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ON BRITTLE FRACTURE IN F.C.C.-METAL

Peter Panfilov and Alexander Yermakov

Laboratory of Strength, Urals State University, 620083 Ekaterinburg, Russia

ABSTRACT

Fracture behavior of inherently brittle and environmentally induced brittle f.c.c.-metals, including evolution of cracks in bulk crystals and thin foils for transmission electron microscope, is considered in this paper. Refractory iridium and gallium covered aluminum are chosen as model substances. In spite of brittle fracture, these materials show good plasticity, which is considerably limited in polycrystalline aggregates because of low grain boundary strength. In both cases, inclination to brittle fracture takes place in bulk crystals only, while thin foils fail by ductile manner. The causes of brittle crack appearance in f.c.c.-metals and physical model of crack evolution in plastic metal are discussed.

KEY WORDS: brittle fracture, plastic deformation, cracks, f.c.c.-metal, iridium

INTRODUCTION

According to long term experience, ductility of crystalline solid does not allow it to fail by brittle manner, and impossibility of dislocation motion is usually considered as the main physical cause for brittleness [1]. The border situation, when ability of crystal to plastic deformation is substantially limited, has been the subject for many researches aimed on micro mechanisms of brittle fracture [2-5]. Conception of dislocation emission from crack tip turned out to be helpful instrument for study of brittle to ductile transition in materials, where dislocation motion is bothered. The model shows good agreement with mechanical behavior of high strength alloys and intermetallics having face centered cubic (f.c.c.) lattice, but it could not be applied to the pure metals, since no any dislocation barriers, particles of the second phase or strong directed interatomic bonds exist here. Despite this platinum group metal iridium fails by brittle manner under mechanical load and, in addition, meets few empirical criteria for brittle fracture [6]. Also, intercrystalline and transcrystalline brittleness of pure f.c.c.-metals could be induced by liquid metals [7]. Phenomenon of brittle fracture in f.c.c.-metal and its physical model, including crack growth, are discussed in this paper.

Iridium is unique f.c.c.-metal due to highest melting point (2443°C) and chemical inertia [6]. Strong interatomic bonds cause an anomaly in elastic modules, which are similar to ones for refractory body centered cubic (b.c.c.) metals [8,9]. Besides its workability is very poor and fracture modes are determined as brittle intercrystalline fracture (BIF) and brittle transcrystalline fracture (BTF) for poly- and monocrystalline states, respectively [10,11]. The problem of processing iridium was successfully solved, when refining technology, allowed to remove all dangerous impurities from the metal, has been elaborated [12]. However, impurities-free iridium crystals continues to cleave under tension in spite of huge elongation prior the failure (100% at 100°C!) [13], while an inclination to BTF mysteriously vanishes under compression tests [14]. In contrast with single crystals, ever high pure polycrystalline

iridium fails practically without preliminary deformation demonstrating 100% BIF on fracture surface [15]. Therefore, iridium was called inherently brittle f.c.c.-metal the more so it unexpectedly meets empirical criteria for brittle fracture [7,16]. The last means that fracture mechanics could be formally applied to this f.c.c.-metal without any limits, although physical reasons did not allow to do that [17]. Observations of V-shaped cracks on bulk crystals [7] seems like confirmation of inclination to brittleness, however both considerable plasticity of material and big angle of crack opening point to specific character of "iridium brittleness". In any case findings on fracture behavior of iridium are extremely important for understanding the mechanisms of brittle fracture in metallic materials, since it gives an fortunate opportunity to watch simultaneously evolution of cracks and deformation process around them by direct means.

The main cause of liquid metal induced brittleness in polycrystalline f.c.c.-metals is catastrophic decrease of grain boundary strength because of influence of liquid metal atoms [18]. BTF induced by liquid metals is very rare phenomenon, which is only observed in aluminum crystals covered by gallium or mercury, and, therefore, there are not generally accepted physical models of brittle crack growth in literature [7].

