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SLIP AND TWINNING IN THE $[\bar{1}49]$-ORIENTED SINGLE CRYSTALS OF A HIGH-ENTROPY ALLOY

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Using $[\bar{1}49]$*- oriented single crystals of an FCC* $Fe_{20}Ni_{20}Mn_{20}Cr_{20}Co_{20}$ *(at.%) high-entropy alloy subjected to tensile deformation, the temperature dependence of critical resolved shear stresses* $\tau_{cr}(T)$ *and the deformation mechanism of slip and twinning are investigated in the early stages of deformation at* $\varepsilon \le 5\%$ *within the temperature interval* $T = 77–573$ *K. It is shown that* τ_{cr} *increases with decreasing the testing temperature and the* $\tau_{cr}(T)$ *temperature dependence is controlled by the slip of perfect dislocations a/2<110>. The early deformation stages* $\varepsilon \le 5\%$ *are associated with the development of planar slip by pileups of perfect dislocations a/2<110>, stacking faults and mechanical twins, which is observed in the temperature interval from 77 to 423 K. A comparison of the temperature dependence* $\tau_{cr}(T)$ *and the development of mechanical twinning is performed between the* $[\bar{1}49]$*-oriented single crystals of the* $Fe_{20}Ni_{20}Mn_{20}Cr_{20}Co_{20}$ *high-entropy alloy, the single crystals of the austenitic stainless steel,* Fe – 18% Cr – 12% Ni – 2Mo *(wt.%) without nitrogen atoms (Steel 316) and Hadfield steel,* Fe – 13% Mn – (1–1.3)% C *(wt.%).*

Keywords: single crystals of a FeNiMnCrCo high-entropy alloy, plastic deformation, slip, twinning, dislocations.

INTRODUCTION

Recently, the interest of researchers has been focused on designing high-entropy alloys (HEAs) [1–11]. The HEA concept relies on reaching maximum values of the entropy of mixing five or more elements in equal atomic proportions with an aim of achieving high levels of strain hardening in single-phase materials. This results in heavy lattice distortion and sluggish diffusion, which in turn provides an increase in strength properties, both at the yield stress and in the course of plastic deformation, and their stability in the single-phase state of a material over a wide temperature range $T = 77–973$ K [1, 3]. Using an FCC polycrystalline FeNiMnCrCo HEA, it is shown that at high-level stresses at the yield stress these alloys are characterized by good plasticity and ductile fracture at low testing temperatures [3]. The formation of stacking faults (SF) and the development of mechanical twinning accompanied by slip in the early stages of plastic deformation in HEA polycrystals give rise to a strong strain hardening effect [1, 3, 4]. As far as HEA single crystals are concerned, the development of twinning deformation in them has not been investigated so far. We are familiar with two studies on single-crystal $CoCrFeNiAl_{0.3}$ HEAs [9, 10], which report investigations of the [001]-orientation in tension and the $[\bar{1}49]$-orientation in compression, and one study of a single-

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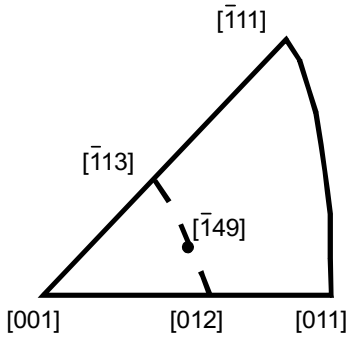

Fig. 1. Position of the [$\bar{1}$49]-orientation in the stereographic triangle.

crystal FeNiMnCrCo HEA, dealing with the [$\bar{5}$91]-orientation in compression [8]. These works report data on determination of the lattice parameter and shear modulus and on investigation of the structure type developing in HEAs in the course of their deformation [8–10]. The purpose of this work is to study the deformation mechanisms – slip and twinning – in the early stages of plastic deformation at $\varepsilon \leq 5\%$, the critical resolved shear stresses τ_{cr} to slip and twinning within a wide temperature range $T = 77$–423 K in the FCC $Fe_{20}Ni_{20}Mn_{20}Cr_{20}Co_{20}$ (at.%) HEA single crystals oriented along the [$\bar{1}$49]-direction under tensile deformation (FCC – face-centered cubic lattice) (Fig. 1). The [$\bar{1}$49]-orientation was selected due to the following considerations: [$\bar{1}$49]-crystals are oriented to allow the deformation to develop via slip or twinning in one system. The Schmidt factors for slip and twinning in the [$\bar{1}$49]-orientation are $m_{sl} = 0.49$ and $m_{tw} = 0.4$, respectively, with $m_{sl} > m_{tw}$ [12]. Hence, [$\bar{1}$49]-oriented single crystals are more favorable for the deformation to develop via slip than via twinning; they are good candidates for identifying the role of slip in determining the temperature dependence of τ_{cr} in the $Fe_{20}Ni_{20}Mn_{20}Cr_{20}Co_{20}$ HEA crystals.

