Vol. 9 №. 3 **Trans. Nonferrous Met. Soc. China** Sep. 1999

Article ID: 1003 - 6326(1999)03 - 0623 - 06

Dislocation evolution with rheological forming of metal①

Luo Yingshe(罗迎社)[1], Guo Xiaogang(郭小刚)[1], Mei Yafu(梅亚夫)[2], Hu Yungui(胡云贵)[3]
1. *Research Institute of Fundamental Mechanics and Materials Engineering,*
2. *College of Mechanical Engineering, Xiangtan University, Xiangtan 411105, P. R. China;*
3. *The Department of Engineering Mechanics, Nebraska-Lincoln University, NE, USA*

Abstract: It is known that some internal defects exist in metal materials. Preliminary attempt to relate dislocation evolution with metal rheological forming was done. By the attempt, it is learned that the evolution of dislocation density $\rho(x, y, t)$ is essentially the change of n independent internal variables $q_\alpha(\alpha = 1, 2, \cdots, n)$ with material. The preliminary research in theory and experiments showed that dislocations piling-up could be avoided. One can improve the internal microstructure and mechanical properties of products by rheological forming method.

Key words: dislocation; metal; rheological forming **Document code:** A

1 INTRODUCTION

Until now, for most of the applied engineering materials, there would be some internal defects in them; for examples, impurities, cracks, cavities, point defects such as displaced atom or foreign matter atoms, and line defects such as dislocation, plane defects caused by piled dislocation in layers and crystal boundary, etc. All these materials can be called bodies with defects.

Usually, defect has dual characteristics. On one hand, those defects in the material are studied as some unfavorable rheological factors, because some components may lose effectiveness as time increases, even though the defects seem negligibly small. However in this paper, on the other hand, some kinds of defect such as dislocation, are studied as a favorable rheological factor. The theory and experiments revealed that[1] if no evolution of dislocation density, then no any rheological deformation. It is the line defects existing in crystals that would slide easily under the function of shearing, and so cause the rheological formation. Conversely, by metal rheological formation some internal defects such as cavities, microcracks etc will vanish, therefore the mechanical properties of products will be improved. In fact, both of rheological deformation and rheological fracture are stemmed from dislocation evolution. In order to prevent the source of dislocation fracture, one may create certain condition, for instance, in order to prevent dislocations from piling-up and be favorable for undergoing rheological deformation, one may heat the material to a proper temperature, and choose a proper deformation rate.

It is well known that since the concept of dislocation was proposed in 1934, the research on dislocation theory has gone through three stages[2,3]. The first was from 1934 to 1955, the statics and dynamics of single dislocation were mainly studied, aiming at solving the related stress field. The second was from 1955 to 1978, beginning the study of continuous distributed dislocation theory and continuous distributed dynamics. The third was from 1979 onwards, beginning with the research on Lasota's theory[3] and geometry theory of defect standardized field. The common characteristics of all the above theories are that the displacement vectors caused by strain field and stress field are

---

① Project 19632030 supported by the National Natural Science Foundation of China and project 97JKY2034 supported by Science and Technology Committee of Hunan Province Received Aug. 24, 1998; accepted Mar. 23, 1999

added up according to the principles of vector addition, so as to cause the addition of strain tensor and stress tensor on the strain field and stress field, therefore there would be two sets of constitutive equations and two sets of physical constants, which is difficult to measure.

From the point of view of mechanics, researchers called the above content as rheological solid mechanics combining micro and macro levels. In recent years, it is getting more and more deeper and wider in development combined with macro-meso-micro levels. Some significant achievements on the constitutive equations of metal polycrystal of the infinite deformation have been obtained[4].

2 THEORETICAL STUDY ON EVOLUTION OF DISLOCATION DENSITY

Up to now, there are four kinds of theoretical models on evolutional dynamic of dislocation density. They may be divided into two sorts: Holt model[5], low energy[6] dislocation structure model and Kratochvil model[7] belong to the first sort; the characteristic of this kind of model is that the dislocation flow J would be defined by stress tensor or dislocation energy, although the theory structure is complete, the two sets of physical constants are difficult to be measured so that it is not easy to use. The second one is Walgraef-Aifantis' dynamic model of reaction-diffusion[8], except some constants and parameters, the others can be described by dislocation density $\rho(x, y, t)$.