THE CAUSE OF BRITTLE BRACTURE IN IRIDIUM

Simplest arguments have been laid in the basis of presented physical model. Brittle fracture becomes possible when a crystal cannot be plastically deformed under mechanical load, inasmuch as either it is undeformable material or its resource of plasticity has been exhausted during preliminary deformation. Without a doubt, the second way should be chosen for high plastic f.c.c. –metal, whose fracture surface looks like BTF or mixture of BTF and BIF. Strong interatomic bonds are only one inherent distinction between refractory iridium and other f.c.c.–metals, but namely this circumstance causes low mobility of dislocations and as aftermath the highest value of yield stress at room temperature. Octahedral slip of $<110>$ dislocations is the dominant deformation mechanism in iridium, therefore its strong hardening under load could not be connected with alternative ones like mechanical twinning or non-octahedral slip [19]. Consequently, mobile $<110>$ dislocations should create the obstacles for octahedral slip, what, however, it is impossible in normal f.c.c.-metal ever at low temperatures [20,21]. Experiments have shown, that all huge plasticity of single crystalline iridium is realized on the easy slip stage when the sole permitted dislocation configuration is dislocation nets, whose dense seems like concentration of dislocations in irradiated metals [22]. In this case, the obstacles are $<110>$ dislocations themselves, which have been situated in the net. Stress rising allows to hammer $<110>$ dislocations in the net practically without any limit, until the time when tensile stress becomes enough for crack creation. Evidently, this condition never meets under compression, and, as a results, grain boundary (GB) free iridium behaves like normal f.c.c. –metal [23].

BIF is the cause of poor plasticity of polycrystalline iridium, but no differences between crystallographic characteristics of GBs in iridium and normal f.c.c. –metal have been revealed [24]. This is an experimental fact that at room temperature polycrystalline samples fail after easy slip stage, when $<110>$ dislocations gain a tend to move through GBs. Neck region begins to form in polycrystalline aggregates at 400°C, since transition of dislocation structure from the net to small angle boundaries

become possible because of the rising of dislocation mobility [25]. However, necking to a point and flowing neck are observed in severe deformed (GBs free) iridium and its alloy at 1000°C, where GBs do not bother this transformation [26]. Anomalously wide diffusion zone, contained high concentration of vacancies, is the reason why GBs in iridium are opaque for <110> dislocations [27]. Workability of iridium could be improved by means of doping GBs by heavy metallic impurities, despite this never suppress BIF in material [28].

CRACK GROWTH IN BULK IRIDIUM CRYSTALS

Detail experimental data on transcrystalline cracks in iridium single crystals under tension and bending were published in [29,30]. It was shown that tiny cracks, whose length has been measured as 0.01?0.03 mm, appear on the edges of preliminary deformed crystal near notch like defects. They had V-shape with 10?15° angle of opening and advanced along a normal to tensile axis. No any deformation tracks were revealed near crack edges. More long cracks became sensitive to crystallographic orientation of sample, however this did not influence on their shape. So, cracks in crystals stretching along <110> changed their growth direction from 90° to 60°?70°, while cracks in samples with tensile axis of <100> continued to grow in the same direction. Simultaneously, thin deformation tracks appeared near crack edges, but this did not lead to crack tip blunting. Their orientation and shape were similar to geometry of octahedral slip bands advancing around notches in single crystals of normal f.c.c.-metal under tension along <110> and <100> directions, respectively. Deformation tracks ceased to leave from the edges of upper part of cracks, when crack length became over 0.1 mm. It important to note that such cracks, whose motion causes the failure, appeared before the moment of separation. Sometimes, long cracks could branch on two or more parts, each of them laid either primary or secondary cleavage planes ({100} and {210}, {110}). Under bending, cracks also possessed V-shape, but deformation tracks were observed near the longest cracks when tensile deformation became considerable.

