₀W₂₀ RHEAs. The yield stress of Nb₂₅Mo₂₅Ta₂₅W₂₅ and V₂₀Nb₂₀Mo₂₀Ta₂₀W₂₀ RHEAs decreased steadily as temperature increased (Fig. 12). The resistance to high temperature softening of the Nb₂₅Mo₂₅Ta₂₅W₂₅ and V₂₀Nb₂₀Mo₂₀Ta₂₀W₂₀ RHEAs is likely due to slow diffusion of elements, which directly relates to the T_m of RHEAs [49].

At 800 and 900 °C, Chen et al. [105] studied the high temperature strength of ZrTiHfNbMo_x RHEA ($x = 0.5, 1.0, 1.5$). At 800 and 900 °C, the yield stress of the ZrTiHfNbMo_{0.5} RHEA is 585 MPa and 288 MPa with plasticity of 50% respectively. When the molybdenum content of ZrTiHfNbMo_x RHEA increases to $x = 1.0$ at 800 °C and 900 °C, the yield stress increases significantly to 1099 MPa and 804 MPa, with plasticity greater than 40% and 50%. If the molybdenum content is increased to $x = 1.5$ at 800 °C and 900 °C, the yield stress rises to 1155 MPa with plasticity <15% and 897 MPa with plasticity > 25%. Chen et al. [105] attributed the high temperature strength to controlling the molybdenum content addition in the ZrTiHfNbMo_x RHEAs. For the high temperature compressive study of Cr containing CrMoNbV RHEA by Feng et al. [100] at 300–1000 °C. The yield strength of CrMoNbV is 1062 MPa at 1000 °C. The rhenium containing WRe-TaMo RHEA by Wan et al. [82] deformed at 1600 °C and at a strain rate of 0.001 s⁻¹ showed a yield strength of

172 MPa, peak compressive strength of 244 MPa, and plasticity >25%, which is attributed to the high melting temperature.

Similarly, by substituting Re with Nb and Ti, Wan et al. [117] studied the high temperature compression properties of WTaMoNbTi RHEA at 1600 °C at a strain rate of 0.001 s⁻¹. The WTaMoNbTi RHEA showed a yield strength of 173 MPa and a peak strength of 218 MPa with plasticity above 25%. Wan et al. [117] demonstrated that adding Ti to WTaMoNb RHEA improved its high temperature strength and plasticity. Furthermore, the addition of Ti increased the lattice constant, thereby intensify the effect of solid solution hardening. The effect of Ti content on the high temperature mechanical properties of Ti_xZrVNb RHEAs with varying elemental compositions by Huang et al. [85] revealed high strength. At 600 °C, the TiZrVNb, Ti_{1.5}ZrVNb and Ti₂ZrVNb RHEAs exhibited high strength of 923, 886, and 922 MPa with plasticity >25%. However, the yield strengths decreased at 800 °C to 233, 236, and 211 MPa for the TiZrVNb, Ti_{1.5}ZrVNb and Ti₂ZrVNb RHEAs due to high temperature softening. Table 5 summarizes the yield strength of other hot deformed RHEAs at different strain rates, strains, and temperatures.

In summary, the addition of Cr, Al, W, V, Nb, Mo, Ta, Zr, Ti, and Hf contributed to the high temperature strength and plasticity of RHEAs. The addition of Mo has shown a good ability to softening resistance of RHEAs at higher temperatures than RHEAs without Mo. However, the excessive addition of Mo tends to decrease the high temperature yield strength. W and Mo have strong strengthening effects, which induce phase boundaries that act as dislocation movements

Table 5. Summary of yield strength of hot deformed RHEAs at 800–1400 °C.

Alloy composition	Strain	Strain rate (s ⁻¹)	Temperature (°C)	Yield strength (MPa)	Plasticity (%)	References
HfMoNbTaTiZr	–	10 ⁻³	800	1007	>20	[87]
			1000	814	30	
			1200	556	30	
AlNb ₂ TiV	0.3	10 ⁻³	800	694	>30	[118]
			1000	612	>30	
NbTiVZr	0.69	10 ⁻³	800	500	>40	[119]
			1000	200	50	
Zr _{2.0} TiHfVNb _{2.0}	0.69	10 ⁻³	800	647	50	[81]
			900	476	50	
			1200	165	50	
HfNbTaTiZrW	–	10 ⁻³	800	535	>35	[120]
			1000	295	>35	
			1200	92	>35	
HfNbTaTiZrMoW	–	10 ⁻³	800	577	>35	[120]
			1000	409	>35	
			1200	345	>35	
TiAl ₂₀ VNbMo	–	10 ⁻³	800	624	>50	[57]
			1000	180	>50	
NbMoCrTiAl	–	10 ⁻³	800	860	0.135	[121]
			1000	594	>0.15	
			1200	105	>0.24	
NbTaW _{0.5} (Mo ₂ C) ₂₀	–	10 ⁻³	1200	1026	22.3	[122]
			1400	697	>30	
C _{0.25} Hf _{0.25} NbTaW _{0.5}	–	10 ⁻³	1000	868	50	[123]
			1200	792	50	
			1400	749	50	
Re _{0.5} MoNbW(TaC) _{0.5}	0.48	10 ⁻³	1000	937	15.9	[124]
			1200	901	35	
Ti ₂ VNbMoZr ₂	–	10 ⁻³	800	977	>50	[125]
Zr ₁ NbMoTaW	–	10 ⁻³	1000	555	25	[98]
(VNbTiTa) ₉₀ Si ₁₀	–	10 ⁻³	1000	646	>50	[103]
			1200	133	>50	
TiNbTaZrHf	–	10 ⁻³	800	1351	38.5	[126]
			1000	508	44	
			1200	105	>50	
ZrTiHfNb _{0.5} Ta _{0.5} O _{0.2}	–	10 ⁻³	800	966	>50	[127]
				537	>50	
HfNbTiVSi _{0.5}	0.69	10 ⁻³	800	875	50	[55]
			1000	240	50	
ZrTiHfNb _{0.5} Ta _{0.5}	0.69	10 ⁻³	800	170	50	[78]
			900	112	50	
ZrTiHfNb _{0.5} Mo _{0.5}	0.69	10 ⁻³	800	539	50	[78]
			900	274	50	
W _{0.4} MoNb _{1.3} TaTi	0.69	10 ⁻³	800	1125	44.7	[76]
			1000	905	37.2	
			1100	425	48.3	
TiVNbMoTaW	–	10 ⁻³	800	791	>10	[128]
			1000	753	>25	
			1200	659	10	
V ₁ NbMoTa	–	10 ⁻³	800	933	>51	[113]
			1000	811	>51	

