Texture Evolution and Dynamic Recrystallization of Zr–1Sn–0.3Nb–0.3Fe–0.1Cr Alloy During Hot Rolling

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Abstract

In the present work, the sheets of Zr–1Sn–0.3Nb–0.3Fe–0.1Cr alloy were hot rolled with different reductions (10%, 30%, 50%, and 60%) at 1023 K and 1073 K. The microstructure evolutions including grain microstructure, texture, and dislocation were investigated, using electron backscattering diffraction and transmission electron microscope. The results showed that dislocation slip, twinning, and dynamic recrystallization (DRX) were the main deformation mechanisms. DRX was found to be promoted by larger reduction and higher rolling temperature. The predominant texture formed during hot rolling was basal $\langle 0001 \rangle // \text{ND}$, whose intensity reached peak value after 30% reduction hot rolling. While the intensity of DRX texture $\langle 10\text{--}10 \rangle // \text{ND}$ and $\langle 1\text{--}210 \rangle // \text{ND}$ increased with increasing reduction and temperature. This study provided an effective way to tailor the texture and microstructure of the alloy, for optimizing process parameters.

Keywords Zr–1Sn–0.3Nb–0.3Fe–0.1Cr alloy · Hot rolling · Texture evolution · Recrystallization

1 Introduction

Zirconium alloys are extensively used as cladding materials in nuclear reactors because of superior mechanical characteristics, low neutron absorption cross section, and excellent corrosion resistance [1–6]. Microstructure and texture of zirconium alloy have a great influence on its service performance, which includes mechanical properties, irradiation-induced resistance, corrosion resistance, hydride orientation distribution, and creep rate, etc. [7–12]. Therefore, tailoring microstructure including texture and grain size plays a vital role in the fabrication of zirconium alloy, which includes several heat-treatment

and deformation stages. Being a hexagonal close-packed structure crystal structure, the zirconium alloy gives a high intensity of texture following heat treatment and deformation [11]. It was reported that radial basal texture was competent to improve the hydride cracking resistance of Zr-alloy clad tube in heavy water reactors. While in the light water reactors, the radial basal texture improved the resistance to corrosion cracking and brought desirable anisotropic strength [10]. Some previous studies investigated the texture components of cold-rolled and hot-rolled Zr alloys. They found that the $\langle 10\text{--}10 \rangle$ parallel to rolling direction with $\langle 0001 \rangle$ tilted approximately 30° from the normal direction texture was thought to form during cold rolling, while the $\{0001\}$ and $\langle 1\text{--}210 \rangle$ textures parallel to the rolling direction were the dominant texture in the annealed sheets [13–15]. Also, another literature indicated that the basal-normal texture was formed in as-rolled ZIRLO™ sheets [16]. In recent years, some studies were carried out on the investigation of the deformation mechanism in pure Zr and Zr alloys. The dislocation slip was thought to be responsible for most cases in deformation of ZIRLO™ [16–18]. While the dynamic recrystallization of α phase was considered as an important microstructure evolution mechanism at 1123 K, it was supported by the research concerning the hot deformation behavior of Zr–1Nb–1Sn alloy [19, 20]. In some other cases, twinning, originated from external stresses in Zr alloys, was responsible for the deformation of zirconium

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alloys [21–23]. All of those mechanisms above gave a strong influence on the microstructure and texture evolution in Zr alloys.

The addition of Nb and Sn has been reported to enhance irradiation-induced creep resistance of Zr alloys [7, 9]. Zr–1Sn–0.3Nb–0.3Fe–0.1Cr alloy, also called as N18 in China, was developed based on Zr–Sn and Zr–Nb systems to meet higher fuel burn-up requirements in pressurized water reactor in China. The corrosion resistance and hydrogen absorption resistance of Zr–1Sn–0.3Nb–0.3Fe–0.1Cr alloy, as literature reported, were better than Zr–Sn alloys such as Zr–2 and Zr–4. In addition, in the Zr–Nb system, especially Zr–2.5Nb alloy, the content of Nb exceeds the solution limit of the service temperature, which undermines the corrosion resistance of the alloy. In contrast, the N18 was more competent to the harsh service environment [24–28]. Generally, Zr–1Sn–0.3Nb–0.3Fe–0.1Cr alloy sheets are used as cladding materials, whose service properties are very sensitive to its microstructure. It was recognized that the texture evolution and recrystallization behavior of the Zr alloys sheet depended on the process parameters such as reduction in hot rolling and temperature [29]. Absolutely, the addition of more alloying elements has obvious effect on recrystallization. Therefore, it is important to reveal the relationship between hot-rolling conditions and recrystallization and texture of Zr–1Sn–0.3Nb–0.3Fe–0.1Cr alloy for tailoring its microstructure and properties. However, there were few reports focusing on the investigation of texture evolution and deformation mechanism during its hot-rolling process.