2.1 Dynamic physical model of evolution on dislocation density

Because dislocations exist in materials with temperature field and stress field, the plastic deformation of metal material is a kind of irreversible process which is far from a balanced state of thermodynamic. From the point of view of physics, plastic deformation is a macro-reactivity of dislocation density and movement (i.e. the evolution of dislocation density), and is a kind of dislocation evolution. When we treat a set of dislocations' state as a system, the physical model is displayed as follows.

The irreversible essential of plastic deformation is the irreversible movement of dislocation. Hence the evolutional system of the dislocation density also has irreversible characteristic. This irreversible characteristic must destroy the irreversibility of time inversion, so that the dislocation dynamic equations should be parabolic, and not the hyperbolic.

As the temperature field exists in the system, the dislocation movement under the effect of thermoactivation is controlled by diffusion processes, therefore the dislocation has certainly some diffusibility. They can be increased in the shearing processes and submerged in crossed slips and climbing motion. The characteristics of dislocation are comparable to chemical reactions, so that the evolutional system of dislocation is a complex coupling system of chemical reactions and diffusion, the evolution of dislocation density is a comprehensive acting result of the chemical reactions and diffusion[2,8,9].

According to the principle of total mass conservation in the irreversible thermodynamics,

$$\frac{\partial \rho}{\partial t} = \nabla[D \nabla \rho] + \frac{\partial_i m}{\partial t} \tag{1}$$

and

$$\frac{\partial_i m}{\partial t} = X \tag{2}$$

This is to say the substance produced by chemical reaction equals generalized force

$$X = \nabla\left(\frac{\mu}{T}\right) \tag{3}$$

where m is the mass of the evolutional system, t is time, D is evolution coefficient, T is temperature, μ represents chemical potential function, generally speaking, μ is a nonlinear function of dislocation density ρ, therefore $\mu = \mu(\rho)$.

2.2 Evolution dynamic equations of dislocation density

If T is treated as parameter, from Eqn.3 one can obtain the generalized force

$$X = T^{-1} \nabla \mu \tag{4}$$

where $\mu(\rho)$ expanded into polynominal of ρ, it should have minimum value, if one considers the nonlinear mutual effect among dislocations, the expanded equations of $\mu(\rho)$ may become as the following

$$\mu(\rho) = d\rho^4 - c\rho^3 - b\rho^2 \tag{5}$$

Under the conditions of assuring that Eqn.5 has minimum value and has compendious state in the real space, Eqn.5 has and only has two kinds reasonable forms

$$\mu(\rho) = d\rho^4 - b\rho^2 \tag{6}$$

and $$\mu(\rho) = d\rho^4 - c\rho^3 - b\rho^2 \tag{7}$$

It is easy to see that Eqn.6 has two compendious states $\rho = \pm (b/d)^{1/2}$, at this time symmetrical cracks can possibly happen, while Eqn.7 has no compendious state, it is impossible to have symmetrical cracks.

Put Eqns.6, 7 into 4, and divide by the parameter with the highest power of ρ, one can obtain

$$X_1 = \rho^3 - \tilde{B}(T)\rho \tag{8}$$

and

$$X_2 = \rho^3 - \tilde{C}(T)\rho^2 - \tilde{B}(T)\rho \tag{9}$$

Put Eqn.8 and Eqn.9 respectively into Eqn.1, we can obtain the evolutional dynamic equations expressed in dislocation density $\rho(x, y, t)$, i.e.

$$\frac{\partial \rho}{\partial t} = \nabla(D\nabla\rho) - \tilde{B}(T)\rho + \rho^3 \tag{10}$$

and

$$\frac{\partial \rho}{\partial t} = \nabla(D\nabla\rho) - \tilde{B}(T)\rho - \tilde{C}(T)\rho^2 + \rho^3 \tag{11}$$