and therefore strengthen RHEA at high temperatures. The addition of Ti and Zr into RHEA simultaneously helps with high temperature strength, plasticity and density reduction. Excessive additions of Ti, Zr, and Hf weaken the high temperature yield strength due to their melting points and self-diffusion activation energies. Cr also enhances the high temperature strength of RHEA through the formation of second-phase particles and solid solution strengthening. Research progress has been made in creating a balance in the toughening and high temperature strength of RHEA by introducing carbides, silicides, and oxides (Tab. 5). As seen from reported literature, carbides and oxides containing RHEAs tend to have better strength and plasticity at higher temperatures. Good softening resistance and plasticity at higher temperatures can be achieved for RHEAs through optimization of elemental compositions or the addition of composite particulates.

9 Summary

The deformation behavior of RHEAs at high temperatures have been reviewed in this work. Based on the outcome of this review, the conclusions can be summarised as follows:

- All the probable mechanisms influencing deformation in metallic alloys, such as WH, DRX, and DRV, have also been observed in refractory high entropy alloys. The DDRX and CDRX occurred concurrently in some of the hot deformed RHEAs.
- Sharp drops in flow stress occur in RHEA, which may be attributed to the unlocking of dislocations pinned by the Cottrell atmosphere or short-range order.
- The resistance to high temperature softening of RHEAs can be attributed to slow diffusion of elements, high thermal stability, and second phase strengthening mechanism. A balance between toughness and strength at higher temperatures for RHEAs can be achieved through optimization of elemental contents or the addition of composite particulates to induce a dual-phase structure.
- The microcracks in RHEA based composites are mostly intergranular. The appearance of microcracks at the sides of carbide reinforced RHEAs confirms their load-bearing effect during the high temperature deformation process.
- Constitutive modelling can be used to accurately predict flow stress of hot worked RHEAs. Hot deformation processing maps can be employed to avoid microstructural defects and improve the mechanical and functional properties of RHEAs during hot deformation processing.

10 Future research perspective

- Constitutive modelling provides insights into the microstructural mechanisms influencing the deformation process. Among the constitutive models, the Arrhenius model is commonly used in the modelling of flow curves in RHEAs. The use of improved Johnson-Cook, Zerilli-Armstrong, and other promising models can be investigated since they are rarely reported.

- The use of hot-processing maps to obtain the safe processing window for the hot forming of RHEAs is rarely reported. Apart from the flow curve shape in determining the softening behaviour in RHEAs, cracks and other defects might occur during the hot forming process. Therefore, more studies required on hot-processing maps of RHEAs.
- Studies on the optimization of hot deformation properties and deformation behaviour of composite reinforced RHEAs is expected to constitute a major part of future work because it has been reported to have high yield strength and plasticity at elevated temperatures. Also, the effects of different temperatures and high strain rates on the deformation behaviour of carbide reinforced refractory high entropy alloy.
- The low angle grain boundary formed to bridge bulged grain boundaries, which led to the formation of fine grains. Investigation into the effect of different strain rates on the evolution of low angle grain boundary misorientation can be explored.
- More extensive research is needed on the dislocation motion mechanisms of hot deformed powder metallurgy RHEAs since only one article was found reporting on the dislocation motion mechanisms.

Declaration of competing interests

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References

1. J.W. Yeh, S.K. Chen, S.J. Lin, J.Y. Gan, T.S. Chin, T.T. Shun, C.H. Tsau, S.Y. Chang, Nanostructured high-entropy alloys with multiple principal elements: novel alloy design concepts and outcomes, *Adv. Eng. Mater.* **6** (2004) 299–303
2. A. Takeuchi, T. Wada, H. Kato, Solid solutions with bcc, hcp, and fcc Structures formed in a composition line in multicomponent Ir Rh Ru W Mo system, *Mater. Trans.* **60** (2019) 2267–2276
3. J. Tang, J.L. Xu, Z.G. Ye, X.B. Li, J.M. Luo, Microwave sintered porous CoCrFeNiMo high entropy alloy as an efficient electrocatalyst for alkaline oxygen evolution reaction, *J. Mater. Sci. Technol.* **79** (2021) 171–177
4. B. Cantor, I.T.H. Chang, P. Knight, A.J.B. Vincent, Microstructural development in equiatomic multicomponent alloys, *Mater. Sci. Eng. A* **375–377** (2004) 213–218
5. O.N. Senkov, G.B. Wilks, D.B. Miracle, C.P. Chuang, P.K. Liaw, Refractory high-entropy alloys, *Intermetallics* **18** (2010) 1758–1765
6. M.C. Gao, Progress in high-entropy alloys, *JOM* **66** (2014) 1964–1965

