

Some Thoughts on Stress Corrosion Cracking of (7xxx) Aluminum Alloys

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Abstract: Stress corrosion cracking may be defined as a spontaneous failure of the metal resulting from the combined effects of corrosion and stress. It is general agreement that SCC requires three conditions: first, the alloy must be susceptible, second, there must be a specific environment (water or seawater for aluminium alloys); and third, there must be a tensile stress. The susceptibility of the wrought aluminium alloys to SCC is related to many environmental (chemical), Metallurgical (Microstructural) and Mechanical factors. Crack initiation may start at some surface discontinuities, corrosion pits, or may be developed by intergranular corrosion or slip dissolution. This review describes the models of environmental-related crack propagation of aluminum alloys (7xxx) during the last few decades. Acknowledge on effects of different factors on the susceptibility to SCC permits to propose valuable mechanisms on crack advancement. The reliable mechanism of cracking gives a possibility to propose the optimum chemical composition and thermal treatment conditions resulting in microstructure the most suitable for real environmental condition and stress state.

Keywords: Environmental, mechanism, microstructure, propagation.

1. Introduction

Stress corrosion cracking (SCC) is kind of environment-related failure, which may cause a dramatic degradation of mechanical properties, and failure of material exposed corrosion environment under stresses below yield stress value. Appearance of SCC in practical application has caused that the Al-Zn-Mg-(Cu) alloys at their peak strength are not recommended for aggressive environments.

The susceptibility of any aluminum alloy to SCC is related to many environment (chemical), metallurgical (microstructure) and mechanical factors, many of them being interrelated and difficult to describe in a more systematic way. SCC of Al-Zn-Mg-(Cu) alloys has a very complex behavior. Crack initiation may start at some surface discontinuities, corrosion pits, or may be developed by intergranular corrosion or slip dissolution [1] and the incubation of cracks is undoubtedly associated with local damage of an oxide layer by crack propagation is very complex behavior so there are two groups of models, on the other hand, proposed as involved in crack propagation anodic mechanisms and cathodic mechanisms. That is, during corrosion, both anodic and cathodic reactions must occur, and the phenomena that result in crack propagation may be associated with either type. The most obvious anodic mechanism is that of simple active dissolution and removal of material from the crack tip. The most obvious cathodic mechanism is hydrogen evolution, absorption, diffusion and embrittlement. The breaking of interatomic bonds at crack

tip occurs either by chemical salvation and dissolution or by mechanical fracture [2].

2. Models

The proposed mechanisms which try to explain the processes of degradation of mechanical properties and environmental-related of aluminum alloys can be classed into two basic categories: anodic mechanisms and cathodic mechanisms [2].

2.1. The Brittle Film Rupture Model (or a Champion – Logan Model)

It is a classical dissolution models [3]. In this model the stress acts to open the crack and to rupture the protective surface film. After the localized plastic deformation at the crack tip makes the rupture of the passivating film, exposed is the bare metal that dissolves rapidly, resulting in crack extension. Some investigators assume that once the propagation starts the crack tip remains bare because the rate of film rupture at the crack tip is greater than the rate of repassivation [4]. Others think that the crack tip repassivates completely and is periodically ruptured by the emergence of slip steps [2], [5], [6]. Davis [7] investigated the 7075-T651 alloy and observed that: (1) crack propagation was always associated with acidification at the crack tip, (2) increase in stress intensity caused rapid jumps in the corrosion potential only when the stress intensity was above KISCC and (3) the corrosion potential drifted in the active direction if extensive crack occurred. Some zones of quasi-cleavage were observed on the fracture surface. In his opinion, these results indicated that a film - rupture anodic dissolution mechanism was operating during SCC of aluminum alloys. Burleigh [8] cited some other papers on the model of a brittle rupture of the passive film. The schematic of the process is shown in the Fig. 1.

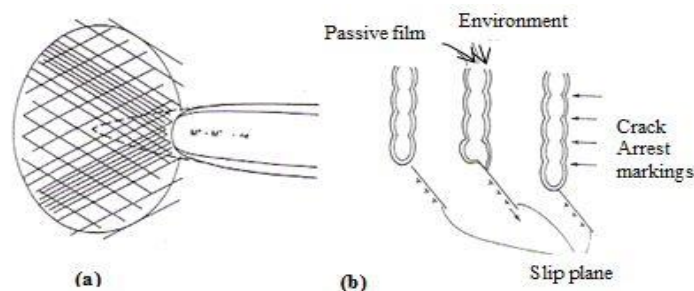


Fig. 1. Schematic representation of crack propagation by the film-rupture model; (a) Crack tip as a result of continues deformation, (b) Crack tip passivates and is ruptured repeatedly [9].