Both of them are nonlinear diffusion-reaction type dynamic equations. Let $\rho = \rho_i + \rho_m$, then put it into Eqn.10, according to the duality relation of source and sink of immovable dislocations and movable dislocations, and the principle of high power item of mutual effect and reactivity, we can get the evolutional dynamic equations as the following to describe the movable and immovable dislocation densities. Subscript i represents immovable dislocation, m represents movable dislocations.

$$\left.\begin{aligned} \frac{\partial \rho_i}{\partial t} &= \nabla(D_i\nabla\rho_i) + g_1(\rho_i) - \tilde{B}(T)\rho_i + \rho_m\rho_i^2 \\ \frac{\partial \rho_m}{\partial t} &= \nabla(D_m\nabla\rho_m) + \tilde{B}(T)\rho_i - \rho_m\rho_i^2 \end{aligned}\right\} \tag{12}$$

where

$$g_1(\rho_i) = \rho_i^3 + \rho_i\rho_m^2 \tag{13}$$

represents the source of immovable dislocations caused by the retention of the dislocation increment. The model 12 is called Walgraef-Aifantis' diffusion-reaction type dynamic equations[8].

From Eqn.11, we can also obtain

$$\left.\begin{aligned} \frac{\partial \rho_i}{\partial t} &= \nabla(D_i\nabla\rho_i) + g_2(\rho_i) - \tilde{B}(T)\rho_i - \tilde{C}(T)\rho_m\rho_i + \rho_m\rho_i^2 \\ \frac{\partial \rho_m}{\partial t} &= \nabla(D_m\nabla\rho_m) + \tilde{B}(T)\rho_i + \tilde{C}(T)\rho_m\rho_i - \rho_m\rho_i^2 \end{aligned}\right\} \tag{14}$$

where

$$g_2(\rho_i) = \rho_i^3 + \rho_m^2\rho_i - \tilde{C}(T)\rho_i^3 \tag{15}$$

It is easy to see that the difference of model 12 and 14 stems from the choice of Eqns.6 and 7.

2.3 Discussion about properties of models

It appears two kinds of nonlinear evolutional dynamic equations of dislocation density, model 12 and 14 given in the above section; now let us discuss the properties of the models.

Based on the general principle of the least free energy, the low energy dislocation structure means that the driving power of dislocation patterns formed is the decrease of free energy in the dislocation system, while the dislocation patterns is a kind of stable low energy dislocation structure. Therefore we can obtain from the generalized forces 8 and 9

$$X_2 - X_1 = -\tilde{C}(T)\rho^2 < 0 \tag{16}$$

that is to say $X_2 < X_1$. According to the principle of low energy dislocation structure, model 14 is the more reasonable fine dynamic model of the evolution of dislocation density.

According to the stable theory, the model 12 is the result of losing stability (that is, $C(T) = 0$) of the above model's structure. In this sense, the evolutional dynamic models of dislocation density is model 14. It is worth pointing out that the parameters b, c, d, or $B(T)$, $C(T)$, given in the above section should be

$$\mu = \mu(\rho, \varepsilon, T) \tag{17}$$

so that we can obtain

$$\left.\begin{aligned} b &= b(\varepsilon, T) \\ c &= c(\varepsilon, T) \\ \underset{\sim}{d} &= \underset{\sim}{d}(\varepsilon, T) \end{aligned}\right\} \qquad (18)$$

$$\left.\begin{aligned} \underset{\sim}{B} &= \underset{\sim}{B}(\varepsilon, T) \\ C &= C(\varepsilon, T) \end{aligned}\right\} \qquad (19)$$

It is not difficult to see that those parameters are really dependent on the functions of strain ε and temperature T. Eqn.17 and the formulae below

$$\left.\begin{aligned} e &= e(\varepsilon, T, q_\alpha) \\ \sigma &= \sigma(\varepsilon, T, q_\alpha) \end{aligned}\right\} \qquad (20)$$

have the same forms[10], whilst Eqn.20 is the expression for extracting internal energy and stress discussed in the imaginary ε_{T_q} state space on the internal variable theory, hence the meanings of ρ and $q_\alpha(\alpha = 1, 2, \cdots, n)$ must be the same, therefore we can extend the evolution of dislocation density ρ into the changement of n independent internal variables q_α. The specific physical meaning of q_α is very extensive, it depends on the internal structure and the structural state of special material under the specific condition, in a word, it indicates some movement or structure rearrangement inside the material, for metal, it indicates all sorts of evolution and development of multi-crystal internal structure, such as dislocation, lattice skewing, and recrystallization. This kind of evolution and development is actually a sort of chaotic dynamic process[11].