7. M.H. Tsai, J.W. Yeh, High-entropy alloys: a critical review, *Mater. Res. Lett.* **2** (2014) 107–123
8. S. Praveen, J.W. Bae, P. Asghari-Rad, J.M. Park, H.S. Kim, Ultra-high tensile strength nanocrystalline CoCrNi equi-atomic medium entropy alloy processed by high-pressure torsion, *Mater. Sci. Eng. A* **735** (2018) 394–397
9. C.L. Chen, Sutrisna, Study of NiFeCoCr medium entropy alloy as a binder phase on W-Mo heavy tungsten alloy by secondary ball milling, *Intermetallics* **138** (2021) 107320
10. Y.S. Lee, R.J. Cava, Superconductivity in high and medium entropy alloys based on MoReRu, *Physica C* **566** (2019) 1353520
11. S. Uporov, E. Sterkhov, I. Balyakin, Magnetocaloric effect in ScGdHo medium-entropy alloy, *J. Supercond. Nov. Magn.* (2022) 1539–1545
12. M.O. Bodunrin, B.A. Obadele, L.H. Chown, P.A. Olubambi, Corrosion behaviour of Al-Fe-Ti-V medium entropy alloy, *IOP Conf. Ser. Mater. Sci. Eng.* **272** (2017)
13. D.B. Miracle, O.N. Senkov, A critical review of high entropy alloys and related concepts, *Acta Mater.* **122** (2017) 448–511
14. Q. Li, X. Bao, S. Zhao, Y. Zhu, Y. Lan, X. Feng, Q. Zhang, The influence of AlFeNiCrCoTi high-entropy alloy on microstructure, mechanical properties and tribological behaviors of aluminum matrix composites, *Int. J. Metalcast.* **15** (2021) 281–291
15. M. Tsai, J. Yeh, High-entropy alloys: a critical review, *Mater. Res. Lett.* **2** (2014) 107–123
16. O.N. Senkov, D.B. Miracle, K.J. Chaput, J.P. Couzinie, Development and exploration of refractory high entropy alloys – a review, *J. Mater. Res.* **33** (2018) 3092–3128
17. S. Praveen, H.S. Kim, High-entropy alloys: potential candidates for high-temperature applications – an overview, *Adv. Eng. Mater.* 1700645 (2018) 1–22
18. L. Raman, A. Anupam, G. Karthick, C.C. Berndt, A.S.M. Ang, S.V.S. Narayana Murty, D. Fabijanic, B.S. Murty, R.S. Kottada, Strengthening mechanisms in CrMoNbTiW refractory high entropy alloy, *Mater. Sci. Eng. A* **819** (2021) 1–10
19. S.A. Alvi, *Synthesis and Characterisation of High Entropy Alloy and Coating* (Lulea University of Technology, Sweden, 2019)
20. V. Bhardwaj, Q. Zhou, F. Zhang, W. Han, Y. Du, K. Hua, H. Wang, Effect of Al addition on the microstructure, mechanical and wear properties of TiZrNbHf refractory high entropy alloys, *Tribol. Int.* **160** (2021) 1–11
21. M. Sahlberg, D. Karlsson, C. Zlotea, U. Jansson, Superior hydrogen storage in high entropy alloys, *Sci. Rep.* **6** (2016) 1–6
22. C. Zhang, A. Song, Y. Yuan, Y. Wu, P. Zhang, Z. Lu, X. Song, Study on the hydrogen storage properties of a TiZrNbTa high entropy alloy, *Int. J. Hydrogen Energy* **45** (2020) 5367–5374
23. R. Floriano, G. Zepon, K. Edalati, G.L.B.G. Fontana, A. Mohammadi, Z. Ma, H.W. Li, R.J. Contieri, Hydrogen storage in TiZrNbFeNi high entropy alloys, designed by thermodynamic calculations, *Int. J. Hydrogen Energy* **45** (2020) 33759–33770
24. J. Montero, G. Ek, L. Laversenne, V. Nassif, G. Zepon, M. Sahlberg, C. Zlotea, Hydrogen storage properties of the refractory Ti-V-Zr-Nb-Ta multi-principal element alloy, *J. Alloys Compd.* **835** (2020) 155376
25. H. Li, L. Zhao, Y. Yang, H. Zong, X. Ding, Improving radiation-tolerance of bcc multi-principal element alloys by tailoring compositional heterogeneities, *J. Nucl. Mater.* **555** (2021) 1–10
26. T. Li, L. Yiping, C. Zhiqiang, W. Tongmin, L. Tingju, Opportunity and challenge of refractory high-entropy alloys in the field of reactor structural materials, *Acta Metall. Sin.* **57** (2021) 42
27. I.A. Ivanov, A. Ryskulov, A. Kurakhmedov, A. Kozlovskiy, D. Shlimas, M.V. Zdorovets, V.V. Uglov, S.V. Zlotski, J. Ke, Radiation swelling and hardness of high-entropy alloys based on the TiTaNbV system irradiated with krypton ions, *J. Mater. Sci.: Mater. Electr.* **32** (2021) 27260–27267
28. X. Shang, S. Bo, Y. Guo, Q. Liu, ZrC reinforced refractory-high-entropy-alloy coatings: compositional design, synthesis, interstitials, and microstructure evolution effects on wear, corrosion and oxidation behaviors, *Appl. Surf. Sci.* **564** (2021) 1–11
29. S.B. Hung, C.J. Wang, Y.Y. Chen, J.W. Lee, C.L. Li, Thermal and corrosion properties of V-Nb-Mo-Ta-W and V-Nb-Mo-Ta-W-Cr-B high entropy alloy coatings, *Surf. Coat Technol.* **375** (2019) 802–809
30. S.K. Bachani, C.J. Wang, B.S. Lou, L.C. Chang, J.W. Lee, Fabrication of TiZrNbTaFeN high-entropy alloys coatings by HiPIMS: effect of nitrogen flow rate on the microstructural development, mechanical and tribological performance, electrical properties and corrosion characteristics, *J. Alloys Compd.* **873** (2021) 1–15
31. A. Ostovari, J. Pasandideh, A. Abdollahzadeh, N.A. Shaburova, E. Trofimov, On the application of NbTaTiVW refractory high entropy alloy particles in the manufacturing process of WC based matrix body drill bits, *Int. J. Refract. Metals Hard Mater.* **99** (2021) 1–7
32. R.R. Eleti, T. Bhattacharjee, A. Shibata, N. Tsuji, Unique deformation behavior and microstructure evolution in high temperature processing of HfNbTaTiZr refractory high entropy alloy, *Acta Mater.* **171** (2019) 132–145
33. F.J. Humphreys, M. Hatherly, Hot deformation and dynamic restoration, in: *Recrystallization and Related Annealing Phenomena*, edited by M. Hatherly and F.J. Humphreys, 2nd edn. (Elsevier, Oxford, 2004), pp. 415–451
34. R.R. Eleti, *Deformation Mechanisms and Microstructure Evolution in HfNbTaTiZr High Entropy Alloy during Thermo-mechanical Processing at Elevated Temperatures* (Kyoto University, Japan, 2019)
35. C.M. Sellars, W.J.M. Tegart, Hot workability, *Int. Metall. Rev.* **17** (1972) 1–24
36. T. Sakai, J.J. Jonas, Plastic deformation: role of recovery and recrystallization, *Mater. Sci. Technol.* (2001) 7079–7084
37. X.M. Chen, Y.C. Lin, D.X. Wen, J.L. Zhang, M. He, Dynamic recrystallization behavior of a typical nickel-based superalloy during hot deformation, *Mater. Des.* **57** (2014) 568–577
38. R.D. Doherty, D.A. Hughes, F.J. Humphreys, J.J. Jonas, D. Juul Jensen, M.E. Kassner, W.E. King, T.R. McNelley, H.J. McQueen, A.D. Rollett, Current issues in recrystallization: a review, *Mater. Sci. Eng. A* (1997) 219–274
39. M. lei Hu, W. dong Song, D. bo Duan, Y. Wu, Dynamic behavior and microstructure characterization of TaNbHfZrTi high-entropy alloy at a wide range of strain rates and temperatures, *Int. J. Mech. Sci.* **182** (2020) 105738
40. F. Dong, Y. Yuan, W. Li, Y. Zhang, P.K. Liaw, X. Yuan, H. Huang, hot deformation behavior and processing maps of an equiatomic MoNbHfZrTi refractory high entropy alloy, *Intermetallics (Barking)* **126** (2020) 1–10