2.2. The Active – Path Intergranular SCC Model (Pre-existing Active Path Mechanism)

It is assuming that intergranular SCC can occur by an active-path process that results from a difference in the microchemistry of the material at the grain or interface boundary [2], [10], [11]. As reported by Dix, corrosion occurs along localized paths, producing fissures. Components of tensile stress normal to the path create a stress concentration at the base of then localized fissures. These preferentially corroded paths may represent relatively low resistance to corrosion or may be anodic. As reported in [12] in Al alloys such pre-existent paths are associated with grain boundaries. At sufficient concentration of stress, the fissures open further, thus exposing fresh unfilmed metal to the corrosion attack. Because this metal is anodic, increase in current flow would be expected, and hence there is an acceleration until protective films are reformed. Jacobs [13] proposed a modified model for the 7075 Al alloy which was similar in its broader aspects to the early model of Dix and required corrosion along a continues anodic path combined with mechanical tearing of the metal at the crack tip. According to the author, the initiation of a stress corrosion crack can take place at the interface between the pitted precipitates particles and the Al matrix. The crack initiating at this interface or at the base of a pit propagates intergranularly via a series of

chemical-mechanical (corrosion-mechanical) steps. Peel and Poole [14] proposed the model of SCC based on the proposal that grain boundary particles may act as sacrificial anodes and thereby retard stress corrosion crack growth. This point of view was supported by Poulou [13].

2.3. The Corrosion Tunnel Model

Assumes that a fine array of small corrosion tunnels forms at emerging slip steps. These tunnels grow in diameter and length until the stress in the remaining ligaments cause ductile deformation and fracture. The crack thus propagates by alternating tunnel growth and ductile fracture. Cracks propagating by this mechanism should result in grooved fracture surfaces with evidence of microvoid coalescence on the peaks [2]. The detailed models are illustrated in the Fig. 2.

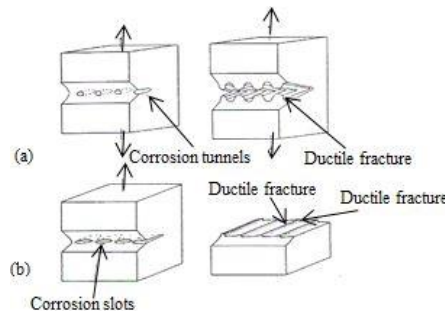


Fig. 2. Corrosion tunnel models: (a) initiation of a crack by formation of corrosion tunnels at tip steps and ductile deformation and fracture of remaining ligament, (b) Schematic diagram of the tunnel mechanism of SCC and flat slot formation [2].

2.4. The Adsorption – Enhanced Plasticity Models

These are the models which base at similar fracture processes observed frequently for SCC, hydrogen embrittlement (HE) and liquid metal embrittlement (LME). Because chemisorption is common to all three, this process was believed to be responsible for the environmentally induced crack propagation. Based on fractographic studies, it was concluded that cleavage fracture is not an atomically brittle process, but occurs by alternate slip at the crack tip with formation of very small voids ahead of the crack. It was also proposed that chemisorption of environmental species facilitated the nucleation of dislocation at the crack tip, promoting the shear processes responsible for brittle, cleavage-like fracture [2].

2.5. The Film – Induced Cleavage Model

Emphasizes the role of surface layers which can initiate micro-cracking. The model assumes that a surface film could induce cleavage fracture by the following stages: (1) a thin surface film or layer forms on the surface, (2) a brittle crack initiates in this layer, (3) the brittle crack crosses the film/matrix interface with little loss in velocity, (4) once in the ductile matrix, the brittle crack will continue to propagate, (5) this crack will eventually blunt and arrest, after which the entire process repeats itself. The hypothesis that a brittle crack will continue to propagate after it has entered the normally ductile matrix is a critical point. This allows a thin surface layer to induce brittle crack propagation over distances much greater than the film thickness [15], [16]. It was proposed that transgranular environmentally related controlled crack propagation can occur by a combination of oxidation-related (e.g. slip dissolution) and brittle fracture mechanisms. The crack front moves forward by an oxidation process that is controlled by the same rate-determining steps as those in the slip-dissolution model but when the film rupture event occurs after an increment in strain, the crack may rapidly penetrate a small amount into the underlying ductile metal matrix [1]. The extent of the additional film induced cleavage component of crack advance may be governed by the state of coherency and bonding between the surface film and matrix, the fracture toughness of the

substrate, the film thickness and initial velocity of the cleavage crack.