In the present study, hot rolling was carried out on Zr–1Sn–0.3Nb–0.3Fe–0.1Cr alloy sheets as well as the texture and microstructure of as-rolled sheets were characterized by electron backscattering diffraction (EBSD) and transmission electron microscope (TEM). The influence of hot-rolling parameters on microstructure evolution of this alloy was investigated in detail aiming to reveal the relationship between them clearly.

2 Experimental

Zr–1Sn–0.3Nb–0.3Fe–0.1Cr alloy was obtained as a plate that has been hot rolling cogged and carried out a short time annealed. The initial specimens with a dimension of 80 mm × 50 mm × 5 mm were machined out in the way that the lengthwise direction is parallel to the rolling direction. The chemical composition of the alloy is given in Table 1. In order to investigate the effects of hot deformation temperature and reductions in thickness on the microstructure evolution, the

Table 1 Chemical compositions of Zr–1Sn–0.3Nb–0.3Fe–0.1Cr alloy (wt%)

Sn	Nb	Fe	Cr	Zr
1.00	0.30	0.30	0.10	Bal.

hot-rolling experiments with various reductions (10%, 30%, 50%, and 60%) of the specimens were investigated at 1023 K and 1073 K. The specimens were hot rolled using a rolling mill with diameter 420 mm and rolling velocity of 0.36 m/s. The hot-rolling temperature of the Zr–1Sn–0.3Nb–0.3Fe–0.1Cr alloy was chosen according to its $\alpha \rightarrow \beta$ transus temperature, which was around 1048–1058 K. Temperatures of 1023 K (higher temperature in the α -phase region) and 1073 K (above the $\alpha \rightarrow \beta$ transus) were used during the hot-rolling experiments. After hot rolling, the specimens were quenched in water.

As shown in Fig. 1, each sample was taken from the intermediate area of as-rolled sheet. The microstructure of samples was examined by a scanning electron microscopy (SEM, JELL-6300). Thin-foil samples for EBSD were prepared by wire electrical discharge machining (WEDM) and conventional grinding from as-rolled sheets. The samples were twin-jet electro-polished in a solution of 10 vol% perchloric acid and 90 vol% ethanol at about -30°C and 35 V.

Microstructures were characterized by field emission transmission electron microscopy (TecnaiG2 S-TWIN TMP F20). Samples for TEM testing were prepared by a similar method as the EBSD samples. Thin-foil samples were twin-jet electro-polished in a solution of 10 vol% perchloric acid and 90 vol% ethanol at about -30°C and 35 V.

3 Results and Discussion

3.1 Microstructure Evolution

The bright-field images of Zr–1Sn–0.3Nb–0.3Fe–0.1Cr alloy are shown in Fig. 2. In the initial state, only dislocations were observed although dislocation density was low, while twins and sub-grains could not be found. When the sheets were rolled with the reduction of 10%, it was obvious that the quantity of dislocations increased and twinning occurred. Twins could also be observed in the sheet rolled with the reduction of 30%. However, sub-grains could be found when

Fig. 1 Schematic sample of Zr–1Sn–0.3Nb–0.3Fe–0.1Cr rolling sheet

Fig. 2 TEM bright-field images of Zr–1Sn–0.3Nb–0.3Fe–0.1Cr alloy rolled at 1073 K with different reductions: **a** initial state, **b** 10%, **c** 30%, **d** 60%

the reduction increased to 60%, instead of twins. HCP metals and alloys are usually difficult to deform than the cubic one, because of fewer movable slip systems, especially at room temperature [23]. Hence, twinning occurred where dislocations were difficult to slip, to provide additional deformation. As the reduction increased, the density of dislocations boosted sharply, accompanied by dislocations pileup and tangle, which led to the generation of sub-grains. In the rolling process of Zr–1Sn–0.3Nb–0.3Fe–0.1Cr alloy, DRX played an important role in the evolution of microstructure under the effect of temperature and large reduction.