41. J. Pang, H.H. Zhang, L. Zhang, Z. Zhu, H. Fu, H. Li, A. Wang, Z. Li, H.H. Zhang, A ductile Nb₄₀Ti₂₅Al₁₅V₁₀-Ta₅Hf₃W₂ refractory high entropy alloy with high specific strength for high-temperature applications, *Mater. Sci. Eng. A* **831** (2022) 1–9
42. Z. Savaedi, R. Motallebi, H. Mirzadeh, A review of hot deformation behavior and constitutive models to predict flow stress of high-entropy alloys, *J. Alloys Compd.* **903** (2022) 1–21
43. H.Y. Diao, R. Feng, K.A. Dahmen, P.K. Liaw, Fundamental deformation behavior in high-entropy alloys: an overview, *Curr. Opin. Solid State Mater. Sci.* **21** (2017) 252–266
44. J. Brechtel, S. Chen, C. Lee, Y. Shi, R. Feng, X. Xie, D. Hamblin, A.M. Coleman, B. Straka, H. Shortt, R.J. Spurling, P.K. Liaw, A review of the serrated-flow phenomenon and its role in the deformation behavior of high-entropy alloys, *Metals (Basel)* **10** (2020) 1–72
45. A.H. Chokshi, High temperature deformation in fine grained high entropy alloys, *Mater. Chem. Phys.* **210** (2018) 152–161
46. E.P. George, W.A. Curtin, C.C. Tasan, High entropy alloys: a focused review of mechanical properties and deformation mechanisms, *Acta Mater.* **188** (2020) 435–474
47. E.P. George, W.A. Curtin, C.C. Tasan, High entropy alloys: a focused review of mechanical properties and deformation mechanisms, *Acta Mater.* **188** (2020) 435–474
48. E.S. Vieira, J.A.N.F. Gomes, A comparison of scopus and web of science for a typical university, *Scientometrics* **81** (2009) 587–600
49. O.N. Senkov, G.B. Wilks, J.M. Scott, D.B. Miracle, Mechanical properties of Nb₂₅Mo₂₅Ta₂₅W₂₅ and V₂₀Nb₂₀Mo₂₀Ta₂₀W refractory high entropy alloys, *Intermetallics* **19** (2011) 698–706
50. O.N. Senkov, J.M. Scott, S.V. Senkova, F. Meisenkothen, D.B. Miracle, C.F. Woodward, Microstructure and elevated temperature properties of a refractory TaNbHfZrTi alloy, *J. Mater. Sci.* **47** (2012) 4062–4074
51. C.C. Juan, M.H. Tsai, C.W. Tsai, C.M. Lin, W.R. Wang, C.C. Yang, S.K. Chen, S.J. Lin, J.W. Yeh, Enhanced mechanical properties of HfMoTaTiZr and HfMoNbTaTiZr refractory high-entropy alloys, *Intermetallics (Barking)* **62** (2015) 76–83
52. O.N. Senkov, C.F. Woodward, Microstructure and properties of a refractory NbCrMo_{0.5}Ta_{0.5}TiZr alloy, *Mater. Sci. Eng. A* **529** (2011) 311–320
53. O.N. Senkov, C. Woodward, D.B. Miracle, Microstructure and properties of aluminum-containing refractory high-entropy alloys, *JOM* **66** (2014) 2030–2042
54. N.N. Guo, L. Wang, L.S. Luo, X.Z. Li, R.R. Chen, Y.Q. Su, J.J. Guo, H.Z. Fu, Hot deformation characteristics and dynamic recrystallization of the MoNbHfZrTi refractory high-entropy alloy, *Mater. Sci. Eng. A* **651** (2016) 698–707
55. Y. Zhang, Y. Liu, Y. Li, X. Chen, H. Zhang, Microstructure and mechanical properties of a refractory HfNbTiVSi_{0.5} high-entropy alloy composite, *Mater. Lett.* **174** (2016) 82–85
56. W. Guo, B. Liu, Y. Liu, T. Li, A. Fu, Q. Fang, Y. Nie, Microstructures and mechanical properties of ductile NbTaTiV refractory high entropy alloy prepared by powder metallurgy, *J. Alloys Compd.* **776** (2019) 428–436
57. Z.Q. Xu, Z.L. Ma, M. Wang, Y.W. Chen, Y.D. Tan, X.W. Cheng, Design of novel low-density refractory high entropy alloys for high-temperature applications, *Mater. Sci. Eng. A* **755** (2019) 318–322