The FCC $Fe_{20}Ni_{20}Mn_{20}Cr_{20}Co_{20}$ (at.%) HEA single crystals were grown by the Bridgeman method in an argon atmosphere from the workpiece blanks melted in a resistance furnace. In order to achieve a homogeneous distribution of elements in the bulk of the workpieces, they were re-melted three times. The crystals were homogenized in a helium atmosphere at 1470 K for 48 hrs. Then dog-bone specimens were cut from them in an electric-spark cutting machine. The damaged surface layer was ground off mechanically and then polished in 200 ml of an H_3PO_4+50 g CrO_3 electrolyte. The specimens were quenched into water after heat treatment in a helium atmosphere at $T = 1473$ K for 1 hour. The orientation was determined in a DRON–3M diffractometer. The electron microscopy examinations were performed in a JEOL 2010 electron microscope at an accelerating voltage of 200 kV. The surfaces of deformed specimens were examined in a Keyence VHX–2000 optical microscope. The mechanical tests within the temperature range from 77 to 573 K were carried out in an Instron 5969 universal testing machine at the strain rate $4 \cdot 10^{-4}$ s^{-1}. The chemical composition of single crystals after their quenching was determined by the X-ray fluorescence method: Fe = 20.60%, Ni = 19.47%, Co = 21.51%, Cr = 18.97%, Mn = 19.45% (at.%); in the atomic concentration of the elements the chemical composition of the $Fe_{20}Ni_{20}Mn_{20}Cr_{20}Co_{20}$ HEA coincides with that of the FeNiMnCrCo HEA determined in [8]. The critical resolved shear stresses were calculated using an expression for slip $\tau_{cr}^{sl} = \sigma_{0.1} * m_{sl}$ ($\sigma_{0.1}$ – axial stresses at the yield point, $m_{sl} = 0.49$ – the Schmidt factor of slip for the [$\bar{1}$49]-oriented single crystals [12]) and for twinning $\tau_{cr}^{tw} = \sigma_{\varepsilon} * m_{tw}$ (σ_{ε} – stresses for the strain value 5%, after which TEM examination reveals twinning, $m_{tw} = 0.4$ – the Schmidt factor for twinning in the [$\bar{1}$49]-oriented single crystals [12]).

EXPERIMENTAL RESULTS AND DISCUSSION

Using the methods of X-ray diffractometry and electron microscopy of thin foils, it has been found that the as-grown $Fe_{20}Ni_{20}Mn_{20}Cr_{20}Co_{20}$ (at.%) HEA single crystals after quenching at 1473 K for 1 hour represent a substitution

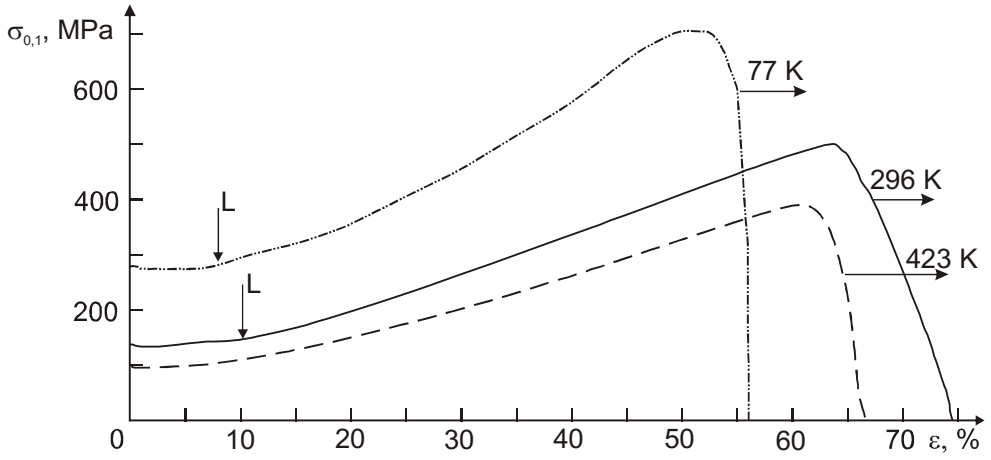

Fig. 2. Flow curves from the $[\bar{1}49]$-oriented single crystals of a $Fe_{20}Ni_{20}Mn_{20}Cr_{20}Co_{20}$ (at.%) HEA under tensile deformation.