EVOLUTION OF CRACKS IN THIN FOILS OF IRIDIUM FOR TRANSMISSION ELECTRON MICROSCOPE

Evolution of cracks in thin foils of iridium (and aluminum) was considered in [29–31]. Smallest objects in iridium foils, which could be determined as cracks, had the length approximately of 10-2 ?m . They and more longer cracks (up to 0.5 ?m in length) emitted perfect dislocation with <110> Burgers vectors from tips, which had a tend to move far away, but this never caused distinct blunting of cracks. In so doing, thin region of stacking fault leaved ahead the crack after passing of each dislocation. The growth of number of emitted dislocations led to rising the power of stacking fault and it transformed into twin lamella having clear visible “zebra” contrast. At that both stacking faults and twins extended from crack tip on the distance of 1 ?m . As a rule, in iridium foils emitted dislocations have been stopped on highly dense dislocation nets, situated on the distance of 3?7 ?m from the edge, and they began to look like dislocation pileups, while in aluminum emitted dislocations never met obstacles and flew in depth of foil without any trouble. Sometimes, transformation of emitted dislocation pileup in to a braid was observed in iridium. The crack length could reach the value of 1 ?m , if material around it did not contain highly dense nets. They always had “square” tips and began to emit perfect dislocations normally to initial growth direction. After that crack changed its growth direction on perpendicular one. As a re-

sult, crack obtained broken or zig-zag profiles. It is well known that motion of such crack causes the failure of thin foils of ductile metals [32–34], therefore in this case “brittle” iridium behaves like normal f.c.c.–metal.

FRACTURE BEHAVIOR OF GALLIUM COVERED ALUMINIUM CRYSTALS

Gallium (both in solid and liquid state) covered aluminum crystals (GCACs) failed under tension of 15–20%, while in air they shown elongation prior the failure of 35–50%, at that neck began to form at 15?20% [35]. Despite considerable elongation, no deformation relief and necking were observed in gallium covered aluminum crystals. According to metallographic study, gallium did not induce visible intensification of octahedral slip in the samples, although their yield stress was considerably lower than for pure aluminum crystal. Also, presence of gallium on the surface never led to activation of mechanical twinning and non-octahedral slip. V-shaped (iridium like) transcrystalline cracks (length is about 0.02?0.05 mm) were only revealed on the samples covered by gallium in solid state. Cracks appeared after preliminary deformation of 8?10 % on the power defects like notches or reper lines. Some of them could be grown up, but deformation tracks never lost their edges as it was reported in [7]. Appearance of slip band near the crack led to its transformation to pore-like defect, which grew in length and in width, simultaneously. Analysis has shown that the length of cracks transforming to pores is approximately in two times longer than for V-shape cracks. Pores were situated in narrow strip, where localization of plastic deformation took place. The dangerous crack formed as a result of junction of such pores along cross section of crystal covered by gallium in both solid and liquid state. Therefore, there are three stages of crack evolution in aluminum crystals covered by solid gallium. The first, growth of V-shape (iridium like) cracks; the second, transition from V-shape crack to pore and its growth; and, the third, junction of pores into dangerous crack.

DISCUSSION

It was shown that brittle fracture takes place in strengthened f.c.c.–metal only. Single crystals fail after considerable elongation, while poor plasticity of polycrystalline samples is synonymously connected with low cohesive strength of GBs. Any way, the cause of brittleness in plastic f.c.c.–metal is exhaust of resource of plasticity during preliminary deformation, if any dangerous defects, where cracks could appear, are absent in material. BTF in f.c.c.–metal becomes possible, when storage of plastic deformation occurs on the easy slip stage and sole permitted dislocation configuration is dislocation net. At that the crystal cannot lose ability to be deformable at the one moment because of heterogeneity of distribution of plastic deformation in the sample. It means that material is brittle in vicinity of cracks only, whereas it continues to be plastic in any other places. Therefore, neither crack appearance nor ever crack growth induce instantaneous separation of crystal. During deformation, volume of “brittle” substance is increasing until the motion of single crack causes failure of crystal. This is the reason why fracture behavior of iridium crystals is distinguished from fracture both non-metallic and b.c.c.–metallic crystals at temperatures lower than the point of brittle to ductile transition (BDT), although fracture surfaces of iridium, tungsten, molybdenum, iron, and silicon crystals in brittle state look similar.