58. T. Yan, E. Yu, Y. Zhao, Constitutive modeling for flow stress of 55SiMnMo bainite steel at hot working conditions, *Mater. Des.* **50** (2013) 574–580
59. X.M. Chen, Y.C. Lin, M.S. Chen, H.B. Li, D.-X. Wen, J.L. Zhang, M. He, Microstructural evolution of a nickel-based superalloy during hot deformation, *Mater. Des.* (2015) 41–49
60. C.W. Park, M.S. Choi, H. Lee, J. Yoon, H.R. Javadinejad, J.H. Kim, High-temperature deformation behavior and microstructural evolution of as-cast and hot rolled-21S alloy during hot deformation, *J. Mater. Res. Technol.* **9** (2020) 13555–13569
61. Q. Yang, M. Ma, Y. Tan, S. Xiang, F. Zhao, Y. Liang, Initial β grain size effect on high-temperature flow behavior of tb8 titanium alloys in single β phase field, *Metals (Basel)* **9** (2019) 1–19
62. Y. Wan, J. Mo, X. Wang, Z. Zhang, B. Shen, X. Liang, Mechanical properties and phase stability of WTaMoNbTi refractory high-entropy alloy at elevated temperatures, *Acta Metall. Sin. (English Letters)* **34** (2021) 1585–1590
63. Q. Liu, G. Wang, Y. Liu, X. Sui, Y. Chen, S. Luo, Hot deformation behaviors of an ultrafine-grained MoNbTaTiV refractory high-entropy alloy fabricated by powder metallurgy, *Mater. Sci. Eng. A* **809** (2021) 1–10
64. M. Azarbaras, M. Aghaie-Khafri, J.M. Cabrera, J. Calvo, Dynamic recrystallization mechanisms and twinning evolution during hot deformation of Inconel 718, *Mater. Sci. Eng. A* **678** (2016) 137–152
65. Y.C. Lin, M.S. Chen, J. Zhong, Constitutive modeling for elevated temperature flow behavior of 42CrMo steel, *Comput. Mater. Sci.* **42** (2008) 470–477
66. H. Wu, S.P. Wen, H. Huang, K.Y. Gao, X.L. Wu, W. Wang, Z.R. Nie, Hot deformation behavior and processing map of a new type Al-Zn-Mg-Er-Zr alloy, *J. Alloys Compd.* **685** (2016) 869–880
67. S. Miura, Y. Murasato, Y. Sekito, Y. Tsutsumi, K. Ohkubo, Y. Kimura, Y. Mishima, T. Mohri, Effect of microstructure on the high-temperature deformation behavior of Nb-Si alloys, *Mater. Sci. Eng. A* 510–511 (2009) 317–321
68. E.P. George, D. Raabe, R.O. Ritchie, High-entropy alloys, *Nat. Rev. Mater.* **4** (2019) 515–534
69. T. Li, Y. Lu, Z. Li, T. Wang, T. Li, Hot deformation behavior and microstructure evolution of non-equimolar Ti₂ZrHfV_{0.5}Ta_{0.2} refractory high-entropy alloy, *Intermetallics (Barking)* **146** (2022) 1–11
70. O.N. Senkov, J.P. Couzinie, S.I. Rao, V. Soni, R. Banerjee, Temperature dependent deformation behavior and strengthening mechanisms in a low density refractory high entropy alloy Al₁₀Nb₁₅Ta₅Ti₃₀Zr₄₀, *Materialia (Oxf)* **9** (2020) 1–16
71. Z. Yu, B. Xie, Z. Zhu, B. Xu, M. Sun, High-temperature deformation behavior and processing maps of a novel AlNbTi₃VZr_{1.5} refractory high entropy alloy, *J. Alloys Compd.* **912** (2022) 1–11
72. Z.C. Bai, X.F. Ding, Q. Hu, M. Yang, Z.T. Fan, X.W. Liu, Unique deformation behavior and microstructure evolution in high-temperature processing of a low-density TiAlVNb₂ refractory high-entropy alloy, *J. Alloys Compd.* **885** (2021)
73. H. Liang, H. Guo, Y. Nan, C. Qin, X. Peng, J. Zhang, The construction of constitutive model and identification of dynamic softening mechanism of high-temperature deformation of Ti-5Al-5Mo-5V-1Cr-1Fe alloy, *Mater. Sci. Eng. A* **615** (2014) 42–50