2.6. The Adsorption – Induced Brittle Fracture

Model is based on the hypothesis that adsorption of environmental species lowers the interatomic bond strength and the stress required for cleavage fracture. This model is frequently referred to as the stress-sorption model, and similar mechanisms have been proposed for hydrogen embrittlement and LEM. This model predicts that crack should propagate in a continuous manner at a rate determined by the arrival of the embrittling species at the crack tip [2]. Theories of stress sorption cracking were reviewed by Uhlig [17] who said that this mechanism proceeds not by chemical or electrochemical dissolution of metal at the tip of a crack but instead by weakening of already strained metal atom bonds through adsorption of environmental constituents. Then the surface energy is reduced.

2.7. The Slips Dissolution Model

Relates crack propagation to the oxidation that occurs when the protective film at the crack tip is ruptured. Once the film is ruptured, crack-tip advancement is governed by oxidation on the bare surface, i.e. a mixture of the dissolution of the exposed metal matrix and film reformation. The crack advance will be maintained provided the film rupture process reoccurs because of the action of a strain rate at the crack tip. The model is associated with an evidence that SCC occurs most rapidly at potentials near the active to passive transition. The corrosion attack is then concentrated at some anodic points and under load the passive film over these points is easily broken. Then the strain at the tip of a fissure will continually to be broken or more likely will never be reformed. If the sides of this progressing crack remain passive, which they should since they are not strained, the corrosion will propagate rapidly through the sample and failure will eventually result. The strain will produce slip steps which will help rupture the passive film [1], [2].

2.8. The Tarnish – Rupture Model

In the original model, a brittle surface film forms on the metal, fracturing under the applied stress. Fracture of the film exposes bare metal, which rapidly reacts with the environment to reform the surface film. The crack propagates by alternating film growth and fracture. This hypothesis was later modified to explain intergranular SCC based on the assumption that the oxide film penetrated along the grain boundary ahead of the crack tip. Again, crack propagation consists of alternating periods of film growth and brittle-film fracture. Film growth requires transport of species across the film, and, as a result, the thickness of the film is limited in the absence of stress [18], [19].

2.9. Magnin Proposed The Dissolution /Absorption Enhanced Plasticity Model

That emphasized the role of corrosion-deformation interaction at crack tip [20], [21]. The following steps are included: (1) a localized anodic dissolution on the most dense slip planes at the very crack tip after depassivation, (2) an enhanced localized plasticity because of localized corrosion on the above planes; more mobile dislocations can be injected because of stress concentration due to localized dissolution and because of adsorption and absorption of hydrogen which can lower the critical shear stress, and blunting is then reduced, (3) emitted dislocations interact with obstacles such as precipitates, grain boundaries, or dislocations configurations previously formed when the stress is applied (e.g. Lomer-Cottrell locks); two zones are formed, an enhanced plasticity zone blocked by a previously hardened zone, and they induce formation of pile-ups where the local stress increases, (4) K_{1c} is locally reached for the obstacles strong enough, and is lowered by absorption of hydrogen, and then a crack's embryo forms by a kind of Stroh mechanism at the obstacle. The above models are more or less related to anodic dissolution. Up to late seventies only these models were considered as valuable for explaining the crack propagation in the 7xxx

series of aluminum alloys. Now the most popular has become another group of models which are strictly associate with hydrogen evolution and/or its further adsorption and absorption, directly resulting in the crack propagation from some stress raisers. These models (schematically shown in the Fig. 3) claim that SCC is the result of hydrogen-induced subcritical crack growth. claim that SCC is the result of hydrogen-induced subcritical crack growth. Because the anodic reaction must have a corresponding cathodic reaction, and the reduction of hydrogen is predominant cathodic reaction, hydrogen-induced subcritical crack growth can be responsible for subcritical crack growth in corrosive environments. Interactions between a discretized dislocation array and the crack tip under an applied stress produce a maximum stress ahead of the crack tip, to which hydrogen is driven under the stress behind the tip. When the hydrogen concentration reaches a critical value, a microcrack is nucleated because either the local cohesive strength is reduced, dislocation motion is blocked in the hydrogen enriched zone, or both. The microcrack arrests about 1 μ m ahead of the original location of the tip and these processes then repeat leading to discontinuous microcracking [20].