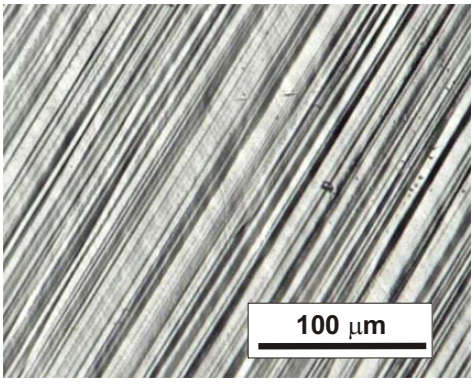

Fig. 3. Optical microscopy image of the surface of the $[\bar{1}49]$-oriented $Fe_{20}Ni_{20}Mn_{20}Cr_{20}Co_{20}$ (at.%) HEA single crystals deformed in tension to 30% at $T = 77$ K.

solid solution based on an FCC-lattice with the lattice parameter $a_0 = 0.3595$ nm and do not contain dispersed oxide and carbide particles. Figure 2 presents the flow curves for the $[\bar{1}49]$-oriented single crystals of FeNiMnCrCo HEA within the temperature interval from 77 to 423 K. At all testing temperatures, the deformation ε in the flow curves evolves within two linear stages with different values of the strain hardening coefficient $\Theta = d\sigma/d\varepsilon$. In the first stage at $\varepsilon \leq 10\%$ the deformation develops as a Lüders band with the strain hardening coefficient $\Theta_I = d\sigma/d\varepsilon$ equal to zero. At all of the testing temperatures, on the specimen surface there is a one slip system to the end of the Lüders band. At $\varepsilon > 10\%$, there is transition to the second linear stage with the predominating single slip system (Fig. 3), while the electron microscopy examination reveals the presence of twinning, whose interaction with slip results in an increased values of Θ_{II}. With decreasing testing temperature, plasticity ε_{pl} of the $[\bar{1}49]$-oriented single crystals decreases. In the FeNiMnCrCo HEA polycrystals with the grain sizes $d = 4.4$ and 155 μm ε_{pl}, on the contrary, increases (Table 1). This difference in the temperature dependence of plasticity in FeNiMnCrCo HEA single- and polycrystals is associated with the formation of localized macroscopic strain bands in the $[\bar{1}49]$-crystals oriented for single shear in the late stages of deformation and their absence in polycrystals.

In order to reveal the mechanisms of deformation in early stages of plastic flow, in this work we performed electron microscopy investigations of the specimens deformed to 5% at the testing temperatures 77, 300 and 423 K.

TABLE 1. Comparison of Mechanical Properties of Single- and Polycrystals of $Fe_{20}Ni_{20}Mn_{20}Cr_{20}Co_{20}$ HEAs and Single Crystals of Austenitic Steels

Alloy	Orientation	T, K	T_{cr}^{sl}, MPa	$(\tau_{cr}^{sl}/G)\cdot10^{-3}$	ε_{pl},%	τ_{cr}^{tw}, MPa	$(\tau_{cr}^{tw}/G)\cdot10^{-3}$
ВЭС	$[\bar{1}49]$	77	137	1.6	56	112	1.3
		296	69	0.82	75	56	0.7
		423	49	0.59	67	40	0.5
HEA polycrystal, d = 4.4 μm [3]		77	193	2.3	90	280	3.3
		293	114	1.36	60	–	
		473	95	1.13	45	–	
		673	92	1.09	38	–	
HEA polycrystal, d = 155 μm [3]		77	101	1.2	110	190	2.2
		293	49	0.58	80	–	
		473	35	0.41	63	–	
		673	32	0.38	68	–	
Steel 316 single crystal [14]	$[\bar{1}23]$	77	175	2.2	55	–	
		300	80	1	85	–	
Hadfield steel single crystal [15, 16]	$[\bar{1}23]$	77	350	4.3	50	350	4.3
		300	148	1.85	45	230	2.9

Note. G = 84 GPa – shear modulus of the FeNiMnCoCr HEA [8], G = 80 GPa – shear modulus of Steel 316 and Hadfield steel [17].