3 EXPERIMENTAL RESEARCH

3.1 Dislocation research

It has been pointed out that the deformation and fracture are two different kinds of results of dislocation evolution. During the metal's thermo-rheological forming, one should test and choose out of proper forming force, deformation rate or forming velocity, according to working instruction or new successful technique, for example, computer simulation of forming process under certain temperature condition and velocity condition. If one choose the above factors improperly, it may lead to dislocations piling-up even to fracture. The reasons which caused this kind of phenomenon are that during the deformation process of metal crystal, the movable dislocations shaped just like a knife edge are often obstructed by the crystal boundaries, sub-grain boundaries, the second phase and fixed dislocation. They can not move and hence pile up, as shown in Fig.1.

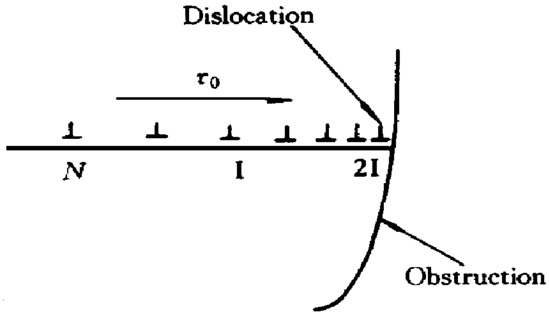

Fig.1 Dislocation piling-up scheme

It is not difficult to obtain[12]

$$\tau = N\tau_0 \qquad (21)$$

where τ is the force applied to obstructing by leading dislocation whose value equals the reacting force by the obstruction, τ_0 is the outside stress, N is the number of dislocations in the piling group. Eqn.21 shows that there is a stress concentration at the forward position in the piling group. If this stress concentration cannot be relaxed, the forward position in the piling group will become fracture source of dislocation. Fig.2 shows the micrograph of the Ti (Ti-6Al-4V) alloy forming product, which was formed under improper temperature 800 ℃. Dislocation pilings and micro-cracks usually occur on the crystal boundary as shown in Fig.2, thus it is easy to form a fracture source of dislocation, and proper deformation temperature should be 950 ~ 970 ℃.

By qualitative analysis, if there are sufficient piling dislocations in front of the obstruction, the stress concentration in piling group increases to a certain value, it may stimulate the dislocation source of the neighboring crystals to move, make the deformation extend from one crystal to another, or might destroy some immovable obstruction, this is the yield process produced in the material. In order to carry out metal's rheological forming, to prevent it from producing fracture source of dislocation, it is very important to study the effect on rheological

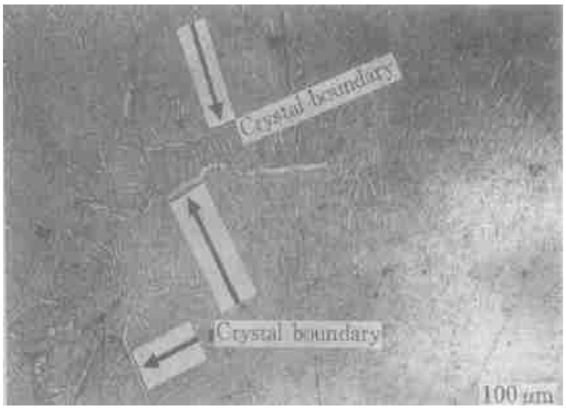

Fig.2 Micrography of Ti-6Al-4V alloy

stress by temperature and strain rate.