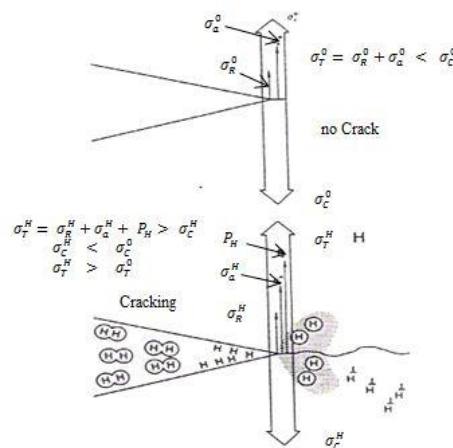


Fig. 3. Schematic representation of phenomenon of hydrogen.

2.10. The Pressure Model

Suggested that concentration of hydrogen in lattice could reach large supersaturations as a result of charging and further recombination of molecular hydrogen in some defects of microstructure (cavities, interfaces) [9]. The hydrogen would then recombine to form bubbles of hydrogen gas. These bubbles would put a strain on the lattice and help cause premature failure. Generally, such phenomenon may be expected at large hydrogen fugacities (large hydrogen pressures or cathodic polarization) or quenching of specimen. However, in many cases hydrogen cracking may occur at external hydrogen pressures well below those which would create supersaturations in metal [2].

2.11. The Early Hydride Model

Proposes that hydrogen forms a brittle hydride which can be easily cracked. Such a model may be applicable only in metals in which such hydrides may form and have been detected [2], [22]-[24]. Hydrogen embrittlement by stress-induced formation of hydrides (or high concentration solid solutions) is observed under conditions in which the hydrides can form at a rate sufficient to preclude other forms of failure. Another condition for hydride embrittlement is that the hydride be a brittle phase that undergoes cleavage fracture. Under an applied stress, the chemical potentials of the solute hydrogen and the hydride are reduced at tensile stress concentrations, such as crack tips. Diffusion of hydrogen to these elastic singularities and precipitation of hydrides then occurs. The phase change is accompanied by a decrease in the critical stress intensity for crack propagation because the hydrides are generally brittle phases. The

crack may propagate into the hydride, the formation of which is accompanied by a high, compressive local stress field, and that can give a rise to phase transformation toughening. However, the greatly decreased K_{IC} allows the crack propagation when the applied stress is only moderately increased, and the crack propagates by cleavage until the hydride-solution boundary is reached. At this point, the crack enters a ductile phase with a high K_{IC} and the crack stops until more hydride is formed [25] (see Fig. 4).

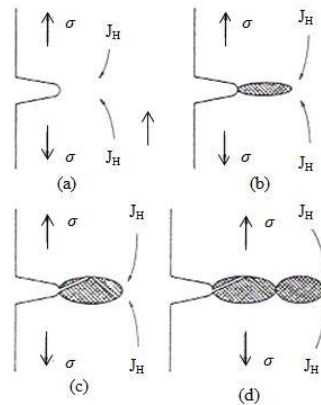


Fig. 4. Schematic representation of HE mechanism by formation of metallic hydrides [9].

2.12. Hydrogen – Induced Plasticity Models

Assume that hydrogen atoms weaken interatomic bonds at crack tips and thereby facilitate the injection of dislocations (alternate slip) from crack tips. Crack growth occurs by alternate slip at crack tips which promotes the coalescence of cracks with small voids nucleated just ahead of the cracks [20], [26]. The surface adsorption model is analogous to the Griffith's theory of brittle fracture (and some above mentioned models that yet take into account other than hydrogen environmental species). The idea is that hydrogen adsorbs onto the crack tip walls and lowers the surface energy, and hence the work of fracture [2].