Since within the range of strains from 0.1 to 10% the deformation in the flow curves of the $[\bar{1}49]$-oriented single crystals develops as a Lüders band, so at ε = 5% only 50% of the gage section undergo deformation. For this reason, in order to study the dislocation structure of the $[\bar{1}49]$-oriented single crystals foils were cut from the deformed part of the specimen.

In order to determine the stacking fault energy γ_0 in the FeNiMnCrCo HEA from the triple dislocation nodes [13], in this work we specially investigated the dislocation structure of the $[\bar{1}11]$-oriented single crystals deformed to 5% at T = 300 K, in which deformation was developed in a few systems and deformation triple dislocation nodes were observed. The value of γ_0 experimentally obtained from the triple dislocation nodes in single crystals of the FeNiMnCrCo HEA was found to be equal to 0.019 – 0.021 J/m^2; this value is consistent with the data for γ_0 in the $Fe_{20}Ni_{20}Mn_{20}Cr_{20}Co_{20}$ HEA obtained earlier in [7, 8].

The investigations of dislocation structure of the $[\bar{1}49]$-oriented single crystals of the FeNiMnCrCo HEA deformed to 5% demonstrate that in the temperature interval T = 77–423 K in the early deformation stages of these crystals, dislocation pileups, stacking faults and imperfect twins are observed in them (Fig. 4). The formation of dislocation pileups is associated with the difficulty of cross slip of dislocations due to low stacking fault energies γ_0 and high resistance of the substitution atoms to dislocation motion.

Twinning in the $[\bar{1}49]$-oriented single crystals of the FeNiMnCrCo HEA deformed to 5% is observed in one system within the temperature interval from 77 to 423 K and is developed simultaneously with slip (Figs. 4*b* and 5). At the testing temperatures 77 and 300 K, the twins are found to be very thin, t = 10 nm (Fig. 5*a* and *b*), and have a defective structure (Fig. 4*b*). As the testing temperature is increased, the thickness of twins increases and becomes t = 20–25 nm (Fig. 5*c*) at T = 423 K,

Thus, the investigations of dislocation structure of the $[\bar{1}49]$-oriented single crystals of the FeNiMnCrCo HEA deformed to 5% strain in the temperature interval from 77 to 423 K demonstrated that the onset of plastic deformation in them at σ > 0.1% occurs due to slip in one system and then due to formation of stacking faults and twins.

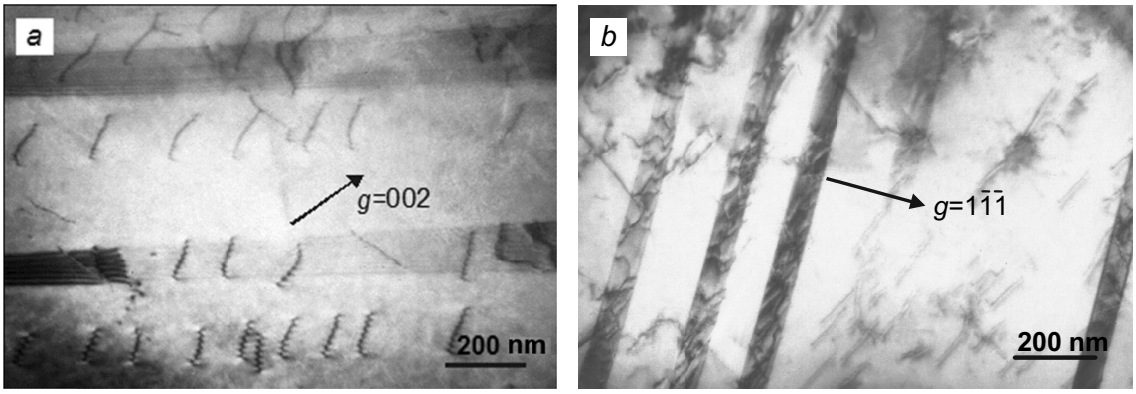

Fig. 4. TEM images of dislocation pileups, stacking faults (*a*) and imperfect twins (*b*) in the $[\bar{1}49]$-oriented single crystals of the $Fe_{20}Ni_{20}Mn_{20}Cr_{20}Co_{20}$ (at.%) high-entropy alloy deformed in tension to 5% at $T = 77$ K.