The volume fraction of DRX grains and the average grain size of as-rolled sheets after different reduction percentages is presented in Fig. 3. As widespread known, recrystallization is prone to be benefited by a larger reduction and higher deformation temperature. There is a rising tendency in volume fraction of DRX grains with the increasing reduction percentage in Zr–1Sn–0.3Nb–0.3Fe–0.1Cr alloy in Fig. 3. It

should be pointed out that the rising tendency was not very obvious until the hot-rolling reduction reaching the critical deformation (about 30%). With a low deformation level, stored energy was low correspondingly. It was revealed that the recrystallization mechanism was hard to actuate (fraction of DRX below 5%) due to insufficient kinetics and thermodynamic conditions. Afterward, the increasing tendency became faster when the reduction percentage increased to 50%. Based on this phenomenon, it can be speculated that an evident incubation period may exist in the dynamic recrystallization and nucleation of recrystallization requires sufficient stored energy in the deformed matrix. The temperature has a dramatic influence on DRX behavior. With the increase in rolling temperature, the fraction of DRX grains increased sharply. Fraction of DRX reached about 30% in the sheets after 60% reduction at 1073 K, which was much higher than that of the sheets rolled at 1023 K (approximate 15%), but the finest grain size at both temperatures reached 4.3 μm .

Fig. 3 Volume fraction of DRX **a**, average grain size **b** of Zr-1Sn-0.3Nb-0.3Fe-0.1Cr alloy after hot rolling with different reductions

As to this trend, more deformed grains were broken up to finer lamellas during severe deformation. Moreover, DRX process was more active at a higher temperature. With the reduction percentage over 30%, further increases were found in the gaps of DRX fraction of the samples rolled at different temperatures. It revealed that DRX mechanism tended to be more active at higher rolling temperature in this alloy, resulting in an increase in the volume fraction of DRX.

The DRX maps with grain boundaries of the sheet after 50% and 60% reductions at 1073 K are shown in Fig. 4. The blue grains are DRX grains, the yellow grains represent recovery ones, and deformed grains are red color. The observed plane was the rolling plane. The internal average misorientation angle within the grains is always used to clarify grains; if the average angle in a grain exceeds 7.5° , the grain is classified as a deformed one. The misorientation of substructured grains is between 1° and 7.5° in the present work. All the remaining grains are classified as recrystallized. The grain boundary type maps are presented in Fig. 5,

whose observed plane is the rolling plane. Same as the previous, the high-angle grain boundary (HAGBs) and low-angle grain boundary (LAGBs) are, respectively, highlighted by blue and red lines. It could be seen that the amount of LAGBs and HAGBs boosted as the increasing reduction simultaneously in the microstructure of as-rolled sheets. During the rolling deformation, LAGBs usually emerge in the area around the grain boundaries, suggesting relatively high dislocations density and stored energy of this area. Also, a large amount of LAGBs generated in the deformed grains with the increase in reduction, contrasting by their absence in the DRX grains. It indicated that the increase in the reduction to 60% led to a transformation from LAGBs to HAGBs and the formation of DRX grains was expected to be influenced by this transformation. According to Figs. 4 and 5, the recrystallization grains emerge at grain boundaries triple junctions firstly due to the high stored energy. And neither abnormal DRX grains nor obvious HAGBs' migration is observed in the figure. The grain size of DRX

Fig. 4 DRX maps of Zr-1Sn-0.3Nb-0.3Fe-0.1Cr alloy after hot rolled at 1073 K with different reductions: **a** 50%, **b** 60%. The blue grains are DRX grains, the yellow grains represent recovery ones, and deformed grains are in red

Fig. 5 Grain boundary maps of Zr–1Sn–0.3Nb–0.3Fe–0.1Cr alloy after hot rolling at 1073 K with different reductions: **a** 10%, **b** 60%. The HAGBs and LAGBs are, respectively, highlighted by blue and red lines

Fig. 6 Fraction of LABG of Zr–1Sn–0.3Nb–0.3Fe–0.1Cr alloy after hot rolling with different reductions

grains is finer about 0.3–3.2 μm than deformed ones about 3.2–15 μm , occasionally approaches 20 μm .