2.13. The Mostly Acknowledged Hydrogen Enhanced Local Plasticity (or Localized Plasticity)

Model „HELP” is associated with an evidence that distribution of hydrogen can be highly non-uniform under an applied stress. Thus, locally the flow stress can be reduced, resulting in localized failure by ductile processes, while the total macroscopic deformation remains small [25]. Birnbaum thinks that the high local stress field at the tip of a crack reduces the chemical potential of solute hydrogen, and as a result of diffusion, the hydrogen concentration is locally increased. The resistance to dislocation motion and thus the flow stress is decreased by the presence of hydrogen and slip occurs at stresses well below those required for plastic deformation [26], [27]. The reason for the enhanced dislocation motion is not yet established. One mechanism is based on „elastic shielding” of dislocations by hydrogen atmospheres; the hydrogen atmospheres move together with dislocations and, as two dislocations approach each other, their hydrogen atmospheres reconfigure themselves that reduces the energy of entire system. Lynch proposed a little different, adsorption-induced localized-slip process [26]. The essence of the proposed mechanism for cleavage-like motion is that absorbed hydrogen atoms weaken interatomic bonds at crack tips and thereby facilitate injection of dislocations (alternate slip) from crack tips. Crack growth occurs by alternate slip at crack tips which promotes coalescence of cracks with small voids nucleated just ahead of cracks. The proposed mechanism for ductile cracking also involves nucleation of voids ahead of cracks, but coalescence of cracks with voids occurs by egress of dislocations from near-crack-tip sources. The adsorption mechanism is supported by results obtained at high velocities, effects of environmental conditions on spacing of brittle striations, no effect of dissolved hydrogen ahead of cracks growing in inert environments,

and no dissolution for sub-critical crack growth in moist air. The localized-slip mechanism is confirmed by evidence of large strains and extensive slip on planes intersecting crack fronts, small dimples covering fracture surfaces (at least for overaged specimens), and no exact parallelism of fracture and $\{100\}$ planes. The view of the model is shown in the Fig. 5.

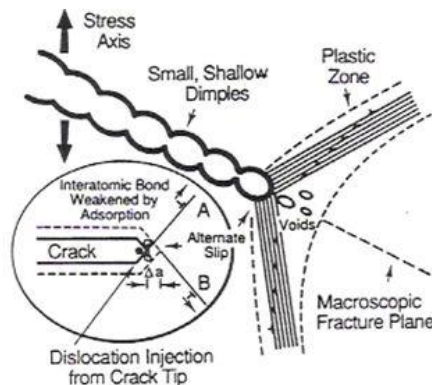


Fig. 5. View of adsorption enhanced developed by Lynch [26].

2.14. The Next Model Considered Is the Hydrogen Enhanced Decohesion

Mechanism that attributes the embrittlement to a decrease in the atomic bond strength resulting from the local concentration of hydrogen. Thus, the fracture is cleavage, which occurs when the applied stress exceeds the cohesive stress. In intergranular fracture, the relevant parameters are the cohesive energy and cohesive force of the grain boundary which are postulated to be decreased by the presence of hydrogen as well as the segregation of many other solutes [25], [26], [28]-[30] (see Fig. 6).

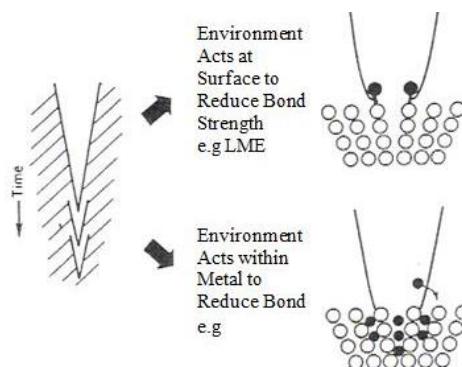


Fig. 6. The role of environmental species in leading to breaking metal bonds associated with brittle crack propagation [8].