74. Y. Wu, Y. Liu, C. Li, X. Xia, Y. Huang, H. Li, H. Wang, Deformation behavior and processing maps of Ni3Al-based superalloy during isothermal hot compression, *J. Alloys Compd.* **712** (2017) 687–695
75. T. Sakai, A. Belyakov, R. Kaibyshev, H. Miura, J.J. Jonas, Dynamic and post-dynamic recrystallization under hot, cold and severe plastic deformation conditions, *Prog. Mater. Sci.* **60** (2014) 130–207
76. W. Wei, T. Wang, C. Wang, M. Wu, Y. Nie, J. Peng, Ductile W0.4MoNb_xTaTi refractory high-entropy alloys with excellent elevated temperature strength, *Mater. Lett.* **295** (2021) 1–4
77. Q. Liu, G. Wang, Y. Liu, X. Sui, Y. Chen, S. Luo, Hot deformation behaviors of an ultrafine-grained MoNbTaTiV refractory high-entropy alloy fabricated by powder metallurgy, *Mater. Sci. Eng. A* **809** (2021) 140922
78. R. Feng, B. Feng, M.C. Gao, C. Zhang, J.C. Neufelind, J.D. Poplawsky, Y. Ren, K. An, M. Widom, P.K. Liaw, Superior high-temperature strength in a supersaturated refractory high-entropy alloy, *Adv. Mater.* **33** (2021) 1–9
79. Y. Zhang, J.P. Liu, S.Y. Chen, X. Xie, P.K. Liaw, K.A. Dahmen, J.W. Qiao, Y.L. Wang, Serration and noise behavior in materials, *Prog. Mater. Sci.* **90** (2017) 358–460
80. X.W. Liu, Z.C. Bai, X.F. Ding, J.Q. Yao, L. Wang, Y.Q. Su, Z.T. Fan, J.J. Guo, A novel light-weight refractory high-entropy alloy with high specific strength and intrinsic deformability, *Mater. Lett.* **287** (2021) 1–4
81. Y. Chen, Y. Li, X. Cheng, C. Wu, B. Cheng, Z. Xu, The microstructure and mechanical properties of refractory high-entropy alloys with high plasticity, *Materials* **11** (2018) 1–11
82. Y. Wan, Q. Wang, J. Mo, Z. Zhang, X. Wang, X. Liang, B. Shen, WReTaMo refractory high-entropy alloy with high strength at 1600 °C, *Adv. Eng. Mater.* **24** (2022) 1–31
83. N. Yurchenko, E. Panina, D. Shaysultanov, S. Zherebtsov, N. Stepanov, Refractory high entropy alloy with ductile intermetallic B2 matrix/hard bcc particles and exceptional strain hardening capacity, *Materialia (Oxf)* **20** (2021) 1–7
84. S. Wu, D. Qiao, H. Zhang, J. Miao, H. Zhao, J. Wang, Y. Lu, T. Wang, T. Li, Microstructure and mechanical properties of CxHf0.25NbTa W0.5 refractory high-entropy alloys at room and high temperatures, *J. Mater. Sci. Technol.* **97** (2022) 229–238
85. T. Huang, S. Wu, H. Jiang, Y. Lu, T. Wang, T. Li, Effect of Ti content on microstructure and properties of Ti_xZrVNb refractory high-entropy alloys, *Int. J. Minerals Metall. Mater.* **27** (2020) 1318–1325
86. C. Yang, H. Bian, K. Aoyagi, Y. Hayasaka, K. Yamanaka, A. Chiba, Synergetic strengthening in HfMoNbTaTi refractory high-entropy alloy via disordered nanoscale phase and semicoherent refractory particle, *Mater. Des.* **212** (2021) 110248
87. K.K. Tseng, C.C. Juan, S. Tso, H.C. Chen, C.W. Tsai, J.W. Yeh, Effects of Mo, Nb, Ta, Ti, and Zr on mechanical properties of equiatomic Hf-Mo-Nb-Ta-Ti-Zr alloys, *Entropy* **21** (2019) 1–14
88. A. Chaudhuri, A. Sarkar, R. Kapoor, J.K. Chakravarty, R.K. Ray, S. Suwas, Understanding the mechanism of dynamic recrystallization during high-temperature deformation in Nb-1Zr-0.1C alloy, *J. Mater. Eng. Perform.* **28** (2019) 448–462