Sharp V-shaped transcrystalline cracks in iridium crystals may be called the brittle cracks in f.c.c.–metal, since, the first, growth of one from them leads to BTF, and, the second, they have appropriate geometry. The cracks in GCACs on the first stage of evolution are also brittle ones, inasmuch as they look like that. An absence of deformation tracks near crack tips, at least, on initial and final stages of evolution, allows concluding that no plastic deformation occurs in crack tip or brittle crack in f.c.c.–metal does not emit dislocations from the tip. On the contrary, crack edges are “sources” of octahedral slip bands, if tensile stress has been applied to iridium crystal, however this should be considered exclusively as showings of residual plasticity of material far away from crack location. Indeed, brittle crack begins immediately transform to pore, as soon as octahedral slip appear near the tip, as it takes place in GCACs, where inclination to BTF shows in thin gallium reach layer near crystal surface¹. Hereafter evolution of crack in GCACs includes elements of both brittle and ductile crack growth: sharp pores, laid on low index {100} and {111} planes, are jointing in dangerous crack. Such behavior is similar to ductile fracture in neck region of f.c.c.–metal [20]. Certainly, it does not mean that material ahead brittle crack has completely lost an ability to plastic deformation or dislocations cannot be generated at crack tip, however, it is absolutely clear that contribution of these processes to fracture behavior should be insufficient.

TEM observations give extremely important information about fracture process on micro scale. It seems to be that generation of perfect dislocations accompanies the breaking of interatomic bonds and form of crack surface is f.c.c.–metal. So, atomic scale crack, which does not look like atomically sharp crack, emits perfect dislocations. In contrast with physical model of crack considered in [3–5,33], dislocation emission does not lead to visible crack tip blunting and arrest of crack. Emitted dislocations are flowing into the foil, if the power obstacles are absent ahead. Highly dense net bother them to come through, and pileup of emitted dislocations begins to turn into a braid. As a result, new generated dislocations have not opportunity to leave crack tip what should cause the stoppage of crack in thin foil of f.c.c.–metal. This hypothesis sounds like physical model of BTF in bulk iridium crystal transferred on 2D media. Indeed, in both cases dislocation nets do not allow iridium behaves like plastic substance. Also, it is very interesting that twin lamella near crack is a track remained by moving <110> dislocations in the thin region of foil. Unfortunately, the findings cannot be used for explanation of brittleness in f.c.c.–metals, since iridium thin foils show typical ductile fracture behavior like aluminum, silver and gold [31–34,36]. Gallium does not change fracture mode of aluminum foils, too [7]. Another word, this chapter confirms conclusion that brittle crack in f.c.c.–metal does not emit perfect dislocations, although it conserves an ability to generate them, but new born dislocations cannot move from crack tip because of highly dense net or pileup.

The fact that high plastic iridium meets empirical criteria for brittle fracture [8,16] should be considered as casual coincidence, although some relations between strong interatomic bonds (it holds third position in melting points among the metals) and inclination to brittleness would exist. Despite this BTF happens in reality, and, therefore, fracture mechanics can be used for description of transcrystalline cracks in iridium crystals. Naturally, mechanical models should be applied to the area, where material has being in brittle or undeformable state, since the role of plasticity at crack

¹ It may be supposed that transition from brittle crack to pore happens when it comes through this layer

tip is insufficient in this case only. At that consideration of plasticity with a help of “coefficients in equations”, which have not definite physical content, looks preferably than discussion on dislocation emission from brittle crack, inasmuch as this event never occurs in f.c.c.-metal. The more so, real mechanisms of plasticity ahead crack cannot be correctly described in the frames of fracture mechanics [37].

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