3. Discussion

Some authors considered simultaneous action of both anodic dissolution and hydrogen embrittlement. Magnin [21] suggested the SCC mechanism of a 7150 Al-Zn-Mg-Cu alloy in NaCl solution be related to both anodic dissolution and hydrogen embrittlement, through for action of „critical effect”. On smooth samples, the role of anodic dissolution is to produce critical defects (pits or localized dissolution in slip bands) which favours hydrogen effects after localized H^+ reduction and plasticity. At slow strain rates, the relative influence of these mechanisms depends on electrochemical potential and on the temper of the alloy. Similar conclusion of concurrent action of hydrogen embrittlement and anodic dissolution, was recently given by Gerberich [31]. Lee [32], [33] mentioned that from the equilibrium aspect of crack propagation, a correlation between threshold stress intensity and temperature can be developed based upon the anodic dissolution model. He thinks that the mechanism for crack advancement of high strength aluminum alloys is

presumably determined by applied stress intensity and gradually moves from anodic dissolution in stage I to hydrogen embrittlement in stage II. Haag [34] reviewed some possible models and stated that in distilled water the hydrogen embrittlement is the most likely mechanism for cracking whereas in aqueous chloride environments the anodic dissolution of the grain boundary precipitates appears to be rate controlling. The other proposed models were rather seldom and included e.g. nucleation of microvoids [35] and crack blunting with hydrogen bubbles [36]. The main evidence in favor of the hydrogen models is supported by a lot of evidence (usually indirect). One of the most popular is an experiment of Gruhl [30] who observed an advancement in air of an artificial crack made outside the tube, inside which the NaCl solution was present. He claimed that hydrogen was evolved inside the tube, diffused through the walls and mobilized the crack. Another support was a discontinuous crack propagation [8], [37], [38] at a rate similar in gaseous and aqueous environments. SCC and HE are influenced by electrochemical potential. Gest and Troiano [37] studied the hydrogen permeation and observed its increase at both cathodic and anodic side of potentials so that SCC occurring at anodic polarization may be also associated with hydrogen absorption. Next evidence of hydrogen importance is an observed recovery of ductility. In a lot of papers [32], [37], [39]-[43] observed the recovery of ductility of specimens tested in laboratory air and attributed this effect either to desorption of hydrogen or its internal rearrangement at some traps, e.g. Cr-rich inclusions and at rather large particles, like η and T-phase precipitates in 7075 alloy, both of which may act as innocuous traps [42]. The partial only recovery after testing in aqueous solutions was explained by concomitant, irreversible nature of hydrogen trapping at the Al_2CuMg -type precipitates [44]. Holroyd and Hardie [45] suggested that recovery of plasticity after aging at room temperature and TG failure at the free corrosion potential support the hydrogen embrittlement as a major source of SCC at these conditions. All measurements made in WVSA were attributed to hydrogen entry and its further effect [46], [47]. Bond [27] investigated the effects of hydrogen in the 7050 and 7075 Al-Zn-Mg alloys and concluded that introduction of either dry or water-vapor-saturated hydrogen atmospheres during deformation increased the dislocation mobility and crack propagation rate in both alloys. In his opinion, solute hydrogen enhanced dislocation mobility and reduced flow stress, and promoted highly localized ductile fracture. Scamans [48] studied the AlZn6Mg3 alloy and observed the recovery of ductility of the commercial alloy and chromium containing alloy during storage in laboratory air at room temperature after previous pre-exposure to moist atmosphere. He concluded that the embrittlement was due to the deep penetration of atomic hydrogen which reduced the grain boundary cohesion, and similar effect must occur during the IG stress corrosion failure. Gest and Troiano [40] and Albrecht, Bernstein and Thompson [49] investigated a high strength Al-Zn-Mg-Cu alloy. They concluded that both the temperature dependence and strain rate dependencies were characteristic of a strain aging type of hydrogen embrittlement. Speidel [50] also suggested that hydrogen entry was responsible for the sub-critical crack growth rate and environmental-related failure of Al-Zn-Mg alloys. He concluded that both hydrogen embrittlement and stress corrosion cracking have similar features: the effects of heat treatment and microstructure on cracking, similar effects of strain rate and temperature. The microstructure of grain boundaries was discussed and said to be in favor of hydrogen models. Haynie and Boyd [51] suggested that hydrogen may be involved in the SCC mechanism by accelerating the localized attack at the grain boundaries. Scamans made high-resolution fractography of stress-corrosion cracks in AlZnMg and AlZnMgCu alloys and observed the discontinuous failure together with fracture morphology. He attributed the effects observed to the hydrogen-induced grain boundary separation, associated with tensile decohesion following the hydrogen penetration ahead of the crack tip or the absorption of hydrogen at the crack tip [22]. Gruhl claimed that at the grain boundaries a super concentration of hydrogen occurred as a result of their different microstructure [52]. Pathania and Tromans [53] after investigation of the AlZn3.6Mg2.6 in salt/chromate solution observed that pre-charging reduced the subsequent SCC initiation