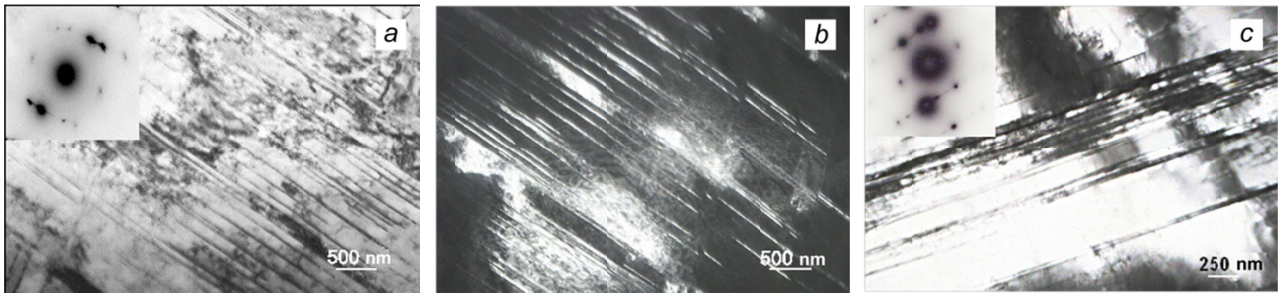

Fig. 5. Twinning in the $[\bar{1}49]$-oriented single crystals of the $Fe_{20}Ni_{20}Mn_{20}Cr_{20}Co_{20}$ (at.%) high-entropy alloy deformed in tension to 5%: bright-field image, $T = 300$ K (*a*), dark-field image in the twin reflection, $T = 300$ K (*b*), bright-field image, $T = 150$ K (*c*); diffraction patterns, (110)-zone axis (inserts to (*a*) and (*c*)).

Figure 6 shows the dependence of critical resolved shear stresses $\tau_{cr}^{sl}(T)$ on testing temperature. It is evident that it is typical for slip deformation of FCC substitutional solid solution and consists of two temperature intervals [18–20]. At $T < 373$ K, there is strong temperature dependence $\tau_{cr}^{sl}(T)$, which is even stronger than that of $G(T)$, while at $T > 373$ K, the $\tau_{cr}^{sl}(T)$ plot is consistent with the $G(T)$ dependence [18]. In accordance with [18], $\tau_{cr}^{sl}(T)$ could be given by

$$T_{cr}^{sl}(T) = \tau_{cr}^{S}(T) + \tau_{cr}^{G}(T)\,, \qquad (1)$$

where $\tau_{cr}^{S}(T)$ and $\tau_{cr}^{G}(T)$ are the thermally activated and athermic components of τ_{cr}, respectively. For comparison, Fig. 6 presents the values of τ_{cr} for the $[\bar{1}23]$-oriented single crystals of the austenitic steel 316 without nitrogen atoms and Hadfield steel with low values of $\gamma_0 = 0.02$–0.025 J/m^2 [14–16] and the values of τ_{cr} obtained for the FeNiMnCrCo HEA polycrystals with the grain size $d = 155$ nm [3]. The estimates of τ_{cr}^{sl} for polycrystals of the FeNiMnCrCo HEA were obtained using $\sigma_{0.1} = \tau_{cr}M$ (M is the Taylor factor equal to 3.06) [3]. It is evident from Fig. 6 that the values of τ_{cr} in the $[\bar{1}23]$-oriented single crystals of Steel 316 without interstitial atoms and Hadfield steel are found to be 38 and 213 MPa higher at 77 K and 11 and 79 MPa higher at 300 K, respectively, than the values of τ_{cr}^{sl} at the above-mentioned temperature in the $[\bar{1}49]$-oriented crystals of the FeNiMnCrCo HEA. Note that the ratio