89. O.N. Senkov, J. Gild, T.M. Butler, Microstructure, mechanical properties and oxidation behavior of NbTaTi and NbTaZr refractory alloys, *J. Alloys Compd.* **862** (2021) 1–14
90. Y. Jia, L. Zhang, P. Li, X. Ma, L. Xu, S. Wu, Y. Jia, G. Wang, Microstructure and mechanical properties of Nb-Ti-V-Zr refractory medium-entropy alloys, *Front. Mater.* **7** (2020) 1–11
91. Y. Guo, J. He, W. Lu, L. Jia, Z. Li, The evolution of compositional and microstructural heterogeneities in a TaMo0.5ZrTi1.5Al0.1Si0.2 high entropy alloy, *Mater. Charact.* **172** (2021) 1–11
92. R.R. Eleti, A.H. Chokshi, A. Shibata, N. Tsuji, Unique high-temperature deformation dominated by grain boundary sliding in heterogeneous necklace structure formed by dynamic recrystallization in HfNbTaTiZr BCC refractory high entropy alloy, *Acta Mater.* **183** (2020) 64–77
93. B.S. Murty, J.W. Yeh, S. Ranganathan, *High-Entropy Alloys*, 1st ed. (Butterworth-Heinemann, UK, 2014)
94. Y. Zhang, Y. Liu, Y. Li, X. Chen, H. Zhang, Microstructure and mechanical properties of a refractory HfNbTiVSi0.5 high-entropy alloy composite, *Mater. Lett.* **174** (2016) 82–85
95. Q. Liu, G. Wang, Y. Liu, X. Sui, Y. Chen, S. Luo, Hot deformation behaviors of an ultrafine-grained MoNbTaTiV refractory high-entropy alloy fabricated by powder metallurgy, *Mater. Sci. Eng. A* **809** (2021) 1–10
96. H.Y. Yasuda, Y. Yamada, K. Cho, T. Nagase, Deformation behavior of HfNbTaTiZr high entropy alloy single crystals and polycrystals, *Mater. Sci. Eng. A* **809** (2021) 1–10
97. M. Lei Hu, W. Dong Song, D. Bo Duan, Y. Wu, Dynamic behavior and microstructure characterization of TaNbHfZrTi high-entropy alloy at a wide range of strain rates and temperatures, *Int. J. Mech. Sci.* **182** (2020) 1–12
98. T.T. Li, W. Jiao, J. Miao, Y. Lu, E. Guo, T. Wang, T.T. Li, P.K. Liaw, A novel ZrNbMoTaW refractory high-entropy alloy with in-situ forming heterogeneous structure, *Mater. Sci. Eng. A* **827** (2021) 1–9
99. Q. Wei, G. Luo, R. Tu, J. Zhang, Q. Shen, Y. Cui, Y. Gui, A. Chiba, High-temperature ultra-strength of dual-phase Re0.5MoNbW(TaC)0.5 high-entropy alloy matrix composite, *J. Mater. Sci. Technol.* **84** (2021) 1–9
100. R. Feng, B. Feng, M.C. Gao, C. Zhang, J.C. Neufelind, J.D. Poplawsky, Y. Ren, K. An, M. Widom, P.K. Liaw, Superior high-temperature strength in a supersaturated refractory high-entropy alloy, *Adv. Mater.* **33** (2021) 1–9
101. Y.V.R.K. Prasad, K.P. Rao, S. Sasidhara, *Hot Working Guide*, 2nd edn. (ASM International, Ohio, 2015)
102. S.A. Sajjadi, A. Chaichi, H.R. Ezatpour, A. Maghsoudlou, M.A. Kalaie, Hot deformation processing map and microstructural evaluation of the Ni-based superalloy IN-738LC, *J. Mater. Eng. Perform.* **25** (2016) 1269–1275
103. Z.Q. Xu, Z.L. Ma, Y. Tan, M. Wang, Y. Zhao, X.W. Cheng, Effects of Si additions on microstructures and mechanical properties of VNbTiTaSix refractory high-entropy alloys, *J. Alloys Compd.* **900** (2022) 1–9
104. X. Li, A. Fu, Y. Cao, S. Xu, S. Gao, High-temperature mechanical properties and deformation behavior of carbides reinforced TiNbTaZrHf composite, *J. Alloys Compd.* **894** (2022) 1–10