times. The SCC path was intergranular indicating that hydrogen accumulated more easily in the boundary regions because of localized corrosion and hydrogen reaction in these areas. In their opinion, these results were entirely consistent with the lattice decohesion model developed by Oriani. The intensive search of hydride phases was carried out. Ciaraldi [24] observed the formation and rupture of a stress-induced hydride at the grain boundaries (AlH₃) for a high purity AlZn_{5.6}Mg_{2.6} alloy exposed to moist air. However, they did not detect such hydride for specimen failed by SCC and concluded that entirely different mechanism must be responsible for SCC in this alloy. The formation of magnesium hydride was also reported by Tuck during exposure of Al-Zn-Mg alloy to water vapor [23]. Chene [37] observed intergranular behavior following RT charging, limited to a zone few grains deep from the surface. This fracture appearance was considered in terms of possible formation of an aluminum- or magnesium-based hydride at room temperature. Polyanskii [35] reviewed the role of hydrogen embrittlement in the corrosion cracking. In this review he reported on some evidence of hydrogen contribution to SCC: (1) the effects of some „hydrogen poisons”, (2) reversible influence of hydrogen on mechanical properties, (3) increase in HE with decrease of a strain rate, (4) brittleness of external area where the concentration of hydrogen was greater than critical for embrittlement. In conclusion he suggested that hydrogen embrittlement was either the only or primary mechanism of SCC of Al alloys. Speidel [50] also suggested that hydrogen entry was responsible for the sub-critical crack growth rate and environmental-related failure of Al-Zn-Mg alloys. He concluded that both hydrogen embrittlement and stress corrosion cracking had similar features: the effects of heat treatment and microstructure on cracking, similar effects of strain rate and temperature. According to [54] the degree of corrosion fatigue damage was highly dependent on the amount of hydrogen at the surface. The observed fracture morphology was dependent mainly on both available hydrogen and cyclic stress level, and the hydrogen transport may be rate limiting. It was claimed that solid solution dissolved hydrogen interacted with mobile dislocations resulting in serrated yielding. Łunarska wrote that at anodic and free corrosion potential of the P/M Al-Zn-Mg alloy the local dissolution processes were more detrimental than those related to hydrogen embrittlement that occurred at cathodic potentials [55]. She said that the susceptibility to SCC was associated with propensity to pitting. Pickens and Christodolou [56] studied some P/M alloys and concluded that most of the crack advance in 7091 alloy was by hydrogen embrittlement. This conclusion was based at the observation of the greater crack velocity in Mode I than Mode III. They also claimed that good resistance to SCC was achieved because of trapping of hydrogen by large GB precipitates and by easier hydrogen recombination at some of them. The another evidence for hydrogen models to be valid include [37]: specific environment which exists at the pit which favors the hydrogen evolution, similar susceptibility of Al alloys to SCC and HE, similar effect of microstructure on susceptibility to SCC and HE, presence of hydrogen at the grain boundaries, similar fracture in water vapor saturated air (WVSA) and in NaCl solutions, possibility of accelerated strain-induced hydrogen transport, dependence of SCC on the mode of deformation, temperature dependence of crack velocity, strain rate relation of crack velocity, small effect of corrosion, effect of electrochemical polarization, existence of a SCC threshold stress, greater susceptibility of pickled specimens, and higher hydrogen concentration at the grain boundaries.

4. Conclusion

There is considerable disagreement in the literature on the mechanism of SCC in some aluminum alloys(7xxx). Some time ago the anodic dissolution models were the most popular, and at present hydrogen induced cracking is favoured by most researchers. Presumably, both mechanisms, electrochemical and hydrogen – associated ones, may contribute into cracking, depending on the mode and value of stress, alloy microstructure, and environmental condition. However, while these two mechanisms dominate the research literature, they are not the only proposed mechanism. Rupture of a brittle passive film and other

mechanisms are also viable alternatives. Thus, while the stress corrosion cracking phenomenon of some aluminum alloys (7xxx) is well documented, the actual mechanism is still disputed.

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