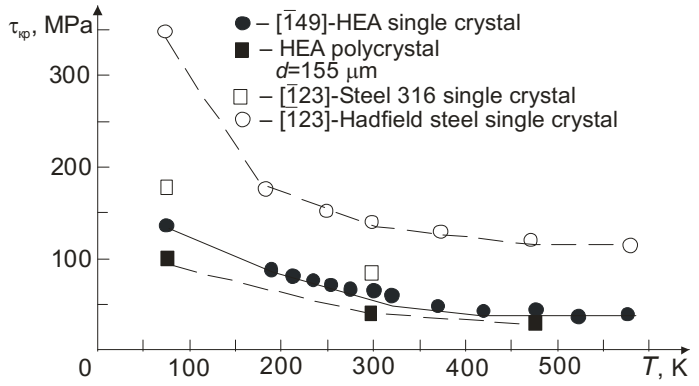

Fig. 6. Temperature dependence of critical resolved shear stresses τ_{cr}^{sl} for the $[\bar{1}49]$-oriented single- and polycrystals of the FeNiMnCrCo alloy and for the $[\bar{1}23]$-oriented single crystals of Steel 316 and Hadfield steel deformed in tension.

$\tau_{cr}^{S}(77\ K)/\tau_{cr}^{G}(300\ K)$ for the single crystals of Steel 316, Hadfield steel, and FeNiMnCrCo HEA are found to be close to each other and vary within 2.2–2.4 (Fig. 6). For FCC-solid solutions of Cu–Al, Cu–Mn and austenitic stainless steels, it was shown that high values of $\tau_{cr}^{S}(T)$ were associated with a thermally activated surmounting of the short-acting barriers on the order of a Bürgers vector by a slip dislocation $a/2\langle 110\rangle$ [3, 21, 22]. Since in the high-entropy crystals the processes of thermally activated dislocation motion $\tau_{cr}^{S}(T)$ and athermic slip $\tau_{cr}^{G}(T)$ turn out to be similar to those observed earlier in the austenitic steels and in Hadfield steel, where the onset of plastic deformation at $\varepsilon \le 0.2$–1% is associated with dislocation slip [19, 23] rather than mechanical twinning, we might assume that in FeNiMnCrCo HEAs the onset of plastic deformation would be due to the motion of perfect dislocations $a/2\langle 110\rangle$, which is supported by the studies of dislocation structure of the $[\bar{1}49]$-oriented single crystals deformed to 5%.

The above comparison of the values of τ_{cr}^{sl} of the single- and polycrystals of FeNiMnCrCo HEAs demonstrates that the values of $\tau_{cr}^{sl}(T)$, obtained using the $[\bar{1}49]$-oriented single crystals, appear to be close to those of coarse – grained polycrystals (Fig. 6, Table 1). Hardening of polycrystals is described by a well known Hall–Petch relation [24]

$$\sigma_{0.1} = \sigma_0 + kd^{-1/2}, \qquad (2)$$

where σ_0 is the resistance of substitution atoms to dislocation motion in a polycrystal with infinitely large grains, d is the grain diameter, and k is the constant describing the resistance of grain boundaries to deformation. An analysis of relation (2) suggests that if the values of $\tau_{cr}^{sl}(T)$ for single-crystals and coarse-grained polycrystals are equal, then the grain-boundary contribution $kd^{-1/2}$ into hardening in the case of coarse-grained polycrystals of FeNiMnCrCo HEAs appears to be negligible, and hardening in these polycrystals would be controlled by the motion of dislocations in the bulk of the grain.

A combination of heavy lattice distortion and low value of γ_0 in the $[\bar{1}49]$-oriented single crystals of FeNiMnCrCo HEAs favors the development of twinning in the early stages of deformation in a wide temperature range.

Earlier in single crystals of austenitic stainless steels without interstitial atoms with low value γ_0, oriented to a single shear, no twinning deformation was observed since the beginning of plastic flow in the temperature range T = 120–300 K [14, 19]. In single crystals of austenitic stainless steel without nitrogen atoms with low γ_0 at T = 300 K twinning deformation was observed in the $[\bar{1}11]$-oriented single crystals only, following the 18–20% preceding slip deformation [25]. From the onset of plastic flow at $\varepsilon \ge 1.5\%$, twinning in the crystals, oriented for a single shear, at T = 300 K was observed in austenitic stainless steel with the concentration of nitrogen up to 0.5 wt.% and in Hadfield steel [14, 23, 25]. Table 1 presents the values of critical resolved shear stress for the case of twinning τ_{cr}^{tw} in the $[\bar{1}49]$-oriented crystals of FeNiMnCrCo HEAs. It is evident that the maximum values of τ_{cr}^{tw} are observed at T = 77 K, they