Figure 6 shows the fraction of LAGB. As can be observed, the fraction of LAGBs decreased with the increase in reduction percentage. It was worth to note that the fraction of LAGBs dropped sharply with the reduction percentage increasing from 30% to 50%, which properly could be

explained by the rapidly increasing amount of recrystallization structure.

The mechanism of formation of DRX grains is schematically given in Fig. 7 (the bold blue lines and the red lines, respectively, represent HAGBs and LAGBs). The LAGBs preferentially emerged around grain boundaries which can be seen. Dislocation density increased during the rolling process, resulting in the increase in stored energy, and then dislocations slipped and piled up at grain boundary. The LAGBs in the deformed matrix increased sharply. DRX nuclei formed due to high stored energy so that they usually appeared around grain boundaries, especially triple junctions. Then, the new DRX grains formed and grew on these nuclei. In order to consume the stored energy, the LAGBs around the new DRX grains might tend to merge and transform to HAGBs, causing the growth of new grains. Consequently, DRX grains showed a slight expansion in the sheets after 60% reduction. Generally, continuous DRX and discontinuous DRX were two main DRX mechanisms, the former was recovery process and the latter was classic nucleation and growth process. The character of continuous DRX was that dislocations were absorbed by subgrain boundaries continually, resulting in the new grains formed by the formation of new HAGBs. The discontinuous DRX

Fig. 7 Schematic mechanism of the formation of DRX grains

had two distinct stages of nucleation and growth driven by the stored strain energy [30, 31]. Based on the facts that new DRX grains appeared around grain boundaries, there was an obvious growth process. Therefore, the DRX process in Zr–1Sn–0.3Nb–0.3Fe–0.1Cr alloy sheet during hot rolling is considered to be a discontinuous DRX. Meanwhile, new LAGBs might emerge around the grown DRX grains resulting in the repetition of this recrystallization process. According to the literature, similar phenomenon also appears in other HCP metals and their alloys. It was reported that the nucleation of pure magnesium strongly depends on the amount of strain accumulation on each grain [32]. As well as substructures merged and a large fraction of LAGB evolved to HAGBs with the increase in strain was also observed in pure titanium and Ti–5Al–5Mo–5 V–3Cr–1Zr alloy [33, 34].

3.2 Sliding and Twinning Deformation

Sliding and twinning are two important mechanisms in metal deformation. Figures 8 and 9 show the Schmid factor distribution of slip systems of as-rolled samples. The prismatic systems, first-order pyramidal, and second-order pyramidal ($\{11\text{--}22\} \langle 11\text{--}23 \rangle$) slip systems were found to be active in the present investigation. The $\langle a \rangle$ prismatic

system served as the principal slip system of the deformed Zr alloy. However, the second-order pyramidal was dominant during hot-rolling process as the c -axis compression deformation came along. Any c -direction strain could be accommodated by the slip on the first-order pyramidal and second-order pyramidal slip systems [23, 34]. As the rolling temperature increased, more soft orientation grains with the second-order pyramidal slip systems and prismatic systems appeared (especially the latter one). This phenomenon might be related to lower critical shearing stress at a higher temperature and more DRX behavior caused by the start of these slip systems. It should be pointed out that the basal slip system was almost inoperative in the present investigation though it was a major slip system in the majority of other HCP metals and alloys. However, as to pure Zr and Zr alloys, they were characterized by hard slip along $\langle 0001 \rangle$ and easy slip along $\langle hkl0 \rangle$ [16]. There are two principal ways to change the microstructure, changing the texture by grain rotation and changing the dislocation density. Therefore, it is worth noting that most of the coarse deformed grains show soft orientation with prismatic systems, first-order pyramidal, and second-order pyramidal slip systems, which might contribute to the formation of basal texture [34].

Fig. 8 Schmid factor distribution of different slip systems of Zr–1Sn–0.3Nb–0.3Fe–0.1Cr alloy sheets after hot rolling at 1023 K with different reductions: **a** basal slip, **b** prismatic slip, **c** first-order pyramidal slip, **d** second-order pyramidal slip

Fig. 9 Schmid factor distribution of different slip systems of Zr–1Sn–0.3Nb–0.3Fe–0.1Cr alloy sheets after hot rolling at 1073 K with different reductions: **a** basal slip, **b** prismatic slip, **c** first-order pyramidal slip, **d** second-order pyramidal slip