3.2 Effect on rheological stress by temperature and strain rate[13, 14]

We conducted the experiment on aluminum alloy plate A6061, at the different temperature levels of testing, we can see clearly that heating carries complex effects on rheological properties of the material, owing to the different tensile speed and different drawing strain rate. In the formula $\sigma = K\varepsilon^M$, K is a material constant, we treat the stress as the function of strain-rate, the sensitivity index M of strain-rate is not a constant, it should be a function of deformation.

3.3 Test of products properties

Proper rheological deformation temperature and speed cannot only prevent the dislocations from piling-up and the production of source of dislocation fracture, but also eliminate some internal defects, for examples, smashing crystals, dispelling holes and micro-flaws, etc. Therefore the internal structure in product material can be improved, and the mechanical properties can also be improved. Figs.3(a), (b) and (c) show the material metallographical structure before and after thermo-rheological forming of Ti alloy impeller, respectively. It is not difficult to see that before deformation the material is needle constitution of $\alpha + \beta$, the hub part with larger deformation shows equiaxial transformed constitution of $\alpha + \beta$, and some new extended α less than 0.5 mm in length. The vane part with the largest deformation shows complete equiaxial $\alpha + \beta$ constitution. It is thus clear that the process of thermo-rheological forming is a smashing process of crystals and improving process of internal structure and mechanical properties continually.

It is because rheological forming eliminates some internal cracks, smashes crystals, therefore the mechanical property index of products has been improved. Table 1 and Table 2 give the test results of Ti-6Al-4V alloy product as some examples.

At the same time, we produced a kind of parts with thin wall and high ribs. The parts were made of a kind of high strength, low

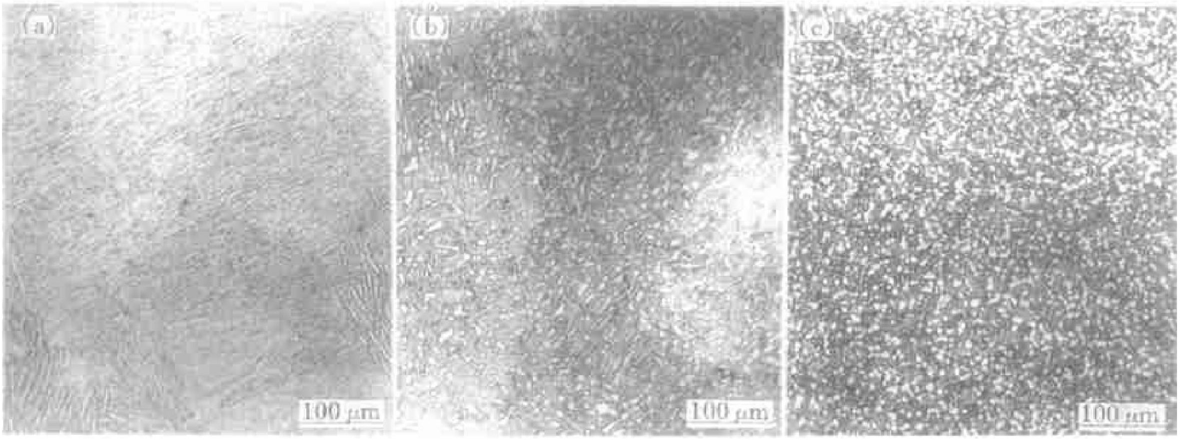

Fig.3 Material micrographies of Ti-6Al-4V alloy impeller made by rheological forming method

(a) —Raw material; (b) —Hub part; (c) —Vane part

Table 1 Room temperature test results(20 ℃)

Test items	σ_b/ MPa	$\sigma_{0.2}$/ MPa	δ/ %	ψ/ %
Prescribed data	≥930.00	≥830.00	≥9.00	≥25.00
Test data	1 080.44	1 030.27	13.30	55.30
Improved ratio/ %	116.13	124.13	147.78	221.20

Table 2 High temperature test results(300 ℃)