decrease with the increasing temperature so that $\tau_{cr}^{tw}(T = 77\ K)/\tau_{cr}^{tw}(T = 296\ K) = 2$. Such a strong dependence of τ_{cr}^{tw} in the $[\bar{1}49]$-oriented single crystals of FeNiMnCrCo HEAs is controlled by the temperature dependence of τ_{cr}^{sl}, since at $\varepsilon = 0.1–5\%$ the onset of plastic deformation is associated with slip in single- and polycrystals of austenitic stainless steels and in Hadfield steel [19, 23]. It should be noted that in the $[\bar{1}49]$-oriented single crystals of FeNiMnCrCo HEAs the values of τ_{cr}^{tw} are low; they are 3–4 times lower than those in the $[\bar{1}23]$-oriented crystals of Hadfield steel (see Table 1). In polycrystals of FeNiMnCrCo HEAs, τ_{cr}^{tw} for twinning are higher than those in their $[\bar{1}49]$-oriented single crystals, but lower than those in single crystals of Hadfield steel (Table 1). In the $[\bar{1}49]$-oriented single crystals of FeNiMnCrCo HEAs, twinning develops practically from the very beginning in one system simultaneously with slip deformation in the stage where deformation develops as a Lüders band, without hardening (see Fig. 2). In polycrystals of FeNiMnCrCo HEAs with the grain size $d = 4.4$ and 155 μm twinning is preceded by the 20–35% slip deformation, which results in a considerable increase in the level of deforming stresses; this in turn gives rise to an increase in τ_{cr}^{tw} in polycrystals due to the interaction of twinning with slip [3].

Generally, twinning in FCC-crystals, oriented in the center of the stereographic triangle, develops under tension when the crystal axis reaches the [001] – $[\bar{1}11]$ symmetral, when the conditions are available for activation of several slip systems necessary for the formation of twins [26, 27]. In the $[\bar{1}49]$-oriented single crystals of FeNiMnCrCo HEAs, oriented in the center of the stereographic triangle for a single shear, twinning begins at the strain less than 5%, before the crystal axis reaches the [001] – $[\bar{1}11]$ symmetral, when the variation in the initial orientation is negligible and slip occurs in a single system at the maximum Schmidt factor value. Hence, twin formation occurs by a different mechanism, which does not require several slip systems.

The development of twinning without availability of slip in several systems can occur via a 'glide source' mechanism [28], when splitting of the perfect dislocations, $a/2\langle 110\rangle$, into the partial Shockley's dislocations, $a/6\langle 211\rangle$, follows the Heidenreich – Shockley reaction in the (111) plane

$$a/2\langle 101\rangle = a/6\langle 211\rangle + a/6\langle 112\rangle,$$

$$b = b_1 + b_2, \tag{3}$$

and the value of splitting of a perfect dislocation, $a/2\langle 110\rangle$ is found as follows (according to [29]):

$$d = \frac{Gb_1^2}{8\pi\gamma_{eff}},$$

$$\gamma_{eff} = \gamma_0 \pm \frac{(m_2 - m_1)}{2}\sigma b_1 + \frac{(\tau_1 - \tau_2)}{2}b_1. \tag{4}$$

Here G is the shear modulus of the FeNiMnCrCo HEA, b_1 is the Burgers vector of a Shockley partial dislocation, γ_{eff} is the effective stacking fault energy, which depends on the initial stacking fault energy γ_0, the value of applied stresses σ, factor $Q = \frac{(m_2 - m_1)}{2}$, where m_1 and m_2 are the Schmidt factors for the leading b_1 and trailing b_2 Shockley dislocations, respectively [29], the sign of applied stresses «±», and the physical factor $\frac{(\tau_1 - \tau_2)}{2}b_1$, which takes into account the difference in friction forces acting on the trailing b_2 and leading b_1 Shockley partial dislocations [29].