Twinning can provide additional deformation in Zr alloys. The common twins in Zr alloys are {11-22}, {10-11}, {10-12}, and {11-21}. The first two types of twin can produce a compress strain along the c -axis and the latter two can bring a tensile strain along the c -axis [23]. In the present study, the {11-22} <-1-123> compressive twins are the major twinning during hot-rolling process and a few {10-11} <-10-1-2> compressive twins grains also occur, while there is almost no sign of other two types of tensile twins in the microstructure (only a few tensile twins are observed in the sheets after server deformation). The fraction of twin grains is shown in Fig. 10. It can be seen that the fraction of twin reaches the peak after 30% deformation and then drops gradually with the increase in rolling reduction. The slight deformation generated few twins. With the increase in reduction, more twins were required to provide additional deformation. Different from the slip mechanism, the twinning mechanism is not significant during severe deformation. Therefore, the fraction of twins shows a decreasing trend with the reduction increased from 30% to 60%. When the reduction percentage increased to 60%, the twinning almost disappeared completely. As expected, less fraction of twinning appeared in the samples rolled at a higher temperature when slip was easier to be active.

Fig. 10 Volume fraction of twin of Zr–1Sn–0.3Nb–0.3Fe–0.1Cr alloy sheets after hot rolling with different reductions

3.3 Texture Evolution

The change in volume fraction of basal texture <0001>//ND, <10-10>//ND, and <1-210>//ND texture is presented in Fig. 11. The components of the texture of as-rolled sheets are sensitive to rolling reduction and temperature. The basal

texture was found to be the main texture, which was consistent with the results in the literature [15]. The proportions of $\langle 10\text{-}10 \rangle // \text{ND}$ and $\langle 1\text{-}210 \rangle // \text{ND}$ texture were much lower than that of basal texture after hot rolling. Furthermore, there were more than 50% grains still exhibiting nearly random orientation.

It can be seen that the volume fraction of basal texture increased to peak at 30% reduction (34% at 1023 K, 42% at 1073 K) and then decreased rapidly till 50% reduction, followed by a slight lift stretching to 60% reduction. The formation of basal texture in Zr alloy was caused by grain rotation during hot-rolling deformation, which was supposed to be contributed by slip and twinning. There was almost no sign of DRX at 10% reduction as slip and twinning were preferred mechanisms. This led to small rotation of deformed grains, which also meant slight changes in the initial orientation. Therefore, texture was mostly inherited from the initial state and the texture intensity was relatively weak. With the increase in the reduction from 10 to 30%, the basal texture was strengthened by the rotation and deformation of the grains. However, the basal texture is usually formed in the deformed matrix and some broken grains while DRX grains show totally different orientations. As to the sheets after higher reduction, dynamic recrystallization plays a crucial role in the deformation, which indirectly weakens basal texture. So, the fraction of basal texture decreased in accordance with the decrease in deformed structure and increase in DRX structure. The volume fraction of basal texture decreased in the sheets after 50% and 60% reduction. This indicated that the $\langle 0001 \rangle // \text{ND}$ orientation of the deformed grains was weakened by deformation. This was supported by studies on other HCP alloys, which presented similar trends (the basal texture weakened by lager deformation or DRX process). Texture weakening of AZ31 magnesium alloy occurred and increased with the increase in annealing temperature due to the increase in recrystallization grains [35]. The formation of ultrafine grains during deformation could decrease the basal texture intensity and

thereby inducing a significant texture weakening effect [36]. In the present study, the texture weakening was also considered to be caused by the increase in DRX. In general, the basal texture was strengthened by the slip while weakened by DRX process. So that the change in texture content was the result of the combined interaction of deformation and DRX, which resulted in the increase in basal texture fraction with the reduction increased from 50 to 60%.