Test items	σ_b/ MPa	$\sigma_{0.2}$/ MPa	δ/ %	ψ/ %
Prescribed data	≥650.00	≥530.00	≥9.00	≥35.00
Test data	780.38	650.20	14.00	64.50
Improved ratio/ %	120.06	122.68	155.56	184.29

plastic aluminium alloy material (LY12) by thermo-rheological creep forming method. The so-called creep forming is to keep stress as constant, let the thin equal thickness wall forming increasing with t. Before this time, it was produced by machine cutting, cold or warm extrusion, it was easy to produce dislocation piling that would lead to fracture and internal cracks at the transitional area where the thick and large cross-section and thin one meet, because the thinnest area is only 2 mm thick, but about 100 mm high[15]. As we adopted a new method of thermo-rheological creep forming technique, choosing proper forming temperature and forming speed, heating the dies and the processing material at the same time, we are able to prevent dislocations from piling-up, and ensure the internal material quality. Fig.4 (a) shows the metallographical structure of thick and large cross-section of the part, Fig.4 (b) shows the metallographical structure at the interface of thick and thin cross-sections of the part, crystals are small and distributed uniformly, hence the internal quality at the transitional area has been ensured.

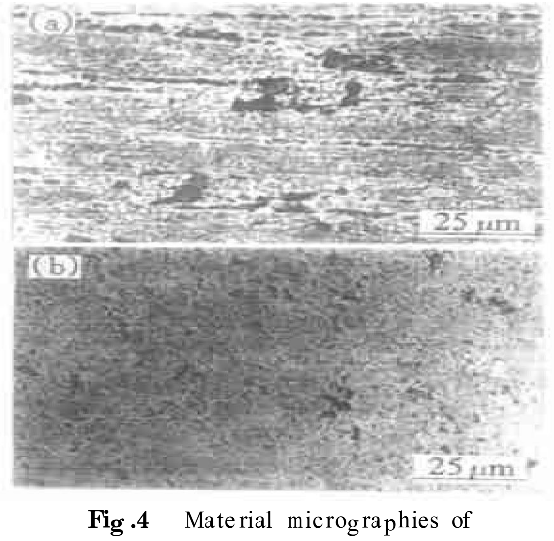

Fig.4 Material micrographies of aluminium alloy material (LY12)
(a) —Thick and large cross-section;
(b) — Meeting place of thick and thin cross-sections

REFERENCES

1 Luo Y S and Wang C. In: Xu B Y *et al* eds, Int Conf on Microstructures and Mechanical Properties of New Engineering Materials. Beijing: Inter Aca Pub, 1995, 213 ~ 216.

2 Dong L K. Fractal Dynamics. Shenyang: Liao Ning Sci & Tech Press, 1994.

3 Aida K and Dominic G B E. A Gauge Theory of Dislocation and Disclination. Berlin: Springer, 1983.

4 Wang R. J Mech & Prac, 1992, 14: 1 ~ 5.

5 David L H. J Appl Phys, 1970, 41: 3197 ~ 3201.

6 Wilsdorf D K. Mater Sci Eng, 1986, 81: 141 ~ 161.

7 Kratochvil J. Review Phys Appl, 1988, 23: 419 ~ 429.

8 Daniel W and Elias C A. Int J Eng Sci, 1985, 23: 1351 ~ 1372.

9 Nicolis G and Prigogine I. Self-Organization in Nonequilibrium Systems. New York: John Wiley & Sons, 1977.

10 Luo Y S. J Xiangtan Univ, 1989, 11: 153 ~ 161.

11 Luo Y S, Luo K W and Zhou J P. Journal of Hunan University, 1995, 22: 98 ~ 103.

12 Shi D K. Dislocation and Material Strength. Xi′an: Xi′an Jiaotong Univ Press, 1988.

13 Luo Y S and Toshihiko M. Forg Stamp Techn, 1996, 21: 32 ~ 35.

14 Liu C L, Zhang P and Pan Yong. Trans Nonferrous Met Soc of China, 1995, 5: 84 ~ 87.

15 Luo Y S and Luo Z H. Hot Working Tech, 1997, 126: 11 ~ 13.

(Eidted by Huang Jinsong)