An analysis of relation (4) suggests that during tensile deformation in order to realize the 'glide source' it is necessary for γ_{eff} to tend to zero. Primarily, this is achieved when the external stresses, and hence, the Schmidt factor m_1 for dislocation b_1 is higher than m_2 is for b_2, $m_1 > m_2$, and $Q < 0$ (geometrical factor). Secondly, the resistance of the lattice of the FeNiMnCrCo HEA to the motion of dislocation b_1 is smaller than that to the motion of dislocation b_2:

$\tau_1(b_1) < \tau_2(b_2)$ (physical factor). The $[\bar{1}49]$-orientation lies on the line passing through the poles [012] – $[\bar{1}13]$ (see Fig. 1), within which the Schmidt factors are $m_1 = m_2 = 0.43$ and $Q = 0$ [29]. Therefore, in the $[\bar{1}49]$-crystals the 'geometric factor' cannot ensure the action of the 'glide source'. Splitting of a perfect dislocation b into Shockley partial dislocations b_1 and b_2 is assumed to occur due to the physical factor, when dislocation b_1 appears to be similar to a screw dislocation and weakly interacts with the elastic field stresses from the substitution atoms, while the trailing dislocation b_2 has a substantial edge component, so $\tau_2(b_2)$ turns out to be larger than $\tau_1(b_1)$ [16, 23, 30]. This favors separation of the twinning dislocation b_1 from b_2, and the perfect dislocation a/2<110> loses stability of splitting into Shockley partial dislocations a/6<211>. *In situ* observations of the motion of perfect a/2<110> and partial a/6<112> Shockley dislocations in FeNiMnCrCo HEAs in the column of a transmission electron microscope validate the development of twinning in these alloys via the 'glide source' mechanism [6]. It was shown in [6] that a heavy lattice distortion in FeNiMnCrCo HEAs increases the resistance to the motion of perfect dislocations a/2<110>, while Shockley partial dislocations a/6<112> turn out to be very mobile, which favors easy twin nucleation [6, 28]. As a result, there develops an intrinsic stacking fault [29, 31]. The thickness of twins can further increase due to overlapping of the stacking faults emerging via the same mechanism. It was revealed by a TEM examination that the $[\bar{1}49]$-oriented single crystals of FeNiMnCrCo HEAs contain stacking faults and thin imperfect twins (see Fig. 4). Lengthy stacking faults can form via the 'glide source' mechanism [28], when a perfect dislocation a/2<110> splits into Shockley partial dislocations b_1 and b_2, slipping in the (111) plane. If a pileup of a/2 <110> dislocations experiences slip, stacking faults are formed in the slip plane containing a high density of partial dislocations. The thickness of twins can increase due to overlapping of imperfect twins on top of each other [26, 27]. Such imperfect twins turn out to be more effective obstacles for the perfect dislocations and twins of other systems to overcome, which might be the reason for high values of Θ_{II} in the second stage of hardening.

SUMMARY

The experimental investigations of the $[\bar{1}49]$-oriented single crystals of the $Fe_{20}Ni_{20}Mn_{20}Cr_{20}Co_{20}$ (at.%) HEA have demonstrated that within the temperature interval $T = 77$–573 K under tensile deformation the temperature dependence of critical resolved shear stresses $\tau_{cr}^{sl}(T)$ is controlled by slip and its behavior is characteristic for single crystals of austenitic stainless steel without nitrogen atoms.

In the $[\bar{1}49]$-oriented $Fe_{20}Ni_{20}Mn_{20}Cr_{20}Co_{20}$ HEA single crystals, a combination of heavy lattice distortion and low stacking fault energy, $\gamma_0 = 0.019$–0.021 J/m^2, similar to the case of Hadfield steel with a high concentration of interstitial atoms – carbon, gives rise to twinning from the onset of plastic flow at $\varepsilon \leq 5\%$ within a wide range of temperatures from 77 to 423 K. The critical resolved shear stresses τ_{cr}^{tw} for twinning strongly depend on the testing temperature: $\tau_{cr}^{tw}(T = 77\ \text{K})/\tau_{cr}^{tw}(T = 296\ \text{K}) = 2$. The strong temperature dependence of $\tau_{cr}^{tw}(T)$ is determined by the temperature dependence of $\tau_{cr}^{sl}(T)$.

At $T < 300$ K, the twins are thin (twin thickness 10 nm) and their structure is imperfect, as is the case in single crystals of Hadfield steel. As the testing temperature is increased to 423 K, twin thickness increases to 20–25 nm. From the very onset of plastic flow, twinning develops via the 'glide source' mechanism.

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