105. Y.W. Chen, Y.K. Li, X.W. Cheng, C. Wu, B. Cheng, High-temperature mechanical properties and microstructure of ZrTiHfNbMo_x ($x=0.5, 1.0, 1.5$) refractory high entropy alloys, *IOP Conf. Ser. Mater. Sci. Eng.* **359** (2018) 1–5
106. A.N. Behera, R. Kapoor, A. Sarkar, J.K. Chakravartty, Hot deformation behaviour of niobium in temperature range 700–1500°C, *Mater. Sci. Technol.* **30** (2014) 637–644
107. K. Wu, G.Q. Liu, B.F. Hu, C.Y. Wang, Y.W. Zhang, Y. Tao, J.T. Liu, Effect of processing parameters on hot compressive deformation behavior of a new Ni-Cr-Co based P/M superalloy, *Mater. Sci. Eng. A* **528** (2011) 4620–4629
108. A. Momeni, The physical interpretation of the activation energy for hot deformation of Ni and Ni-30Cu alloys, *J. Mater. Res.* **31** (2016) 1077–1084
109. C. Shi, X.G. Chen, Evolution of activation energies for hot deformation of 7150 aluminum alloys with various Zr and V additions, *Mater. Sci. Eng. A* **650** (2016) 197–209
110. Y.C. Lin, X.M. Chen, A critical review of experimental results and constitutive descriptions for metals and alloys in hot working, *Mater. Des.* **32** (2011) 1733–1759
111. G.R. Johnson, W.H. Cook, A computational constitutive model and data for metals subjected to large strain, high strain rates and high pressures, in *The Seventh International Symposium on Ballistics*, Den Haag, Netherlands (1983) pp. 541–547
112. J.J. Jonas, C.M. Sellars, W.J.M.G. Tegart, Strength and structure under hot-working conditions, *Metall. Rev.* **14** (1969) 1–24
113. M. Wang, Z.L. Ma, Z.Q. Xu, X.W. Cheng, Designing VxNbMoTa refractory high-entropy alloys with improved properties for high-temperature applications, *Scr. Mater.* **191** (2021) 131–136
114. T. Sakai, M. Ohashi, K. Chiba, J.J. Jonas, Recovery and recrystallization of polycrystalline nickel after hot working, *Acta Metall.* **36** (1988) 1781–1790
115. Z.D. Han, N. Chen, S.F. Zhao, L.W. Fan, G.N. Yang, Y. Shao, K.F. Yao, Intermetallics effect of Ti additions on mechanical properties of NbMoTaW and VNbMoTaW refractory high entropy alloys, *Intermetallics* **84** (2017) 153–157
116. O.N. Senkov, S. Gorsse, D.B. Miracle, High temperature strength of refractory complex concentrated alloys, *Acta Mater.* **175** (2019) 394–405
117. Y. Wan, J. Mo, X. Wang, Z. Zhang, B. Shen, X. Liang, Mechanical properties and phase stability of WTaMoNbTi refractory high - entropy alloy at elevated temperatures, *Acta Metall. Sin.* **34** (2021) 1585–1590
118. X.W. Liu, Z.C. Bai, X.F. Ding, J.Q. Yao, L. Wang, Y.Q. Su, Z.T. Fan, J.J. Guo, A novel light-weight refractory high-entropy alloy with high specific strength and intrinsic deformability, *Mater. Lett.* **287** (2021) 1–4
119. Y. Jia, L. Zhang, P. Li, X. Ma, L. Xu, S. Wu, Y. Jia, Microstructure and mechanical properties of Nb-Ti-V- Zr refractory medium-entropy alloys, *Front Mater.* **7** (2020) 1–11
120. M. Wang, Z. Ma, Z. Xu, X. Cheng, Microstructures and mechanical properties of HfNbTaTiZrW and HfNbTaTiZr-MoW refractory high-entropy alloys, *J. Alloys Compd.* **803** (2019) 778–785
121. H. Chen, A. Kauffmann, B. Gorr, D. Schliephake, C. Seemüller, J.N. Wagner, H.J. Christ, M. Heilmaier, Microstructure and mechanical properties at elevated temperatures of a new Al-containing refractory high-entropy alloy Nb-Mo-Cr-Ti-Al, *J. Alloys Compd.* **661** (2016) 206–215
122. S. Wu, D. Qiao, H. Zhao, J. Wang, Y. Lu, A novel NbTaW0.5(Mo2C)x refractory high-entropy alloy with excellent mechanical properties, *J. Alloys Compd.* **889** (2022) 1–12
123. S. Wu, D. Qiao, H. Zhang, J. Miao, H. Zhao, J. Wang, Y. Lu, T. Wang, T. Li, Microstructure and mechanical properties of CxHf0.25NbTaW0.5 refractory high-entropy alloys at room and high temperatures, *J. Mater. Sci. Technol.* **97** (2022) 229–238
124. Q. Wei, G. Luo, R. Tu, J. Zhang, Q. Shen, Y. Cui, Y. Gui, A. Chiba, High-temperature ultra-strength of dual-phase $\text{Re0.5MoNbW(TaC)0.5}$ high-entropy alloy matrix composite, *J. Mater. Sci. Technol.* **84** (2021) 1–9
125. T. Li, J. Miao, Y. Lu, T. Wang, T. Li, Effect of Zr on the as-cast microstructure and mechanical properties of lightweight Ti2VNbMoZrx refractory high-entropy alloys, *Int. J. Refract. Metals Hard Mater.* **103** (2022) 1–8
126. X. Li, A. Fu, Y. Cao, S. Xu, S. Gao, High-temperature mechanical properties and deformation behavior of carbides reinforced TiNbTaZrHf composite, *J. Alloys Compd.* **894** (2022) 1–10
127. Y. Chen, Y. Li, X. Cheng, Z. Xu, C. Wu, B. Cheng, M. Wang, Interstitial strengthening of refractory $\text{ZrTiHfNb0.5Ta0.5O}_x$ ($X=0.05, 0.1, 0.2$) high-entropy alloys, *Mater. Lett.* **228** (2018) 145–147
128. Z.D. Han, N. Chen, S.F. Zhao, L.W. Fan, G.N. Yang, Y. Shao, K.F. Yao, Effect of Ti additions on mechanical properties of NbMoTaW and VNbMoTaW refractory high entropy alloys, *Intermetallics* **84** (2017) 153–157

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