Separating the coarse grains (grain size larger than $3.2 \mu\text{m}$) and fine grains (grain size smaller than $3.2 \mu\text{m}$), the IPF maps with crystallographic orientation of the sheets after 60% reduction at 1073 K are shown in Fig. 12. The majority of the coarse grains and some of the broken fine grains are red, while the fine DRX grains present green and blue. It can be seen that most of the deformed grains tend to be oriented as $\langle 0001 \rangle // \text{ND}$ (including the broken lamellas) while the orientation of refined DRX grains is prone to be $\langle 10\text{-}10 \rangle // \text{ND}$ and $\langle 1\text{-}210 \rangle // \text{ND}$. It indicated that the recrystallization grains owned mainly $\langle 10\text{-}10 \rangle // \text{ND}$ and $\langle 1\text{-}210 \rangle // \text{ND}$ orientations. Furthermore, the texture components showed that the fraction of $\langle 01\text{-}10 \rangle // \text{ND}$ texture was slightly higher than that of $\langle 1\text{-}210 \rangle // \text{ND}$ texture. Meanwhile, with the increase in reduction from 10% to 60%, the fraction of these two types of textures decreased firstly and then increased, reaching the bottom after 30% reduction. In the sheets after 10% reduction, the fraction of $\langle 01\text{-}10 \rangle // \text{ND}$ and $\langle 1\text{-}210 \rangle // \text{ND}$ textures was relatively high due to the remaining of this type of grains after a slight rolling deformation. However, the deformation tended to mitigate these two types of textures and promoted the transformation of basal texture. Thus, with the increase in reduction from 30% to 60%, the fraction of $\langle 01\text{-}10 \rangle // \text{ND}$ and $\langle 1\text{-}210 \rangle // \text{ND}$ textures was very low at 30% reduction and then increased. The component of texture revealed that the texture intensity was higher in the sheets rolled at higher temperature. On one hand, the rotation of grains was more severe and the texture intensity tended to be higher for the reason that slip in Zr–1Sn–0.3Nb–0.3Fe–0.1Cr alloy was easier at higher temperature. On the other hand, the increase in recrystallization grains strengthened the textures of $\langle 10\text{-}10 \rangle // \text{ND}$ and $\langle 1\text{-}210 \rangle // \text{ND}$. As mentioned above, the second-order pyramidal slip systems were found to contribute more to the deformation at higher rolling temperature; this may attribute to the increase in texture intensity.

Therefore, the crystal texture of as-rolled Zr–1Sn–0.3Nb–0.3Fe–0.1Cr sheets can be separated to three parts, basal texture (deformed grains), DRX texture, and other parts (almost random orientation). The deformed matrix consisted of coarse deformed grains, broken lamellas, and a few twin grains. The soft orientation of major slip systems dominated in deformed grains and was enhanced with the increase in reduction. These grains contributed to the basal texture of hot-rolled sheets. While twinning was

Fig. 11 Changes in volume fraction of $\langle 0001 \rangle // \text{ND}$, $\langle 10\text{-}10 \rangle // \text{ND}$, and $\langle 1\text{-}210 \rangle // \text{ND}$ textures with different reductions

Fig. 12 IPF maps animation of crystal orientation of the sheet after 60% reduction at 1073 K

not very dominant than other deformed grains and mostly presented $\{11\text{-}22\} \langle -1\text{-}123 \rangle$ orientation. DRX grains usually presented $\langle 10\text{-}10 \rangle // \text{ND}$ and $\langle 1\text{-}210 \rangle // \text{ND}$ orientation, which was regarded as the DRX texture.

4 Conclusions

The microstructure evolution of Zr–1Sn–0.3Nb–0.3Fe–0.1Cr alloy during hot rolling was examined through EBSD and TEM analysis. The effect of hot-rolling conditions on the texture intensity and dynamic recrystallization behavior of Zr–1Sn–0.3Nb–0.3Fe–0.1Cr alloy was investigated. The conclusions are as follows:

1. The DRX process and twinning appeared during the hot rolling of Zr–1Sn–0.3Nb–0.3Fe–0.1Cr alloy sheets. The majority of DRX grains presented $\langle 10\text{-}10 \rangle // \text{ND}$ and $\langle 1\text{-}210 \rangle // \text{ND}$ orientation. The fraction of DRX reached to a peak value, about 30%, at 1073 K and 60% reduction. The $\{11\text{-}22\} \langle -1\text{-}123 \rangle$ type compression twin was found during hot-rolling process. The maximum fraction of twins was about 4.3% with 1023 K and 30% reduction, which was a lower temperature and a medium reduction.
2. With 30% reduction, the basal texture intensity of alloy was 34% at 1023 K, 42% at 1073 K, which was the maximum intensity. An increasing trend was found in the fraction of DRX texture ($\langle 10\text{-}10 \rangle // \text{ND}$ and $\langle 1\text{-}210 \rangle // \text{ND}$ texture) with the increase in reduction and temperature.

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